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ENVIRONMENTAL AND SERVICE EXPOSURE
EFFECTS ON THE FATIGUE AND
FRACTURE BEHAVIOUR OF
AIRCRAFT GAS TURBINE MATERIALS

by

RAJKUMAR THAMBURAJ, B.E., M.ENG.

A THESIS SUBMITTED TO THE
FACULTY OF GRADUATE STUDIES IN
PARTIAL FULFILLMENT OF THE REQUIREMENTS
FOR THE DEGREE OF
DOCTOR OF PHILOSOPHY

CARLETON UNIVERSITY
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submitted by R. Thamburaj, B.E., M.Eng., in partial
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VOLUME I

POSSIBILITIES OF LIFE EXTENSION FOR
CANADIAN FORCES AIRCRAFT
ENGINE COMPONENTS: J85-CAN40
COMPRESSOR AND TURBINE MATERIALS

Abstract

Volume 1

This section describes a detailed experimental and statistical analysis carried out to determine the risks and benefits of implementing a damage tolerance based lifing procedure for the General Electric J85-CAN40 engine. This is a 'post-facto' investigation, using components of this engine retired by the Canadian Forces and the data available regarding these used parts. The emphasis is on the 5th stage compressor disc, which is one of the most crack-prone components of this engine.

A number of procedures have been demonstrated by which the "safe life" and the "safe inspection interval" for this component can be calculated reliably. The possibility of improving the existing lifing procedure through a damage tolerance approach, and the limitations of the damage tolerance approach are discussed on the basis of experimental observations and information available from field experience.

Also described in this volume are the results of a study carried out to determine the feasibility of rejuvenating retired J85-CAN40 turbine wheels by a re-heat treatment procedure.

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Nomenclature

a	crack length
a_i	initial crack length
a_c	critical crack length
a_{insp}	apparent crack depth
A_1, A_p	constants in deterministic equation
B	specimen thickness
B_n	specimen thickness at sidegroove
C_1, C_p	random variables
C, n	material constants in Paris Law
D	cross head displacement
f	frequency (Hz)
$f(t_1)$	failure probability at time t_1
F, P	load (kN)
$F(t)$	cumulative failure probability
$h(t)$	Hazard Rate
H_p, h, H_w	DCB specimen dimensions
K, K_I	stress intensity factor
k_a	standard normal deviate
l	lower bound or origin of Weibull distribution
M	moment of inertia (m^4)
N	RPM
N_i	crack initiation life (hours)
N_d	life to dysfunction (hours)
N_f	life to failure
N_{pr}	crack propagation life

N_{PR}	residual life time
N_t	inspection time
O_i	order number of i^{th} failure in Weibull analysis
$P(F)$	component failure probability
PN_R	probable number of rejection sites in a component
PN_F	probable number of failure sites or indications in a component not rejected and causing failure
R	stress ratio
R_t	reliability function
R_y	cyclic plastic zone radius
S	standard deviation
SF	safety factor
t_f	failure time
ω_i	yearly flying rate
W	specimen width
z_i	total number of LCF cycles
β	Weibull slope or shape parameter
σ	applied stress

$\sigma_{min}, \sigma_{max}$	maximum and minimum stresses
σ_{ys}	yield strength
σ_{yc}	cyclic yield stress
$\Delta\sigma$	stress range
λ	geometry factor
μ	mean
η	characteristic life or scale parameter
δ	load line displacement

VOLUME 1

CHAPTER 1

LOW CYCLE FATIGUE, ENGINE STRUCTURAL INTEGRITY AND
THE RETIREMENT FOR CAUSE PHILOSOPHY

The most common failure mode in engine components such as compressor and turbine discs which are of interest to this study is low cycle fatigue. Fatigue behaviour is complex, and is affected not only by variables like loads, geometry, structural response, manufacture, etc., but also by material behaviour. The development of fatigue damage, i.e., the initiation and propagation of fatigue cracks, is further complicated by the variability in the response of materials to this complex total environment. Scatter in the fatigue performance of materials exposed to a given service environment is a well documented observation. The early, unpredicted appearance of fatigue damage as a detectable crack, followed by its propagation to cause failure if undetected, has a potentially detrimental effect on the function and safety of operational structures, not to mention the serious economic consequences.

While a number of empirical models have been developed¹ to predict fatigue crack initiation and propagation behaviours, the analysis techniques do not seem either as well developed or universally understood as the comparable static strength design procedures. However, despite the complexities of fatigue life prediction, some degree of damage tolerance built into the structure can provide the opportunity to safely extend its serviceability beyond the stage of fatigue crack initiation.²

According to such a damage tolerant lifing procedure, components would be inspected after a certain number of hours of service exposure and those that were cracked would be retired.^{1,2} The components that were found to be crack free would be returned to service for a predetermined period of time after which they would be reinspected. Once again, cracked components would be retired and uncracked parts would return to service. Much of the value of this approach depends upon calculating this "Return-To-Service Interval" which should assure that cracks which went undetected at prior inspection will not propagate to failure in service during the interval between inspections. When correctly implemented, this procedure, popularly known as "Retirement For Cause" represents a considerable improvement over the traditional procedure of retiring all critical components of a certain kind when, say, 1 in 1000 of these is found to have a detectable crack.^{3,4} The parts would be retired for having incurred quantifiable damage, rather than because an analytically predetermined minimum design life had been exceeded. Thus the primary goal of the Retirement For Cause (RFC) approach is safe utilization of inherent component life capacity.

Perhaps the three most important techniques upon which the RFC procedure is based are: (a) fracture mechanics calculations of crack propagation rates; (b) quantification of inspection uncertainty; (c) determination of critical crack sizes. The basic principles involved in these techniques are discussed as follows.

1.1 Fracture Mechanics Concepts

The stress field ahead of cracks that are present or which may develop in service can be characterized by the stress intensity factor, (K), which is proportional to the applied stress (σ) and the square root of crack size (a). That is,

$$K \propto \sigma\sqrt{\pi a} \quad (1.1)$$

Unstable fracture can occur when K equals the fracture toughness of the material (K_c). In the simplest cases, residual fracture strength (σ_f) for a specific crack size or critical crack size (a_c) for a specified stress may be determined from eqn. (1.1),

$$\begin{aligned} \sigma_f &= K_c / \sqrt{\pi a_c \lambda} \\ a_c &= (1/\pi \lambda) (K_c / \sigma_f)^2 \end{aligned} \quad (1.2)$$

provided that fracture occurs before net section yield. Fracture toughness (K_c) is a function of material properties, section size and operating environment, while λ is a factor that depends on the shape of the crack and structure and the gradient of stress.

When stable crack growth can occur by fatigue, the residual life or failure time (t_f) of a part containing an initial crack of size 'a' can be obtained by integrating the crack growth equation. In general, the rate of crack growth with respect to time is given by

$$da/dt = f(K) \quad (1.3a)$$

Often, an approximation by a simple power law is used to describe this functional relationship for a fatigue process, as given below:

$$da/dt = C(\Delta K)^n \quad (1.3b)$$

where the constants C and n depend upon loading mode, type of crack growth, material properties and environment. By rearranging equation (1.3b) and integrating, failure time can be calculated as follows:⁵

$$t_f = \int_0^{t_f} dt = \int_{a_i}^{a_c} \frac{da}{C(\Delta K)^n} \quad (1.4a)$$

$$= \frac{1}{C(\sigma\sqrt{\pi\lambda})^n} \int_{a_i}^{a_c} \frac{da}{a^{\frac{n}{2}}} \quad (1.4b)$$

$$= \frac{(n/2-1)^{-1}}{C(\sigma\sqrt{\pi\lambda})^n} \left[\left(\frac{1}{a_i}\right)^{\frac{n}{2}-1} - \left(\frac{1}{a_c}\right)^{\frac{n}{2}-1} \right] \quad (1.4c)$$

In principle, equations (1.1)-(1.4) can be used to determine the "crack propagation life" of a component based on known or postulated crack sizes. Unfortunately, this is complicated by various uncertainties, particularly those associated with non-destructive inspection.

1.2 Evaluation of Inspection Uncertainty

Inspection Uncertainty. The key element of an inspection method is the ability to discriminate between imperfections less than the allowable severity and the imperfections greater than the allowable severity. An ideal inspection will reject all material units containing imperfections above a selected severity and pass all material units containing only imperfections below the selected severity. Real inspections have a finite transition region (Fig. 1.1) between the imperfection severity which has a high probability of causing a rejection and the imperfection severity which has a high probability of passing the inspection.⁶ It is the extent of this transition region, called the inspection uncertainty, that must be reduced in order to improve an inspection method.

Basically, there are two types of error that are of concern in nondestructive inspections. Type I is the acceptance of defective material units; Type II is the rejection of sound material units. For any inspection method, the frequency of Type I errors can be reduced to any desired level by setting the rejection level lower. Unfortunately, this increases the frequency of Type II errors. It is the frequency of this second type of error which limits the practical sensitivity that can be achieved by an inspection. Only by reducing the inspection uncertainty can both the frequency of Type I and Type II errors be reduced simultaneously.

Quantitative Modelling.^{3,7} Perhaps the oldest method of describing NDE reliability is to quote a crack size, (a), for which there is a 90% probability of detection (POD) with a statistical confidence level of 95%, i.e., an "a - 90/95". Since this approach does not describe the NDE behaviour at crack sizes smaller (or larger) than "a - 90/95", it cannot be used for RFC decisions. Furthermore, determining this parameter from a finite amount of data results in so much variability that it is believed to be of little practical value in providing even a qualitative assessment of NDI reliability.^{3,7}

One of the most common approaches to quantitative modelling is the "(POD - a) relationship". Conventionally, data are presented as the fraction of cracks within a specific size range which were detected, i.e., the "POD", plotted against crack size, "a". Various algebraic models can then be used to describe this relationship using regression techniques to evaluate the model coefficients.

When data is available in copious quantities, this approach is acceptable, and is relatively easy to use. However, when there is a paucity of data, the scheme of grouping data according to crack size range becomes untenable.

An alternative approach to determining model coefficients is to use binary data analysis. POD data are treated as either 0 (component accepted) or 1 (component rejected). Any model describing the "POD vs. a " relationship can then be postulated and usually a two parameter model is adequate. The model coefficients are evaluated by determining coefficients which maximize the probability of obtaining the observed (POD - a) data pairs. Various functional forms of the "POD - a " model have been evaluated using this technique which is believed to have great promise.³

Under Retirement For Cause, the decision to return a part to service will be based not on how large a crack actually is, but on the size it appears to be. A second method for NDI modelling directly addresses this sizing problem. By plotting the indicated crack size on the y-axis vs. actual crack size on the x-axis, as shown in Fig. 1.2, the variability in this decision parameter becomes apparent. This procedure does not avoid the necessity of modelling; however, since some relation between apparent and actual crack sizes must be determined, often a linear relationship suffices.

A third method is to "normalize" the indicated crack size by dividing by actual crack size. The resulting non-dimensional representation can be illustrated by a probability plot as shown in

Fig. 1.3. Symbol sizes in this figure represent the relationship of actual crack sizes to each other. By sampling from this distribution, a "POD vs. a " curve may be generated.

Quite often, results from a set of inspection experiments are used to create a model, using which a Monte-Carlo simulation is performed. The simulation process permits repeating "experiments" under fixed and known conditions, and a large data base can be generated to determine the "(POD - a)" curve.

1.3 RFC Procedures

1.3.1 Deterministic Fracture Mechanics (DFM) Approach⁵

In the deterministic approach, a conservative assumption is made that very large flaws of size a_0 exist, where a_0 has some undefined small chance of going undetected by inspection or other procedures. NDI constraints define the maximum flaw size that could be missed, and load and stress analysis establish maximum steady and cyclic stresses in areas of concern. DFM analysis determines maximum crack propagation at the greatest cyclic and steady stresses from the largest initial flaw that might be present. The RFC parameters (i.e., inspection intervals, critical crack sizes, etc.) are then established by selecting an appropriate safety factor based on engineering judgement. The disadvantage of the DFM approach is that it is based on a "worst case" estimate. Since the maximum crack propagation rate and maximum initial flaw size are used for calculating the inspection interval, it may be overly conservative in many cases and not lead to effective utilization of component life.

1.3.2 Probabilistic Fracture Mechanics (PFM) Approach^{5,8}

Many satisfactory components, which might be rejected, might be saved by removing the arbitrary and unrealistic constraints imposed by worst case assumptions and allowing the model to account probabilistically for actual variation in critical parameters based on test or field experience. Various probabilistic structural analysis⁹ and probabilistic fracture mechanics^{8,10} representations are possible which calculate the probability of failure $P(F)$, rather than an exact life or strength, by quantifying the variability of key engineering materials and inspection parameters. One convenient representation⁵ is to write the component failure probability $P(F)$ in terms of several key conditional failure probabilities. That is,

$$P(F) = \text{Probability of Component Acceptance} \\ \times \text{Probability of a Non-Reject Defect Causing Failure}$$

Mathematically,

$$P(F) = \exp(-PN_R) [1 - \exp(-PN_F)] \quad (1.5)$$

where PN_F is the probable number of component failure sites or indications in a component not rejected and causing failure and PN_R is the probable number of rejection sites in a component. These probable numbers are calculated from

$$PN_R = \int_0^{\infty} pn(a) P\left(\frac{R}{a}, S\right) da \quad (1.6)$$

and

$$PN_F = \int_0^{\infty} pn(a) [1 - P\left(\frac{R}{a}, S\right)] P\left(\frac{F}{a}\right) da \quad (1.7)$$

where $pn(a) da$ represents the pre-inspection material quality or the probable number of defects of size $a \pm \frac{da}{2}$, $P\left(\frac{R}{a}, S\right)$ the rejection

probability of a component given that an actual defect size 'a' exists and that inspection rejection level is set at size S, and $P(\frac{F}{a})$ the probability of component failure from an existing defect of size 'a'. Thus, the failure probability depends strongly on three distributions. The contrast between deterministic and probabilistic engineering analysis is shown in Fig. 1.4. The probabilistic approach is illustrated further in Fig. 1.5.

Besuner and Tetelman⁵ have presented an example of the application of probabilistic fracture mechanics to the lifing of a cracked turbine rotor spindle. This example is summarized here as follows.

The residual lifetime N_{PR} of the rotor spindle is assumed to be related to three parameters through a fracture mechanics model. The parameters are (1) applied stress range; (2) the material crack growth rate parameter C and (3) initial area of the largest spindle defect projected on the fracture plane, A_i . Known or assumed variations in these parameters should be translated into variations in residual lifetimes and the residual lifetime distribution used to design cyclic lifetimes, inspection intervals and fatigue reliability estimation.

For simplicity, it is also assumed that all previously observed scatter in crack growth rate for a given ΔK can be simulated by random variation in C rather than by some complex, joint dependent variation of C and n. Thus, it may be stated that,

$$\log C = \bar{N}(\mu_c, S_c) \quad (1.8)$$

according to which, log C is a random variable with median and mean μ_c and standard deviation S_c .

It may be further assumed that the stress ($\Delta\sigma$) variations as well as the initial flaw size (A_1) distributions follow a log-normal distribution.

If the components are subjected to very similar usage, then a parameter y may be defined as,

$$y = N_{PR} = N_{PR}(C, \Delta\sigma, A_1) \quad (1.9)$$

To obtain an analytical solution for the distribution of fatigue life the following statistical theorem is used.

If $X_1 = N(\mu_1, S_1^2)$ (1.10)

$$X_2 = N(\mu_2, S_2^2) \quad (1.11)$$

and $Y = P_1 X_1 + P_2 X_2 + P_3$ (1.12)

where P_1 , P_2 and P_3 are constant, then

$$Y = N(\mu_y, S_y^2) \quad (1.13)$$

$$\mu_y = P_1 \mu_1 + P_2 \mu_2 + P_3 \quad (1.14)$$

and $S_y^2 = (P_1^2 S_1^2 + P_2^2 S_2^2)$ (1.15)

The deterministic fracture mechanics eqn. (1.4c) mentioned earlier may be written in terms of $A_1 (= \pi a_1^2)$ as,

$$N_{PR} = \frac{(A_1/\pi)(1/\sqrt{2})^{n/2}}{C \Delta\sigma^n (4/\pi)^{n/2} (n/2-1)!} \quad (1.16)$$

Then using the known numerical distribution parameters for C , $\Delta\sigma$ and A_1 and the deterministic fracture mechanics relationship of eqn. 1.16, the life distribution may be determined as,

$$\log N_{PR} = N(\mu_N, S_N^2) \quad (1.17)$$

where
$$\mu_N = \log B + p\mu_C + q\mu_\sigma + r\mu_A \quad (1.18)$$

and
$$S_N = (p^2 S_C^2 + q^2 S_\sigma^2 + r^2 S_A^2)^{1/2} \quad (1.19)$$

The number of cycles at a given failure probability a , is simply

$$N_{PR} = 10^{(\mu_N + k_a S_N)} \quad (1.20)$$

and k_a is the standard normal deviate (i.e., the number of standard deviations from μ_N) corresponding to a particular value " a " of the standard normal distribution function.

These relationships may be combined with Monte-Carlo simulation programmes which would obtain a life distribution according to random variations in $\Delta\sigma$, C and A_1 . The crack initiation life distribution can be obtained through a similar simulation procedure.⁸

Implementation of PFM requires extensive collection of statistical data to develop appropriate probability distributions for inspection, service loads, local stress concentrations and material crack propagation and fracture toughness properties. Large uncertainties in any input parameter necessitate using conservative bounds of its probability distribution, thereby reducing the reliable life extension possible from the RFC concept. However, the use of conservative bounds of a parameter's probability distribution, rather than of a parameter's value, may often safely allow significantly more life extension for the PFM approach than for the DFM approach.

1.3.3 Combined Analysis (CA)

Besuner et al.¹¹⁻¹⁵ have formulated a modified procedure which makes more direct use of past operating experience as reflected in

inspection information to establish the RFC strategy. This method fits or calibrates an incomplete engineering model against available in-service data.

For example, in order to calculate the life distribution for a population of turbine rotors, an RFC procedure based on PFM would utilize a conservative calculation of the distribution of crack propagation lives, N_p , to establish life extension allowables. PFM would require and rely upon a complete engineering model including, for example, an accurate stress analysis. CA can tolerate some error in the stress analysis if adequate substitute information is available, such as field inspection data. CA would utilize all available in-service data on propagation life N_p and initiation life N_i along with laboratory data on engineering models which relate N_i , N_p and failure life, N_f . CA procedures also continuously account for new service or test data for refitting and, if needed, reformulating the incomplete engineering models with the larger data set. Thus, errors in the model or in materials data are not as influential in CA/RFC life extension evaluations as they are in PFM/RFC approaches because CA provides continual calibration with actual experience. The basic approach of incorporating a "safety" or "fudge" factor in the design calculation to explain actual test or field experience is a common design approach. The CA approach simply provides a more formal and systematic basis for incorporating actual performance data into the life prediction. The CA approach is illustrated in Fig. 1.6.

Besuner et al.¹¹⁻¹⁵ have formulated a step-by-step procedure for performing combined analysis on turbine rotors and an example which is based on field inspection data is given below.

1. An analytical model is chosen to calculate an estimated failure life, N_f . A generalized, deterministic equation is used.

$$N_f = \frac{A_i}{\sigma^{B_i}} + A_p \log(\hat{a}_c/a_i) / \sigma^{B_p} \quad (1.21)$$

such that $\hat{a}_c = (K_c / (2.5\sigma))^3$, and where A_i , A_p , B_i and B_p are constants which represent the best analytical and laboratory evaluation of N_f by the analyst. The exponent 3 in this equation accounts for the stress decrease below the crack surface (an exponent of 2 would be correct for crack geometries subject to uniform stress) and the factor 2.5 is representative of a stress intensity factor analysis of the rim crack.

2. A rotor disc is inspected at time N_t and the largest value of the apparent crack depth $\hat{a}_{insp}$ measured. $\hat{a}_{insp}$ is derived from the calibrated inspection signal given by

$$\log(\hat{a}_{insp}/a_t) = G_{au}(\mu, S) \quad (1.22)$$

where G_{au} refers to a normal distribution with parameters μ , the mean, and S , the standard deviation. The value of b taken to be either 0.1 or 0.2 reflects positive bias in the inspection. For example, $\mu = 0.1$ implies that, typically, the crack size at time, t (a_t) will be overestimated by a factor of $10^{0.1} = 1.26$. Also, S is taken to be either 0.2 or 0.4 and is the logarithmic standard deviation of $\hat{a}_{insp}/a_t$ and reflects inspection uncertainty in a similar manner.

3. Substituting the value of $\hat{a}_{insp}$ in the deterministic equation (1.21) for life,

$$N_t = A_i / \sigma^{B_i} + A_p \log(\hat{a}/a_i) / \sigma^{B_i} \quad (1.23)$$

the value of σ may be solved for.

4. Using the same equation and the computed value of σ , the life N_f which is the predicted life to failure (when the crack size reaches the critical limit a_c) is calculated.

5. A life extension may then be calculated as

$$N_c = \frac{(N_f - N_i)}{SF} \quad (1.24)$$

where, SF is a safety factor used to compensate for the fact that N_f and N_i are estimates.

The RFC procedure mentioned above may be programmed using Monte-Carlo simulation techniques. Improvements over this simple approach may be made by treating the constants A_i and A_p in the deterministic equation (1.21), as random variables C_i and C_p which simulate the variation of crack initiation and crack growth respectively, at a given σ . Variations in material condition, such as surface roughness or local hardness and composition can give rise to these random variations.

The cumulative probability distribution of C_i and C_p are assumed to be log-normal so that $\log C_i$ and $\log C_p$ are Gaussian or normally distributed variables. Similarly, σ and K_c may be assumed to be normally distributed variables and these aspects are then programmed into the Monte-Carlo simulation programme.

Thus, there are basically three approaches to accurately assess structural reliability. The deterministic, "worst case" assumptions are often both technically unrealistic and economically unacceptable. The probabilistic (PFM) approach does not necessarily

make the component more or less reliable; what it does do is to quantify the impact of uncertainties that actually exist so that there is a quantitative basis for engineering decisions. When no field data exists, PFM can be used to quantify the structural reliability. The PFM approach can be employed with any amount of field data, but if insufficient materials and inspection performance data exist or inaccurate analyses are utilized, the variability in key parameters may be so large that the overall reliability results are unacceptable or inconclusive. The third approach, called combined analysis, formally combines only the field data and engineering analyses most credible to the analyst to quantify structural integrity when neither the data base nor the engineering (PFM) approaches alone can do so.

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Chapter 1

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Fig. 1.1

Comparison of Inspection Uncertainty, Characterized by the Probability of Rejection as a Function of Flaw Size for Ideal and Real Inspections⁵

Fig. 1.2

Actual Crack Size vs. Indicated Crack Size³

Fig. 1.3

Probability Plot Showing Ratio of Indicated to Actual Crack Size vs. Cumulative Percent. Symbol Sizes are Proportional to Actual Crack Size³

DETERMINISTIC ENGINEERING ANALYSIS

"TYPICAL" ANALYSIS INDICATES ADEQUATE SAFETY MARGIN

"WORST-CASE" ANALYSIS INDICATES NEGATIVE SAFETY MARGIN OR FAILURE

t = "TYPICAL"

w = "WORST-CASE"

PROBABILISTIC ENGINEERING ANALYSIS

ESTIMATES FAILURE PROBABILITY

Fig. 1.4

Contrast Between Deterministic and Probabilistic Engineering Analysis⁸

Fig. 1.5

Statistical Variation of Strength (σ_f) and Applied Stress (σ): Failure Possible Over the Crosshatched Region⁸

SURVEY RELEVANT ROTOR POPULATION AND FAILURE HISTORY,
RECORD PRESENT AGES, RETIREMENT AGES, USAGE (DUTY
CYCLE) PARAMETERS, AVAILABILITY OF SPARE ROTORS.

IDENTIFY KEY FAILURE MODE (CREEP AND/OR FATIGUE),
CAUSE (BORE INCLUSION/MICROSTRUCTURE), AND PRIMARY
ENGINEERING PARAMETERS (STRESS, TEMPERATURE, ETC.).

FORMULATE A TIME-TO-FAILURE RANDOM
VARIABLE PROBABILITY DISTRIBUTION MODEL
 $P(\text{FAILURE BEFORE } N_E \text{ YEARS}) = F(N_E) = 1 - \exp(-N/\beta)^\alpha$
(THE WEIBULL MODEL)

ESTIMATE FATIGUE LIFE SCATTER (α) FROM STATISTICAL
ANALYSIS OF LAB DATA, COLLECTED FLAW MEASUREMENTS,
DUTY CYCLE VARIATIONS, ETC., THERE ARE NOT ENOUGH
IN-SERVICE FAILURES TO ESTIMATE α FROM FAILED DATA.

PERFORM DAMAGE ANALYSIS OF ROTOR POPULATION TO EXPRESS β ,
THE CHARACTERISTIC (TYPICAL) LIFE IN TERMS OF PRIMARY
ENGINEERING PARAMETERS AND THEN TO ESTIMATE β FOR EACH ROTOR.

CALCULATE THE CONDITIONAL FAILURE PROBABILITIES AND RESULTING
EXPECTED COSTS ASSOCIATED WITH VARIOUS OPTIONS FOR EACH
ROTOR, IMPLEMENT THE MINIMUM COST OPTION.

Fig. 1.6

Illustration of Combined Analysis Approach¹¹⁻¹⁵

CHAPTER 2

LIFING PHILOSOPHY OF J85-CAN40 ENGINES

2.1 INTRODUCTION

Gas turbine engine manufacturers and users alike have been debating the possibility of using damage tolerance concepts since the late 70's^{1,2}. These concepts are being considered for use both at the design stage for stress analysis and lifing procedures, and to establish intervals for inspection and repair during service.

In 1982, the Structures and Materials Laboratory of NAE-NRC decided to embark on a long term research programme with a view to assessing and evaluating the feasibility of using damage tolerance based maintenance methodologies in aircraft turbine engine components.

The primary aim of this programme was to develop an understanding of the concepts involved and to evaluate any risks associated with these maintenance methodologies. Approximately at the same time, the Laboratory became aware of the fact that the Canadian Forces (CF) in co-operation with its maintenance contractor (Orenda Division of Hawker Siddeley Canada Inc.) were retiring a large number of compressor discs, spacers and turbine discs from J85-CAN40 engines. This was seen as an ideal opportunity to investigate options for the life extension of these parts at a laboratory level, using fracture mechanics based damage tolerance concepts. Upon examining the life cycle management history of these retired components in some detail, it became evident that CF, through manufacturer's (General Electric) recommendations had

used a damage tolerance approach in establishing the life limits for the compressor discs.⁴ Components were used well beyond their original safe life limits and a large number of compressor discs and spacers contained cracks at retirement. Cracks as long as 19 mm were observed in some of the retired components. It was also clear from the limited information available that the engine manufacturer had employed probabilistic techniques to define safe inspection intervals for the maintenance of these engines.

Many mathematical analogs have been devised to model distributions of probabilistic parameters in real life. Some of the better known distribution models include the normal distribution, the binominal distribution, the Poisson distribution and the gamma distribution. According to General Electric (G.E.) the operational fatigue failure data of aircraft engine components such as compressor and turbine discs is best described by a Weibull distribution function.⁵ An appropriate analysis can provide useful information with regard to assessing maintenance schedules, hardware support costs and logistical supply systems, etc.⁵ The Weibull distribution is a cumulative probability distribution which provides a simple graphical tool to evaluate the above mentioned problem areas. Weibull analysis was also extensively used by G.E. to determine the maintenance schedules for J85-CAN-40 engines although the details of these analyses were not given. It is the purpose of this study to understand and apply Weibull analysis methods to establish life limits and safe inspection interval for J85-CAN40 engines.

2.2 .Historical Background of J84-CAN40 Case Study

Traditionally, North American engine manufacturers such as General Electric (G.E.) and Pratt and Whitney establish safe life limits of rotating components by the number of low cycle fatigue (LCF) cycles to crack initiation in one in one thousand of these components. (-3 σ strength). A safe life approach implies that when a set of discs and spacers reaches the pre-determined LCF life limit, they are automatically discarded, regardless of whether a crack has developed or not. In the case of J85-CAN40 engines, however, systematic NDI inspection periods were employed to separate the cracked parts from uncracked parts and uncracked components were re-installed for further service.⁵ Application of this type of maintenance methodology suggests that G.E. had departed from its traditional safe life approach and instead, used a damage tolerance approach in J85-CAN40 engines. Prior to outlining the details of the damage tolerance approach used, it is important to consider the reasons why G.E. deviated from its traditional lifing philosophy in the case of J85-CAN40 engines.

The J85 turbine was first designed by G.E. in the 1950s for missile applications, with a 15 hour design life limit and was later modified for aircraft engine application. By 1982 there were 13,000 J85 class turbines in service in various military aircraft in 29 countries around the world.⁶ The United States Air Force (USAF) uses the J85-5 model in their T-38 trainer aircraft, whereas in Canada the J85-CAN40 model is used in the CF114 Tutor (a trainer aircraft) and an afterburner version (J85-CAN15) in the CF-5 aircraft.

J85-CAN40 turbines were first purchased by CF in the 1960's. At the time of the purchase, no safe limits were specified by the engine manufacturer for the different rotating components. The safe life limits were eventually made available by G.E. in 1970. These safe lives are believed to have been determined by G.E. in a probabilistic manner, based on their experience with the USAF J85-5 model. A safe limit of 3,000 hours was assigned to the fifth stage compressor disc in both J85-5 and J85-CAN40 engines in 1970, Table 2.1.

In 1973, Orenda first observed cracks in a fifth stage spacer and by 1976-77 cracking was as a major problem with fifth - seventh stage compressor discs, and the spacers attached on either side of these discs. Cracks in these components were developing much earlier than their specified safe life limits of 1970. In 1978, G.E. completed an engineering programme to re-evaluate the recommended compressor and turbine rotor LCF limits for J85-CAN40 engines. The study was performed by evaluating the Tutor aircraft mission description provided by CF and comparing it to various missions flown by USAF for which there was a broad data base available. The precise details of the mission description supplied by CF to G.E. prior to 1978 remain uncertain at this stage. However, G.E. analysis of this unspecified data base suggested that the J85-CAN40 mission was less severe than the mission assumed for baseline comparison (USAF J85-5), Table 2.1. The safe lives of all J85-CAN40 components were then factored upwards by 13%, Table 2.1. Between June 1978 and September 1979, G.E. continuously revised the LCF limits of USAF J85-5 engine components as may be seen from Table 2.1.

In November 1981, G.E. suggested a major reduction in the safe life limits of J85-5 engine components. For example, in a fifth stage compressor disc, safe life was reduced from 4,400 hours (September 1979) to 3,200 hours (November 1981), Table 2.1. In addition, between 1979 and 1981 G.E. determined that the CF Tutor mission was actually 1.8 times more severe than the USAF T-38 mission, which reduced the safe life limit of J85-CAN40 engines even further, Table 2.1. It appears that the severity of the CF Tutor mission was deduced from strip chart flight recordings supplied by CF to G.E. The details of the analytical methods used by G.E. to develop suitable templates for interpreting the CF recorded data and deriving a typical J85-CAN40 mission profile remain inaccessible. The severity of a mission is apparently related to the number of throttle excursions over an average mission flight time. In conclusion, the J85-CAN40 safe life problem arose due to an initial overestimation of J85-5 safe life limits and an underestimation of J85-CAN40 mission severity relative to J85-5 engines. In November 1981, the operating safe lives of J85-CAN40 compressor and rotor assemblies were thus dramatically reduced, e.g., the safe life of a fifth stage compressor disc was reduced from 4680 to 1777 hours, Table 2.1.

In November 1981, a large number of J85-CAN40 components had either exceeded or were approaching the revised safe life limits. At this juncture, CF had either the option of grounding their fleet or replacing the life expired components with new parts. Replacements were generally expected to arrive from the G.E. manufacturing facility

approximately 20 months after placing orders. Obviously, neither of these options was feasible to CF since their primary objective was to keep the fleet operational. Therefore, it appears that between September 1979 and November 1981, G.E. had considered the use of fracture mechanics concepts for component life cycle management in order to systematically phase out components affected by cracking. G.E. then supplied information regarding fracture critical locations, critical crack lengths and safe inspection intervals of various components in J85-5 engines. In the case of J85-CAN40, the J85-5 data base was simply modified by considering appropriate factors such as severity of CF mission profile and average time of a typical CF flight. G.E. also carried out a limited amount of fracture mechanics and spin pit testing on a few selected service exposed components in order to verify their predictions regarding safe inspection intervals for J85-CAN40 engines. This type of maintenance methodology obviously gave the engine user some time in between ordering, manufacturing and installing the new components.

The purpose of this project was to establish a technology base within Canada that could be tapped, in the event that similar problems are experienced in the future. In developing this, a major effort and an in-depth engineering analysis of a variety of variables was required and these details can be found elsewhere.⁷ The J85-CAN40 engine is being treated as a work-horse in developing this technology base since a large number of retired components are available for analysis. It should be noted that since July 1984, Orenda has been

replacing the service exposed components of this engine with new parts having improved LCF configuration. The type of maintenance methodology mentioned earlier is not in use anymore, and critical components are retired immediately upon expiry of the safe life limits.

2.3 Statistical Principles^{8,9}

Many of the recommendations made by G.E. were based on statistical techniques which were used to determine the probabilities of failure in critical components. Therefore, prior to discussing these recommendations, the fundamental definitions and concepts of statistical failure analysis are explained below.

Life Distributions. If (t) were to represent the random variable equal to the component failure time during service, a plot between specific time intervals and the number of components (i) failed within that interval will yield a failure distribution curve. Figure 2.1 shows a typical failure distribution curve for turbine engine discs, where the failure criterion is defined as time to crack initiation during service. Providing the overall number of components (N) , operating in service is known, a failure distribution curve can be replotted in the form of a probability distribution or a probability density curve, where the failure probability $f(t_i)$ within a time interval of $t_i - 1$ and t_i is defined as,

$$f(t_i) = \frac{\text{no. of components } (n_i) \text{ failed}}{\text{Total number of components } (N)} \quad (2.1)$$

The shape of the probability density curve will be similar to that of the failure distribution curve. A probability density curve can also be represented algebraically as a function of time, $n(t)$. Through an integration procedure, an algebraic function can further be used to determine the cumulative failure probability, $F(t)$, where

$$F(t) = \sum f(t_i) \quad (2.2)$$

or

$$F(t) = \int_0^t \phi(t) dt \quad (2.3)$$

and the maximum value of $F(t)$ equals unity. In practice, probability density curves for various types of failures can be described by different algebraic functions that form the basis of differentiating between a number of probability distributions generally referred to in the literature.

Apart from failure probabilities, industrial applications of statistics also deal with the problem of reliability of in-service components within a specified time interval. Reliability, $R(t)$, is defined as the probability that a component performs properly for the period of time intended, where

$$R(t) = 1 - F(t) \quad (2.4)$$

The reliability function thus indicates the survival probability of a component at any point in time. To expand the role of the reliability function in failure analysis, the concept of the hazard rate, $h(t)$, has further been invoked for industrial application. The hazard rate, $h(t)$, is the probability a component will fail in the next instance of time divided by its reliability at that instant, where

$$h(t) = \frac{\phi(t)}{1 - F(t)} = \frac{\phi(t)}{R(t)} \quad (2.5)$$

The hazard rate thus gives an indication of the instantaneous failure rate of components in use. To explain, consider the case of turbine engine discs manufactured from inclusion-infested and inclusion-free heats of an alloy. Assume that a disc from each heat has identical values for $\phi(t)$ at this very moment, i.e., both discs have the same probability of failing in the next instant. The instantaneous failure rate or the hazard rate, $h(t)$, will be greater for the inclusion-infested disc since its $R(t) = 1 - F(t)$ will be much less than that of the inclusion-free disc. The hazard rate thus considers the effectiveness of the reliability or earlier cumulative failure probability in a given set of components.

The choice of an appropriate life distribution usually depends on what the engineering expert chooses for the hazard rate, $h(t)$, of the components. It should be noted that the hazard function, $h(t)$, is treated as a constant in a number of probability density functions such as normal (Gaussian), exponential, etc. In practice, however, $h(t)$ may either rise or fall over a period of time, depending upon the type of component and its environs. The uniqueness of a Weibull distribution function lies in its ability to account for such a change in the hazard function over a period of time.

Weibull Distribution.^{8,9} In a Weibull distribution, the hazard rate is described by a simple power law function,

$$h(t) = (\beta n - 1) \left[\left(\frac{t - 1}{n - 1} \right) \right]^{\beta - 1} \quad (2.6)$$

and the cumulative failure probability distribution is expressed as,

$$F(t) = 1 - \exp \left[- \left(\frac{t-l}{n} \right)^\beta \right] \quad (2.7)$$

where,

t = age of component in time of cycles

l = lower bound or origin of distribution in time or cycles

n = characteristic life or a scale parameter or failure life in time or cycles at 63.2% cumulative failure probability

β = Weibull slope or shape parameter

and

$F(t)$ = cumulative failure probability.

Notice that in equation (2.6) for $\beta > 1$ the hazard function will rise and for $\beta < 1$ the hazard function will decline as time increases. In the special case, where $l = 0$, equation (2.7) reduces to,

$$F(t) = 1 - \exp \left[- \left(\frac{t}{n} \right)^\beta \right] \quad (2.8)$$

Upon taking logarithms twice on both sides of equation (2.8), we get,

$$\ln \ln \frac{1}{1 - F(t)} = \beta \ln t - \beta \ln n \quad (2.9)$$

By transforming the variables such that

$$Y = \ln \ln \frac{1}{1 - F(t)} \quad \text{and} \quad Z = \ln t$$

a linear relationship of the following form results

$$Y = \beta Z - \beta \ln n \quad (2.10)$$

In practice β and n are the parameters to be determined from the experimental or the field data base. However, prior to drawing any concrete conclusions regarding the lifing philosophy of a given component, it is customary to ensure that the experimental or the field lifing data fits well with the Weibull distribution function. The slope of the line, β ,

in equation (2.10) is particularly significant, since it can provide a clue to the physics of the failure in question. The relationship between various values of the slope and typical failure modes is shown in Fig. 2.2.⁴ These differences in slope can be extended to identify mixtures of failures, lines or modes.¹⁰ Another advantage of Weibull analysis is that it may be useful even with small data bases. The experience of turbine engine manufacturers suggests that the Weibull distribution most frequently provides the best fit of the type of data experienced in the gas turbine industry.¹⁰

Weibull Analysis. Suppose that N discs are placed on fatigue life test and the observed failure times of k discs out of N are recorded as $t_{(1)}, t_{(2)}, t_{(3)}, \dots, t_{(k)}$, where $k < N$. In order to ensure that a Weibull distribution is applicable to the data base collected, an estimate of the proportion of the cumulative failure probability $F(t_i)$ as a function of time t_i is first carried out. A number of schemes for estimating $F(t_i)$ are in common use,^{8,9} where

$$F(t_i) = [(R_i - \frac{1}{2})/N] \times 100$$

or $F(t_i) = [R_i/(N+1)] \times 100$ (mean ranks)

or $F(t_i) = [(r_i - 0.3)/(N + 0.4)] \times 100$ (median ranks).

and R_i is the rank number obtained by ranking the failed samples from low to high failure times and $R_i \text{ max.} = k$. Substituting $F(t_i)$ for $F(t)$ in equation (2.9) such that $Y_i = \ln \ln \frac{1}{1 - F(t_i)}$ and plotting Y_i v/s $\ln t_i$, a series of points distributed along a straight line will be obtained when the times to failure approximate a Weibull distribution.

Having established that a Weibull distribution describes the component failure life data, the next step involves the estimation of component design life at -2σ or -3σ probability of failure. Turbine engine manufacturers stress that all data, i.e. failure and non failure, must be taken into account while performing the Weibull analysis for design purposes.²⁻⁴ Therefore, life times of all N components in our test example need to be recorded for further analysis. This procedure alters the rank number R_i as will be seen later. The first step of analysis requires that all lifing data, failure and non failure, be arranged in an ascending order of Time Since New (TSN). Each component is assigned an order number O_i and it is important that this step be completed correctly, since incorrect order numbers will invalidate the following analysis. The data can now be condensed by considering the order numbers of the failed units alone for the rest of the analysis.

General Electric has formulated a simple but stepwise workable format for conducting the Weibull analysis in accordance with equation (2.8).¹¹ The rank number R_i for each failed component is calculated by using the expression,

$$R_i = \frac{N + 1 - R_i - 1}{N + 2 - O_i} + R_i - 1 \quad (2.11)$$

where $R_0 = 0$ and O_i = order number of the i th failure. The remaining step involves the calculation of cumulative percentage failure probability. (% F), where

$$\%F_i = A R_i - B \quad (2.12)$$

$$A = \frac{100}{N-1} \left[2 \left(1 - \frac{1}{N} \right) - 1 \right],$$

$$B = 100 \left[\frac{2 \left(1 - \frac{1}{N} \right) - 1}{N-1} + 2 \left(-\frac{1}{N} \right) - 1 \right]$$

and A and B are both constants. In the final analysis, the percentage F_i , for each R_i , versus TSN is plotted on a Weibull probability paper and 0.135% F (Probability of failure of 1 in 741 components) is graphically determined. It should be noted when $O_i = R_i$ in equations (11) and (12), the analytical procedure becomes similar to that outlined in statistics text books.^{8,9} This procedure is shown in detail in Appendix C.

The Weibull plot should be inspected to determine how well the failure data fits the straight line. The scatter should be evenly distributed about the line. However, sometimes the failure points will not fall along a straight line on the Weibull plot, and modification of the simple Weibull approach may be required. The bad fit may relate to the physics of the failure or to the quality of the data. There are at least two reasons why a bad fit may occur. First, the origin - if the points fall on gentle curves, it may be that the origin of the age scale is not located at zero (Fig. 2.3). There may also be physical reasons why this will be true. A procedure for determining the origin correction has been given by Abernethy et al.¹⁰

Secondly, a mixture of failure modes can result in a plot of failure points showing cusps in sharp corners. This is an indication that there is more than one failure mode, i.e. a mixture of failure

modes (Fig. 2.4). In this case, it is necessary to conduct a laboratory failure analysis of each failure to determine if separate failure modes are present. If this is found to be the case, then separate Weibull plots should be made for each set of data for each failure mode. If the laboratory analysis successfully categorised the failures into separate failure modes, the separate Weibull plots will show separate straight line fits, that is, very little data scatter.

2.4 Lifing Method

In the case of the compressor discs and spacers, the lifing procedure was based on the probability of formation of cracks of a detectable size at tie bolt holes. This procedure should be understood from the point of view of the following definitions.

Crack Initiation Criterion. Though not precisely stated in the documentation from G.E., this is generally taken to be the condition where a crack is barely detectable by state-of-the-art NDI techniques under ideal testing conditions. For the components of interest, this crack size limit is about 0.8 mm. The probability of crack initiation can be represented by a Weibull plot as shown in Fig. 2.5 (line N_1). The initiation life is determined by the 0.135% cumulative failure probability criterion. For example, the predicted initiation life (1981) for the 5th stage compressor disc of the J85-CAN40 engine is 822 h.

Dysfunction Criterion. Once a crack has initiated, it will propagate during continued engine operation at a relatively steady

rate. After a certain crack length, however, the level of vibration becomes high and the rate of propagation may become very rapid with danger of catastrophic failure occurring during service. This critical crack size is referred to as the dysfunction limit. The dysfunction limits were apparently derived by fracture mechanics testing, vibration analysis and field experience. The values given for various compressor components are listed in Table 2.2.

Safe Life: The safe life in the case of the J85-CAN40 compressor components has been based on a dysfunction criterion. The life to dysfunction was determined in a probabilistic manner similar to the initiation life. The probability of reaching dysfunction in a certain number of service hours can be represented by the N_d line in Fig. 2.5. For example, in the case of the fifth stage compressor disc of the J85-CAN40 engine, the predicted (1981) dysfunction life is 1777 h, (Fig. 2.5).

Crack Propagation Interval. This is believed to be related to the number of service hours required to cause propagation of a barely detectable crack to a size that can cause dysfunction in 1 in 741 cases. The precise definition of this parameter has been provided by G.E. However, it was clearly stated by G.E. that though components are not to be used beyond their specified safe life limit, under circumstances such as a decrease in component life limits, parts which are known to be crack free at prior inspection may be operated for a maximum of one propagation interval beyond that inspection with minimum additional risk. Cracked parts were never to be reinstalled. Crack

propagation intervals specified by G.E. for the various components of the J85-CAN40 engine are given in Table 2.3.

Safe Inspection Interval. This corresponds to the number of in-service hours after which the components of the engine are inspected for cracks using NDI techniques. A crack propagation interval may be divided by a safety factor to determine a safe inspection interval. In the case of the J85-CAN40 engine, an inspection interval of 1600-1800 h was used prior to the specification of a crack propagation interval.

An interval close to the crack propagation interval was used after this stage. It appears that "Periodic Evaluations" were also carried out between overhaul periods but these did not involve rigorous NDI evaluation.

Mission Severity Factor (MSF). The MSF is related to parameters such as the duration of the flight and the number of throttle excursions. In 1981, a factor of 1.8 was termed as the mission severity factor for the J85-CAN40 engine relative to J85-5 engines. In 1983, however, G.E. revised their estimate of the mission severity factor since N_d (observed) line was found to lie to the left of the N_d (predicted) line of 1981, (Fig. 2.6). The numerical value of the MSF was assessed to be 1.4 in 1983, thus leading to the revision of the safe lives as shown in Tables 2.1 and 2.3.

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Chapter 2

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Table 2.1

Safe Lives and Mission Severity Factors Supplied by G.E. at
Various Dates for the 5th Stage Compressor Disc

Date	Safe Life J85-5 Hours	Mission Severity Factor	Safe Life CF Hours
May-1970	3,000	1.00	3,000
June 1978	4,500	0.94	4,787
Sept. 1, 1979	4,000	0.94	4,255
Sept. 17, 1979	4,400	0.94	4,680
Nov. 24, 1981	3,200	1.80	1,777
Oct. 12, 1983	3,200	1.40	2,285

Table 2.2

Dysfunction Limits for J85-CAN40 Compressor and Turbine Discs

Component	Dysfunction Definition
4th Stage Compressor Disc	4.5 mm Bolt Hole Crack Radially Inwards
5th Stage Compressor Disc	
6th Stage Compressor Disc	
7th Stage Compressor Disc	5.08 mm Bolt Hole Crack Radially Inwards
8th Stage Compressor Disc	
1st Stage Turbine Disc	Flange Bolt Hole Crack Initially Radially Outward Subsequent to Radially Inward Cracking
2nd Stage Turbine Disc	

Table 2.3

Predicted Safe Lives and Crack Propagation Intervals for J85-5 and
J85-CAN40 Compressor Discs by G.E.

Component	J85-5 (1981-1983)		J85-CAN40 (1981-1983)			
	Safe Life	CPI	1981	1981	1983	1983
	- Hrs.	Hrs.	S.L. Hrs.	CPI Hrs.	S.L. Hrs.	CPI Hrs.
4th Disc	4,200	1,000	2,333	555	3,000	714
5th Disc	3,200	1,000	1,777	555	2,285	714
6th Disc	4,000	1,300	2,222	722	2,857	928
7th Disc	4,000	1,300	2,222	722	2,857	928
8th Disc	5,300	1,300	2,944	722	3,786	928

MATERIAL DATA SCATTER RESULTS
IN CONSERVATIVE LIFE PREDICTION

Fig. 2.1
Distribution of Life to Initiation of Turbine Discs
Under LCF Conditions^{1,2}

1. INFANT MORTALITY

- INADEQUATE BURN-IN, GREEN RUN
- MISASSEMBLY
- SOME QUALITY PROBLEMS

2. RANDOM FAILURES

- INDEPENDENT OF TIME
- MAINTENANCE ERRORS
- ELECTRONICS

3. EARLY WEAROUT

• SURPRISE !

- LOW CYCLIC FATIGUE

4. OLD AGE WEAROUT (RAPID)

- BEARINGS
- CORROSION

Fig. 2.2

Understanding the Physics of Failures Using Weibull Analysis⁵

Fig. 2.3

Bad Fit due to Origin not Located at Zero of the Age Scale¹⁰

Fig. 2.4

Bad Fit due to Mixture of Failure Modes¹⁰

Fig. 2.5

N_1 and N_d lines (1981) for J85-5 and

for CAN-40 with Store Compressor Discs

Fig. 2.6

Revision of Mission Severity Factors

CHAPTER 3

ANALYSIS OF RETIRED J85-CAN40 ENGINE COMPONENTS

While various procedures and restrictions were laid down by G.E. towards the maintenance of the J85-CAN40 engine, a review of service life data on retired components revealed immediately that the user's maintenance techniques had deviated significantly from the specified instructions. This data, summarized in Tables 3.1, 3.2 and 3.3 led in 1982 to the following important observations.

- The compressor and turbine discs had been used well beyond their safe life limits, and up to three times the specified limit in some cases.
- While a maximum of one crack propagation interval in excess of the safe life was specified, the actual usage appears to have extended over several repeated intervals beyond the safe life.

Visual observations of the various retired parts led to various conclusions as well, which are stated below:

- Cracking was most severe in bolt holes of the 5th and 7th stage compressor discs and in spacers on either side of them. The maximum crack sizes indicated in these components is shown in Table 3.2. Cracks as long as 19 mm were found in a 5th stage compressor disc while the dysfunction limit is only 4.5 mm (Fig. 3.1).
- Cracking was less severe in the initial stages of the compressor assembly. Likewise, there were no cracks in the 8th stage disc which represents the final stage of the compressor.
- Cracking was absent in the 1st and 2nd stage turbine wheels.

The fracture critical locations in the engine are indicated schematically (marked x) in Fig. 3.2. It appears that cracking severity rises to a maximum around the fifth stage compressor disc and spacer and diminishes as the eighth stage disc is reached. It is also obvious that the majority of fourth stage and eighth stage compressor discs and all turbine wheels were retired in a crack free condition. The reasons for this crack-free retirement are not clear at this time, although it might have been considered convenient to replace all components at the same time, since they were all life expired based on earlier, safe life concepts.

Further analysis consisted of a detailed experimental study as well as a statistical (Weibull) analysis of data available on service life and crack size distribution in these parts. The experimental techniques used range from optical, scanning and transmission electron microscopy (to characterise the service induced damage), to microhardness measurements, non destructive inspection, and crack propagation analysis by laboratory scale fracture mechanics testing. The experimental findings were used to interpret the results of the statistical analysis. In addition, finite element stress analysis computations were performed to estimate the stress levels in these parts during service and the stress intensity factor levels under which crack propagation occurs.

Until now, attention has been focussed upon the fifth stage compressor disc (fabricated from AM-355, a precipitation hardening, martensitic stainless steel) and the second stage turbine wheel (fabricated from A-286, an iron base superalloy). The fifth stage compressor disc was chosen because it is one of the most crack-prone components in the compressor assembly.

3.1 Compressor Disc Analysis

3.1.1 Fractographic Analysis

Scanning electron microscopy of bolt hole cracks that were pried open clearly established that the mechanism of failure was low cycle fatigue (Fig. 3.3). Other mechanisms such as stress corrosion cracking or hydrogen embrittlement seemed to be absent. No evidence of fretting damage was found either.

Striation spacing measurements were made upon one of the longest cracks present in a fifth stage compressor disc and the results are shown in Fig. 3.4. The striation spacings at small crack lengths are about 0.5 μm at small crack lengths ($<1\text{ mm}$) and are at an almost constant level of $\approx 1.0\text{ }\mu\text{m}$ between 1-10 mm. After about a crack length of 10 mm, the striation spacing appears to rise sharply to $\approx 2.5\text{ }\mu\text{m}$. Thus, there appears to be a significant crack length interval over which the striation spacing and hence the crack propagation rate appears to be approximately constant. Similar investigations were performed on eleven other discs, involving several cracks in each case.

An important observation made on the fracture surfaces was the tendency for striations to be banded (Fig. 3.5). Most often, the number of striations within a band appeared to be ≈ 5 (Fig. 3.5-a) while in rare occasions, about 10 striations were found to be grouped together (Fig. 3.5-b). Each band or group of striations is believed to represent a flight mission with each individual striation representing a manoeuvre in which the engine underwent a major change in rotational speed.

Analysis of crack initiation sites revealed that the observed cracking is often provoked by the presence of sub-surface inclusions or

foreign particles, (Fig. 3.6). Machining defects may also be potential crack initiation sites. Quite often, the initiation site was a corner crack (Fig. 3.7) and the direction of crack propagation was initially at an angle to the radial direction. This may be an indication that some bending stresses are occurring in the component as well. In some cases, multiple crack initiation sites were found (Fig. 3.7).

3.1.2 Optical Microscopy

Rim and bore areas of a fifth stage compressor disc were examined by optical microscopy to identify any signs of service-induced microstructural degradation. Radial sections of the component were prepared which were then divided into five sections from hub to rim. These were mounted in bakelite holders, polished and etched to reveal the microstructure. Immersion etching in a solution of 10 ml HNO_3 , 10 ml acetic acid, 15 ml HCl and 2-5 drops glycerin revealed the martensitic structure while electrolytic etching at 6 V for 10 s with an electrolyte of 10 gm ammonium persulphate in 100 ml of water brought out the prior austenite grain boundaries.

There were no obvious differences in the microstructures of the hub and rim sections. (Fig. 3.8). This is not an unexpected result since the maximum temperature that this component experiences ($\sim 200^\circ\text{C}$) is not severe enough to cause significant overall microstructural degradation.

Considerable crack branching was identified in several cases (Fig. 3.9) and the reasons for this are not clear. A number of what

may be inclusions were present in the material (Fig. 3.8), and these have a significant role in crack initiation when present as sub-surface defects, as mentioned earlier. However, they seem to have no direct effect in promoting crack propagation.

3.1.3 Replica Transmission Electron Microscopy

Plastic replicas were prepared from crack tip locations and examined using a transmission electron microscope. It was found that significant changes had occurred in precipitate morphology around the crack tip regions which experience severe strain cycling. As shown in Fig. 3.10-a, while the precipitate particles away from the crack tip appear to be spherical and uniformly distributed in the vicinity of the crack tip, there is a tendency for the precipitates to be elongated along what may be planes of intense shear (Fig. 3.10-b and c). The appearance of precipitate shearing at approximately 45° to the direction of crack growth supports this speculation. It is conjectured that such changes in morphology may be the result of localised heating along planes of intense shear. While the actual mechanism is not clear at this stage, these changes are at least an indication that the crack tip stress intensities are not small.

3.1.4 Microhardness Measurements

The majority of steels, when subjected to plastic deformation, will strain harden or soften. In a cracked component, such strain hardening or softening can be expected in the plastic zone

ahead of the crack tip. If the loading is cyclic, a small zone of reversed plastic yielding called the "cyclic plastic zone" can be found at the crack tip along with a larger "monotonic plastic zone", as shown in Fig. 3.11. The areas over which hardening or softening has occurred can be revealed by making a series of microhardness measurements in a grid pattern at the crack tip (Fig. 3.12). If R_y should represent the cyclic plastic zone radius, then a theoretical estimate of ΔK is given by^{1,4}

$$R_y = C \left(\frac{\Delta K}{\sigma_{ys}} \right)^2 \quad (3.1)$$

where, σ_{ys} is the yield strength of the material and C is a constant dependent upon the material and type of loading. In the AM-355 compressor disc investigated, a considerably hardened zone was located very close to the crack tip (Fig. 3.13). For a range of crack lengths between 2-6.7 mm, the size of this plastic zone did not vary significantly and R_y was approximately 0.45 mm. Using a value of $\frac{1}{2\pi}$ for C^{1+4} and 840 MPa for σ_{ys} , the value of ΔK is computed as 45 MPa $\sqrt{m}$. In the quantitative sense, this is only a preliminary estimate since the yield strength, as well as the value of C for this material are yet to be established in this study. However, the qualitative observation that the plastic zone size remains approximately constant over a significant crack length interval is important. This confirms the conclusion reached by striation spacing measurements that the crack propagation rates in the fifth stage disc bolt hole cracks tend to be fairly constant over a significant portion of the crack length.

3.1.5 Finite Element Analysis

A preliminary 2-D finite element stress analysis for the fifth stage disc was performed at NAE using the mesh geometry shown in Fig. 3.14 and assuming a rotational speed of 16,500 RPM.⁵ The effects of blade mass, the aerodynamic loads, bolt and washer mass, and interactions due to the spacers on either side have been neglected for this computation. Though a somewhat coarse mesh has been used, the sophisticated 12-node isoparametric elements used for modelling this disc are known to produce reasonably accurate results.⁵ The stress intensity factors for various bolt hole crack lengths were computed by using crack tip singularity elements and by successively releasing nodes located along the crack path.

The stress intensity factors computed in this manner for various bolt hole crack lengths have been plotted in Fig. 3.15. Due to the various assumptions made, the actual value of the stress intensity factor might have been underestimated. It is of interest, however, to note that the value of K tends to remain constant over a significant portion of the crack length. This is consistent with the results obtained through fractography and by microhardness measurements. However, a rigorous 3-D analysis should be performed to confirm these results. In such an analysis, it is necessary to take into account the fact that the fifth stage disc is not rotating, independently, but is actually part of a complex bolted assembly containing a number of other discs and spacers.

3.1.6 Crack Propagation Analysis by Fracture Mechanics Testing

An essential element in a study of this nature is laboratory testing of the disc material under simulated service conditions in order to determine typical lives for crack initiation as well as for dysfunction. Laboratory simulation can be performed using specimens machined from the actual disc material at much lower cost and effort than spin pit testing. Significant progress has been made in determining crack propagation rates in the fifth stage compressor disc material using fracture mechanics principles. Details of this investigation may be found in Reference 12. The manner in which the crack growth rate data from this study may be used in calculating a safe inspection interval is discussed in a later section.

A parallel study using spin-pit crack growth rate testing of retired discs has also been performed at the University of Toronto.¹³

3.1.7 Quantification of Inspection Uncertainty

In an earlier section, the philosophy of incorporating statistical estimates of inspection uncertainty into a 'Retirement For Cause' lifing procedure was described in detail. It is of considerable value to derive some such estimate for the case of the compressor components of the J85-CAN40 engine, particularly since cracks as long as 19 mm have been found to exist in these. It was strongly suspected that due to the limitations of the NDI technique adopted by the engine user, components with defects just below the detection limit were being inadvertently reinstalled into the engines thereby leading to the development of very large cracks. An experimental study was therefore

undertaken to determine, (a) the longest crack that is likely to be missed by the production type NDI used by Orenda; (b) the longest crack that is likely to be missed by NDI under ideal laboratory conditions; (c) the confidence with which cracks of various lengths can be detected; and, (d) the average error involved in measuring the lengths of cracks which are well above the detection limits and the accuracy with which small crack extensions can be measured for the case of a propagating crack.

However, it should be noted that in reality the detection limit is not a single, absolute number. Real life inspections are far from being ideal and hence the cracks which go undetected will have a size distribution with a certain mean and standard deviations.

Experimental Procedure. The initial part of this study was conducted on a fifth and a seventh stage compressor disc, both of which had developed several cracks at bolt holes as a result of service exposure. Initially, information was available only on the number of cracks present in each disc and their approximate length, and their locations in terms of bolt hole number was often not known. It was therefore decided to investigate all 40 bolt hole locations through metallographic techniques and attempt to correlate the crack length measurements made through optical metallography with the values measured by Orenda using a production type fluorescent penetrant inspection technique. Details of the technique used are given in Table 3.4.

The metallographic technique involved cutting out small areas around each bolt hole present, mounting these coupons in bakelite

holders and mechanically polishing with abrasive paper and diamond paste by standard metallographic procedures. Care was taken to avoid excessive grinding which would efface small corner cracks. Crack lengths in the polished specimens were measured optically, using a traveling microscope. The results are shown in Tables 3.5 and 3.6. In the case of the fifth stage compressor disc, a total of 35 cracks were found while only 25 had been reported by Orenda. It was attempted to arbitrarily match the reported lengths with the lengths measured in this study (Table 3.5), and it appears that the cracks that were missed are most likely to have lengths in the range 0.17-1.30 mm. This result is shown graphically in Fig. 3.16.

In the case of the seventh stage disc, the correlation was made easier by the fact that information relating bolt hole numbers to reported crack lengths was available. In Table 3.6 which compares optically measured crack lengths to reported values for various bolt holes in the seventh stage disc, it may be seen that the reported crack lengths are by no means very accurate in all cases. However, in this disc, no cracks went undetected by Orenda's FPI technique.

These results prompted a more rigorous study of inspection uncertainty under ideal laboratory conditions. Support for this was provided by DND through the Aircraft Maintenance Development Unit (AMDU). AMDU was provided with a fifth and seventh stage disc for NDI evaluation. Three NDI techniques, namely, radiography, magnetic particle inspection (MPI) and liquid penetrant inspection (LPI or FPI) were evaluated. Because of the shallow depth of the holes, the use of

LPI with a digital micrometer for crack length measurement was determined to be the most suitable. Details of the LPI technique used by AMDU are also given in Table 3.4. As this technique is operator sensitive, AMDU agreed that the same experienced technician would conduct all of the inspections. The bolt hole numbering scheme shown in Fig. 3.17 was employed. Each bolt hole from both discs was examined on either side and thus, a total of 160 independent inspections were made. The intent of each inspection was to detect the presence or absence of a crack and if a crack was detected, determine its size as accurately as possible. Both compressor discs were subsequently returned for verification of the NDI indications.

Upon receiving these discs, bolt hole areas in which cracks had been detected by AMDU were cut out and the crack faces were pried open allowing crack lengths and crack geometries to be measured by scanning electron microscopy (SEM). In the case of corner cracks, the dimension measured was the crack length along the face of the disc.

The bolt holes where no cracks were detected by AMDU were cut out, mounted in bakelite holders, polished and inspected by optical microscopy as explained earlier. Using this technique, seven unidentified cracks were found in the seventh stage disc and three were found in the fifth stage disc. Among the undetected cracks, the longest was 1.75 mm on the serial numbered side of bolt hole #19 on the fifth stage disc. The crack lengths reported by Orenda and AMDU are compared against the actual crack lengths (as measured by optical microscopy and scanning electron microscopy) in Tables 3.7 and 3.8.

Most of the cracks which went undetected were small corner cracks and would be visible on one side of the disc only, thus making their detection very difficult. It was also obvious from observations of the fracture surfaces that the cause for cracking in these bolt holes was low cycle fatigue as well.

Probability of Detection Analysis. The probability of detection (POD) for various intervals of crack length may be determined simply as the ratio of the number of cracks in that interval that were detected by NDI to the number of cracks in that interval that are actually present. For example, in the 0-0.5 mm crack length range, since only one out of seven cracks were detected, the POD is 1/7. It is customary to evaluate a confidence interval for detection for a particular range of crack lengths and use the lower bound of this interval as the required limit in the probability of detection. In this study, the MIL-A-83444 requirement⁷ for probability of detection (90 percent probability of detection at a 95 percent confidence level) has been selected as the required level in the probability of detection. This has been adopted since there are no separate specifications available for cold section gas turbine discs. The results of this analysis are shown in Figs. 3.18 and 3.19.

The analysis of the AMDU and Orenda results was based on the assumption that the data is binomially distributed. This is thought to be a valid assumption since the data has the following characteristics which are required of a process to be described by the binomial distribution:⁸

- a. The result of each NDI inspection can be classified into one of two categories: success or failure.
- b. Each inspection is independent of all others.
- c. The inspection is performed a fixed number of times.

In order to obtain the lower 95 percent confidence limit for the group probabilities of detection, the overlapping-interval method described by Packman et al.⁹ was used. It should be noted the plots shown in Fig. 3.19 have been made with a barely sufficient number of data points for each crack length range and it would be desirable to repeat this analysis using additional data from several more discs.

In Fig. 3.20, the effectiveness of the LPI/digital micrometer combination for accurate crack length measurement is illustrated. It is evident there is some scatter, both conservative and non-conservative at all crack lengths. An average error in crack length measurement of 0.34 mm (0.013 in.) was calculated as the absolute value of the cumulative error divided by the number of readings. Clearly, this level of error would make this technique quite inaccurate for the purpose of monitoring crack propagation rates in spin pit testing.

It is felt that the apparently lower sensitivity of the LPI measurements reported by Orenda could be a reflection of the fact that these measurements were made at the time of retirement and the objective at this stage was to maintain records rather than to achieve the most accurate measurements possible. In most cases reported, crack lengths had been rounded off to the nearest fraction of an inch. This, in itself, could introduce significant error.

It should also be noted that the type of emulsifier used in the LPI technique could affect the accuracy of results. An oil based emulsifier removes penetrant from the surface quickly and can decrease sensitivity. The dwell time is not strictly adhered to. Other factors affecting detection results are the level of control of the test conditions, and the levels of operator skill, training and experience. From this point of view, it would seem favourable to adopt a technique such as eddy current NDI which, whether manual or automated, would be less operator sensitive than LPI. The eddy current option, though rejected initially, should be re-examined in the light of these results.

3.1.8 Statistical Analysis

The following subsections describe the results of Weibull analysis performed on service life data available on retired fifth stage compressor discs. The purpose of this analysis was to determine the principles upon which G.E. developed the N_1 (predicted) initiation life limit, the N_d (observed) dysfunction life limits as well as the crack propagation interval for this engine.

The analysis is based on data regarding 201 discs that were retired between August 1977 and October 1983. 84 of these had developed cracks and out of these 49 had cracked to the dysfunction limit and above.

Prediction of Crack Initiation (N_1) Life. It should be noted that G.E. only supplied a N_1 (predicted) life limit and not an N_1 (observed)

life limit. This predicted life limit could have been developed based on a data base generated by conducting a number of constant amplitude LCF tests using push-pull type specimens. In these tests, a fixed number of constant amplitude cycles could be assumed to represent one mission cycle and the number of missions required to cause crack initiation can then be deduced from the information generated. In addition, spin pit testing could have been carried out using realistic mission profiles to generate the crack initiation data. Since the precise methods used by G.E. to obtain this life limit are not known, an independent approach was used to try and approximate the N_1 (predicted) life limit line. This approach is described as follows.

It is possible that crack propagation started on the first cycle following reassembly of some components that were said to be crack free at a particular inspection time. It is, however, unlikely that this condition occurred on all parts and equally unlikely that all parts cracked to the observed lengths on the last cycle just before retirement. It was therefore assumed, for the purpose of this analysis, that in those discs that had cracked to the dysfunction limit, cracks that were barely detectable by NDI were already present at previous overhaul and these went undetected. In such an event, the difference between TSN (Time Since New) and TSO (Time Since last Overhaul) of these discs can be regarded as time to crack initiation. The life distribution plot of TSN-TSO for 49 discs that had cracked to dysfunction is shown in Fig. 3.21. The Weibull analysis results of this data is presented in Fig. 3.22.

It may be seen from Fig. 3.22 that the line using TSN-TSO falls slightly to the right of the N_i (predicted) line supplied by G.E. This is to be expected, since it has been demonstrated in an earlier section that cracks as long as 1.75 mm can go undetected by the FPI technique in the overhaul of these engines. The N_i (predicted) line given by G.E. is presumably based on crack initiation to a length of 0.8 mm (1/32 in.) which is the initiation criterion used by most engine designers. The line drawn using TSN-TSO on the other hand predicts the life to the detection limit for the NDI techniques used in this programme, i.e. a crack of 1.75 mm. Thus, the TSN-TSO approach provides an estimate of the cycles to detection rather than the cycles to initiation. This difference is clearly indicated in Fig. 3.22.

Another approach used to estimate the N_i line involves a combination of experimental and statistical techniques as below.

Eleven bolt hole regions having cracks ≈ 6.35 mm were cut out of different fifth stage discs and the cracks in them were pryed open. The fracture surfaces were studied by SEM and a striation count was made in each case to estimate the number of striations between the tip of the crack and the point of initiation (i.e., 0.8 mm). The total number of striations obtained in this manner were in each case divided by a certain factor representing the number of striations or crack growth cycles per mission. This provided an estimate of the number of missions since crack initiation. Factors of 5, 10 and 15 were used, since fractographs of a number of bolt hole crack surfaces had indicated that in each "band" of striations a maximum of 10-15 occurred.

while it was more common to count about 5 in each. The values obtained by dividing the total number of striations by these factors were then subtracted from the respective TSNs. If the further assumption is made that each Tutor aircraft mission lasts for about an hour, the resulting figures would then represent the time to 0.8 mm crack initiation in each disc for assumptions of 5, 10 and 15 LCF cycles per mission. This assumption was made based on the information provided to us on the Tutor aircraft mission profiles (Fig. 3.23). The Weibull plots were then made for these three cases as shown in Figs. 3.24-3.26. The N_i (predicted) line of G.E. is also shown for comparison. It is seen that the assumption of 5 striations per mission yields the most accurate results which is in agreement with the fractographic observations.

However, the number of data points used to prepare these plots is quite limited and it would be desirable to confirm the results by performing a similar analysis on several more cracked discs.

Prediction of the N_d (Observed) Line. The N_d (observed) line could be readily determined by conducting Weibull analysis based on the TSN of the 49 discs that had cracked to dysfunction. The life distribution of the TSNs for these discs is shown in Fig. 3.27 and the Weibull plot in Fig. 3.28. It may be seen that this procedure produces a line very close to the N_d (observed) line given by G.E.

However, it should be noted that the 49 dysfunction data points were taken from those discs in which cracks 4.5 mm or greater had formed during service. Strictly speaking, the N_d line should be plotted by using the TSN values of a statistically significant number

of discs which contain a maximum crack size of 4.5 mm and not greater. Data on discs with crack lengths longer than the precise dysfunction crack size may not form an integral part of the population required to estimate the dysfunction limit.

It is of course possible, in the case of cracks greater than the dysfunction size to correct the TSN for the excess crack propagation using the cycle-counting, fractographic method explained earlier. But this would entail a scale of effort not within the scope of this study.

It was therefore attempted to analyze TSN data of only those discs containing maximum crack lengths of 3.175 mm, 4.5 mm and 6.35 mm in three separate trials. The TSN of uncracked discs and other discs containing cracks smaller than the preset failure criterion were included in ordering and ranking of the data. The Weibull analysis results for the 3.175 mm, 4.5 mm and 6.35 mm crack lengths are presented in Figs. 3.29, 3.30, and 3.31, respectively. The N_1 and N_d (observed) lines given by G.E. in 1983 are also plotted in these figures for comparison purposes.

It is evident that all three lines are to the right of the N_d (observed) line of G.E. and fall virtually on the same place. Since a certain finite number of cycles are required for crack propagation from 3.175 mm to 4.5 mm and then to 6.35 mm, one would expect at least some separation between these lines - if the analysis were accurate. The observed merging of these lines is believed to be the result of insufficient data for each individual case. Also, it can be seen that

most of the data points lie well above the 0.1% or 0.135% failure line in Figs. 3.29-3.31. The reliability of interpolating these data for decision making at 0.1% failure probability is thus rendered highly questionable.

However, a considerable amount of additional data can be generated if it is realized that each one of the discs being analyzed contains 40 bolt holes and hence 40 individual sources of data since the loading conditions are identical in each case. Cracking in each bolt hole occurs independently of other cracks in other bolt holes. Thus, with TSN data on ≈ 200 discs, more than 8000 separate data pairs can be generated to analyze bolt hole cracking probabilities. The results of using this technique are discussed in the following section.

In the absence of possibilities to analyze data in the manner stated above, Monte-Carlo simulation techniques may be applied to generate additional data. Basically, a model is created based on existing information and computer simulation with random number generators is used to create a large amount of additional data. Such methods should be used in conjunction with the Kolmogorov-Smirnov test¹⁰ to distinguish whether or not a data point created belongs to the actual life distribution of the component concerned.

The slope, β , of all N_d lines plotted for J85-5 and J85-CAN40 fifth stage compressor discs lie between 9-9.8 (Table 3.9), whereas a β value of ≈ 3 is generally associated with the physics of LCF problems in the literature, Fig. 2.2. It should, however, be noted that any reference to LCF problems in the turbine engine literature generally deals.

with crack initiation rather than classical crack propagation studies. The β values obtained for the N_i lines plotted in this study are much closer to 3 (Table 3.9) and hence the β values obtained for the N_d lines are not totally unexpected.

Prediction of a Crack Propagation Interval. It was initially felt that the crack propagation interval (CPI) should be related to the predicted N_i and N_d distributions at 0.1% or 0.135% F. For example,

$$\text{CPI}(-3\sigma) = \text{TSN}(N_d) \text{ at } 0.1\% \text{ F} - \text{TSN}(N_i) \text{ at } 0.1\% \text{ F}$$

According to this relation, 1 disc in 1000 will be expected to contain a dysfunction crack size after one CPI beyond initiation life.

Applying this calculation to the Weibull plots (N_i and N_d observed) of the J85-CAN40 fifth stage compressor discs gives the -3σ CPI as $\approx 2285 - 1067 \approx 1200$ h (Fig. 3.22), whereas G.E. gave the -3σ CPI as 555 h. It is possible that a safety factor was used and quite often, a factor of 2 is generally used to calculate the return to service interval in damage tolerant designs.¹² Based on this, the CPI would then be ≈ 600 h which is a much closer estimate of the predicted value. Various other methods of determining the CPI were investigated in this study and these are detailed below.

1. It was mentioned earlier that a Weibull plot of TSN-TSO (Fig. 3.22) yields an estimate of cycles to detection rather than cycles to initiation. It was felt that it would be more practical to consider the interval between the N_d (observed) line and the TSN-TSO plot at the 0.1% F level as a measure of the crack propagation interval. In this manner, a safer estimate of the crack propagation interval would be obtained because in actual practice it is

critical to know the interval between the detection limit and the dysfunction limit rather than the interval between the initiation life and the dysfunction life. Based on this, the propagation interval may be calculated from Fig. 3.22 as 2285 h (N_d observed at 0.135% F) - 1600 h ($N_{\text{detection}}$ at 0.135% F) = 685 h which is close to the detection crack propagation interval of 555 h predicted by G.E. In this case, it should be noted that a safety factor is not used since the calculations are based on actual operating experience.

2. It was established earlier that an assumption of 5 striations per mission was a reasonable estimate. Then the crack propagation interval in hours may be calculated as:

$$\text{CPI} = \frac{\text{Dysfunction cracksize} - \text{Limit of Detection}}{(\text{Striation spacing}) \times \text{No. of striations per mission}} \quad (3.5)$$

For a dysfunction line of 4.5 mm, a limit of detection of 1.75 mm, an average striation spacing of 1 μm (between crack length of 0.8 mm and 4.5 mm) and 5 striations per mission, the crack propagation interval is given by,

$$\text{CPI} = \frac{4.5 - 1.75}{1.0 \times 5} \times 10^3 = 550 \text{ h},$$

which is strikingly close to the estimate of 555 h provided by G.E.

The striation spacings between 0.8 mm and 1.75 mm are only of the order of 0.7 μm on an average as stated earlier and hence, the interval between N_1 (0.8 mm) and N_d may be estimated as,

$$\left[\frac{1.75 - 0.8}{0.7 \times 5} + \frac{4.5 - 1.75}{1.0 \times 5} \right] 10^3 = 830 \text{ h}$$

It can be seen that the interval between the N_1 (predicted, 1981) and N_d (predicted, 1981) lines of G.E. at the 0.135% F level is

about 955 h (Fig. 2.5) which is not very far from this estimate.

Allowing for the errors involved in statistical analysis, this is considered to be a reasonable correlation.

3. Another approach was to plot TSO (Time Since Overhaul) vs cumulative failure probability. Since crack propagation to the dysfunction limit occurs in these discs during the time between last overhaul and the time of retirement, it was felt that the TSO plot of discs that had cracked to dysfunction could be used to derive the crack propagation interval as well.

In the TSO plot shown in Fig. 3.32, which was supplied by General Electric, it appears that the maximum propagation interval at 0.135% F would be ~250 h. (However, it would be erroneous to interpret the plot in this manner, since the data indicates a strong tendency for the plot to have some curvature. In Fig. 3.32, at least two different slopes (I and II) can be distinguished. This curvature can be corrected for by a method suggested by Abernethy et al.¹¹ The method consists of subtracting a correction factor from all the TSO values and replotting the Weibull line (Fig. 3.33). The point at which the line intersects the 0.135% F line is determined and the correction factor is added back to the value of the TSO at that point. The method determining the correction factor may be found in Appendix B. For the 5th stage compressor disc TSO data, the correction factor was found to be 505 h and the TSO value for the replotted Weibull line at the 0.135% F level was 50 h, giving a CPI of $505 + 50 = 555$ h, as predicted by G.E. and by the methods stated earlier.

4. An estimate of the CPI can also be obtained by fracture mechanics testing of pre-cracked compact tension (CT) specimens machined from the respective discs. Under constant ΔK conditions between the detection limit (1.75 mm) and the dysfunction limit (4.5 mm), the propagation life N_{pr} can be calculated¹³ from fracture mechanics principles as,

$$N_{pr} = \frac{4.5 - 1.75}{C(\Delta K)^n} \quad (3.6)$$

The ΔK value was demonstrated earlier to be in the vicinity of 45 MPa $\sqrt{m}$ for a significant crack length interval. However, at the time of this study precise information on C and n at temperatures typical of service operation for this component ($\sim 200^\circ\text{C}$) were not available. A study is underway at NAE to generate this data.

6. As mentioned earlier, each one of the 40 bolt holes in the 5th stage disc may be regarded as an independent source of data and Weibull plots may be prepared as in the case of the earlier types of analysis. The Weibull plots for crack lengths of 3.175 mm, 4.5 mm and 6.35 mm are shown in Fig. 3.34 and were generated from over 8,000 bits of data. The most striking aspect of this plot is that the data points for a crack length of 3.175 mm fall along a line well separated from that containing the data points for 4.5 mm and 6.35 mm. The plots for the latter two crack lengths have almost merged into and, perhaps indicating that once the dysfunction crack size of 4.5 mm is reached, crack propagation to 6.35 mm may occur over a fairly small time interval. The separation between 3.175 mm

line and the plots for 4.5 mm or 6.35 mm appears to be 200-300 h at the 0.135 F level. The plots for smaller crack lengths have not been prepared due to lack of data, but it is logical to assume that the propagation interval from the detection limit of 1.75 mm to 1/8" (3.2 mm) may be at least another 200-300 h. Thus, the crack propagation interval from the detection limit to the dysfunction limit may be expected to lie between 400-600 h. This result supports the conclusions derived earlier. It should also be noted that in the plot of Fig. 3.34, the data points plotted are clustered very closely near the 0.135% F line which gives higher confidence in the results than in the case of earlier analyses where virtually no data points fell near the 0.135% F level.

In the case of J85-CAN40 components, G.E. suggested that parts which are known to have been crack-free at prior inspection may be operated for a maximum of one CPI beyond that inspection with minimum additional risk. These concerns can be rationalized in terms of continued microstructural degradation of turbine engine components with increasing service. At the end of one CPI, the microstructure of a component may be very different from that of the virgin material. In such an event, the crack propagation rates of these components after one CPI will vary considerably and their dysfunction life distribution will be quite different from that of the virgin material. It is thus suggested that a CPI will continuously change with increasing service. In particular, the CPI of a turbine disc population should be continuously updated with increasing service, since stress-temperature

induced microstructural degradation is bound to occur in hot section components. Whether significant microstructural degradation occurs in cold section components, such as compressor discs and spacers, remains uncertain at this stage.

3.2 Analysis of J85-CAN40 Turbine Components

The turbine wheels for the J85-CAN40 engine are fabricated from A-286, which is a precipitation hardenable nickel-chromium-iron alloy. The strengthening precipitate in this alloy is believed to be of the γ' type, with the basic composition $\text{Ni}_3(\text{Al}, \text{Ti})$. In the conventionally heat treated condition, the strengthening precipitate is in the form of fine spherical particles. However, upon prolonged exposure to temperatures around 600 - 700°C, this face-centred-cubic precipitate can transform into hexagonal η phase (Ni_3Ti) which occurs in the form of fine needles.

In this section, an experimental study that was carried out to characterize in-service microstructural degradation in the turbine wheels of the J85-CAN40 engine has been described. All the work has been carried out on the second stage turbine wheel and will be extended to the first stage wheel as well. The possibility of repairing the microstructural damage by re-heat treatment of these components has been assessed.

3.2.1 Microstructural Characterization

The study was carried out on two second stage wheels which had undergone 1953 h and 3737 h of service respectively. Neither one

of these wheels had any reported occurrence of in-service cracking, but the wheel that had been service exposed for 3737 h had some surface discolouration.

As in the case of the compressor disc analysis, the microstructural characterisation of this component involved the use of a number of techniques. Optical microscopy indicated that the bore section had an equiaxed grain structure while at the mid-section, a necklace-type structure dominates (Fig. 3.35). The latter microstructure essentially consists of larger warm-worked grains surrounded by a string of finer recrystallised grains at the site of the coarse grain boundaries. The necklace structure could be found near the rim section also, but to a significantly lesser extent than at the mid-section. Therefore, the average grain size in these wheels appeared to decrease from the bore towards the mid-section (due to the large number of fine, recrystallised grains present in the mid-section) and then increased towards the rim (Fig. 3.36). There was no obvious sign of microstructural degradation that could be detected in these wheels by optical microscopy.

Plastic-carbon replicas were prepared from hub and rim sections of both wheels and viewed by transmission electron microscopy. However, this technique was not useful in resolving the very fine strengthening particles present in this alloy.

Thin foil transmission electron microscopy was resorted to for better resolution. The foils were thinned by jet polishing with a solution containing 94 parts glacial acetic acid and 6 parts perchloric acid and viewed in a Phillips EM200 electron microscope.

Foils taken from very near the rim indicated that very little plastic deformation had occurred at this location. Very limited dislocation activity was seen in these regions (Fig. 3.37). However, on proceeding only a small distance radially away from the rim (towards the bore), evidence of significant plastic deformation became evident. There appeared to occur a transition region where some grains are deformed intragranularly while others remain intact (Fig. 3.38). At the bore itself, evidence of severe plastic deformation was found, with the occurrence of intense planar slip (Fig. 3.39).

An important observation from these foils was that in both discs, in regions that had undergone significant plastic deformation (as evidenced by a high level of dislocation activity), a needle-like precipitate was always present. These precipitates appeared to occur along slip bands and in areas of intense dislocation activity (Fig. 3.40). The progressive coarsening of these precipitates from the rim towards the bore is shown in the scanning electron micrographs of Fig. 3.41. The nature of this phase is not clear and is being investigated. Since the undeformed regions in both discs appeared to be free of this needle-like phase, there seems to be strong evidence to assume that these precipitates form as a result of a strain-induced transformation.

There are no essential differences between the γ' precipitate at the rim section of these two discs (Fig. 3.37). These findings indicate that microstructural degradation in the turbine wheels of the J85-CAN40 engine consists mainly of low cycle fatigue damage. The elevated temperature exposure by itself seems not to have created any significant variations in microstructure.

2

MICROCOPY RESOLUTION TEST CHART
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These conclusions are consistent with microhardness measurements made at the hub and rim sections of the two wheels studied (Fig. 3.42). The hardness values tended to remain more or less constant over the entire radial distance in both wheels indicating that matrix precipitate dissolution or coarsening at the rim is quite minimal at the service temperature for these components ($\sim 550^{\circ}\text{C}$).

3.2.2 Rejuvenation by Re-heat Treatment

Having characterized the type of microstructural degradation that occurs in these components, it was attempted to restore the original microstructure by a re-heat treatment procedure. It was expected that the re-heat treatment would cause the needle-like phase to dissolve and re-precipitate as γ'' and also considerably reduce the intensity of the dislocation sub-structures found in the service-exposed components.

One of the standard heat treatments for A-286 ($1000^{\circ}\text{C}/1 \text{ h/OQ} + 718^{\circ}\text{C}/16 \text{ h/AC}$) was employed. Specimens taken from the hub and rim of both discs were subjected to this heat treatment and thin foils prepared from the re-heat treated specimens were examined by transmission electron microscopy. The results are shown in Fig. 3.43. It is clear that the needle-like phase and the intense dislocation activity present near the rim regions are virtually eliminated in both discs by this re-heat treatment. Some remnants of prior plastic deformation are present in the bore regions (Fig. 3.44) but this is not critical since cracking in these wheels has been reported by the manufacturer to originate from bolt holes near the rim circle.

In order to be commercially feasible, this rejuvenation technique should be such that the extent of surface oxidation and the level of dimensional distortion as a result of re-heat treatment are minimal.

In order to investigate the extent of dimensional distortion which can occur during re-heat treatment of these components, service exposed Stage I and Stage II turbine wheels were subjected to a 1000°C/1h/Argon Quench + 718°C/10h/Argon Quench schedule. The 1000°C and 718°C exposures were in argon.

Examinations of the re-heat treated wheels indicated significant surface oxidation, perhaps due to insufficient purity in the argon atmosphere used. The wheels were also inspected for dimensional distortion effects which were measured in terms of the error in (a) flatness, (b) cylindricity of the hub and (c) perpendicularity of the axis of the wheel to the plane of the wheel. The results before and after re-heat treatment are shown in Table 3.10.

It appears that the extent of distortion which occurs in the Stage II wheel is quite limited as far as the in-plane flatness is concerned. The deviations before and after heat treatment do not suggest that excessive distortion will occur in this component.

However, the Stage I wheel appears to have suffered significant distortion with respect to perpendicularity of the axis to the plane of the wheel. It is most likely that this is a result of the hub section geometry. The Stage I wheel has a long, integrally machined annular shaft which forms an extension of the hub, and this extension has very limited restraint to minimize distortions in the presence of thermal

stresses created during the quenching operation. The re-heat treatments were carried out without any jigs or fixtures which would promote the overall rigidity of the component during heat treatment.

It is strongly felt that further work should be carried out to design such tooling which is necessary if the re-heat treatment schedules are to be carried out with a minimum level of distortion.

References

Chapter 3

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Table 3.1

Total Number of Discs Cracked in J85-CAN40

Compressor Assembly Between

December 1976 - May 1984

4th Stage Disc	2
5th Stage Disc	109
6th Stage Disc	19
7th Stage Disc	81
8th Stage Disc	6

Table 3.2

Service Life Data on Retired Components from the J85-CAN40 Engine

Component	Safe Life Limit Hours	Safe Inspection Interval Hrs.	Longest Life Component Hrs.	Mean Life (Hrs.)	Average Life Extension (Hrs.)
4th Stage Compressor Disc	2333	555	4428.2	3863	1530
5th Stage Compressor Disc	1777	555	4473	3875	2098
6th Stage Compressor Disc	2222	722	5066.4	3971	1749
7th Stage Compressor Disc	2222	722	5066.4	3812.9	1590.95
8th Stage Compressor Disc	2944	722	5066.4	4080	1136
1st Stage Turbine Disc	1111	555	737.6	2364	1253
2nd Stage Turbine Disc	1111	555	3791.6	2971	1860

Table 3.3

Maximum Crack Sizes Observed in Retired J85-CAN40 Compressor Components

Component	Maximum Crack Size in Service Retired Components - NSI (mm)
4th Stage Compressor Disc	No Record of Cracking
5th Stage Compressor Disc	19.1
6th Stage Compressor Disc	6.4
7th Stage Compressor Disc	7.9
8th Stage Compressor Disc	4.0
1st Stage Turbine Disc	No Record of Cracking
2nd Stage Turbine Disc	No Record of Cracking

Table 3.4

Details of FPI Techniques Used by Orenda and AMDU

AMDU

1. Ultrasonic cleaning using water and mild detergent.
2. CGSB liquid penetrant Type 1, Method B, Group 5/6
 - prewash with ARDROX 954551 solvent
 - apply ARDROX 983P2E penetrant, dwell 30 minutes
 - apply ARDROX 9PR4 emulsifier
 - hydro wash
 - oven dry
 - apply ARDROX 9D6 aqueous developer
 - dwell 15 minutes
3. Inspect before and after dwell time.

ORENDA

1. CGSB liquid penetrant Type 1, Method B, Group 5/6
 - apply ARDROX 987 penetrant, dwell 30 minutes
 - emulsify in ARDROX 9PR8-M, dwell 1.5 minutes $\pm$ 15
 - hydro wash 3 minutes maximum
 - warm air dry
 - apply ARDROX D-499C nonaqueous developer
 - dwell 10 minutes minimum
 - inspect.

NOTE

The essential difference between both techniques is that the AMDU technique uses a water based emulsifier while the ORENDCA technique uses an oil based emulsifier. The emulsifier dwell time using the oil based technique is critical to the sensitivity of the inspection.

Table 3.5
Comparison of Crack Lengths Measured by Orenda and Those Measured by
Optical Metallography in a J85-CAN40 5th Stage Compressor Disc

Crack No.	Crack Lengths	
	N.D.I.--ORENDA (mm)	Optical (mm)
1	0.79 (1/32 in.)	1.55
2	0.79	1.68
3	2.38 (3/32 in.)	1.90
4	2.38	1.94
5	2.38	2.12
6	2.38	2.19
7	2.38	2.30
8	2.38	2.40
9	2.38	2.59
10	2.38	2.61
11	3.18 (1/8 in.)	2.75
12	3.18	3.30
13	3.18	3.38
14	3.97 (5/32 in.)	3.55
15	3.97	3.60
16	3.97	3.69
17	4.76 (3/16 in.)	4.58
18	4.76	4.58
19	4.76	4.88
20	4.76	5.07
21	6.35 (1/4 in.)	5.12
22	6.35	5.72
23	7.94 (5/16 in.)	6.67
24	10.30 (13/32 in.)	7.70
25	10.30	9.10
26		0.17
27		0.40
28		01.45
29		0.48
30		0.52
31		0.66
32		0.98
33		1.11
34		1.21
35		1.30

Table 3.6
Comparison of Crack Lengths Measured by Orenda and Those Measured by
Optical Metallography in a J85-CAN40 7th Stage Compressor Disc

Bolt Hole No.	Crack Lengths	
	N.D.I. (mm)	Optical (mm)
3	3.18 (1/8 in.)	2.21
7	6.35 (1/4 in.)	5.08 1.18*
12	4.76 (3/16 in.)	4.35 1.84*
15	1.59 (1/16 in.)	1.76 0.41*
22	6.35 (1/4 in.)	5.09
27	7.94 (5/16 in.)	7.54
	4.76*(3/16 in.)	3.80*
37	3.18 (1/8 in.)	2.72 1.79*
38	4.76 (3/16 in.)	4.39 0.21*

All cracks are located at the bolt holes and are radially inward unless otherwise specified

*Radially outward

Table 3.7

Crack Length Measured Comparison: Actual Length vs AMDU Measurement
for 5th Stage Compressor Disc

Bolt Hole Number	Serial Side		Non-Serial Side		Orenda
	Actual	AMDU	Actual	AMDU	
4	1.70	1.90	1.21	1.35	3.75
6			2.16	1.00	MISS
7	5.45	4.48	5.47	4.59	4.76
9			.27	MISS	MISS
12	1.37	1.22			MISS
15	2.32	2.43	.72	1.96	3.75
17	5.56	5.54	5.35	5.64	6.35
18	1.25	XX			MISS
19	1.75	MISS	1.89	2.33	3.75
20	1.04	1.13	1.42	.98	3.75
22	0.00	XX			
23	.03	MISS			MISS
25	5.23	5.84	5.96	5.93	6.35
31	.42	.32			MISS
34	3.39	3.72	3.48	4.05	3.75

NOTES:

1. All dimensions in mm
2. MISS - No NDI observation
3. XX - NDI observation but too small to measure

Table 3.8
Crack Length Measured Comparison: Actual Length vs AMDU Measurement
for 7th Stage Compressor Disc

Bolt Hole Number	Serial Side		Non-Serial Side		Orenda
	Actual	AMDU	Actual	AMDU	
1	.88	.49	1.56	.91	MISS
4	1.79	2.46	2.08	1.68	3.97
5	.83	MISS	.59	XX	MISS
7			1.32	MISS	MISS
10	.96	MISS	.53	MISS	MISS
12	1.08	MISS	.29	MISS	MISS
14	1.45	1.20	1.72	2.06	3.97
15	1.01	1.42	.38	MISS	.79
16	.77	XX			.79
17	1.32	1.52, .62	1.20	1.01	.79
18	.38	MISS	.11	MISS	MISS
21	1.24	1.53	1.48	1.40	.79
22	0.00	XX	.80	1.07	MISS
24	0.00	XX			
27	2.03	2.96	1.95	2.52	3.97
29	2.57	2.96	2.30	2.19	3.97
31			1.29	MISS	MISS
32			.47	MISS	MISS
34	4.72	4.45	4.63	4.87	6.35
35			1.35	XX	MISS
36	1.67	2.44	2.30	2.17	3.97
37	1.75	2.30	2.45	2.03	3.18
38	2.84	3.07	2.56	2.52	3.97
39	1.02	1.16	1.58	1.57	1.58
40	1.29	1.62	1.51	.50	1.58

NOTES:

1. All dimensions in mm
2. MISS - no NDI observation
3. XX - NDI observation but too small to measure

Table 3.9
Details of Weibull Analysis Results for 5th Stage Compressor Disc

Trial No.	Aim	Sample Size	No. of Data Points	Lower Bound Assumed	0.135% F	Slope B
1.	To predict N_f line using TSN-TSO of those discs cracked to dysfunction or greater (Fig. 3.28)	201	49	0	888	4.00
2.	To plot the N_d (observed, 1983) line (Fig. 3.34)	201	49	0	2,368	9.19
3.	To plot the N_d line using TSN of discs containing a maximum crack size of 1/8 in. (Fig. 3.35)	150	24	0	2,700	9.95
4.	To plot the N_d line using TSN of discs containing a maximum crack size of 3/16 in. (Fig. 3.36)	161	12	0	2,267	7.86
5.	To plot the N_d line using TSN of discs containing a maximum crack size of 1/4 in. (Fig. 3.37)	183	19	0	2,650	9.80

Table 3.10
Results of Study on Dimensional Distortion Due to Re-Heat Treatment (RHT) of
Stage I and Stage II Turbine Wheels

A. Stage I

	Flatness (inches $\times 10^{-3}$)		Cylindricity (inches $\times 10^{-3}$)		Perpendicularity (inches $\times 10^{-3}$)	
	Before RHT	After RHT	Before RHT	After RHT	Before RHT	After RHT
Nominal	0.0	0.0	0.0	0.0	0.0	0.0
Measured	0.604	1.02	0.545	0.3	1.185	4.35
Tolerance	1.0	1.0	0.300	0.3	1.00	1.00
Deviation	-	0.02	0.245	0	0.185	3.35

B. Stage II

	Flatness (inches $\times 10^{-3}$)	
	Before RHT	After RHT
Nominal	0.0	0.0
Measured	0.246	1.02
Tolerance	1.0	1.0
Deviation	-	0.02

Fig. 3.1

19 mm Crack Found in a 5th Stage Compressor Disc

STAGE 2 DISC

STAGE 3 DISC

STAGE 4 DISC

STAGE 5 DISC

STAGE 6 DISC

STAGE 7 DISC

STAGE 8 DISC

STAGE 2 SPACER

STAGE 3 SPACER

STAGE 4 SPACER

STAGE 5 SPACER

STAGE 6 SPACER

STAGE 7 SPACER

Fig. 3.2

Fracture Critical Locations in J85-CAN40 Compressor

Fig. 3.3

Fraeture Morphology of Bolt Hole Cracks in
5th Stage Compressor Disc

Fig. 3.4

Results of Striation Spacing Measurements

— Long Crack in a 5th Stage Compressor Disc

Fig. 3.5

Banding of Striations Representing Various
Manoeuvres Within a Mission

Fig. 3.6

Crack Initiation at a Sub-surface
Inclusion — 5th Stage Disc

Fig. 3.7

Corner Cracks and Multiple Crack
Initiation — 5th Stage Disc

Fig. 3.8

Microstructure of Compressor Disc Material

(a) Hub/Rim

(b) Inclusions

Fig. 3.9 Crack Branching in 5th Stage Disc Bolt Hole Crack

Fig. 3.10

Precipitate Morphologies at Crack Tip in
5th Stage Compressor Disc

- (a) Away from the Crack Tip
- (b) Near the Crack Tip

Fig. 3.11

Crack Tip Plastic Zones Created During
Cyclic Deformation

Fig. 3.12

Grid Pattern of Microhardness Measurements

Fig. 3.13 Results of Microhardness Measurements at Crack Tip of 5th Stage Compressor Disc Bolt Hole Crack

Fig. 3.14

2-D Finite Element Mesh Used for Modelling
5th Stage Compressor Disc

Fig. 3.15

Results of Finite Element Analysis: Variation of Stress Intensity Factor with Crack Length

Fig. 3.16

Comparison of Crack Lengths Obtained by
Fluorescent Penetrant Inspection and
Optical Metallography for a J85-CAN40
5th Stage Compressor Disc

Fig. 3.17

Bolt Hole Numbering Scheme

Fig. 3.18 Histogram of AMDU Detection Data vs. Measured Actual Crack Lengths

Fig. 3.19 Probability of Detection — Lower 95% Confidence Bands

3.20 Effectiveness of Crack Length Measurement

Fig. 3.21 Life Distribution Plot of $TSN - TSO$ for Discs Cracked to Dysfunction and above

Fig. 3.22

Weibull Plot of $TSN - TSO$ for Discs Cracked to Dysfunction and above

TUTOR AIRCRAFT PROFILE: PILOT TRAINING 37.7 X YFR

Fig. 3.23 Tutor Aircraft Mission Profiles

Fig. 3.24

N_i Line Based on an Assumption of 5 Striations per Mission

Fig. 3.25

N_i Line Based on an Assumption of 10 Striations per Mission

Fig. 3.26

N_i Line Based on an Assumption of .15 Striations per Mission

Fig. 3.27

Life Distribution Plot of *TSN* for Discs Cracked to Dysfunction and above

Fig. 3.28 Weibull Plot of TSN for Discs Cracked to Dysfunction and above

Fig 3.29

Weibull Plot of *TSN* for Discs Cracked to
3.175 mm Only

Fig. 3.30

Weibull Plot of TSN for Discs Cracked to 4.76 mm

Fig. 3.31

Weibull Plot of *TSN* for Discs Cracked to
6.35 mm Only

Fig. 3.32 Weibull Plot of TSO for Discs Cracked to Dysfunction and above

Fig. 3.33

Weibull Plot of *TSO* for Discs Cracked to Dysfunction and above showing Correction Applied for Curvature

TSN (hours)

Weibull plot for Cracks

3.34

Treating Each Bolthole as a Separate Bit of

Data

Fig. 3.35

Microstructural Variations Between Hub and Rim
Sections of A-286 Stage 2 Turbine Wheel

TIME OF 10196-205	1953.5 HRS. SERVICE
TIME OF 10196-205	1953.5 HRS. SERVICE
TIME OF 10196-205	5757.6 HRS. SERVICE
TIME OF 10196-205	5757.6 HRS. SERVICE

Fig. 3.36

Variation of Grain Size from Hub to Rim in
Stage 2 Turbine Wheel of J85-CAN40 Engine

Fig. 3.37 Absence of Dislocation Activity at the Very Rim of Stage 2 Turbine Wheels

Fig. 3.38 Transition Region Showing the Beginning of Marked Dislocation Activity (About 7 mm away from the rim)

Fig. 3.39

Intense Planar Slip at Bore of Stage 2 Turbine Wheels

Fig. 3.40. η -phase Precipitation Along Slip Bands

Fig. 3.41
Progressive Coarsening of Needle-like phase (η)
on Proceeding Towards the Bore

- (A) AT RIM
- (B) 5 MM FROM RIM
- (C) 7.25 MM FROM RIM

Fig. 3.42

Results of Microhardness Measurements in Hub and Rim Areas

Fig. 3.43 Absence of Dislocation Activity and Needle-like phase (η) after Reheat Treatment in Rim Areas

Fig. 3.14 Remnants of Prior Plastic Deformation Present in
Bore Area after Reheat Treatment

CHAPTER 4

ANALYSIS OF J85-CAN40 AND J85-CAN15

MISSION PROFILES

4.1 Development of Mission Profiles

An estimate of the mission profile and the type of manoeuvres that the J85-CAN 15/40 engines are subjected to is useful for the purpose of interpreting service failure data and in designing experiments to simulate service exposure, e.g., spin pit tests and laboratory scale fracture mechanics tests. This section describes the results of an analysis that was conducted on data that was acquired from J85-CAN15 engines aboard the CF-5 aircraft, with a view to develop standardized mission profiles for this engine.

The data that were available were acquired through a Datamyte recorder at a sampling rate of 6 Hertz. The parameters recorded were altitude, power lever angle, compressor inlet temperature, turbine inlet temperature, speed and engine RPM. For the purpose of mission profile analyses only the RPM data was considered.

The RPM data were analyzed by making the assumption that the dwell periods do not affect the severity of usage for cold section components. Therefore, the dwell periods were removed, resulting in a series of peaks and troughs (N levels) of RPM values. The data were "filtered" at 1.0% meaning that RPM variations below 1.0% were not considered for the purpose of this analysis. The filtering was carried out using a computer program written in Fortran and the data was processed using a PDP-11/34 digital computer.

The data had been recorded for eleven different flights that had been flown according to AETE (Aerospace Engineering Test Establishment) directives.¹ Each flight had a separate set of manoeuvres, but these were not complete missions. However, the AETE data was of value in determining the time periods for which each particular manoeuvre was conducted. For example, it could be determined if this data was of value in determining the time periods for which each particular manoeuvre was conducted. For example, it could be determined from this data that a 5 g-roll over, 180° turn would last approximately 25 secs.

Since the eleven flights were not complete missions, it was attempted to build a mission profile using this data in conjunction with a standardized set of mission profiles developed by Canadair.² The set of profiles developed by Canadair consisted of fourteen different standardized sequences of manoeuvres (complete missions) which were not described in complete detail. Using this information, (as specified) to plot a mission profile would not be a rigorous approach due to insufficiency of intermediate data points. Hence, information gained about each one of the various manoeuvres (e.g. formation exercise) from the AETE flights was used to complement the Canadair specifications. The results are given in Figs. 4.1-4.14 which illustrate the RPM variations during fourteen standardized missions. Mission profiles have also been plotted for the case of the data provided by the Canadair specification (without the intermediate data points acquired from the AETE flights) for comparison (Figs. 4.15-4.28).

Basic information regarding the Tutor aircraft (J85-CAN40 engine) mission profiles was also available, but a detailed analysis of the type mentioned above was not possible in this case since supporting information (such as the AETE flight data for the CF-5 aircraft) was not available. However, plots of the available information on these profiles are presented in Figs. 4.29-4.35. The MSF of these profiles relative to the J85-5 profiles had been assessed as 0.94 prior to 1981.

4.2 Assessment of Mission Severity

From the point of view of determining the number of LCF cycles (ground-air-ground cycles and partial cycles within a mission) which occur in each mission, it is of interest to know how many major throttle excursions take place in each mission. If several types of missions are flown with a particular aircraft, each for a different period of time in a year, the yearly average of the number of major throttle excursions (including the ground-air-ground cycle) can be calculated.

These engines undergo hundreds of RPM variations between 80%-100% of maximum RPM and if each of these relatively minor variations were to contribute to crack growth in the compressor components, the life of these components would be far lesser than observed in actual experience. This is because fractographic analysis of cracks in the fifth stage compressor discs of the J85-CAN40 engine indicated that the crack propagation rate per LCF cycle was between 0.5-1.00 μm . If all the perturbations observed in the mission spectrum

were to contribute to crack propagation, the resulting crack lengths per crack propagation interval would be much higher than actually observed. It was therefore decided to consider only the major perturbations in RPM. Since most of the major perturbations seemed to be associated with RPM variations of 25% or more, only these variations were considered for calculating the yearly average of the LCF cycles per mission. There is no data available at this time to clearly establish what level of RPM variations are to be considered as constituting a partial LCF cycle. The choice of 25% has been used here only as a trial to demonstrate a possible procedure to determine mission severity. A similar method is being used by General Electric for estimating mission severities for the F404 engine.³ The calculation procedure for the CF-5 (J85-CAN15 engine) mission profile and that for the Tutor aircraft (J85-CAN40 engine) mission profile is shown in Tables 4.1 and 4.2. Basically, this calculation involves determining the number of RPM fluctuations exceeding 25% of maximum RPM, in each of the 14 mission profiles and obtaining a weighted average, based on the Yearly Flying Rate for each type of profile. It may be seen that the average number of LCF cycles based on the yearly flying rate is about 6.35 (≈ 6) for the CF-5 aircraft. A similar calculation for the Canadair mission profiles plotted without considering the additional data obtained from the AETE flights indicates the average number of LCF cycles as only 1.5. Thus, in the absence of AETE flight data the Canadair mission profile severities might be considerably underestimated.

The average number of LCF cycles/mission for the Tutor aircraft appears to be at least 2.7 or ≈ 3 (Table 4.2). If actual flight data were made available and an analysis carried out as for the CF-5 data, the mission severity may turn out to be somewhat higher.

Considering that there is no data to establish that only RPM fluctuations of the order of 25% or more constitute a partial LCF cycle, and since AETE flight data has been available only for the CF-5 aircraft (J85-CAN15 engines), it is difficult to reach concrete conclusions regarding the actual mission severities for the CT-114 Tutor engines (J85-CAN40). However, the analysis carried out here serves to indicate a procedure by which an estimate of mission severities may be determined if adequate data were available.

References

Chapter 4

1. Aerospace Engineering Test Establishment (AETE) Project Directive 76/50, Low Cycle Fatigue Investigation on the J85-CAN15 Engine Data Collection, Jan. 9, 1979.
2. CANADAIR MAS-219-151, CF-5 Mission Profiles, July 1983.
3. F/A 18 IECMS Handbook, General Electric Company, 1984.

Table 4.1

Canadair CF-5 Mission Profiles: Calculations of the Number of Cycles with RPM
Fluctuations Exceeding 25% of Max. RPM
(including ground-air-ground cycle) based on a weighted average
(yearly flying rate, YFR) for each mission profile

Mission Profile	YFR (w_i) (%)	# times $\Delta > 25\%$	total # of cycles (including ground- air-ground cycle, z_i)	$\frac{z_i \times w_i}{100}$
1. Pilot Familiarization	14	2	2	3×0.14
2. Formation	11	7	8	8×0.11
3. Instrument Training	19	3	4	4×0.19
4. Navigation	5	4	5	5×0.05
5. Reconnaissance	1	0	1	1×0.01
6. Air to Air Refueling	0	0	1	1×0.0
7. Air to Air Gunnery	5	1	2	2×0.05
8. Dart Tow	0.5	6	7	7×0.005
9. Advanced Handling /ACM/BFM	15	3	4	4×0.15
10. Ground Attack (on Manned Range)	14	14	15	15×0.14
11. Ground Attack (on Tactical Range)	10	7	8	8×0.10
12. Ground Attack (Heavy Ordnance)	2.5	8	9	9×0.025
13. Air Displays	1	8	9	9×0.01
14. Test Flight	2	3	4	4×0.02
TOTAL	100%	64	78	6.35

Table 4.2

Tutor Aircraft: Calculation of the Number of Cycles with RPM Fluctuations
Exceeding 25% of Max. RPM

RPM (including ground-air-ground cycle) based on a weighted average
(yearly flying rate, YFR) for each mission profile

Mission Profile	YFR (w_1) (%)	# times $\Delta > 25\%$	total # of cycles (including ground- air-ground cycle, z_i)	$\frac{z_i \times w_1}{100}$
1. Pilot Training	37.7	4	5	5×0.377
2. Formation	6.4	1	2	2×0.064
3. Cross Country	15.0	0	1	1×0.150
4. Navigation	8.1	0	1	1×0.081
5. Instrument Flights	27.6	0	1	1×0.276
6. Test Flights	3.9	3	4	4×0.039
7. Air Displays	1.3	1	2	2×0.013
TOTAL	100%	9	16	2.702

PILOT FAMILIARIZATION 14 X YFR

Fig.4-1 Canadair Mission Profiles Improved with AETE Flight Data:
Pilot Familiarization, 14% Yearly Flying Rate (YFR)

FORMATION (CLOSE OR TACTICAL) 11 % YFR

Fig.4.2 Canadair Mission Profiles Improved with AETE Flight Data:
Formation (Close or Tactical), 11% Yearly Flying Rate (YFR)

INSTRUMENT TRAINING / FERRY FLIGHTS 19 % YFR

Fig. 4.3 Canadair Mission Profiles Improved with AETE Flight Data: Instrument Training/Ferry Flights, 19% Yearly Flying Rate (YFR)

NAVIGATION TRAINING 5% YFR

Fig. 4.4 Canadair Mission Profiles Improved with AETE Flight Data, Navigation Training, 5% Yearly Flying Rate (YFR)

RECONNAISSANCE 1 % YFR

Fig. 4.5 Canadair Mission Profiles Improved with AETE Flight Data: Reconnaissance, 1% Yearly Flying rate (YFR)

AIR TO AIR REFUELLING 0 % YFR

Fig. 4.6 Canadair Mission Profiles Improved with AETE Flight Data:
Air to Air Refuelling, 0% Yearly Flying Rate (YFR)

AIR TO AIR GUNNERY 5 X YFR

Fig. 4.7 Canadair Mission Profiles Improved with AETE Flight Data:
Air to Air Gunnery, 5% Yearly Flying Rate (YFR)

Fig.4.8 Canadair Mission Profiles Improved with AETE Flight Data:
Dart Tow, 0.5% Yearly Flying Rate (YFR)

ADVANCED HANDLING / BFM / ACM 15 % YFR

Fig. 4.9 Canadair Mission Profiles Improved with AETE Flight Data:
Advanced Handling/BFM/ACM, 15% Yearly Flying Rate (YFR)

GROUND ATTACK ON MANNED RANGE 14 % YFR

Fig. 4.10 Canadair Mission Profiles Improved with AETE Flight Data:
Ground Attack on the Manned Range, 14% Yearly Flying Rate (YFR)

GROUND ATTACK ON THE TACTICAL RANGE 10 % YFR

Fig. 4.11 Canadair Mission Profiles Improved with AETE Flight Data:
Ground Attack on the Tactical Range, 10% Yearly Flying Rate (YFR)

GROUND ATTACK CHANNED/TACTICAL RANGED HEAVY ORDNANCE 2.5 % YFR

Fig. 4.12 Canadair Mission Profiles Improved with AETE Flight Data:
Ground Attack on the Manned or Tactical Range with Heavy Ordnance
2.5% Yearly Flying Rate (YFR)

AIR DISPLAYS 1 % YFR

E 02
1.25

Fig. 4.13 Canadair Mission Profiles Improved with A&TE Flight Data:
Air Displays, 1% Yearly Flying Rate (YFR)

TEST FLIGHT 2 X YFR

Fig. 4.14 Canadair Mission Profiles Improved with AETE Flight Data:
Test Flights, 2% Yearly Flying Rate (YFR)

PILOT FAMILIARIZATION 14 % YFR

Fig. 4.15 Canadair Mission Profiles without AETE Flight Data Improvement:
Pilot Familiarization, 14% Yearly Flying Rate (YFR)

FORMATION 11 X YFR

Fig. 4.16 Canadair Mission Profiles without AETE Flight Data Improvement:
Formation (Close or Tactical), 11% Yearly Flying Rate (YFR)

INSTRUMENT TRAINING / FERRY FLIGHTS .19 % YFR

Fig. 4.17 Canadair Mission Profiles without AETE Flight Data Improvement:
Instrument Training/Ferry Flights, 19% Yearly Flying Rate (YFR)

NAVIGATION TRAINING 5 X YFR

Fig. 4.18 Canadair Mission Profiles without AETE Flight Data Improvement:
Navigation Training, 5% Yearly Flying Rate (YFR)

Fig. 4.19 Canadair Mission Profiles without AETE Flight Data Improvement: Reconnaissance, 1% Yearly Flying rate (YFR)

Fig. 4.20 Canadair Mission Profiles without AETE Flight Data Improvement:
Air to Air Refuelling, 0% Yearly Flying Rate (YFR)

Fig. 4.21 Canadair Mission Profiles without AETE Flight Data Improvement:
Air to Air Gunnery, 5% Yearly Flying Rate (YFR)

Fig. 4.22 Canadair Mission Profiles without AETE Flight Data Improvement:
Dart Tow, 0.5% Yearly Flying Rate (YFR)

ADVANCED HANDLING / BFM / ACM 15 X YFR

Fig. 4.23 Canadair Mission Profiles without AETE Flight Data Improvement:
Advanced Handling/BFM/ACM, 15% Yearly Flying Rate (YFR)

GROUND ATTACK ON THE MANNED RANGE 14 X YFR

Fig. 4.24 Canadair Mission Profiles without AETE Flight Data Improvement:
Ground Attack on the Manned Range, 14% Yearly Flying Rate (YFR)

GROUND ATTACK ON THE TACTICAL RANGE 10 X YFR

Fig. 4-25. Ground Attack on the Tactical Range, 10% Yearly Flying Rate (YFR)

GROUND ATTACK ON THE CHANNED/TACTICAL RANGE HEAVY ORDNANCE 2.5 X YFR

Fig. 4.26 Canadair Mission Profiles without AETE Flight Data Improvement:
Ground Attack on the Manned or Tactical Range with Heavy Ordnance
2.5X Yearly Flying Rate (YFR)

AIR DISPLAY 1 X YFR

Fig. 4.27 Canadair Mission Profiles without AETE Flight Data Improvement:
Air Displays, 1% Yearly Flying Rate (YFR)

TEST FLIGHT 2 X YFR

Fig. 4.28 Canadair Mission Profiles without AETE Flight Data Improvement:
Test Flights, 2% Yearly Flying Rate (YFR)

TUTOR AIRCRAFT PROFILE: PILOT TRAINING 37.7 % YFR

Fig. 4.29 Tutor Aircraft Mission Profiles without Flight Data Complement: Pilot Training, 37.7% Yearly Flying Rate (YFR)

TUTOR AIRCRAFT PROFILE: FORMATION 6.4 % YFR

Fig. 4.30 Tutor Aircraft Mission Profiles without Flight Data Complement: Formation, 6.4% Yearly Flying Rate (YFR)

3

MICROCOPY RESOLUTION TEST CHART
NATIONAL BUREAU OF STANDARDS
STANDARD REFERENCE MATERIAL 1010a
(ANSI and ISO TEST CHART No. 2)

TUTOR AIRCRAFT PROFILE CROSS COUNTRY 15 % YFR

Fig. 4.31 Tutor Aircraft Mission Profiles without Flight Data Complements:
Cross Country, 15% Yearly Flying Rate (YFR)

TUTOR AIRCRAFT PROFILE: NAVIGATION 8.1 X YFR

Fig. 4.32 Tutor Aircraft Mission Profiles without Flight Data Complement: Navigation, 8.1% Yearly Flying Rate (YFR)

TUTOR AIRCRAFT PROFILE: INSTRUMENT FLIGHTS 27.6 YFR

Fig. 4.33 Tutor Aircraft Mission Profiles without Flight Data Complement: Instrument Flights, 27.6% Yearly Flying Rate (YFR)

TUTOR AIRCRAFT PROFILE: TEST FLIGHT 3.9 X YFR

Fig. 4.34 Tutor Aircraft Mission Profiles without Flight Data Complement:
Test Flight, 3.9% Yearly Flying Rate (YFR)

TUTOR AIRCRAFT PROFILE: AIR DISPLAYS 1.3 % YFR

Fig. 4.35 Tutor Aircraft Mission Profiles without Flight Data Complement:
Air Displays, 1.2 Yearly Flying Rate (YFR)

SUMMARY AND CONCLUSIONS

Analysis of J85-CAN40 Compressor Components

1. The compressor discs and spacers of this engine were used much beyond the safe limits specified by the manufacturer.
2. The majority of retired compressor discs and spacers were severely cracked in bolt hole areas. Cracking was most severe in the 5th and 7th stage discs and in spacers on either side of them.
3. Fracture was determined to occur as a result of low cycle fatigue damage.
4. These cracks often initiated at sub-surface inclusions. Machining defects may have a role in initiation also. In many cases, cracks appeared to initiate as corner cracks.
5. The cracks propagate under a nearly constant stress intensity factor range (ΔK) and hence a nearly constant propagation rate over a significant crack length interval. This was demonstrated by, (a) striation spacing measurements, (b) microhardness measurements of crack tip plastic zone size; (c) 2-D finite element analysis.
6. Cracks as long as 19 mm were found in the 5th stage discs. It is thought that cracks could have grown to this length due to cracked components being inadvertently reinstalled after an inspection since the NDI procedure used was not capable of detecting all cracks below a certain "detection limit". This detection limit was found to be ≈ 1.75 mm even under ideal laboratory testing conditions.

7. The construction of Weibull cumulative failure probability plots was of considerable use in determining the basis upon which the lifing procedure was specified for these components.
8. The life to initiation (N_i) could be approximated by a Weibull plot using data on TSN-TSO of discs that had cracked to dysfunction and above. However, this appeared to yield a plot of "cycles to detection" rather than "cycles to initiation". The initiation line could be approximated better by a combination of experimental analysis and Weibull analysis. The experimental analysis consisted of determining the number of cycles to crack initiation in a number of discs by striation spacing measurements.
9. The safe limit specified (1983) for these components is based on dysfunction. The dysfunction limit (N_d) could be easily determined through a Weibull plot of TSN's of discs that had cracked to dysfunction and above.
10. The crack propagation interval (CPI) for the 5th stage disc could be determined by the various methods described below.
 - (a) By taking the separation between the N_i (predicted) and N_d (observed) Weibull plots at the 0.135% F level and dividing this by a safety factor of 2.
 - (b) By taking the separation between the N_d (observed) plot and the TSN-TSO (cycles to detection) plot at the 0.135% F level.
 - (c) By using the following equation:

$$CPI = \frac{\text{Dysfunction crack size} - \text{Crack detection limit}}{\text{Striation spacing} \times \text{No. of striations/mission}}$$

(d) By using $\frac{da}{dN}$ vs. ΔK data from laboratory simulation of crack propagation using fracture mechanics tests and calculating CPI as,

$$CPI = \frac{\text{Dysfunction crack size} - \text{Detection limit}}{C(\Delta K)^n \times \text{No. of LCF cycles/mission}}$$

(e) By using a Weibull plot of TSO of discs cracked to dysfunction and above and measuring the TSO value at 0.135% F.

The CPI was determined to be 550-576 h by methods (c), (d) and (e) while the manufacturer's prediction was 555 h.

It is clear that components found to be uncracked at the end of the specified safe limit can be returned to service for at least one crack propagation interval. It may be possible to extend the life of uncracked components through additional propagation intervals, but these additional intervals would have to be calculated taking into account the added microstructural degradation that takes place during these extra hours of service exposure.

Analysis of J85-CAN40 Turbine Components

11. Microstructural degradation in the A-286 2nd stage turbine wheels could be characterized by thin foil electron microscopy and scanning electron microscopy.
12. There appeared to be no temperature induced microstructural degradation such as γ' precipitate dissolution or coarsening at the rim areas.
13. As a result of service exposure, the microstructure in these components undergoes what is most probably a stress-induced transformation with the appearance of needle-shaped η -phase particles. This transformation is most marked at the bore where the stresses are high. At the very rim where the stresses are considerably lower, this η -phase transformation could not be detected. There appears to be a clear correlation between the intensity of dislocation activity and the volume fraction of these needle-shaped particles. The η -phase appeared to coarsen with increase of service exposure time.
14. The η -phase particles and the intense dislocation substructure found in service exposed components could be removed by a re-heat treatment procedure, and the original microstructure restored.
15. The feasibility of applying this re-heat treatment procedure with a minimum of dimensional distortion has been investigated. The results for the Stage 2 turbine wheel appear to be encouraging. In the case of the Stage 1 wheel, special fixtures would be needed to prevent distortional effects during re-heat treatment.

RECOMMENDATIONS FOR FUTURE WORK

Analysis of J85-CAN40 Compressor Components

1. The RFC maintenance technique developed in this study was based on a Deterministic Fracture Mechanics approach. This "worst-case" approach assumed that cracks just below the detection limit are always present in components that are returned to service after an inspection. The rate of crack propagation was also taken from the worst case (longest crack) observed in service.

This is not likely to be the case in most components and hence does not result in full utilization of component life. Further, this method provides no quantitative estimate of risk associated with variations in maintenance parameters.

An improved approach might be to treat the crack detection limit not as a single, absolute number, but in terms of a probability distribution as detailed in Chapter 1 (Volume 1). In this method, components would first be inspected by a specific technique, after which a rigorous metallographic examination is conducted to verify the results of non-destructive examination. If a sufficient number of components are tested in this manner, estimates of the crack size distributions before and after inspection may be obtained. This data would then be used in a simulation model to estimate a life distribution. An inspection interval may then be determined based on a trade-off between cost-savings and risk of uncontained failure.

2. It is strongly suggested that the above analysis be conducted not only with Fluorescent Penetrant Inspection but with more sensitive

methods such as eddy-current inspection and x-ray microfocus techniques as well. This would provide the user with a range of techniques to choose from, depending upon the resources available. A more sensitive NDI technique would assure less risk of uncontained failure in service, while increasing utilization of component life.

3. While much of the work carried out in this study is upon the 5th stage compressor disk, it should be noted that this is not the most crack prone component in the engine. The most failure prone component is actually the fifth stage spacer since of the various components retired so far, the maximum number are fifth stage spacers. A similar analysis needs to be performed on the fifth stage spacers since the inspection interval for the engine as a whole has to be determined based on the intervals required for the most crack prone component. According to the manufacturer's specifications, the CPI for the fifth stage spacer is 832 h, but this would have to be verified.

Analysis of J85-CAN40 Turbine Components

4. It has been demonstrated in the case of these components that microstructural damage which occurs during service can be reversed by a reheat treatment procedure. Further work is needed to establish methods by which this reheat treatment can be achieved with a minimum of dimensional distortion. Special jigs and fixtures would have to be designed particularly in the case of the 1st stage turbine wheel which suffered significant distortion in the preliminary evaluation. Also, mechanical property testing (particularly LCF crack initiation tests) need to be carried out on coupons machined from retired components.

APPENDIX A

GEOMETRY OF J85-CAN40 ENGINE COMPONENTS

Fig. 1

Fifth Stage Compressor Disc of J85-CAN40 Engine

Fig. 2

Geometry of Fifth Stage Disc

Fifth Stage Spacer

Fig. 3

1/2 IN. TYP.

Fig. 4 Second Stage Turbine Wheel of J85-CAN40 Engine

APPENDIX B

METHOD OF CORRECTING A CURVED WEIBULL PLOT

The following analytical method of determining the correction factor for a curved Weibull plot is due to Abernethy et al. ("Weibull Analysis Handbook", AFWAL-TR-83-2079, November 1983).

The data plotted on Weibull paper appears curved if the starting point or origin of the distribution, t_0 is not equal to zero.

The actual value of t_0 in such a case may be determined as,

$$t_0 = t_2 - \frac{(t_3 - t_2)(t_2 - t_1)}{(t_3 - t_2) - (t_2 - t_1)}$$

where, t_1 is the first failure time, t_2 is the time corresponding to the linear halfway distance on the vertical axis between the first and last failure and t_3 is the last failure time.

In the case of the TSO plot for the 5th stage compressor disc,

$$t_1 = 618.2 \text{ h}$$

$$t_3 = 1886.7 \text{ h}$$

$$t_2 = 900 \text{ h (from Fig. B.1)}$$

$$\begin{aligned} \text{Thus, } t_0 &= 900 - \frac{(1886.7 - 900)(900 - 618.2)}{(1886.7 - 900) - (900 - 618.2)} \\ &= 505 \text{ h} \end{aligned}$$

The value of t_0 is subtracted from each of the failure times and the data replotted, which results in Fig. 3.33, presented earlier. In order to know the correct TSO value at the 0.1% F level, the value is read from Fig. 3.33, and the correction factor $t_0 = 505 \text{ h}$ is added to it. Since the replotted line intersects the 0.1% F level line at 50 h, the CPI = $505 + 50 = 555 \text{ h}$ which is exactly the estimate given by the engine manufacturer.

Fig. 1 Weibull Plot of TSO for Fifth Stage Compressor Disc Showing Method of Determining t_0

APPENDIX C

Ordering, Ranking and Calculation of %F from 1983 CF Data for Nd

O ₁	TSN	R ₁	%F	O ₁	TSN	R ₁	%F	O ₁	TSN	R ₁	%F
1	63.8			21	1595.5			41	2520.6		
2	245.2			22	1757.6			42	2554.7		
3	202.2			23	1769			43	2557.3		
4	299			24	1777.5			44	2597.3		
5	324.8			25	1797			45	2626.8		
6	353			26	1818.8			46	2667.7		
7	360.2			27	1827			47	2730.9		
8	380.3			28	1852.3			48	2758.8		
9	389.9			29	1956.9			49	2808.8		
10	417.1			30	1994.1			50	2817.8		
11	603.7			31	2088.7			51	2821.3		
12	635.7			32	2182.3			52	2835.2		
13	716.9			33	2189.4			53	2846.8		
14	723.2			34	2211.5			54	2857.4		
15	768.8			35	2222.1			55	2852.2		
16	1069.3			36	2366.4			56	2956.6		
17	1330			37	2382.4			57*	2965.2	2.614	1.146
18	1337.6			38	2390.7			58	2986.9		
19	1558.6			39	2420			59*	3001	3.999	1.833
20	1582.4			40*	2509	1:239	0.4629	60	3007.3		

O ₁	TSN	R ₁	%F	O ₁	TSN	R ₁	%F	O ₁	TSN	R ₁	%F
61	3016.3			81	3311.2			101	3693.1		
62	3052.9			82	3314.8			102*	3698.1	14.002	6.80
63	3063.9			83	3343.6			103	3725.6		
64	3085.9			84*	3345.5	8.792	4.214	104*	3785.3	15.901	7.74
65	3088.3			85	3356.4			105	3792.3		
66	3126.3			86	3384.8			106	3795.0		
67	3170.9			87*	3390.8	10.457	5.04	107	3797.6		
68	3177.1			88*	3402.5	12.122	5.87	108	3819.8		
69	3178.8			89	3404.9			109*	3844.8	17.881	8.73
70	3180.3			90	3424.7			110*	3877.9	19.861	9.71
71	3182.7			91	3507.1			111*	3879.1	21.841	10.69
72	3209.8			92	3535.8			112	3885.2		
73	3224			93	3545.4			113	3891.5		
74	3230.2			94	3563.8			114*	3892.7	23.86	11.70
75	3233.1			95	3574.9			115	3895.5		
76*	3261.3	5.558	2.912	96	3608.6			116*	2924.9	25.91	12.71
77	3284.2			97	3669.8			117	3930.6		
78	3290.8			98	3682.5			118*	3995.8	27.98	13.74
79	3290.9			99	3684.2			119	4013.3		
80*	3303.3	7.155	3.401	100	3693.1			120	4020.6		

O ₁	TSN	R ₁	ZF	O ₁	TSN	R ₁	ZF	O ₁	TSN	R ₁	ZF
121	4034.7			141	4141.9			161	4279.4		
122*	4036.4	30.13	14.81	142	4154.5			162	4280.5		
123*	4039.1	27.98	15.87	143*	4158.8	56.30	27.80	163	4284.5		
124	4039.4			144	4173			164	4297.3		
125	4040.2			145*	4175.6	58.81	29.05	165	4301.9		
126*	4046.4	34.48	16.97	146	4178.8			166*	4322.7	76.1	37.63
127	4048.6			147	4189			167*	4361.3	79.6	39.37
128*	4049.8	37.71	18.08	148*	4198	61.41	30.34	168*	4366.5	83.1	41.11
129*	4067.1	39.93	19.67	149	4203.6			169	4368.4		
130	4069.1			150	4209.7			170	4371.3		
131*	4072.2	42.21	20.81	151	4222.8			171	4390.8		
132	4073.4			152*	4227.3	64.17	31.71	172*	4392.7	86.93	43.02
133*	4081.8	44.49	21.94	153	4230.1			173*	4398.6	90.76	44.92
134	4082.1			154*	4239.3	66.98	33.11	174*	4405.8	94.6	46.82
135*	4083.6	46.81	23.09	155*	4242.1	69.79	34.50	175*	4410.8	97.98	48.50
136*	4091.7	49.13	24.24	156*	4251.2	72.60	35.90	176*	4424.5	101.83	50.41
137*	4093.7			157	4259.6			177*	4428.2	105.68	52.32
138*	4105.5	51.48	25.41	158	4265.3			178	4438.6		
139*	4119.8	53.83	26.58	159	4270.8			179	4454.8		
140	4138.3			160	4276.8			180	4473.8		

O ₁	TSN	R ₁	%F	O ₁	TSN	R ₁	%F	O ₁	TSN	R ₁	%F
181*	4482.7	110.06	54.50	188*	4542.6	125.0	61.92	195	4858.9		
182	4487			189	4546.2			196	4909.6		
183*	4517.5	114.66	56.78	190	4551.4			197*	4935.8	144.25	71.48
184	4518.2			191	4628.2			198	4946.1		
185*	4523.8	119.50	59.19	192	4739.4			199	5066.4		
186	4532.3			193	4804.1			200*	5247.9	163.5	81.03
187	4541.0			194	4813.6			201	5469.1		

$\%F = AR_1 - B$ where in accordance with eq. (12).

$$A = \frac{100}{200-1} \left[2 \left(\frac{1}{201} \right) - 1 \right] = 0.49656$$

and

$$B = 100 \left[\frac{2 \left(\frac{1}{201} \right)}{201-1} + 2 \left(\frac{1}{201} \right) - 1 \right] = 0.15230$$

Note: * indicate data points from discs where cracks 0.175" or greater had formed during service.

VOLUME 2

ANALYSIS OF J85-CAN15 TURBINE MATERIALS

Abstract

Volume 2

In this volume, a study has been made to microstructure-environment interactions during elevated temperature (650°C) crack growth in Inconel 718, which is used as a turbine disc material for the J85-CAN15 and F-404 engines. Static and cyclic loading conditions ($R = 0.05$, $f = 0.1$ Hz) have been studied using a tapered double cantilever beam (constant-K) fracture mechanics specimen. Four microstructures with a wide range of grain sizes, precipitate particle sizes and grain boundary morphologies were studied. It has been attempted to relate observed differences in crack growth behaviour to possible differences in the micromechanisms of deformation and fracture. Particularly, it has been emphasized that the sensitivity of this alloy to environmental embrittlement is enhanced by a fine grain size and a hyperfine γ'' distribution. Heat treatments which lead to a serrated grain boundary morphology have been shown to improve crack growth resistance.

The feasibility of applying a damage tolerance based lifting technique to Inconel 718 turbine components has been discussed.

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(a) Conventional

(b) Modified

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 $\Delta K = 90 \text{ MPa } \sqrt{\text{m}}, T = 650^\circ\text{C}$

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(a) Conventional

(b) Modified

(c) Merrick

(d) Modified Merrick

2.53 Typical Microstructures Produced by Reheat
Treatments

Nomenclature

a	crack length
a_i	initial crack length
a_c	critical crack length
a_{insp}	apparent crack depth
A_i, A_p	constants in deterministic equation
B	specimen thickness
B_n	specimen thickness at sidegroove
C_i, C_p	random variables
C, n	material constants in Paris Law
D	cross head displacement
f	frequency (Hz)
$f(t_i)$	failure probability at time t_i
F, P	load (kN)
$F(t)$	cumulative failure probability
$h(t)$	Hazard Rate
H_p, h, H_w	DCB specimen dimensions
K, K_I	stress intensity factor
k_a	standard normal deviate
l	lower bound or origin of Weibull distribution
M	moment of inertia (m^4)
N	RPM
N_i	crack initiation life (hours)
N_d	life to dysfunction (hours)
N_f	life to failure
N_{pr}	crack propagation life

N_{PR}	residual life time
N_t	inspection time
O_i	order number of i^{th} failure in Weibull analysis
$P(F)$	component failure probability
PN_R	probable number of rejection sites in a component
PN_F	probable number of failure sites or indications in a component not rejected and causing failure
R	stress ratio
R_t	reliability function
R_y	cyclic plastic zone radius
S	standard deviation
SF	safety factor
t_f	failure time
ω_1	yearly flying rate
W	specimen width
z_i	total number of LCF cycles
B	Weibull slope or shape parameter
σ	applied stress
$\sigma_{min}, \sigma_{max}$	maximum and minimum stresses
σ_{ys}	yield strength
σ_{yc}	cyclic yield stress
$\Delta\sigma$	stress range
ϕ	geometry factor
μ	mean
η	characteristic life or scale parameter
δ	load line displacement

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VOLUME 2

CHAPTER 1

ANALYSIS OF J85-CAN15 TURBINE MATERIALS

PHYSICAL METALLURGY OF INCONEL 718

Inconel 718 is used as a turbine rotor material in the General Electric J85-CAN15 engines for the CF-5 aircraft and the F-404 engines which power the CF-18 Hornet. The development of a damage tolerance based lifing procedure for these components requires a thorough understanding of the physical metallurgy and elevated temperature fracture behaviour of Inconel 718. Various aspects of the physical metallurgy and mechanical behaviour of Inconel 718 are discussed here.

This alloy is a nickel-base precipitation-hardenable nickel-base superalloy developed for use at temperatures up to approximately 700°C.¹ Inconel 718 differs from other nickel-base superalloys in that (1) columbium is substituted for much of the aluminium and titanium, and (2) it has a high percentage of iron, Table 1.1. These differences produce a high strength material with reduced temperature capabilities compared to some nickel-base superalloys but greatly enhanced fusion welding characteristics.²

This alloy has found application in gas turbine engines, rocket motors and nuclear reactors. In most cases, it is utilized in the welded condition to take advantage of its excellent weldability. Integral turbine rotors of Inconel 718 are fabricated by the General Electric Company through the inertia welding process, according to the sequence shown in Fig. 1.1.²

1.1 Microstructure

Like most superalloys, Inconel 718 is a complex alloy involving precipitation of several phases. In the conventionally heat treated condition, it contains the gamma phase (the matrix material), a coherent body-centred tetragonal precipitate referred to as gamma double prime (γ'') and also a coherent face-centred-cubic phase known as γ' which is found in most nickel-base precipitation hardenable alloys.

γ' and γ'' have the basic compositions $\text{Ni}_3(\text{Al}, \text{Ti})$ and Ni_3Cb respectively. Small amounts of CbC and TiN are present also. In addition to these constituents, other phases may appear because of segregation from casting, improper heat treatments or high temperature exposure in service. The time-temperature-transformation diagram shown in Fig. 1.2 illustrates the conditions under which these constituents may form. A description of each of these phases is given below.

Gamma (γ): Gamma is the matrix material and consists of the solid-solution elements nickel, chromium, iron, columbium, molybdenum, titanium and aluminium. It has an FCC structure and is comparatively weak in its solution annealed condition (280 - 420 MPa yield strength).

Gamma Prime (γ'): This phase has an ordered FCC structure and the unit cell, shown in Fig. 1.3 consists of Cb, Al and Ti atoms at the corners and Ni atoms at the center of the faces. The approximate composition of this phase has been reported by Eiselstein,¹ as: 75 Ni, 15 Cb, 5 Ti and 5 Al.

The gamma prime precipitate particles are usually spherical and appear as a fine dispersion within the γ matrix. They appear to form uniformly throughout the matrix during aging. They are coherent with the matrix and begin to form at 650°C after about 50 minutes of aging. The solvus temperature is between 843°C and 871°C.³ The most rapid phase formation is between 732°C and 760°C.^{1,4} The γ' precipitate contributes only about 10-20 percent to the hardening of the alloy and it is believed that γ' transforms to γ'' upon exposure to temperatures around 788 to 815°C for 1h.^{1,5,6}

Gamma Double Prime (γ''): Gamma double prime, like γ' is an A_3B compound. However, it has an ordered BCT structure which is illustrated in Fig. 1.4. In γ'' , Cb atoms are located at the corners and in the center of a cell formed by placing two fcc cells one upon the other.⁷ γ'' transforms to orthorhombic Ni_3Cb after exposure at temperatures above 815°C.¹⁻¹² This transformation can be accelerated by deformation.¹³

The γ'' precipitate forms as discs with an average diameter of 600 Å and a thickness of 50 to 90 Å with normal exposures at temperatures between 650°C and 760°C.³ Above 760°C, it exhibits accelerated coarsening. Since it has been proved conclusively that the major strengthening effect in Inconel 718 comes from this metastable, body centred tetragonal precipitate,^{9,10} the amount and morphology of this phase will have an important effect on the mechanical properties.

Paulonis et al.³ have shown that γ' and γ'' precipitation is relatively sluggish in this alloy. Transmission electron microscopy

showed that a supersaturated solid solution could be retained upon air cooling from 955°C. No visual evidence, superlattice reflections or dislocation pairs indicative of an ordered precipitate could be observed. This is quite different from the behaviour of most γ' strengthened nickel-based superalloys, in which the precipitation of γ' often cannot be suppressed even by water quenching. Paulonis et al.³ also found some evidence indicating that the precipitation of γ' and γ'' may occur through an interdependent nucleation mechanism in which one of the ordered phases nucleates directly upon the other.

Ni₃Cb: This phase has an ordered orthorhombic structure. According to Boesch,⁵ the composition of this phase is Ni₃(Cb.8 Ti.2).

Ni₃Cb forms as needles in the grains and at the grain boundaries, Fig. 1.5. Although Ni₃Cb is considered quite stable, it is not coherent with the matrix. Therefore, it contributes little or nothing to the strength and may harm ductility.^{1,11} However, small amounts of Ni₃Cb at the grain boundaries give rise to a serrated grain boundary morphology as shown in Fig. 1.5 and this improves notch rupture ductility at 650°C.¹²

At temperatures above 982° or 1037°C, Ni₃Cb generally dissolves.¹¹ It appears that heat to heat variations can cause significant variations in the time taken for Ni₃Cb formation in the 760°C to 982°C temperature range. Thus, while Eiselstein¹ reported that it begins to form after 5 h at 926°C, Boesch and Canada⁵ have found that it begins to form after only 15 to 20 min. at 926°C. The start of formation of Ni₃Cb needles is also affected by prior solution

anneal temperature (Fig. 1.6). This is believed to be due to the effect of annealing temperature upon grain size, grain boundary area and the number of grain boundary nucleation sites available.

Laves: This is an A_2B compound which is identical in structure to Fe_2Ti . The laves phase consists mostly of nickel (40-45%), columbium (25%), iron (10%), chromium (10%), molybdenum (10%) and possibly a little titanium. The A atoms are probably Ni, Fe and Cr while the B atoms are probably Cb, Mo and Ti.

Laves phase appears as round, massive particles and is found in heavily cored or segregated areas.¹ Under the microscope, the phase appears as bright, white particles. In Fig. 1.7 it may be seen as the agglomerated phase in association with the Ni_3Cb needles.

Laves phase can form during solidification and during heat treatment. During solidification, it forms interdentritically in segregated areas at relatively low temperatures of 1148° to 1204°C.¹ During heat treatment, at times to 10 h, it can form at low temperatures of 787° to 843°C. This phase is definitely harmful to mechanical properties since it impairs ductility. In addition, it removes alloying elements required for precipitation hardening, and therefore, tends to reduce strength. It is removed by hot working at temperatures above 1093°C.¹¹ Recent developments indicate that the presence of Laves phase can be completely avoided in powder metallurgy processed Inconel 718.¹⁶

CbC and TiN: Titanium nitride forms in very small amounts during solidification and hence is not of much concern.¹¹ It appears under the microscope as a pink phase, massive and angular in shape. It is believed to be stable up to the melting point.^{1,4}

Columbium carbide, however, forms in larger quantities and is apparently not as stable as TiN.¹¹ It forms on solidification as massive, angular particles. These carbides, which appear grey under the microscope, have a tendency to remove columbium and thus reduce strength. It can form in large amounts at grain boundaries.¹

CbC dissolves on annealing above 1038°C and will precipitate on aging from 704° to 982°C as a grain boundary film.¹⁻¹² While it appears that moderate amounts of this phase do not affect the mechanical properties,¹ excessive amounts should be avoided due to the film-like morphology.

M₆C: This form of carbide phase, which can also form in the grain boundaries, is considered to be rather harmless because of the small amounts present and the form in which it precipitates.⁷ It forms sluggishly, in comparison to other phases discussed and forms on long-time exposure to temperatures ranging from 704° to 982°C.

1.2 Factors Influencing the Microstructure and Mechanical Properties of Inconel 718

1.2.1 Chemical Composition

Nickel: The yield strength of Inconel 718 is maximized at a nickel content of about 53 percent.^{7,17} ~~Stress rupture strength is somewhat~~

higher at greater nickel contents, but can only be achieved at the expense of strength.

Chromium: This element promotes oxidation resistance and strengthens the matrix. However, at a level greater than 19 wt. percent, chromium tends to reduce hot workability.¹¹

Molybdenum: Molybdenum in Alloy 718 acts much like chromium. It helps the high temperature strength by solid-solution hardening and since it remains in solid-solution, it does not precipitate to any large extent. Greater than 3 wt. percent in this alloy would impair workability.

Columbium: Columbium in Inconel 718 has a potent effect in increasing strength.⁴ Conversely, ductility is decreased by columbium, particularly in amounts above 5 or 6 percent. As mentioned earlier, columbium forms the γ'' precipitate which is the most significant strengthening constituent.

Titanium: Titanium in Inconel 718 acts much like columbium. It forms the γ' precipitate and raises strength while reducing ductility. However, the effect of titanium is dependent on the aluminium and carbon content, as well as upon the heat treatment variables.^{1,4} These interactions are thought to be a result of effects upon the strengthening precipitates. Stress rupture life reaches a maximum near 1 percent titanium.¹¹

Aluminium: The effect of aluminium on mechanical properties is dependent on titanium and carbon content and heat treatment variables even more than for titanium. While aluminium contributes to the formation of the γ' strengthening precipitate, above the nominal 0.5 wt.

percent, the yield strength of the alloy is decreased. Below 0.5% Al, a similar trend is observed.^{1,4} Ductility and stress rupture life are not improved by aluminium. However, this element may serve to retard the overaging of the γ' .¹⁸

It has been found desirable to choose higher Ti and lower Al rather than a 1:1 ratio.⁷ A common specification is 0.4 to 0.8% Al and 0.65 to 1.15% Ti. For tension limited applications, a recent specification is 0.2 to 0.6% Al and 0.7 to 1.4% Ti.⁷

Minor elements: Carbon will primarily be present as columbium carbide, CbC. Since this carbide takes columbium away, the presence of carbon decreases the strength of Alloy 718.⁴ Increased carbon may also decrease stress rupture life.¹ Thus, carbon is kept at a low value of 0.5 wt. pct. The CbC precipitated at these levels may provide a beneficial effect as a grain boundary pinning agent.

Boron, at a level of 0.005 wt. pct. provides optimum stress rupture life.¹ Boron additions appear to be most effective with low Al and high Ti. Silicon should be kept well below 1 pct. since this element drastically harms hot workability. Increased silicon also reduces stress rupture life,¹ perhaps by embrittling the solid-solution matrix. Nitrogen is found as TiN. Like carbon, nitrogen should be minimized because it will take titanium away from the gamma prime.

1.2.2 Heat Treatment

Commercial heat treatments for Inconel 718 usually consist of a solution anneal in the temperature range 955-1093°C followed by aging at lower temperatures.

High solution treatment temperatures, in the vicinity of 1150°C are to be avoided. Such a high temperature would take almost all second phases into solution and lead to excessive grain growth which would be harmful to strength. Furthermore, excessive amounts of film-like CbC may precipitate upon subsequent aging.

For a single age, maximum strength can be attained by heating at 704-760°C for 10-20 h.¹ Lower aging temperatures can reach higher hardnesses but require excessive time.

A double age, wherein the second aging temperature is about 100°C below the first aging temperature, increases strength by 70-140 MPa. Furthermore, the ductility is not appreciably decreased by the second age. A very slow cool (~10°C/h) from the first aging treatment will produce the same effect as a double age.

The "conventional" heat treatment applied to Inconel 718 consists of a solution anneal at 955°C and a double age at 718 and 621°C (Table 1.2). This low annealing temperature does not dissolve any CbC, not too much of the Ni₃Cb originally present, and a fine grain size (~25 μm) is retained.

Aging at 718°C gives best strength⁴ and continued aging at 612°C increases strength. No CbC films form because CbC is not dissolved on annealing. However, as shown in Fig. 1.8, Ni₃Cb needles are present. Apparently, the low temperature anneal does not dissolve all the Ni₃Cb. This structure would not be ideal for tensile properties. However, a broken-up large second phase in grain boundaries, such as the Ni₃Cb needles shown in Fig. 1.8 may be

beneficial to stress-rupture properties. In creep, such needles may inhibit grain boundary sliding.

The "modified" heat treatment [Table 1.2] produces a significantly coarser grain size ($\approx 200 \mu\text{m}$) than the conventional heat treatment (Fig. 1.9). It is believed that a more homogeneous microstructure is produced by this heat treatment. The fracture toughness and fatigue crack growth resistance of Inconel 718 are improved by the "modified" heat treatment.^{19,20} The lower fracture toughness of the "conventionally" treated material is believed to be due to the presence of coarse Ni_3Cb particles which enhance the premature nucleation and coalescence of microvoids. Such particles are absent in material given the modified heat treatment.

The γ'' particle size distributions produced by the "conventional" and "modified" treatments were investigated by Mills²¹ and the results are shown in Fig. 1.10. The range of γ'' diameters resulting from the modified heat treatment (90-700 Å) was considerably larger than that resulting from the "conventional" heat treatment (80-450 Å). Figure 1.10 also illustrates that the "modified" alloy exhibits a higher percentage of large γ'' platelets (diameters in excess of 300 Å) but essentially the same platelet width in comparison to the conventional γ'' particle size, indicating that γ'' coarsening during the "modified" heat treatment occurs essentially along the diameter. The coarser grain size and the coarser γ'' particle size produced by the "modified" heat treatment lead to a reduction of about 100 MPa in the

650°C yield strength relative to material given the "conventional" heat treatment.

The Wilson heat treatment (Table 1.2) which uses a low annealing temperature of 926°C for 10 h results in larger size precipitates. The somewhat overaged microstructure is believed to improve time dependent notch sensitivity.²² Sadananda and Shahinian²³ found that this treatment lowered creep crack growth rate in Alloy 718 at 650°C by two orders of magnitude in relation to other known treatments. This improvement is believed to result from precipitation of a significant amount of Ni_3Cb which strengthens the grain boundaries.

Notch rupture ductility is improved by Raymond's heat treatment²⁴ also, which employs an intermediate age at 955°C. The intermediate age avoids the formation of precipitate free zones along the grain boundaries. Created by the rapid precipitation of CbC , these zones are known to be harmful to notch ductility.

Another heat treatment to improve notch rupture properties in Inconel 718 is due to Merrick²⁵ and employs an intermediate age at 843°C (Table 1.2). The conditions under which Ni_3Cb precipitates during this heat treatment lead to a "wavy" or "zig-zag" grain boundary morphology. This in turn minimizes the extent of grain boundary sliding, increases the resistance to crack propagation and thereby confers ductility under the conditions of notched stress rupture testing. This intermediate aging can, however, result in a significant lowering of tensile and yield strengths, as a result of γ coarsening.

The zig-zag grain boundaries produced in Inconel 718 are angular in nature and hence are quite different from the well-rounded serrations formed in other nickel-base superalloys as a result of slow cooling through the γ' solvus temperature.²⁶

Muller and Donachie²⁷ have proposed another heat treatment which is similar to that reported by Wilson²² in the sense that considerable γ -phase precipitation is allowed to occur. The 1038°C solution anneal permits better stress relief and recrystallization than the "conventional" heat treatment while the 917°C/10 h intermediate age precipitates δ -phase in the matrix and in the grain boundaries and improves notch rupture ductility.

Serrated grain boundaries can be formed in Inconel 718 by slow cooling from the solution annealing temperature as well, provided that the material stays long enough in the temperature range for δ -phase precipitation.²⁸ Slow cooling may also avoid the formation of precipitate free zones along the grain boundaries. However, the cooling rate should be optimized so as not to cause excessive intragranular γ'' coarsening.

1.2.3 Thermomechanical History

A uniform, fine grain size is generally preferred in forgings of Inconel 718, but due to limitations imposed by equipment, forging size and forging configuration, it is often possible to produce this desired grain size; instead some grain duplexing or coarsening occurs in certain areas throughout the forging. The influence of grain size on mechanical properties is discussed below.

Inconel 718 exhibits the familiar effect of an increase in rupture ductility and a decrease in rupture strength with fineness of grain size in uniform, fully recrystallized grain specimens, Fig. 1.11. The increased rupture ductility at fine grain sizes would partly be a result of an increased resistance to cavity linkage due to the greater number of grain boundary triple points present. Further, σ -phase particles form more readily in fine grained microstructures and act as grain boundary strengtheners.²⁷ The most desirable uniform grain size for optimum properties appears to be approximately ASTM6-7. The following results regarding the effect of grain size were reported by Stroup and Heacox.²⁹

When a large percentage of coarse grain exists in duplexed structures (i.e., about 90 percent), notch brittleness occurs. This is perhaps related to the nature of Ni_3Cb precipitation which occurs during heat treatment. The size of Ni_3Cb plates is directly proportional to grain size; larger plate sizes are associated with larger grain regardless of whether they are recrystallized or non-recrystallized grain. The increased Ni_3Cb plate size is known to reduce ductility and promote notch brittleness.

Decreasing the volume of large Ni_3Cb plates further increases the tendency toward notch brittleness. Higher forging temperatures ($\approx 1120^\circ\text{C}$) promote a larger grain size and decrease Ni_3Cb plate volume. Thus, higher forging temperatures lead to lower rupture ductility and a greater tendency toward notch brittleness.

Grain boundary films of CbC are most noticeable in specimens forged from high ($>1120^{\circ}\text{C}$) temperatures. In duplex-grained samples, the largest percentage of microcracking has been observed at the boundaries of the larger grains suggesting that fracture could initiate at these boundaries at much lower stresses than were necessary at the boundaries of the fine grains.

1.2.4 Service Exposure

Barker et al.³⁰ found that the fine matrix precipitates of F.C.C. gamma prime coarsen upon prolonged exposure to stress and temperature ($650-700^{\circ}\text{C}$) to reveal particles having three distinct shapes - spheroidal, small plates and large plates. The spheroidal particles have the characteristic shape of γ' , the small plates have the shape and orientation of bct γ'' while the large plates form in a Widmanstatten pattern similar to orthorhombic Ni_3Cb . The Ni_3Cb which forms during the long time exposure is thought to be of different chemical composition from the Ni_3Cb platelets formed during initial heat treatment.

Small quantities of sigma phase and another phase designated as α' have been found to form as a result of prolonged stress-temperature exposure. The α' phase is a chromium rich solid solution and being precipitated as discrete particles, it does not impair the ductility of the alloy. Moderate amounts of Laves phase were found at about the same conditions that α' and σ were formed.

Significant overaging of the γ' and γ'' precipitates does not occur after 10,000 h at 650°C/255 MPa or in 200 h at 704°C/255 MPa but does occur at 704°C/255 MPa in 800 h.³⁰ Inconel 718 can therefore be subjected to stressed exposure to temperatures up to 650°C for at least 10,000 h without serious instabilities occurring. Exposures for significantly longer times at 650°C and proportionately shorter times at higher temperatures will probably result in phase changes that will reduce the tensile strengths, however; tensile ductility is not impaired significantly. Cozar and Pineau³¹ have determined that the rate of coarsening of the γ' and γ'' particles can be reduced significantly by causing them to precipitate in a "compact morphology". This precipitation consists of cube-shaped γ' particles coated with a γ'' shell over their six faces (Fig. 1.12). A minimum Ti + Al/Nb ratio between 0.9 and 1 has been observed to be necessary to achieve this. Moreover, the aging temperature has to be chosen in such a way that initially isolated γ' precipitates reach a size of about 200 Å before production of γ'' precipitates. However, it should be noted that the standard composition of Inconel 718 has a Ti + Al/Nb ratio of only about 0.4, and hence it may not be possible to achieve the "compact morphology" in standard compositions.

A grain boundary denudation effect has been found to occur in most cases of long time exposure in which Ni_3Nb was formed at the grain boundaries.²⁹ However, this effect does not promote notch brittleness.

The grain size is also an influential factor in determining the mechanical properties subsequent to prolonged service exposure.²⁹

As mentioned earlier, very large Ni_3Cb plates can form in coarse grained material and contribute to poor notch behaviour. In finer grained material, rupture ductility may actually be improved significantly after exposure due to weakening of the matrix by the $\gamma' \rightarrow \gamma'' \rightarrow \text{Ni}_3\text{Cb}$ transformation.

Rizzo and Buzzanell³² have studied the effect of chemical composition on long term exposure at 650°C for 1500 h. Increasing columbium content showed strong increases in tensile and yield strengths while increasing titanium content showed moderate increases in the strength. Increased aluminium caused significant decreases in the tensile strength after exposure. Silicon content had no apparent effect on strength in the exposed condition.

Increasing columbium and silicon content showed moderate decrease in the elongation and reduction of area values of exposed samples. Titanium and aluminium showed no effect upon the elongation but increasing amounts of these elements did cause a slight decrease in the reduction of area values.

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Chapter 1

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TABLE 1.1
CHEMICAL COMPOSITION OF INCONEL 718

Composition (weight %) for Element:								
C	Cr	Ni	Cb	Mo	Ti	Al	B	Fe
.05	19	53	5	3	1	0.5	0.005	18

TABLE 1.2
HEAT TREATMENTS FOR INCONEL 718

Heat Treatment Name	Schedule
"Conventional"	955°C/1h/AC + 718°C/8h $\xrightarrow{1 \text{ C/min} \text{ FC}}$ 621°C, hold for a total aging time of 18h, AC
"Modified"	1093°C/1h $\xrightarrow{1 \text{ C/min} \text{ FC}}$ 718°C/4h $\xrightarrow{1 \text{ C/min} \text{ FC}}$ 621°C/18h/AC
"Merrick"	1066°C/1h + 843°C/4h + 718°C/8h $\xrightarrow{1 \text{ C/min} \text{ FC}}$ 621°C/10h/AC
"Modified Merrick"	1066°C/1h $\xrightarrow{7 \text{ C/min} \text{ FC}}$ 843°C/4h + 718°C/8h $\xrightarrow{1 \text{ C/min} \text{ FC}}$ 621°C/10h/AC
Wilson	927°C/1h/AC $\xrightarrow{\text{FC}}$ 734°C/48/AC
Raymond	1066°C/1h $\xrightarrow{1.5 \text{ C/min} \text{ FC}}$ 955°C/1h/AC + 760°C/5h $\xrightarrow{1.5 \text{ C/min} \text{ FC}}$ 650°C/8h/AC
Müller & Donachie	1035°C/0.3h $\xrightarrow{\text{FC}}$ 927°C/10h/AC + 720°C/8h $\xrightarrow{\text{FC}}$ 620°C/18h/AC

Fig. 1.1

Schematic Showing the Approach to the Fabrication of an Integral Rotor²

Fig. 1.2

Time-Temperature-Transformation Diagram
for Inconel 718³

● Cb, Al, Ti
○ Ni

Fig. 1.3

Unit Cell of γ'

● Cb, Al, Ti
○ Ni

Fig. 1.4

Unit Cell of γ''

Fig. 1.5

Serration at a Grain Boundary due to Ni_3Cb Precipitation

Fig. 1.6

Influence of Solution Anneal Temperature Upon
Precipitation in Inconel 718¹

Fig. 1.7

Laves Phase in Inconel 718

A

B

50 μm

Fig. 1.8

Microstructure of Conventionally Heat Treated
Inconel 718

Fig. 1.9

—Microstructure of Inconel 718 given "Modified"
Heat Treatment

Fig. 1.10

Difference in γ'' Particle Size Distributions
Produced by "Conventional" and
"Modified" Heat Treatments²¹

Fig. 1.11.

Influence of Grain Size in Rupture Ductility²⁸

Fig. 1. 12

"Compact Morphology" for γ' and γ''
Precipitation³⁰

CHAPTER 2

THE INFLUENCE OF MICROSTRUCTURE AND ENVIRONMENT UPON ELEVATED TEMPERATURE CRACK GROWTH RATES IN INCONEL 718

The influence of heat treatment upon the microstructure of Inconel 718 was discussed in detail Chapter 1. In this section, the resistance of various microstructures to elevated temperature creep and fatigue crack propagation has been investigated. A fracture mechanics approach has been used and the heat treatments selected for this study are such that a wide range of grain sizes, γ'' particle sizes and grain boundary morphologies are covered. The following objectives were sought:

- a. to compare the crack growth resistance of various microstructures at 650°C under static loading as well as constant amplitude cyclic loading conditions at low frequency (0.1 Hz), and determine the feasibility of applying a damage tolerance based lifing technique for Inconel 718 turbine components.
- b. to determine the relative extent to which the presence of an oxygen containing environment accelerates crack growth rates in these microstructures and thereby determine the optimum combination of microstructural features in Inconel 718, for elevated temperature service as a gas turbine disc material.
- c. to identify and understand the micromechanisms of deformation and fracture which give rise to the observed macroscopic crack growth rates.

The material, specimen geometry, test equipment, procedure and the results are described as follows. Elevated temperature tensile and stress rupture data have also been reported in support of the crack growth rate measurements.

2.1 Material

The as-received material was in the form of mill annealed plate, having the microstructure shown in Fig. 2.1(a). The chemical composition of the as-received material reported by the vendor (Aero Space Materials Inc.) is shown in Table 2.1. In this as-received condition, Inconel 718 is obviously fine grained. It was also observed that a number of carbide particle stringers were present, oriented parallel to the rolling direction. The orientation of these stringers with respect to the stress axis in a test specimen or component is very important, since crack propagation rates are usually higher in a direction parallel to the orientation of the stringers. The fracture mechanics specimens used in this study were machined such that the line of crack propagation is parallel to the rolling direction, thus providing a "worst case" estimate of crack growth rates. Prior to specimen machining the specimen blanks were given one of the four heat treatments shown in Table 2.2.

The microstructures produced by these heat treatments are shown in Fig. 2.1(c-f) and it may be seen that the finest and coarsest grain sizes are produced by the Conventional and Modified heat treatments, respectively. The Merrick and the Modified Merrick heat treatments produce the same grain size, which is intermediate between

those resulting from the Conventional and Modified heat treatments (Table 2.3).

It may also be seen from Fig. 2.1 that the grain boundaries are planar for the Conventional, Modified and Merrick heat treatments. The latter treatment was mentioned in Chapter 1 as having potential to form "irregular" or "serrated" grain boundaries. However, for the heat of Inconel 718 used in this study, the Merrick heat treatment resulted in a planar grain boundary structure. A modification of the Merrick treatment is the "Modified Merrick" schedule in which a direct age at 843°C is employed subsequent to the solution anneal at 1066°C. This modification led to the formation of pronounced, angular grain boundary serrations (Fig. 2.1(f)).

The γ'' particle size distributions produced by these heat treatments are shown in Fig. 2.2. The finest γ'' distributions are produced by the Conventional and Modified Merrick heat treatments, while the Merrick heat treatment produces a relatively coarse γ'' distribution. The γ'' particle sizes in the Modified heat treatment appear to be at an intermediate level. It also appears that some localised matrix γ'' coarsening occurs in the case of the Modified and Modified Merrick heat treatments, Fig. 2.2(e,f). This effect is more pronounced in the case of the Modified heat treatment compared to the Modified Merrick heat treatment.

2.2 Specimen Geometry

A tapered double cantilever beam specimen has been used in this study. This type of specimen can produce approximately constant K

or ΔK conditions over a certain portion of the useful crack length and has been used extensively by various investigators to study stress corrosion cracking in a variety of alloys¹⁻⁴. The several advantages of using a constant-K specimen for an investigation of the type intended here are listed below,

1. Crack growth rates can be measured conveniently by simply dividing the length of crack propagation (measured after the test, by scanning electron microscopy) by the number of load cycles (under cyclic loading) or by the number of hours of elevated temperature exposure (for static loading), provided that the crack extension occurs over the portion of the specimen in which constant-K conditions exist.
2. In a compact tension specimen which has been used by most investigators in evaluating elevated temperature creep-fatigue crack growth rates, the stress intensity factor rises continuously with crack length and hence the propagation rates for a certain stress intensity factor are measured based on the behaviour of the material over a very small crack extension.

In a constant-K specimen, crack extension is allowed to occur over several millimeters and hence, the crack growth rates measured are an average of the behaviour of the material over a significant crack length. In essence, a constant-K test amounts to conducting several tests with a rising-K specimen and averaging the results. Much less scatter in data may therefore be expected with data generated using constant-K specimens than with compact tension specimens.

3. In a number of cases, other methods of crack growth rate measurement may be difficult or even impossible. For example, optical measurement of crack growth rates has the following disadvantages.
 - a. large man-hour requirement
 - b. does not measure average crack front location unless crack front is straight
 - c. requires smooth and highly polished surfaces (very difficult to achieve in a side-grooved specimen)
 - d. requires visual access to the specimen face and hence impossible to use if the specimen is contained within a large resistance heated furnace and an environmental chamber as well.

Compliance methods or the potential drop technique can present problems as well, when used with a rising-K specimen.

1. The instrumentation needed for compliance measurements such as clip gages and capacitive transducers are usually limited by temperature and to the author's knowledge, such instrumentation is not available for use at temperatures above 760°C .
2. Extensometers to measure Load-line or Crack Opening Displacement may be useful at higher temperatures but due to the several mechanical linkages present in these instruments the accuracy and sensitivity of measurements is usually poor.
3. Precise measurement of crack length by the compliance method would be difficult at frequencies above 0.5 Hz^5 .
4. The compliance vs. crack length relationship for a rising-K specimen is highly non-linear⁵.

5. In the presence of a very corrosive environment it would be impossible to use all-metal extensometers for crack growth rate measurement.
6. The production of conductive corrosion debris can lead to errors in applying the potential drop technique.
7. The electric potential drop technique is also likely to produce errors in crack length measurement if conductive debris is formed between the crack faces, in the presence of corrosive environments.

Though hold time effects associated with cyclic loading have not been investigated in this study, it is worth mentioning that such effects are studied much more accurately and conveniently with a constant-K specimen since the hold time can be applied at a certain constant-K. If this is not the case it becomes very difficult to compare the results of various investigators.

For a number of such reasons, it was decided to conduct the tests required for this study under constant-K conditions, using a constant-K, tapered double cantilever beam specimen.

The use of a tapered DCB specimen is not the only way to achieve constant-K conditions. A recent idea⁶ has employed the fact that for a centre crack in a finite sheet, K increases with the crack length if a constant load is applied at the ends of the sheet, but decreases if the load is applied at points on the crack surfaces midway between the crack tips. By applying point loads through pins a small distance from the crack plane, a balance is found between these two effects.⁵ However, this type of specimen has the advantage that for

many materials, only modest values of K are obtained before the maximum allowable pin loads are reached or crushing of the specimen at the loading point begins.

If a servo-hydraulic rig is used for applying loads, and if, there is some means of monitoring the length of the crack, the load can be reduced as the crack grows in such a way as to keep ΔK , and therefore the growth rate constant. Monitoring can be done optically using a traveling microscope, in which case, the load is shed manually⁷, or using the electric potential drop technique in which case automatic control may be applied using a computer.⁸ A standard specimen geometry such as the compact tension specimen may be used. But, optical monitoring and manual load shedding can be, as mentioned earlier, cumbersome and difficult. On the other hand, the potential drop technique is said to suffer from stability and interference⁵ and can lead to difficulties in corrosive environments.

2.3 Design of Specimen Geometry

In earlier investigations¹⁻⁴ a number of tapered DCB geometries have been designed (Fig. 2.3). But as mentioned earlier, most of these were designed for the study of stress corrosion cracking and very few for the measurement of elevated temperature crack growth rate determination. Most of the geometries shown in Fig. 2.3 have one common disadvantage from the point of view of application to the study of elevated temperature crack growth behaviour in superalloys - that they are quite large in size ranging from 140 mm (5.5") to more than

254 mm (10") in length. Specimens of these sizes are expensive to machine, require large environmental chambers and furnaces and pose difficulties in maintaining a uniform specimen temperature. A large specimen and hence a greater useful crack length will allow more tests to be carried out with one specimen but, this will also mean that the specimen will be exposed to elevated temperature for a lengthened period of time. If excessive, this can lead to temperature induced microstructural changes in the material and consequently, not produce the desired data. It was therefore attempted to design a "compact" tapered DCB specimen for the purpose of this study.

This specimen was designed along the lines described by Srawley and Gross^{9,10} and Gallagher.¹¹ The design parameters are indicated in Fig. 2.4. Using Fig. 2.5 and choosing $H_p/e = \tan\theta = 0.4$, $W/H_p = 317.5$ mm (12.5") and $H_p = 6.35$ mm (0.25"), we get, $W = 79.37$ mm (3.125"), $e = 15.87$ mm (0.625") and $W/e = 5.0$. From Fig. 2.6, it may be seen that for $H_p/e = 0.4$, and $W/e = 5.0$, the value of $KBW^{3/2}/P$ (where B is the specimen thickness) is almost independent of K or ΔK over a substantial range of crack lengths. The fracture mechanics specimen designed according to these dimensions is shown in Fig. 2.7 and has been side-grooved to ensure stability in the direction of propagation. The values of the overall specimen thickness (B) and the thickness at the root of the sidegroove (B_n) have been chosen as 12.7 mm (0.5") and 5.08 mm (0.2"), based on the results of Marcus¹² who demonstrated that side-grooves in a tapered DCB specimen should have an included angle of 45° and a B_n/B ratio of less than 0.5 to minimize chances of the specimen breaking off at the arms (Fig. 2.8).

The maximum bending stress in a tapered DCB specimen occurs at the neck. A simple calculation shown as follows was performed to determine if general yielding can occur in the neck region for the test conditions employed.

The yield strength of Inconel 718 at 650°C is generally about 1000 MPa (140 Ksi) in the conventionally heat treated condition.¹³ It is required to know what loads when applied to the specimen would cause bending stresses in excess of this value. Thus,

$$\sigma = \frac{Mc}{I} = \frac{F \cdot n \cdot c}{I}$$

At the neck of the specimen of Fig. 2.7,

$$n = 12.7 \text{ mm (0.5")}$$

$$c = 12.7 \text{ mm (0.5")}$$

$$I = \frac{(12.7)^2 (12.7)^3}{12} = 2167.8 \text{ mm}^4$$

$$\text{for } \sigma = 966 \text{ MPa (140 Ksi), } F = \frac{966 \times 2167.8}{(12.7)(12.7)} = 12.97 \text{ KN (2912 lbs)}$$

Loads in the vicinity of this value are likely to cause some general yielding at the neck of this specimen. It should also be noted that heat treatments such as the Wilson heat treatment or the Modified heat treatment which lead to somewhat lower yield strengths than the conventional heat treatment would cause yielding at lower loads than specified above.

In this simplified calculation the influence of the side-groove upon the moment of inertia and the stress concentration due to the side-groove have not been taken into account. It would be desirable to perform a rigorous 3-D finite element stress analysis to evaluate the maximum loads that can be applied to this specimen before yielding occurs.

Several analytical expressions have been derived to calculate the stress intensity factor for a tapered DCB specimen.¹⁴⁻¹⁸ These are listed in Table 2.4, along with the computed load in Kilonewtons for $K = 45 \text{ MPa } \sqrt{\text{m}}$, for a crack length of 25.4mm (1"). It is seen that the first five expressions give loads that are in close agreement, while a somewhat anomalous value results from the last expression. For the purpose of this study stress intensity factors were computed by expressions 1-5.

According to Fig. 2.6, for the design parameters chosen for this specimen, K should be approximately constant from $a/W = 0.2 \rightarrow 0.55$, i.e. from $a = 15.87 \text{ mm}$ to 43.7 mm (0.625" to 1.72"). This was verified by a finite element stress analysis as well as through experimental techniques.

The finite element analysis was performed by J. Kacprzyński of the Structures of Materials Laboratory, National Research Council of Canada. The 2-D mesh employed is shown in Fig. 2.9. The variation of stress intensity factor along the length of the specimen for a load of 4.54 Kg (1000 lbs) is shown in Fig. 2.10. Clearly the predicted constant- K crack length interval through finite element analysis (12.5 mm to 44.5 mm) agrees closely with the analytical prediction mentioned earlier.

Further verification was carried out experimentally by striation spacing measurements. A type 304 stainless steel tapered DCB specimen of the design geometry was cycled to failure at room temperature, at a ΔK of $45 \text{ MPa } \sqrt{\text{m}}$ (over the constant- K range), an

R-ratio of 0.05 and a frequency of 1.0 Hz. Subsequently, the fracture surface was studied by scanning electron microscopy and striation spacings were measured along the length of the fracture surface. These results are also shown in Fig. 2.10, and it may be observed that the striation spacings tend to be nearly constant over the $a = 12.5$ mm (0.5") to $a = 44.5$ mm (1.75") crack length interval. Thus, the specimen designed provides a useful constant-K region of nearly 32 mm (1.25").

2.4 Specimen Calibration and Crack Growth Monitoring

The only environmental conditions investigated in this study were laboratory air and vacuum and in principle, it should be possible to apply the electric potential drop technique or a high temperature compliance technique for continuous crack growth rate monitoring. However, the PD technique is yet to be developed at NAE for elevated temperature application and only very recently have sensitive methods become commercially available for measuring compliance changes at temperatures in the vicinity of 650°C . At the time that this investigation was begun, the only instrumentation readily available for crack growth monitoring were extensometers of the type shown in Fig. 2.11, which were fabricated by Applied Test Systems Inc.

This instrument is equipped with upper and lower crossheads, analogous to the arms of a clip gage and can be clipped onto grooves machined in test specimen (Fig. 2.12). Upon loading the specimen, the displacement at these grooves is picked up by the crossheads and

transmitted through the extension rods attached to them to an LVDT arrangement which creates an electrical signal proportional to the displacement. The electrical signal was calibrated against a standard clip gage so that displacements could be measured directly.

Figure 2.13 illustrates the variations in extensometer crosshead displacement (ΔD) with respect to variations in crack length (Δa) for a load of 6.0 KN. This calibration has been carried out at 650°C for Inconel 718, subjected to three different heat treatments. The procedure for generating these calibration plots was as follows,

- a. The specimen was fatigue pre-cracked at room temperature at $\Delta K = 45 \text{ MPa } \sqrt{\text{m}}$, $R = 0.05$ and $f = 1.0 \text{ Hz}$.
- b. This crack was extended for a few millimetres at the same cyclic loading conditions, except with $f = 0.1 \text{ Hz}$ and $T = 650^\circ\text{C}$ in laboratory air taking note of the crosshead displacements at the beginning and end of this interval.
- c. The crack was extended further a few millimetres under the same temperature, stress intensity range as well as R-ratio but this time in vacuum at $f = 0.1 \text{ Hz}$ (a low pressure of $2 \times 10^{-6} \text{ Torr}$ could be created using an environmental retort to contain the test specimen and load train. A description of the supporting equipment needed to create these conditions follows). Once again, at the end of this interval, the crosshead displacement was measured.
- d. This procedure was repeated several times until the useful crack length (constant-K region) was covered, and the fracture surface was examined by scanning electron microscopy.

e. The transitions between cyclic loading in air and vacuum could be clearly determined by differences in the fracture surface topography. In this manner, the crack lengths corresponding to the crosshead displacements mentioned earlier were determined.

It was soon realized, however, that the several mechanical linkages present in this extensometer and the loose manner in which the crossheads hinged upon the specimen made the crosshead displacement measurements far from being reproducible and accurate. This device was therefore abandoned in favour of a similar instrument (also purchased from Applied Test Systems Inc.) in which the crossheads were mechanically fastened to the loading pins as shown in Fig. 2.14. In this case, load line displacement is measured.

The calibration results obtained with this improvement are shown in Fig. 2.15. The measurements were found to be much more reproducible than with the clip-on device. However, the scatter in the data points shown on Fig. 2.15 indicates that errors of up to $\pm 2\text{mm}$ in crack length measurement can still occur. Thus, this instrument was used in this study more as a means of knowing the approximate position of the crack front from outside the environmental test chamber during testing than for accurate dynamic crack growth monitoring.

2.5 Test Equipment

An MTS Model 312.32 servo hydraulic test system of 250 KN capacity was used to carry out the cyclic and static load crack propagation tests in this study. The test frame was equipped with a Model

661.23A load cell of 250 KN capacity, a Model 204.71 actuator with a stroke of 152 mm (6") and a compatible electronic control console. (Figs. 2.16(a-c)). The clevis-type specimen grips and pull rods were fabricated from cast 713C for high temperature strength. The pull rods had cooling passages machined in them such that cooling could be achieved by water circulation during high temperature testing.

Environmental influences were studied using a cylindrical environmental chamber fabricated from Inconel 600. This retort was equipped with a vacuum port, inlet/outlet gas ports, thermocouple and LVDT sealing glands. Sealing was provided at the top and bottom by welded metal bellows. The bellows type packing glands had double O-ring seals and the upper bellows were water cooled. The bottom flange of the retort was removable and the load train assembly was introduced into or removed from the retort through the bottom opening. Radiation shields located on the pull rods protected the extensometer LVDT from excessive heating. A copper coil was also provided on the retort around the location of the LVDT and water was continually circulated through this coil, the pull rods and the upper bellows assembly during testing. The environmental chamber could be evacuated to a pressure of 2×10^{-6} Torr at 650°C.

Specimen heating was achieved using a three-zone resistance heated split furnace with Kanthal Al heating elements equipped with type-K chromel-alumel temperature control thermocouples. The temperature gradient in the furnace at 650°C was within $\pm 3^\circ\text{C}$ over a distance of 4 inches along the vertical axis and within $\pm 2^\circ\text{C}$ over a

distance of 3 inches. The specimen was located as centrally as possible in this hot zone of the furnace and specimen temperature was monitored by spot welding a type-K chromel-alumel thermocouple to the specimen surface. The uniformity of temperature along the length of specimen was verified by placing thermocouples within holes drilled in the side of the specimen in the location shown in Fig. 2.17. At 650°C, the variation in the temperature readings of these three thermocouples was within $\pm 1^\circ\text{C}$.

2.6 Test Parameters and Procedure

Elevated temperature (650°C) propagation tests were conducted under the four types of loading conditions given below,

- a. Cyclic loading in air,
- b. Cyclic loading in vacuum,
- c. Static loading in air,
- d. Static loading in vacuum.

Testing was carried out at three $K/\Delta K$ values, namely 45, 67.5 and 90 $\text{MPa}\sqrt{\text{m}}$ giving a total of twelve loading conditions. For cyclic loading, an R-ratio of 0.05 and a low frequency of 0.1 Hz were used since it was desired to compare the results with those obtained by other investigators who have used similar loading conditions. All the fatigue crack growth testing has been under constant-amplitude conditions and hold-time effects have not been studied.

Material of each one of the four heat treatment conditions given in Table 2.2, was tested under the twelve loading conditions mentioned earlier. It was found convenient to test each specimen at a certain $K/\Delta K$ value in one of the two sequences given below.

Sequence 1.

- a. Fatigue in air
- b. Fatigue in vacuum
- c. Creep in air
- d. Creep in vacuum.

Sequence 2.

- a. Fatigue in vacuum
- b. Creep in air
- c. Fatigue in vacuum
- d. Creep in vacuum
- e. Fatigue in air

It was found essential to carry out the tests in one of these two sequences to ensure that the transitions between loading conditions are clearly distinguishable by virtue of differences in fracture surface topography. The number of cycles or hours of loading were such that each loading condition caused 6-8mm of cracking over the constant-K regime in the specimen. Step c in sequence 2 was an exception to this, however, and was not usually more than 2mm in length since this was a repeat of stage a, and was interposed again between steps b and d only to clearly distinguish between them. The fracture surface of a specimen of Inconel 718 tested according to sequence 2 is shown in Fig. 2.18. Due to errors associated with the extensometry, it was not always possible to contain the test within the constant-K regime. In such cases, the last part of the sequence was repeated with another specimen, under the correct conditions of K or ΔK .

As mentioned earlier, elevated temperature tensile and stress rupture data were also generated to support the crack growth rate information. Tensile tests were conducted at 650°C and the stress rupture tests at 650°C and 690 MPa.

Failed specimens were given detailed metallographic analysis. Scanning Electron Microscopy (SEM) was used to define the dominant fracture modes under various test conditions. SEM as well as optical microscopy were used to study the regions below the fracture surfaces in the DCB specimens to identify the damage mechanisms occurring in the crack tip plastic zone. In addition, thin foils were prepared from material immediately below these fracture surfaces and viewed through a Phillips EM300 transmission electron microscope with a view to identify the micromechanisms of deformation in the vicinity of the crack tip.

2.7 Results and Discussion

2.7.1 Low Stress Intensity Factor ($45 \text{ MPa } \sqrt{\text{m}}$)

Cyclic loading in vacuum: The fracture mode for all microstructures was transgranular. As shown in Fig. 2.19(a-d), a striated appearance was observed on all fracture surfaces. No evidence of intergranular cracking is obvious, indicating that even at a frequency of 0.1 Hz very little creep cavitation occurs in the material. Magnified images of the striations are shown in Fig. 2.20(a-b) for two of the heat treatments.

It is interesting to note that the crack propagation rates (Table 2.5) for all four microstructures tend to be about the same.

This result is not surprising since crack propagation in the Stage 2 (ductile, striated) mode observed here should be essentially independent of microstructure, depending mostly upon the elastic modulus. It was found that the heat treatments applied here do not significantly change the elastic modulus, and hence the similar rates of crack growth.

A close correlation could be obtained between the macroscopic crack growth rates observed and the fatigue striation spacings in this case. Based on the micrographs shown in Fig. 2.20, an average striation spacing of $0.5 \mu\text{m}$ was calculated which agrees quite closely with the macroscopic growth rates of 0.4 to $0.55 \mu\text{m/cycle}$ mentioned in Table 2.5.

Polished and etched cross sections of regions below the fracture surfaces revealed further information regarding the mechanisms of failure. In the case of the Conventional heat treatment, considerable cracking of carbides takes place in the vicinity of the crack tip (Fig. 2.21(a)), and this leads to transgranular crack initiation. It is also possible that a limited amount of intergranular cavitation occurs in this case, as indicated in Fig. 2.21(b-c). However, this effect does not appear to significantly influence the mode of crack propagation, which as mentioned earlier is predominantly transgranular (Fig. 2.21(d)).

A cross section of a specimen given the Modified heat treatment and tested under the above condition is shown in Fig. 2.22 and it appears that the fracture mechanism is essentially the same as

is the case of the Conventional heat treatment - i.e. cracks initiate at slip bands or cracked particles in the face of extensive matrix deformation and propagate in the Stage II striated mode until they link up with the main crack. This observation appears to be consistent with the absence of any significant variation in the macroscopic crack growth rates, between various microstructures.

Cyclic Loading in Air: In the presence of air, the fracture mode changed to a predominantly intergranular process for the Conventional, Merrick and Modified Merrick heat treatments (Fig. 2.23). A predominantly transgranular, striated appearance was still observed in the case of the Modified heat treatment. The crack growth rates in all cases were at least an order of magnitude higher than those for vacuum, with the Modified heat treatment exhibiting the best resistance. The Conventional and Merrick heat treatments led to relatively poor resistance with crack growth rates being about twice that for the Modified heat treatment.

The superior crack growth resistance due to the Modified heat treatment could be due to a number of factors. Primarily, the coarse grain size due to the 1093°C solution anneal of this heat treatment, would lead to a much lower level of grain boundary sliding in comparison with the case of the Conventional heat treatment. This apparently leads to reduced levels of creep cavitation since, as mentioned in Chapter 2 (Appendix D), the rate of cavity growth is proportional to the extent of grain boundary sliding. The lack of intergranular cracking in the specimen given the Modified heat

treatment may also indicate that the level of grain boundary oxidation is very limited in this material. In the case of the other heat treatments the change from transgranular to intergranular crack growth (in air, compared to vacuum) appears to be due an enhancement of the grain boundary sliding mechanism of creep and/or significant grain boundary oxidation.

It is also possible that the higher solution annealing temperature of the Modified heat treatment dissolves some of the large carbide particles (present in the Conventionally heat treated material) and leads to a distribution of smaller carbide particles that are less prone to cracking. It is also well known that the Modified heat treatment leads to a much lesser quantity of the brittle Laves phase precipitate than the Conventional treatment.

The striation spacings observed on the fracture surface in the case of the Modified heat treatment were in reasonable agreement with the observed crack growth rates. Striation spacings of about 1.33 μm were observed (Fig. 2.24) while the crack growth rate was $\sim 1.4 \mu\text{m/cycle}$.

Polished and etched cross sections of fractured specimens indicated that in the case of the Modified heat treatment, fatigue crack growth in air still occurs by similar mechanisms as in vacuum, i.e. due to crack initiation at slip bands and at cracked carbide particles (Fig. 2.25). In contrast, the Conventional and Merrick heat treatments led to extensive grain boundary oxidation (Fig. 2.26(a-b)) and what appears to be internal cavitation (Fig. 2.26(c-d)). It is also possible that carbide cracking is enhanced due to particle oxidation (Fig. 2.27).

It was reported from Table 2.5, that the Modified Merrick heat treatment resulted in fatigue crack growth resistance almost as high as the Modified heat treatment. Yet, the mode of failure for the case of the Modified Merrick heat treatment is intergranular. It therefore appears that significant amounts of intergranular oxygen embrittlement may be occurring in this material, but the process of failure is inhibited by the presence of a serrated grain boundary morphology. As mentioned in earlier chapters, the Modified Merrick heat treatment was designed to precipitate Ni_3Cb (δ -phase) particles at the grain boundaries and cause an irregular or serrated grain boundary morphology.

It is possible that the extent of intergranular oxygen embrittlement or cracking occurring in this alloy is influenced significantly by the grain size, with greater embrittlement at finer grain sizes. A fine grained material may present a greater grain boundary area along which rapid, stress-assisted diffusion of oxygen can occur.¹⁹ Also, for a given amount of intergranular oxidation, the available ductility may be exhausted more rapidly in a fine grained material due to the greater amount of grain boundary sliding which can occur. The presence of Ni_3Cb particles at the grain boundaries which can give rise to a serrated grain boundary morphology and inhibit sliding is likely to offset the detrimental effect of a finer grain size.

While the effects of grain size and grain boundary morphology are important, there have been a number of investigations²⁰⁻²⁵ in which

indirect evidence of an inverse relationship between matrix γ' or γ'' particle size and the sensitivity to environmental embrittlement has been demonstrated. When the matrix precipitate is very fine, dislocations cut the precipitate particles instead of bypassing them and are confined to move in co-planar arrays, causing planar pile-ups and consequently, intense stress concentrations at grain boundaries and other obstacles to their motion. If at the same time, the grain boundary cohesive strengths have been weakened as a result of adsorption of an aggressive environmental species, intergranular cracking would be promoted.

Apart from this adsorption mechanism, it is also thought that in materials exhibiting predominantly planar slip character, oxygen diffusion²⁶⁻²⁸ into the material can occur quite rapidly along planes of intense deformation at elevated temperatures and be swept to the grain boundaries as a result of a solute-dislocation interaction. In the presence of more homogeneous slip, the number of dislocations arriving at a grain boundary would be reduced and consequently, the stress concentrations and oxygen transport to the grain boundaries occurring due to a dislocation "sweep-in" mechanism will be reduced as well.

Another type of microstructure-environment interaction has been suggested by Raj and Ashby²⁹ and involves the diffusion of oxygen along the grain boundaries - this being promoted by a shear stress on the grain boundary and the sliding of one grain over another.

Yet another embrittlement mechanism, suggested by Woodford³⁰ is based on the immobilization of grain boundaries by the environmental species. Intergranular cavities then develop readily in the presence of grain boundary sliding due to the lack of grain boundary mobility for accommodation. This effect is likely to be felt more in a material in which grain boundary sliding occurs readily upon elevated temperature loading. Thus, a fine grained material or one in which grain boundary strengthening agents such as carbide particles or serrations are few would be particularly susceptible. An added deleterious effect can arise from the fact that an enhancement in grain boundary sliding can occur in an air environment (as compared to vacuum) due to a reduction of grain boundary cohesive strength by gaseous adsorption effects.³¹⁻³² Also, grain boundary cavities, once nucleated, can be stabilized by the diffusion of gaseous species into the cavities.

Based on the various findings reported earlier, it may be deduced that the type of microstructures that would be most susceptible to environmental embrittlement would be those that ~~are~~ fine grained, have very few grain boundary strengthening agents, and/or exhibit pronounced planar deformation characteristics at elevated temperature.

In this study, Transmission Electron Microscopy of thin foils prepared from regions immediately below fracture surfaces indicated that planar deformation was predominant (Fig. 2.28) in the case of the Conventional and Modified Merrick heat treatments which produced the finest γ particle size distributions (Fig. 2.2). Material of the

Merrick or Modified heat treatment did not show any sign of intense planar deformation though several foils were examined. It appears therefore that elevated temperature deformation in Inconel 718 given the Modified and Merrick heat treatments occurs in a relatively homogeneous mode.

The especially poor resistance of material given the Conventional heat treatment to elevated temperature crack growth in air may thus be interpreted as being due to the fine grain size and the very fine γ'' morphology which promote environmental embrittlement according to the mechanisms reviewed earlier.

The poor resistance of the Merrick heat treatment, on the other hand, is best explained by a lack of grain boundary strengtheners and enhanced grain boundary sliding in the presence of an air environment under elevated temperature loading conditions. It should also be noted that the growth of creep cavities during dislocation creep due to accommodative grain shear would be enhanced in the presence of coarse, overaged matrix γ'' precipitates, as in this material.

Static Loading in Vacuum: In the absence of an oxygen rich atmosphere, creep crack growth in all cases occurred in a predominantly intergranular fashion (Fig. 2.29). In the case of the Modified heat treatment, however, occasional incidences of transgranular, faceted failure were also observed.

The crack growth rates were lowest for the Modified Merrick heat treatment and highest for the Conventional treatment. The Merrick and the Modified heat treatments seemed to produce nearly the same crack growth rates.

It should be noted that though there are relative variations in crack growth rate, the magnitude of these differences are in all cases very much lower than in air (Fig. 2.30). For example, at $K = 45 \text{ MPa } \sqrt{\text{m}}$ the difference in crack growth rates between material of Conventional and Modified heat treatments is 1.58 mm/h in air, but only 0.056 mm/h in vacuum. Thus, the influence of microstructure, per se, in the absence of an oxygen-rich environment is very limited when the differences in macroscopic crack growth rates in vacuum are compared to those in air. It follows then that the substantial differences in crack growth rates observed in air are a reflection of the manner and extent to which the microstructure leads to environmental sensitivity.

It is not clear at this time whether the variations in crack growth rates observed in vacuum are purely microstructural effects since some environmental effects may still be active at a very limited level. Much of the evidence in this study indicates that at 650°C , environmental effects are very much suppressed at the vacuum conditions employed, but this would have to be verified by further testing under ultra-high vacuum conditions.

An important observation made on polished and etched cross sections of specimens tested in static load under vacuum was that the secondary cracks tended to be very sharp and there appeared to be very limited grain displacement, (Fig. 2.31(a,b)). As will be seen in the subsequent section, this behaviour was in striking contrast with that observed for static loading conditions in air. Also, there was very little oxidation debris associated with these cracks even though they

had been exposed to an oxidizing atmosphere after the crack growth period in vacuum had been completed.

Static Loading in Air: The fracture mode in all cases was

intergranular (Fig. 2.32). In air, creep crack growth was most rapid in the case of the Conventional and Merrick heat treatments as shown in Table 2.5. The Modified and the Modified Merrick heat treatments produced about the same level of crack growth resistance and the growth rates were about half those observed with the Conventional and Merrick heat treatments. As mentioned in the earlier discussion regarding crack growth in air under cyclic loading conditions, the superior resistance of the Modified and the Modified Merrick heat treatments can be attributed to grain size, grain boundary morphology and matrix γ'' particle size effects upon creep and/or environmental effects. In confirmation of earlier findings, a marked occurrence of planar deformation was found only in the case of the Conventional and Modified Merrick heat treatments which produced very fine γ'' particle distributions (Fig. 2.33). It is interesting to note that the Modified Merrick heat treatment produces almost as good crack growth resistance as the Modified heat treatment, in spite of the more aggressive environmental reactions possible due to the fine matrix γ'' particle size.

While cracks which occurred under static loading conditions in vacuum had very little oxidation debris associated with them even after a period of stress-free exposure to oxygen at elevated temperature (during the subsequent creep in air or fatigue in air test portions) cracks associated with static loading in air exhibited

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considerable oxidation damage (Fig. 2.34) indicating that the type of oxidation occurring here is a stress-assisted phenomenon. Further, the grain boundary separations occurring in air seemed to be, on an average, significantly higher than in vacuum (Figs. 2.31 and 2.35).

2.7.2 Intermediate Stress Intensity Factors ($67.5 \text{ MPa } \sqrt{\text{m}}$)

Cyclic Loading in Vacuum: The fracture mode in all cases investigated was the transgranular, striated mode (Fig. 2.36). The crack growth rates are certainly higher than at $45 \text{ MPa } \sqrt{\text{m}}$ and as before, appear to be independent of microstructure (Table 2.5). The striation spacings in each case were measured from Fig. 2.37. The spacings ranged between $1.1\text{--}1.5 \text{ } \mu\text{m/cycle}$ which once again is in good agreement with the observed crack growth rates of $1.1\text{--}1.4 \text{ } \mu\text{m/cycle}$.

It should be noted that at this higher stress intensity factor some ductile tearing through a void sheet mechanism can occur along with the striated mode of failure. This is shown in Fig. 2.38.

Cyclic Loading in Air: Crack growth rates were most rapid for the Conventional heat treatment, while the Modified and Merrick heat treatments seemed to have similar levels of crack growth resistance. The fracture mode was in all cases predominantly intergranular, but yet, striations could be observed on the grain surfaces (Fig. 2.39). These striations are shown more clearly in Fig. 2.40, and it is interesting that the striation spacings measured from these figures ($3\text{--}4 \text{ } \mu\text{m/cycle}$) still agree quite closely with the measured crack growth rates of $3.8\text{--}4.3 \text{ } \mu\text{m/cycle}$ for the Modified and Merrick heat treatments.

The explanations given for the inferior fatigue crack growth resistance due to the Conventional heat treatment at lower stress intensity factors may still apply to the case of intermediate and higher stress intensity factors. Therefore, these will not be repeated here.

Static Loading in Vacuum: The fracture mode observed at the intermediate stress intensity factors was intergranular, but yet, in all cases, the fracture surfaces revealed that this was occurring by microvoid linkage along the grain boundaries. This feature was not observed at the lower stress intensity factor tests (Fig. 2.41). Even more significant was the observation that the macroscopic crack growth rates actually dropped to significantly lower levels than those observed at the lower stress intensity factor (Table 2.5). The reasons for this decrease in crack growth rate with increase in stress intensity factor is not understood at this time and should be investigated in future studies. It is suspected that this is due to a crack tip blunting effect, arising from stress relaxation around the crack tip as a result of significant plastic yielding in the region of the side groove. The DCB specimen designed for this study appears to be best suited for studies at the lower range of stress intensity factors, if an LEFM approach is to be applicable.

Crack growth rates in vacuum appear once again to be most rapid for the Conventional heat treatment and least rapid for the coarse grained microstructure produced by the Modified heat treatment.

Static Loading in Air: The worst behaviour was exhibited in the case of the Merrick heat treatment which led to crack growth rates about

three times as rapid as the Modified heat treatment (Table 2.5). The fracture mode was clearly intergranular in all cases (Fig. 2.42) without any of the microvoid coalescence along grain boundaries observed in the case of crack growth in vacuum. It is quite evident from Fig. 2.42b that the grain boundaries are severely oxidized during this cracking process.

2.7.3 High Stress Intensity Factors ($90 \text{ MPa } \sqrt{\text{m}}$)

Cyclic Loading in Vacuum: The trends observed were similar to those at low and intermediate stress intensity factors. Once again crack growth rates measured were essentially microstructure independent with the fracture being transgranular and striated in all cases (Table 2.5, and Fig. 2.43). The striation spacings observed ($2-3 \text{ } \mu\text{m}/\text{cycle}$) are in close agreement with the observed crack growth rates of $2.1-2.8 \text{ } \mu\text{m}/\text{cycle}$ (Fig. 2.44). It is important to note the consistency for various tests under the same conditions (Table 2.5). This demonstrates the accuracy with which crack growth rates can be measured with the techniques used in this study.

Cyclic Loading in Air: Fatigue crack growth rate in air occurred in an intergranular fashion (Fig. 2.45) with the Modified as well as the Modified Merrick heat treatments giving rise to the best resistance to crack growth. Clearly, the serrated grain boundary morphology due to the Modified Merrick heat treatment contributes a beneficial effect in terms of crack propagation resistance over a wide range of stress-intensity factors. Changing the grain boundary morphology from

to serrated seems to produce a beneficial effect similar to that obtained from increasing the grain size.

Striations were observed on grain surfaces in this case as well, and the striation spacings measured though not in precise agreement with the macroscopic crack growth rates observed, appear to be reasonably close in magnitude. For example, in the case of the Modified heat treatment (Fig. 2.45b) the observed crack growth rate of 7.7 m/cycle is not very far off from the average striation spacing of $\sim 10 \mu\text{m}$.

Static Loading in Vacuum: Once again crack growth rates much lower than the values at the lower stress intensity factor were observed. Further, there appeared to be virtually no increase in crack growth rate in going from 67.5 MPa $\sqrt{\text{m}}$ to 90 MPa $\sqrt{\text{m}}$. The fracture mode was intergranular and as in the case of the intermediate stress intensity factor tests, extensive microvoid coalescence was observed on the grain boundaries (Fig. 2.46). The Modified heat treatment gives rise to a crack growth rate only half as rapid as the Merrick heat treatment. As explained earlier, the decrease in crack growth rates with increasing stress intensity, leading to almost a situation of crack arrest, is believed to be due to a crack tip blunting effect due to the onset of gross plastic yielding.

Static Loading in Air: The fracture mode was predominantly intergranular in all cases, except for the case of the Modified Merrick heat treatment (Fig. 2.47). In this case, a significant amount of transgranular faceting was observed (Fig. 2.47d) and in some areas, an

almost completely transgranular fracture appearance was found (Fig. 2.48).

The crack growth rates produced by the Modified Merrick and the Modified heat treatment appear to be of the same level (Table 2.5) while the Merrick heat treatment causes the worst resistance.

2.7.4 Tensile Tests

The results of tensile tests conducted at room temperature and at 650°C are shown in Table 2.6. At room temperature, the grain size difference between the Conventional and Modified heat treatments does not give rise to any substantial difference in yield or ultimate tensile strength. The coarse γ precipitate due to the Merrick heat treatment does seem to lower the yield strength considerably. The Modified Merrick heat treatment produces strengths that are at an intermediate level between those due to the Modified and Merrick heat treatments.

At 650°C, the Modified heat treatment does produce a somewhat lower level of strength compared to the Conventional heat treatment. The Modified Merrick heat treatment as well as the Modified heat treatment seem to produce similar levels of strength with the former exhibiting somewhat higher ductility.

The results of fractographic studies of tensile specimens tested at 650°C tend to strengthen the explanations offered earlier regarding microstructure-environment interactions during elevated temperature crack growth. In the case of the Modified as well as the Modified Merrick heat treatments surface crack initiation appeared to

occur as a result of slip band cracking (Figs. 2.49-2.50). Only in the case of the Conventional heat treatment did any sign of intergranular surface embrittlement exist (Fig. 2.51). In all cases, the failure mode in the interior of the specimens was transgranular, ductile failure.

This finding is another indication that stress-induced-intergranular embrittlement is enhanced in microstructures that are fine grained and have a very fine matrix γ'' distribution. As a result of its somewhat coarse grain size and the serrated grain boundary morphology, the Modified Merrick heat treatment seems to produce better resistance to intergranular cracking than the Conventional - in spite of the fine γ'' precipitates which promote intense planar deformation.

2.7.5 Stress Rupture Tests

The results of stress rupture tests conducted at 650°C and 690 MPa are shown in Table 2.7. The Modified heat treatment leads to significantly improved stress rupture life, presumably due to the limited amount of grain boundary sliding which occurs in the coarse grained microstructure. However, the rupture ductility seems to be lowest for this heat treatment. The Merrick as well as the Modified Merrick heat treatment seems to produce relatively low rupture lives though the rupture ductility seems somewhat improved. The γ'' coarsening due to the Merrick heat treatment may be partly responsible for the lower rupture life due to this treatment.

Fractographic analysis of stress rupture tested specimen provided additional confirmation of the previously mentioned correlation between γ'' precipitate particle size and slip planarity.

As shown in Fig. 2.52, planar slip is most pronounced only in the case of the Conventional and the Modified Merrick heat treatments which produce a fine γ'' particle distribution. In a relative sense, planar slip appears to be suppressed in the Merrick and Modified heat treatments.

2.8 Feasibility of Applying a Damage Tolerance Based Lifing Method

The primary purpose of this study has been to determine the feasibility of applying damage tolerance based life extension procedures to the Inconel 718 turbine discs of the J85-CAN15 and F-404 engines. There is no information available at present regarding the operating stress levels and the dysfunction criteria for these components. However, from the data given in Table 2.5, it is clear that even in the case of heat treatments relatively resistant to crack propagation, several millimeters of crack growth can occur within a few hours or in within a few hundred cycles of service exposure. The crack growth rates determined here have not considered hold time effects or the effect of additional environmental damage such as sulphidation and hence, in reality, the situation may be even worse.

Hence, a damage tolerance based lifing philosophy designed upon the time for a barely detectable crack to grow to a dysfunction size seems very hazardous to apply to these components. However, it may be possible to rejuvenate these components from time to time by a

reheat treatment or a HIP treatment before any cracking has occurred. This would have to be investigated further in future studies.

2.9 A New Heat Treatment for Inconel 718 Turbine Discs

Inconel 718 turbine discs are usually given the Conventional heat treatment to retain a fine grain size and thereby maximize the strength properties. It has been clearly demonstrated that in this condition, crack propagation rates can be quite rapid in Inconel 718 once crack tip stress intensity factors of $45 \text{ MPa } \sqrt{\text{m}}$ and above are reached (which can occur at very small crack lengths in highly stressed discs of advanced engines). It therefore seems safer to adopt heat treatments that produce better crack propagation resistance even though this may be at the loss of some strength. The coarse grained microstructure of the Modified heat treatment may be better from this point of view, but excessive grain coarsening may cause a deleterious effect as well, by lowering the elevated temperature crack initiation life and the creep ductility.

It was felt that an optimum heat treatment for this application would have to produce,

- a. a grain size intermediate between those produced by the Conventional and Modified heat treatments;
- b. a serrated grain boundary morphology as produced by the direct aging technique in the Modified Merrick heat treatment;
- c. a γ'' particle size that is comparable to that produced by the Modified heat treatment. A very fine γ'' particle

size would promote environmental attack whereas very coarse γ'' would lead to considerable loss of tensile and creep strengths. An optimum size distribution is therefore needed.

Through numerous trials, the heat treatments listed below were arrived at, which may be the solutions to arriving at an optimum microstructure.

- a. $1066^{\circ}\text{C}/2-24\text{h} \xrightarrow[\text{cool}]{\text{furnace}} 843^{\circ}\text{C}/4\text{h}/\text{AC} + 926^{\circ}\text{C}/1\text{h} \xrightarrow[\text{cool}]{\text{furnace}}$
- $718^{\circ}\text{C}/4\text{h} \xrightarrow[\text{cool } 1^{\circ}\text{C}/\text{min}]{\text{furnace}} 612^{\circ}\text{C}/12\text{h}/\text{AC}.$
- b. $1066^{\circ}\text{C}/2-24\text{h} \xrightarrow[\text{cool}]{\text{furnace}} 843^{\circ}\text{C}/4\text{h}/\text{Ac} + 955^{\circ}\text{C}/1\text{h} \xrightarrow[\text{cool } 1^{\circ}\text{C}/\text{min}]{\text{furnace}}$
- $718^{\circ}\text{C}/4\text{h} \xrightarrow[\text{cool } 1^{\circ}\text{C}/\text{min}]{\text{furnace}} 612^{\circ}\text{C}/12\text{h}/\text{AC}.$

Basically, this technique involves allowing some amount of grain coarsening to occur during the solution anneal at 1066°C and employs direct aging at 843°C to precipitate Ni_3Cb at the grain boundary serrations. Since a certain amount of matrix γ'' precipitation and excessive coarsening can occur during this intermediate age, the specimen is reheated to temperatures between 926°C - 955°C to redissolve any overaged matrix γ'' and then aged in a manner that will develop a γ'' particle distribution similar to that in the Modified heat treatment. Typical microstructures produced by such heat treatment technique are

shown in Fig. 2.53. Further research is required to thoroughly test the effect of these heat treatment modifications and to evaluate their merits and demerits.

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Table 2.1

Chemical Composition (wt.%) of Inconel 718 Used for
Fracture Mechanics Testing

Ni	Cr	Cb+Ta	Ti	Al	Mo	Fe	C	Si	Mn	P	Cu	S
53.58	17.75	5.26	1.05	0.52	3.07	17.9	.04	.32	.13	.013	.22	.001

Table 2.2

Heat Treatments Used for Crack Growth Rate Study

Heat Treatment Name	Schedule
"Conventional"	955°C/1h/AC + 718°C/8h $\xrightarrow{1^\circ\text{C/min}}^{\text{FC}}$ 621°C, hold for aging time of 18h, AC
"Modified"	1093°C/1h $\xrightarrow{1^\circ\text{C/min}}^{\text{FC}}$ 718°C/4h $\xrightarrow{1^\circ\text{C/min}}^{\text{FC}}$ 621°C/16h/AC
"Merrick"	1066°C/1h + 843°C/4h + 718°C/8h $\xrightarrow{1^\circ\text{C/min}}^{\text{FC}}$ 621°C/10h/AC
"Modified Merrick"	1066°C/1h $\xrightarrow{7^\circ\text{C/min}}^{\text{FC}}$ 843°C/4h + 718°C/8h $\xrightarrow{1^\circ\text{C/min}}^{\text{FC}}$ 621°C/10h/AC

Table 2.3
Microstructural Variations Due to Heat Treatment

Heat Treatment	Grain Size (μm)	Relative γ' Size	Grain Boundary Morphology
Conventional	10-50	fine	some serrations
Modified	50-200	intermediate	planar
Merrick	50-120	coarse	planar
Modified Merrick	50-120	fine,	pronounced serrations

Table 2.4

Calculation of the Stress Intensity Factor for the Double Cantilever Beam Specimen Used in this Study¹⁴⁻¹⁸

No.	Expression	Computed load in kilonewtons for K = 45 MPa $\sqrt{m}$ for a crack length of 25.4 mm
1	$\frac{KBh^{\frac{1}{2}}}{P} = (12)^{\frac{1}{2}} [a/n + 0.6728 + 0.0377 h^2/a^2]$	5.9
2	$\frac{KBh^{\frac{1}{2}}}{P} = [12 \{a/h + 0.6\}^2 + \frac{1}{3}]^{\frac{1}{2}}$	5.9
3	$\frac{KBh^{\frac{1}{2}}}{P} = 3.46 a/h + 2.38$	5.9
4	$\frac{KBh^{\frac{1}{2}}}{P} = (12)^{\frac{1}{2}} (a/h + 0.64)$	6.0
5	$\frac{KBh^{\frac{1}{2}}}{P} = \{12(a^2/h^2 + \sqrt{5/2} a/h + \frac{1}{2})\}^{\frac{1}{2}}$	6.1
6	$\frac{KBh^{\frac{1}{2}}}{P} = \frac{\lambda}{\left(\frac{H}{e}\right)^{\frac{1}{2}} \left(\frac{a+e}{W}\right)^{\frac{1}{2}}} \left[\frac{a/W}{\left(\frac{H}{e}\right)^{\frac{1}{2}} \left(\frac{a+e}{W}\right)^{\frac{1}{2}}} + 0.7 \right]$	7.0

Table 2.5
Influence of Heat Treatment on Elevated Temperature Crack Growth Rates

Heat Treatment	Cyclic Loading in Vacuum $\times 10^{-3}$ mm/cycle			Cyclic Loading in Air $\times 10^{-3}$ mm/cycle			Static Loading in Vacuum mm/h				Static Loading in Air mm/h		
	$\Delta K=45$ MPa $\sqrt{m}$	$\Delta K=67.5$ MPa $\sqrt{m}$	$\Delta K=90$ MPa $\sqrt{m}$	$\Delta K=45$ MPa $\sqrt{m}$	$\Delta K=67.5$ MPa $\sqrt{m}$	$\Delta K=90$ MPa $\sqrt{m}$	K=45 MPa $\sqrt{m}$	K=67.5 MPa $\sqrt{m}$	K=90 MPa $\sqrt{m}$		K=45 MPa $\sqrt{m}$	K=67.5 MPa $\sqrt{m}$	K=90 MPa $\sqrt{m}$
Modified	0.4	0.9 1.2 2.6	2.1 2.5 2.6	1.4	4.2	7.7	.065	.0021	.0022		1.42	3.52	8.00
Conventional	0.5	1.4	2.6	3.0	7.5	16.3	.145	.0195	*		2.67	4.5	9.06
Merrick	0.52	1.0 1.1	2.6 2.8	2.7	3.8	12.0	.049	.005	.0048		3.0	10.21	11.68
Modified Merrick	0.4	*	2.2	2.0	*	7.2	.0185	*	*		1.5	*	7.56

Table 2.6

Influence of Heat Treatment on Tensile Properties of Inconel 718

Heat Treatment	Room Temperature			650°C Treatment		
	0.2% y.s. (MPa)	UTS (MPa)	%E ₂	0.2% y. (MPa)	UTS (MPa)	%E ₂
Conventional	1138	1386	17.8	952	1131	18.5
Modified	1152	1373	14.5	873	1094	14.3
Merrick	998	1331	*	*	*	*
Modified Merrick	1062	1283	*	855	1028	18.4

Table 2.7
Effect of Microstructure on the Creep Properties of Inconel 718 at 690 MPa and 650°C

Heat Treatment	t_R in hours	$\dot{\epsilon}$ in s^{-1}	% ϵ	%R of A
Modified	176-324(2)	$9.8 \times 10^{-4}(3)$	1.95-3.74(2)	3.76-5(2)
Conventional	173-296(4)	1.1×10^{-3}	5.53-8.30(4)	7.40-19.23(4)
Merrick	160-179(2)	1.4×10^{-3}	1.87-3.53(2)	6.23-8.46(2)
Modified	111-189(4)	$9.4 \times 10^{-5}(2)$	3.16-6.29(4)	5.0-12.18(4)
Merrick	154-190(2)	1.6×10^{-3}	5.80-6.43	9.87-13.44
		9.66×10^{-4}		
		1.02×10^{-3}		
		1.17×10^{-3}		
		1.37×10^{-3}		

Note - Superscripts denote number of tests conducted.

2.1 Influence of Heat Treatment on Microstructure
of Inconel 718

- (a) and (b) as received
- (c) Conventional heat treatment
- (d) Modified heat treatment
- (e) Merrick heat treatment
- (f) Modified Merrick heat treatment

- (e) Localized γ coarsening - Modified
- (f) Localized γ coarsening - Modified "horlick"

2.3 Geometries of Tapered DCB Specimens Used in Various Investigations¹⁻⁴

2.3 Geometries of Tapered DCB Specimens Used in Various Investigations¹⁻⁴

2.4 Design Parameters for a Tapered DCB Specimen

2.5 Crack Length vs Stress Intensity Factor Relationships in Tapered DCB Specimens¹⁰

2.6 Crack Length vs Stress Intensity Factor Relationships in Tapered DCB Specimens¹¹

2.7 Geometry of Tapered DCB Specimen

Specimen 1. 25 deg taper, $b/h = 0.8$.
 Specimen 2. 20 deg taper, $b/h = 0.8$.
 Specimen 3. 25 deg taper, $b/h = 0.9$.

2.8 Influence of Side Groove Geometry on Crack Path

2.9 2-D Finite Element Mesh Used to Analyze Tapered

DCB Specimen

2.10 Results of Finite Element Analysis and Striation Spacing Measurements

ALL DIMENSIONS ARE IN MILLIMETRES

INTEGRAL KNIFE EDGE DESIGN

2.11 Clip-Gage Type Extensometer

2.12 Grooves Machined in Specimen to Accommodate Clip-Gage Type Extensometer

2.13 Calibration Plot of Clip-Gage Extensometer-Crosshead
Displacement vs Crack Length

2.14 Load-Line Extensometer

2.15 Load-Line Displacement vs Crack Length Calibration
for Tapered DCB Specimen of Fig. 10.7

2.16 Test Assembly Used for Elevated Temperature Crack Growth Measurements

Fig. 2.16 (continued)

Fig. 2.16 (continued)

FATIGUE
IN
AIR

CREEP
IN
AIR

FATIGUE
IN
VACUUM

CREEP
IN
VACUUM

FATIGUE
IN
VACUUM

FATIGUE
PRECRACK

2 mm

2.18 Typical Testing Sequence for Elevated Temperature

Crack Growth Rate Measurements

2.19 Fracture Modes for Cyclic Loading in Vacuum,

$$K_I = 6000 \text{ psi} \sqrt{\text{in}}, \Delta K = 45 \text{ psi} \sqrt{\text{in}}$$

(a) Conventional

(b) Modified

(c) Conventional

(d) Modified

2.20 Striation Spacings: Cyclic Loading in Vacuum,
 $T = 650^{\circ}\text{C}$, $AK = 45 \text{ MPa} \sqrt{\text{m}}$

2.1 Regions Below Fracture Surface: Cyclic Loading

In Vacuum, $\Delta K = 45 \text{ MPa}\sqrt{\text{cm}}$, conventional loading

(a) Fracture surface (b) Fracture surface

(c) Fracture surface (d) Fracture surface

2.21 Reflection of low angle fracture surface of Ti-6Al-4V alloy
 in vacuum, $\Delta E = 45 \text{ eV}$, 10 kV , 10 mm at 100°C
 at rest

2.24 Fracture Modes for Cyclic Loading at 10 MPa

$\Delta K = 0.5 \text{ MPa} \sqrt{\text{m}}, f = 60 \text{ Hz}$

(a) Conventional

(b) Modified

(c) Merrick

(d) Modified Merrick

2.24 Striation Spacings in Inconel 718 Given Modified
Heat Treatment and Tested Under Cyclic Loading
in Air: $\Delta K = 45 \text{ MPa}\sqrt{\text{m}}$

2.25 Region Below Fracture Surface of Inconel 718 Given
Modified Heat Treatment and Tested Under Cyclic
Loading in Air $K_{IC} = 55 \text{ MPa} \sqrt{\text{cm}}$

2.2b Regions Below Fracture Surface; $\Delta K = 45 \text{ Mpa } \sqrt{\text{m}}$

(a) and (b) Grain Boundary Oxidation

(c) and (d) Internal Cavitation

2.27 Carbide Cracking, Possible due to Prior Oxidation

2.28

Planar Deformation Behaviour; $K = 45 \text{ MPa } \sqrt{\text{m}}$;

Cyclic Loading in Air

(a) Conventional

(b) Modified Merrick

2.29 Fracture Modes for Static Loading in Vacuum:

(a) $100 \text{ MPa } \sqrt{m}$, $T = 0.01 \text{ m}$

(b) Conventional

(c) T-diffel

(d) Serrif K

(e) T-diffel Serrif K

2.30 Comparison of Crack Growth Rates in Inconel

718 in Air and Vacuum, T = 650°C, $\Delta K = 45$

MPa $\sqrt{\text{m}}$

(a) Static Loading

(b) Cyclic Loading

2.31 Limited Intergranular Separation in Specimens

Tested Under Static Loading in Vacuum

a. Merrick

b. Modified

2.32 Fracture Modes for Static Loading in Air,

$$\Delta K = 45 \text{ MPa } \sqrt{\text{m}}$$

a. Conventional

b. Modified

c. Merrick

d. Modified Merrick

- 2.33 Planar Deformation Behaviour, $\Delta K = 45 \text{ MPa} \sqrt{\text{cm}}$;
 Static Loading in Air
- a. Conventional
 - b. Modified Merrick

2.34 Oxidation Damage During Static Loading in Air,
 $T = 650^{\circ}\text{C}$, $\Delta K = 45 \text{ MPa}\sqrt{\text{cm}}$
 (a) Conventional
 (b) Merrick

2.35 Large Grain Boundary Separation Observed
Developed During Stiffening Process

2.3b Fracture Modes for Cyclic Loading in Vacuum,

$\Delta K = 67.5 \text{ MPa} \sqrt{\text{m}}$, $T = 600^\circ \text{C}$

- a. Conventional
- b. Modified
- c. Merrick

2.37

Striation Spacings for Cyclic Loading in Vacuum:

$\Delta K = 67.5 \text{ MPa } \sqrt{\text{m}}$, $T = 650^\circ\text{C}$

a. Conventional

b. Modified

c. Merrick

2.38 Ductile Tearing During Testing at Intermediate
Stress Intensity Factor

2.39

Fracture Modes for Cyclic Loading in Air,

$\Delta K = 67.5 \text{ MPa } \sqrt{\text{m}}$, $T = 650^\circ\text{C}$

- a. Conventional
- b. Modified
- c. Jerrick

2.40. Striations Observed on Fracture Surfaces:

Cyclic Loading in Air, $K = 67.5 \text{ MPa } \sqrt{m}$, $T = 650^\circ\text{C}$

(a) Conventional

(b) Modified

2.41 Fracture Modes for Static Loading in Vacuum:

$\Delta K = 67.5 \text{ MPa } \sqrt{\text{m}}$, $T = 650^\circ\text{C}$

(a) Conventional

(b) Modified

(c) Merrick

2.42 Fracture Modes for Static Loading in Air,

$\Delta K = 67.5 \text{ MPa } \sqrt{\text{m}}$, $T = 650^\circ\text{C}$

(a) Conventional

(b) Modified

(c) Merrick

2.43 Fracture Modes for Cyclic Loading in Vacuum,

$\Delta K = 90 \text{ MPa} \sqrt{\text{m}}$ $T = 650^\circ\text{C}$

(A) Concentration (c) Herriek

(b) Modified (d) Modified Herriek

2.44

Striation Spacings for Cyclic Loading in Vacuum,

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SUMMARY AND CONCLUSIONS

1. A detailed review of the physical metallurgy of Inconel 718 has been presented. This review indicates that while a number of earlier studies have attempted to develop heat treatments for Inconel 718 to increase strength, or fracture toughness or elevated temperature crack growth resistance, no attempt has been made to develop microstructures optimised for gas turbine disc applications where a combination of high tensile strength, creep strength, LCF life and elevated temperature crack growth resistance would be needed. This is because the bore sections of the disc would experience high stresses and the material at the bore would need to have a high tensile strength to withstand tendencies for burst. The rim sections would operate at relatively high temperature and would need good creep strength as well as high resistance to creep-fatigue crack initiation and propagation.
2. An experimental procedure for determining creep and fatigue crack growth rates in Inconel 718 at 650°C, in air and vacuum, has been described. The tests were conducted (at K , $\Delta K = 45, 67.5$ and $90 \text{ MPa } \sqrt{\text{m}}$) using a tapered double cantilever fracture mechanics specimen designed to provide constant K , ΔK conditions over a 25.4 mm (1.25 in.) crack length interval. The effect of an oxygen bearing environment and microstructure upon elevated temperature crack growth behaviour in this alloy has been discussed.

3. The elevated temperature crack growth resistance of this alloy in air is influenced primarily by the grain size, grain boundary morphology and matrix γ'' particle size.
4. Crack growth rates under both static and cyclic loading conditions were much lower in vacuum than those in air.
5. Under cyclic loading conditions crack growth rates in vacuum were almost independent of microstructure. Some differences were noted under static loading conditions (in vacuum) but these were very small in magnitude relative to the substantial variations with respect to microstructure, noted in air.
6. In air, heat treatments which produced a coarse grains size (Modified heat treatment) and serrated grain boundaries (Modified Merrick heat treatment) were found to be substantially more resistant to crack growth than material of heat treatments which produce a fine grain size (Conventional heat treatment) and a planar grain boundary morphology with very few grain boundary precipitates (Merrick heat treatment).
7. Some of the mechanisms of microstructure-environment interaction developed by earlier investigators seem to be capable of explaining the observed differences in crack growth behaviour in air. Further work is necessary to confirm that these explanations are indeed applicable.

RECOMMENDATIONS FOR FUTURE WORK

1. The microstructural features which are necessary to improve crack propagation resistance in Inconel 718 have been defined and some heat treatments have been developed which combine these desirable features. Further work is now required to evaluate material given these heat treatments with respect to mechanical properties. Data is also required to be generated over a wider range of temperatures (room temperature to 650°C).
2. It is also needed to evaluate the fatigue crack propagation resistance of Inconel 718 under atmospheres containing aggressive species such as H_2S or SO_2 which would be present in the turbine section during engine operation. The crack growth rates determined here are likely to be much slower than would actually occur under the influence of these atmospheres. Further, there is no guarantee that the microstructures exhibiting good crack growth resistance in laboratory air will perform in a superior manner in a H_2S or SO_2 containing environment as well.
3. Rejuvenation treatments as carried out for the A-286 turbine wheels are likely to be feasible for the Inconel 718 turbine wheels as well. Retired Inconel 718 turbine wheels should be available soon from the J85-CAN15 engine and a similar study on these components would be useful.
4. This study on Inconel 718 has concentrated totally upon crack propagation resistance. Equal importance needs to be given to crack initiation behaviour. Microstructural features which promote crack

propagation resistance may have the opposite effect on crack initiation resistance. Further, if rejuvenation treatments are to be applied, it would be essential to establish that crack initiation as well as propagation resistance are restored by the treatment applied.

APPENDIX D

LITERATURE REVIEWS ON FATIGUE AND CREEP FRACTURE

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Nomenclature

a	crack length
a_o	material constant
b	fatigue strength exponent
c	fatigue ductility exponent
C	compliance
C^*	path dependent energy rate line integral
d	grain diameter
D	grain boundary diffusivity
D_B	grain boundary diffusion coefficient
D_v	bulk diffusion coefficient
E	Young's Modulus
f	volume fraction of precipitate
J	energy release rate per unit thickness
K	stress intensity factor
K_{th}	threshold stress intensity factor
ΔK_o	threshold stress intensity for $R = 0$
ΔK_{th}^*	ΔK_{th} for short cracks
l	plastic zone size
n	stress sensitivity exponent
N_f	cycles to failure
P	load
Q	activation energy
t_r	stress rupture life
T	temperature
T_m	melting point
T_i	traction vector

U^*	path independent energy rate line integral
W	strain energy density
W^*	strain energy rate density
α, γ	material constants
δ	crack tip opening displacement
ϵ_0	reference strain
$\Delta\epsilon$	strain range
$\Delta\epsilon_e$	elastic strain range
$\Delta\epsilon_{in}$	inelastic strain range
$\Delta\epsilon_p$	plastic strain range
ϵ'_f	fatigue ductility coefficient
$\dot{\epsilon}_{min}$	minimum creep rate
ρ	average precipitate particle size
ν	frequency
σ_g	gross section stress
σ_{net}	net section stress
σ_m	maximum applied stress
σ_u	ultimate tensile strength
σ'_f	fatigue strength coefficient

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APPENDIX D

Chapter 1

Fundamentals of Cyclic Plastic Deformation and Low Cycle Fatigue at Room Temperature

If a metal is subjected to cyclic loading to produce plastic deformation of a predetermined amount in both tension and compression, the resulting cyclic stress-strain response (CSSR) will produce a hysteresis loop similar to that of Fig. 1.1. The tips of the loops represent the stress and strain limits for each cycle, and the area inside the complete loop is the energy associated with the plastic work per cycle. Since some of the plastic deformation during cyclic loading is reversible, the work dissipated during fatigue failure is always much larger than required for failure in a monotonic tension test¹. The irreversible component ultimately leads to fatigue failure.

1.1 Cyclic Hardening and Softening

Since plastic deformation is not completely reversible, microstructural changes occur during cyclic straining, and lead to changes in flow properties. These changes which will be described in later sections are loosely described as "microstructural damage". Depending upon the initial state and test conditions, the deformation resistance of a material may increase (cyclic hardening), decrease (cyclic softening) or remain essentially unchanged (cyclic stability)². In many cases, all three phenomena may occur, and the sequence may vary, again depending upon the initial state of the sample and test conditions. The final stress-strain response (prior to crack initiation) may be quite different from the initial response.

The initial hardening is usually associated with the generation of a stable dislocation substructure. Later softening may be associated with precipitate resolution (e.g. in Al alloys and nickel-base superalloys) or precipitate shearing by dislocation, channelling and possible overaging in the superalloys³.

Low cycle fatigue tests to determine the life to crack initiation are often conducted by holding the total strain $\Delta\epsilon$ constant (Fig. 1.1). Softening, hardening or saturation can be monitored by measurement of $\Delta\sigma$ for a strain controlled test. A schematic representation of the various cyclic response curves possible is shown in Figs. 1.2-1.4.

1.2 Deformation Modes

Plastic deformation during fatigue can occur predominantly by reversible, "co-planar" slip and by irreversible, "wavy" slip. Planar slip is favoured at low temperatures, high frequency of cycling, low stress and low stacking fault energy⁴.

In planar slip, dislocations are confined to glide in individual slip planes and pile-up against barriers such as grain boundaries and incoherent precipitate particles, creating a long-range elastic repulsion (Fig. 1.5). This leads to an internal "back stress" that contributes to monotonic strain hardening and aids the reversed motion of dislocations during unloading (Bauschinger effect). If strong barriers to planar slip are not present, as in the case of single crystals, the rate of monotonic strain hardening is much reduced

and so is the reversibility of slip at the surface⁵. In general, slip reversibility is considered beneficial to fatigue resistance and ideally, a planar slip material can be cycled indefinitely without change in the stress-strain behaviour. However, it should be noted that in a number of commercial alloys strengthened by precipitates, intense planar slip can cause softening in the slip bands, resulting in strain localization and ultimately in crack initiation^{2,6-8}. For this reason, the nickel-base superalloys which are strengthened by coherent precipitates are considered to have relatively poor fatigue strength⁹.

If the temperature, stress, and stacking fault energy are sufficiently high, or the frequency low, the material may not be able to sustain the high internal stresses created by a planar slip character. This leads to various internal stress relaxation processes such as particle-matrix interface cracking, activation of additional slip systems, dislocation climb or cavitation^{10,11}. Slip is then dispersed on the neighbouring slip planes and the dislocation structure becomes "wavy", or more homogeneous and isotropic.

Such relief of long-range stresses reduces the mechanical reversibility of slip while creating additional dislocation substructure. The increased dislocation density leads to localized cyclic strain hardening. Eventually, the dislocation structure may attain equilibrium and the level of strain hardening will reach a saturation point.

There appears to be some confusion in the literature regarding the terminology associated with slip character. Wells¹² for example, has suggested that planar slip is "heterogeneous" and leads to severe notching. In contrast, McEvily and Johnston¹³ concluded that planar slip was beneficial in "dispersing" slip and thereby reduced the severity of notch peak topography. A further complication comes from the description of planar slip as "coarse" slip and wavy slip as "fine" slip¹⁴.

Gell and Leverant¹⁴ have clarified this by pointing out that these definitions need to be applied differently to two-phase and single phase alloys. Because of the presence of a strengthening precipitate in a two-phase alloy, slip activity once nucleated in a few bands, continues in these bands with little localized work hardening. As a result, deformation is heterogeneous, occurring in widely spaced "coarse" planar bands and both the monotonic and cyclic strain hardening rates are negligible. On the other hand, in a single phase alloy, though slip may still be planar, it is much more "homogeneously" distributed. The homogeneity of deformation is a consequence of local strain hardening in the slip bands and is reflected in the very high cyclic strain hardening rates of these materials. For the same plastic strain range, when there are more active slip bands, there will be smaller shear displacements in each band and crack initiation will be retarded. Thus, in a single phase alloy, though the slip character may be "heterogeneous" on a microscopic scale, the significant extent of slip dispersal which takes place causes the slip to be "homogeneous" on

a macro scale. What would be "coarse" slip in a two-phase alloy may become relatively "fine" in a single phase alloy exhibiting the behaviour mentioned earlier.

As noted earlier, the heterogeneous nature of the planar slip which takes place in several Ni-base superalloys leads to large strain accumulation in the slip bands and early crack initiation. If the planar slip can be more homogeneously distributed as in several single phase alloys, it may be possible to raise the endurance limit of the superalloys¹⁴.

Perhaps, the above discussion can be further clarified by pointing out that whatever may be the definition given to the slip character, the propensity for crack initiation depends primarily upon the reversibility of slip in a band¹⁵. The extent to which slip band displacements are microscopically reversible depends upon the density of active slip systems, dispersal of slip, stacking fault energy, flow stress and shear modulus. Irreversibility of slip is reflected by cyclic strain hardening at low temperatures and may result from intersections between slip systems, dipole accumulation, cell or tangle formation accompanying cross-slip and so forth³.

In wavy slip materials, a unique steady-state stress is achieved on cycling at a given strain level³. Planar slip metals and alloys, however, do not have a unique steady-state and so have a variable cyclic stress-strain relationship which is history dependent. Planar slip materials include both stainless steels and nickel-based superalloys.

It will be seen in a later section how the internal stresses accompanying blocked slip-bands in planar slip materials are sources of grain boundary damage at elevated temperature. In this case, creep cavitation may result from the relaxation of stresses due to impingement of slip on grain boundary particles as piled-up edge dislocations are released by void growth on the tension side of the slip plane. In the presence of an aggressive environment, planar slip character intensifies metal-environment interactions, and can cause significant reductions in fatigue life. The subject of environmental interactions has been dealt with in detail in a subsequent section.

1.3 Mechanisms of the Effect of Vacuum on Plastic Strain in Cyclic Loading

A peculiar effect observed with Cu-7.5% Al specimens is that for the same number of cycles their microhardness is higher in vacuum than in air at any stage of the fatigue process¹⁶ (Fig. 1.6). It is evident that the hardening of the alloy takes place under cyclic deformation, the subsurface being hardest at a depth of about 20 μm (Fig. 1.7). The microhardness of the core also increases but to a much smaller content. Theories¹⁷⁻²² which attempt to explain effects such as these assume that,

- (a) surface hardening is caused by the dominance of near surface dislocation sources over internal sources as a result of the greater ease with which near surface sources act in vacuum.
- (b) mobile dislocations interact with the free surface.

Makhlín²³ has enumerated four possible effects of the interaction of the surface with dislocations. The first is due to the inhibitory action of a thin oxide film on dislocations moving towards the surface. The manner in which it manifests itself depends on differences in the elastic moduli and crystal structures of the film and the substrate. The second surface retardation effect is due to the pinning of surface (Fisher) sources by the oxide film. The third effect is associated with the pinning of dislocations on their way to the surface, and the fourth with changes in lattice parameters and elastic constants of the surface layer. Oxide film formation is not the only cause of interaction between the surface and dislocations. Grosskreutz²⁴ in his experiments on aluminum and gold demonstrated the barrier effect to be characteristic of the surface layer itself, regardless of whether an oxide film is present or not.

In the case of planar slip, vacuum increases the number of slip bands and hardens the surface layer. It is therefore reasonable to suppose that more dislocation sources act in vacuum than in air. This hypothesis was suggested by Alekseev and Grinberg²⁵ and has been proved experimentally by measuring slip band parameters in single crystals of copper strained in air and vacuum in the low amplitude fatigue range²⁶. During the entire fatigue process the number of slip bands at the surfaces of vacuum tested specimens appeared to be an order of magnitude higher than the number in air tested specimens for the same number of cycles. In this study, the number of slip bands in any environment did not change with deformation in the hardening

saturation range but their density increased due to band lengthening. It therefore appears that the number of dislocation sources remain constant as the cumulative plastic strain increases but that it depends strongly on the environment. In the case of cyclic loading, vacuum may be expected to cause a decrease in the initial stress of dislocation sources by diminishing any oxide or absorbed film. Thus, dislocation generation from a greater number of sources is guaranteed throughout the deformation process.

Increased hardening of the subsurface layer has not been observed for the case of wavy slip, in which a cell dislocation structure is formed. This could be because of the ease of cross slip in the bulk of the specimen. Consequently, subsurface sources do not appear to dominate internal sources and the effect of vacuum on subsurface layer hardening is insignificant²⁷.

1.4 Stages in Fatigue Failure

In this review, the fatigue failure process will be treated in two basic stages, namely crack initiation and crack propagation. As observed by Wells et al.⁵, what one chooses to call initiation depends both upon the problem under investigation and the techniques available for examination. Depending upon how crack detection is made - by the naked eye, an eddy current device, a metallographic microscope or an electron microscope crack initiation may be defined as a crack length anywhere between 1/8" and 1000 Å. In this review, it will not be attempted to define a fixed crack size as the criterion of initiation,

but the scale on which the various mechanisms of microcrack initiation operate will be discussed. The subject of microcrack propagation by extension and link up will be treated in the section on crack propagation.

Once initiated, fatigue cracks grow in three stages, as shown schematically in Fig. 1.8. Stage I growth is essentially growth along a crack tip shear plane, probably along a dislocation wall. Single crystals of superalloys show extensive Stage I cracking behaviour²⁹. In polycrystalline materials Stage I cracking is usually confined to the first few grain depths into the materials³⁰. At this point, Stage II growth may take over, with localized crack tip flow occurring in both maximum shear directions at the crack tip ($\pm 45^\circ$). The transition to Stage II fracture is believed to occur as a result of the increasing difficulty of the crack finding favourably oriented slip planes all along its front, as the crack length and consequently the extent of crack tip yielding increase³¹. Stage III growth is a precursor of final unstable fracture.

Stage I fractures produced during high strain low-cycle fatigue are relatively featureless and non reflective at low magnifications, but at higher magnifications show a uniform distribution of equiaxed dimples, similar to those observed on tensile fracture surfaces of ductile materials³². A marked change in appearance is found on Stage I fractures produced at low strain ranges.

The surface is highly reflective and contains distinct fracture steps and river lines similar to those observed on cleavage fracture surfaces of b.c.c. and h.c.p. materials. At higher magnifications it has been found³³ that the regions between fracture steps consist of relatively flat plateaus containing slip offsets. In spite of the resemblance of these fracture surfaces to those obtained by cleavage, it has been shown³⁴ that the crack propagation in Stage I cracking occurs slowly and in a cyclic manner.

In stage II growth, the repeated operation of crack tip flow bands, results in a striated fracture surface³⁴. Some qualitative, mechanistic pictures of crack extension in the Stage II mode are shown in a later section on the mechanisms of cycle dependent fatigue crack propagation. Crack growth during Stage III results in a rough fracture surface and can be considered as crack extension by a combination of Stage II continuum-controlled growth and static fracture modes³⁴.

It has been observed that the same three-stage fatigue crack growth sequence is found in high strain low cycle fatigue tests as well as in low-stress fatigue crack growth tests on pre-cracked specimens where small scale yielding conditions are present at the crack tip³. According to Tomkins³, it is possible that both the Stage I-II transition and the Stage II-III transitions are related to the crack-tip opening displacement (δ) achieving a critical structural size. The Stage I-II transition is believed to occur when δ approaches a dislocation substructure size at which symmetrical operation of the flow bands is possible. Stage II-III transition may occur when δ is of the order of the interparticle spacing.

1.5 Laboratory Simulation of Fatigue Failure

In one important form of simulation a smooth hourglass or cylindrical specimen is used to study the strain and temperature history effects in the notch. A closed-loop servohydraulic testing machine may be used to assist in the strain-temperature-time simulation. Such specimens can be used to describe the processes of crack initiation and early growth; matters related to wave shape, frequency, hold time and environment; and for life prediction. Thus, it forms the basic key to design based on an initiation criterion.

The crack once initiated, propagates in the high strain regime in a manner called 'high-strain' fatigue or "short crack" propagation. Eventually, the crack breaks out of the plastic zone resulting from the notch or fillet geometry and propagates more or less elastically except for a small plastic zone at the crack tip. This behaviour is best modelled with pre-cracked fracture mechanics specimens (e.g. a compact tension specimen).

1.6 Fatigue Crack Initiation at Room Temperature

As noted earlier, crack initiation during cyclic deformation generally occurs in regions where strain has become concentrated or localized. Overall hardening or softening may occur prior to strain localization, but in general, these effects are not critical to the crack initiation process³⁵. Strain localization may occur in the presence of,

- (1) pre-existing stress concentrations such as notches
- (2) steps or notches produced during fatigue by fretting or by wavy slip offsets at surfaces
- (3) localized soft regions such as precipitate free zones (PFZ)
- (4) shearable precipitates
- (5) hard carbide particles and microstructural defects such as inclusions.

Crack initiation by strain localization along slip bands occurs more readily in the wavy slip mode, rather than in the planar slip mode, principally because surface slip is more reversible in the latter and concentrations take longer to develop¹⁴. In the presence of wavy slip, the most severe notches and subsequent cracking generally occur adjacent to grain boundaries or inclusions since, on either side of an interface, the strain often differs markedly³⁵. Significant nucleation or egress of dislocations can occur at the base of notches so produced and such dislocation activity will generally extend the crack.

The formation of a notch-peak topography is observed upon cycling in a high stacking fault energy, wavy slip materials. The lack of constraint at the surface causes the dislocation substructures generated by cycling to extend out of the grains at the free surface to form what are often called "Persistent Slip Bands" (PSBs, Fig. 1.9). Their persistence comes from the fact that removal of the surface topography by electropolishing does not remove the bands which readily reform a notch peak surface profile on recycling. Persistent slip

bands form early in the life and are completely formed soon after a saturation hardening condition is reached. Once formed and stabilized, they remain throughout the life. So, for many alloys, PSBs are present for well over 90% of the life. The PSBs are believed to be soft zones containing bands of small cells and sub-grains³⁶. As they represent weak paths in the material, they are clear crack initiation sites, giving crack initiation within the first few percent of life^{3,37}.

Wavy slip can lead to a variety of other fatigue crack initiation mechanisms at external surfaces, such as development of a notch peak topography by intrusion and extrusion along PFZs, pore formation in zones of distortion and surface instability - all of which are consequences of the irreversibility of wavy slip⁵.

While slip band cracking is the common crack initiation mode in pure and single-phase alloys (and in some two-phase, non-precipitation hardened alloys such as Ti-6Al-4V), many high strength commercial alloys fail by means of cracks which are generated at second phase or impurity inclusions³⁸. In the cast nickel-base superalloys, crack initiation at room temperature is almost always associated with microstructural defects such as brittle phases, casting pores, and coherent annealing twin boundaries⁹. Possible exceptions to this are wrought, recrystallized alloys and carbide free single crystals in which slip band cracking is more predominant. The role of microstructural defects and other microstructural features in crack initiation is reviewed in detail as follows.

1.7 Factors Influencing Fatigue Crack Initiation

1.7.1 Intrinsic Factors

Cracking in Brittle Phases: In the nickel-base superalloys, the various brittle phases present (e.g. carbides, carbo-nitrides, sulpho-carbides and borides) may become cracked at some point during the processing history of the material. Cracks in MC carbides have been observed by Gell and Leverant³⁹ in the as-cast microstructure of directionally solidified MAR-M200 and by Slack⁴⁰ in Udimet 700. At stresses higher than about one fifth of the yield strength, these pre-cracked carbides are the sites of crack initiation⁹, and fatigue life at room temperature is a strong function of carbide size.

Deformation induced cracking can also occur in initially sound, brittle phases. In the planar slip mode, fracture of these phases can occur by impingement of slip bands⁴¹⁻⁴³ or by a sufficiently large tensile strain in the matrix. Carbide fracture can also occur from impact in glass-bead blasted surfaces⁴⁴. Various machining, operations such as grinding and turning may also produce fractured particles at the surface. Any relative motion of the two halves of a fractured particle will propagate slip bands from the crack tip and lead to the initiation of a fatigue crack. According to Gell et al.⁹, the only effective way of suppressing this process at low temperatures is by achieving a surface compressive stress. Thus, the residual compressive stress due to operations such as glass-bead blasting may offset any harmful effects of particle fracture.

Porosity: Gas or shrinkage porosity can be equally harmful as brittle phases in initiating fatigue cracks. In the case of carbide-free single crystals, porosity is believed to be the primary source of fatigue cracking⁹. Surface-connected pores have been found to be more damaging than internal pores and large pores much more damaging than smaller ones.

Coherent Annealing Twin Boundaries: Coherent twin boundaries are often sites of crack initiation in wrought superalloys⁴⁴. It has been generally concluded that the fatigue process is accelerated at such locations when certain orientation relationships are satisfied, and that incoherent steps in the twin boundaries act as dislocation sources, thus concentrating plastic deformation⁵. Although coherent annealing twin boundaries are active sites of deformation, and cracking can be related to this deformation it is accepted that at elevated temperatures where grain boundaries become the primary site of crack initiation, the twin boundaries are no longer of major importance.

Coherent Precipitates: In alloys strengthened by coherent precipitates, the precipitate particles may be sheared or looped and bypassed by dislocations during plastic deformation depending upon the size, spacing and degree of coherency. The strengthening associated with particles that are sheared is usually due to long range order; coherency strains; surface chemical, and/or stacking-fault-energy effects; or modulus hardening³⁴. This strengthening is reduced when the coherent particles are sheared by

by dislocations⁴⁶. This results, successively, in a local decrease in resistance to further dislocation motion, concentration of slip and may lead to the destruction of the strengthening agents⁷. As noted earlier, such concentrated slip may lead to early crack nucleation in slip bands. The degree of softening, magnitude of the stress concentrations produced at the grain boundaries, and the magnitude of the slip step at the surface all depend on the number of dislocations that can be accommodated on a given slip plane. This is controlled by the slip length and the reduction of particle hardening in the slip plane, both of which may be altered by modifications of the microstructure⁴⁷.

Overaging of precipitation hardening alloys causes the precipitates to become partly incoherent and this changes the precipitate-dislocation interaction from one of shearing to that of dislocation looping or bypassing. In the case of Waspaloy it has been demonstrated⁴⁸ that such a change brings about a reduction in strain localization and a more homogeneous distribution of plastic deformation.

Although overaging homogenizes slip, and increases an alloy's resistance to fatigue crack nucleation, it is normally accompanied by a reduction in static strength. Consequently, it may be beneficial to have a dispersion of non-shearable precipitates intermixed with shearable precipitates to disperse the slip. This effect has been successfully demonstrated in 2024 aluminum alloys^{46,50}.

Grain Size: A reduction in grain size has beneficial effects that delay crack nucleation in alloys containing shearable precipitates⁵¹⁻⁵³. A reduction in grain size reduces the slip length and therefore the stress concentration by reducing the number of dislocations in a pile-up. A reduction in slip length also reduces the number of dislocations that can egress at a free surface and thus, the slip-step height and extrusion intrusion size. It is also believed⁵⁴⁻⁵⁵ that the smaller the grain size, the larger the volume fraction of material deformed by multiple slip and the more homogeneous the overall deformation.

Figure 1.10, illustrates the beneficial effect of reducing grain size for a Al-Zn-Mg-Zr alloy containing shearable precipitates⁵¹. Coarse, planar slip and intense slip bands, which were later sites for crack nuclei, occurred early in the life of the coarse grained (Al-Zn-Mg) alloy. Slip in the fine grained material was less intense and crack initiation was delayed. Cycles to initiation in the fine grained alloy exceeded the number of cycles to failure in the coarse grained material.

Figure 1.11 shows the grain size effect in a stress-controlled test for a high purity 7075 alloy (X-7075) aged to contain shearable precipitates⁵³. The yield stress was approximately the same for both materials compared in this figure. Optical examination of the specimen surfaces showed that cracks nucleated much earlier in specimens with the large grain size⁴². However, the slip bands were much more pronounced in specimens with a grain size of

220 μm , as compared to specimens with an average grain size of 30 μm . For the latter specimens, cracks at slip bands could be detected only in grains that were statistically larger than average.

The grain size effect mentioned above has been demonstrated for high temperature alloys such as Alloy 901 and Udimer 700^{54,55}. A reduction in grain size will place an upper limit in terms of temperature of application due to the considerable loss in yield strength that can occur at elevated temperatures in fine grained materials. As will be seen later at elevated temperatures, a fine grained material is also less resistant to time-dependent crack propagation in comparison to a relatively coarse grained material of the same composition. However, for ambient to moderately elevated temperature (<300°C) applications, for example, turbine shafts, and compressor disks, microduplex processing and lower annealing temperatures may be used to refine the grain size and promote fatigue resistance⁵⁴.

The differences in fatigue crack initiation in single crystals and polycrystals is mainly related to differences in slip character. As mentioned earlier, two extremes are possible. If slip is reversible, no defect accumulation, strain hardening, or degradation of the cohesive strength would be expected and crack initiation would have to be governed by effects such as oxidation of the surface slip offsets by the environment. On the other hand, if slip is irreversible, then the possibility exists for strain hardening, dispersal of slip activity onto neighbouring planes, defect

accumulation and a decrease in fracture energy of slip planes. There is evidence that the behaviour of polycrystals is close to the former case and single crystals to the latter⁹. The lesser amount of reversibility in single crystals is thought to be related to the lack of grain boundaries to act as obstacles to dislocations and create a "back-stress", that would aid the reversibility of slip. However, directionally solidified and single crystal forms of cast Ni-base superalloys offer a distinct advantage over polycrystalline materials at elevated temperatures due to the elimination of intergranular cracking. Gell et al.⁹ have also pointed out that while attempting to determine the effect of grain size on LCF crack initiation behaviour, the various heat treatments used to produce a range of grain sizes may simultaneously change the structure of the grain boundary and its immediate vicinity, which could affect resistance to both initiation and propagation. Additional work is therefore needed to separate the effect of grain size from concomitant changes in grain boundary structure. It is also not known at this time whether grain boundary features such as serrations which have a beneficial effect on elevated temperature mechanical properties^{56,57} would influence LCF crack initiation resistance in a favourable manner at ambient temperatures.

Precipitate Free Zones: In many commercial age-hardened alloys, the preferential precipitation of stable incoherent phases at the grain boundaries can lead to a solute depleted Precipitate Free Zone (PFZ) adjacent to the grain boundary. This zone is weaker than the matrix and can be the site of preferential deformation⁵⁸⁻⁶¹, leading to high

stress concentrations at grain boundary triple points and early crack nucleation⁶². The magnitude of the stress concentrations will be a function of the grain boundary length and the difference in shear strength of the age hardened matrix and the soft PFZ⁶¹. Reducing the grain size is believed to be a very effective method of reducing early crack nucleation due to preferential deformation in the PFZ⁶¹. This reduces the slip distance and lowers the stress concentrations at grain boundary triple points. In high purity 7075 alloys, another effective method of reducing slip length within the PFZ is to introduce steps or "ledges" into the grain boundaries by cold deformation. With about 50% cold work, the incidence of grain boundary cracking was considerably reduced and the fatigue life improved at high stress amplitudes⁶². It may also be possible to minimize the effect of precipitate free zones in high strength aluminum alloys by choosing a processing schedule that would cause grain alignment in a certain direction. If the stress axis in the actual component is parallel or perpendicular to the long grain dimension, there will be no shear stress parallel to the grain boundary and preferential deformation within the PFZ will be restricted⁶³.

1.7.2 Extrinsic Factors

Environmental Effects: The fatigue resistance of metals can be profoundly affected by environmental reactions that affect crack initiation and/or propagation. In the case of gaseous environments, oxide films or in some cases, gas adsorption alone has been related to premature crack initiation for some materials⁶⁴. Yet, for a number of

other materials crack initiation does not appear to be affected by gaseous environments and there is a significant amount of controversy on the criteria that govern these phenomena.

In gaseous environments, on the other hand, virtually all corrosive environments affect crack initiation processes, sometimes by a simple phenomenon of pits acting as stress concentrators as well as through much more complex phenomena. This review will be limited to a discussion of the effect of gaseous environments and the reader is referred elsewhere⁶⁴ for a detailed review on the effects of aqueous environments.

Very limited information is available in the literature regarding the influence of gaseous environments, and as stated earlier, the information that has been generated is surrounded by a great deal of controversy. Thompson et al.²⁷ has expressed the opinion that cyclically generated slip bands produced in air become regions of high oxygen concentration because of strain induced vacancy generation. Dissolved oxygen then serves to prevent rewelding of nascent cracks and accelerates the transition from slip band to microcrack (Fig. 1.12).

Fujita⁶⁵ has suggested a similar mechanism in which it was shown how surface slipsteps could be progressively oxidized so that surface diffusion combined with alternating slip displacements could result in a slip-band crack, in the absence of other damage mechanisms. This model is said to apply well to Stage I cracking of planar slip materials at temperatures below the range of intergranular cracking. Other investigators have postulated that metal and alloy surfaces are

strengthened with an oxide film and that when cyclically stressed, dislocations are accumulated in surface regions so that the formation of cavities and voids is enhanced (Fig. 1.13), leading to early crack formation⁶⁶. According to this view, in the absence of oxygen, dislocations readily escape from the surface, and crack nucleation is retarded.

Atmospheric water vapour has also been cited as important to the crack-nucleation process in age-hardened aluminum alloys, water vapour either accelerating the oxidation of emerging slipsteps (thus blocking reverse slip)⁶⁷ or, alternatively, water vapour dissociating to form hydrogen, which embrittles the alloy and causes premature cracking⁶⁸.

Grosskreutz and Bowles⁶⁹ have compared the effect of air and vacuum on cyclically induced surface deformation. Marked reductions in the surface deformation of aluminum single crystals were observed at pressures of 10^{-9} torr vs. atmospheric pressures and they suggested that slip was suppressed in vacuum, as a result of an interaction of the environment with newly formed slipsteps. Later work, however, indicated that the reduced slip intensity observed in vacuum was probably also due to changes in the mechanical properties of the surface oxide film. The stripped oxide, tested in vacuum, exhibited a Young's Modulus approximately four times that exhibited in air, apparently because of the elimination of water vapour in the film⁷⁰. Similar observations of reduced slip intensity in vacuum in the presence of thin oxide films has been reported for nickel-base

superalloys, at elevated temperatures^{71,72}. The effect of an oxidizing environment upon crack nucleation at elevated temperatures will be dealt with in more detail in a later section.

In contrast to the findings mentioned earlier, a number of other investigations have concluded that gaseous environments do not significantly influence the crack-nucleation process. Wadsworth and Hutchings⁷³ showed that the appearance of intrusions, extrusions and microcracks was similar in air and vacuum for an equivalent number of cycles, although further development of these features was very much reduced in vacuum. Hordon⁷⁴, investigating the fatigue behaviour of aluminum in vacuum, has expressed the opinion that not only initiation, but the entire Stage I propagation step is influenced by the environment, although adsorption of oxygen at the tip of a growing crack increases Stage II propagation rates and thus reduces fatigue life. Laird and Smith⁷⁵ maintain that crack initiation in pure metals is unaffected by the environment but observed that the early stages of crack growth are increased by air, the effect being less pronounced as the crack growth proceeds. Achter⁷⁶ and Latanision and Westwood⁷⁷ have also related the environmental effect to the propagation stage rather than the initiation stage.

In an extensive review of the effect of vacuum upon fatigue failure, Verkin and Grinberg⁷⁸ have pointed out that the lack of agreement between theory and experimental results in this area is because environmental effects on the nature of the plastic strain processes that take place (particularly the slip character) during

cycling are not taken into account in these models. In a study of surface and subsurface layers in copper subjected to cyclic loading in ambient air as well as vacuum, it has been shown^{79,80} that fatigue crack initiation depends upon the character of the plastic strain distribution. In the low strain amplitude range, microcracks originate from PSBs in both environments. However, it has been shown that the stage of microcrack initiation in a slip band is about 60 times longer in vacuum than in air for this material^{79,80}. This is because of the relative homogeneity of plastic deformation in vacuum which retards the formation of PSBs and subsequent crack initiation.

In the high strain amplitude deformation of copper, the deformation mode is wavy rather than planar. PSBs are not formed in either environment and the cracks are of mixed character (transgranular and intergranular) in both cases. In this situation, no marked influence of the air environment is observed. In both environments, microcracks arise in the initial stage at the same point of the fatigue life and are practically equal in length and number⁷⁹.

In aluminum, which has a high Slacking Fault Energy and hence considerably lower tendency for cracking along PSBs than copper which is a low SFE material, the fatigue life increases in vacuum over the whole range of deformation amplitudes⁸¹ (including both high and low amplitudes). The degree of increase in life in vacuum is less than that for copper. In this material, X-ray structure analysis and optical microscopy examination have shown that plastic deformation over the entire range of strain amplitudes involves grain rotation and

sub-grain formation, the boundaries of the subgrains being the sites of crack initiation in both environments. In aluminum, even in the low amplitude range, the deformation does not occur in a purely planar mode, but also has some features of wavy slip. The slip homogenization, characteristic of low amplitude fatigue of copper in vacuum does not occur for aluminum. It appears that crack initiation lives are not affected by the environment, and the observed difference in lives is rather an effect of environmental influence on crack propagation rates.

In high strength alloys, a major part of the fatigue life is accounted for by the life to crack initiation⁷⁸. Studies of environmental effects upon crack initiation in these alloys is therefore of considerable importance. However, there is very little information available at present on this subject. It is likely that the effect observed with copper, namely the homogenization effect in vacuum will be observed with complex alloys also.

1.8 Mechanisms of Fatigue Crack Growth at Room Temperature

Fatigue cracks may initiate in the Stage I mode and grow in this crystallographic, faceted mode until a transition to Stage II occurs. Two mechanisms³ of Stage I crack growth are possible, the difference between these being that crack growth occurs along cleavage planes or along slip planes. Crack growth may occur in this crystallographic mode until a transition to Stage II occurs. This transition is believed to occur when the crack tip plastic zone size

exceeds the average grain size of the material. More specifically, the microstructural parameter that determines the faceted growth is related to the mean free path of the dislocations which could depend upon grain boundaries or any other internal obstacles³. Direction of crack growth is normal to the maximum applied tensile stress and in ductile materials the fracture surface is characterised by fine striations. Tomkins³ has observed that striation spacing is often misinterpreted as delineating crack growth rate and although in a number of commercial alloys one striation spacing does represent one increment of crack advance in one cycle, this is not a general observation. In alloys in which striation spacing (μm) has been correlated with macroscopic crack growth rate, careful measurements have shown that μm is consistently larger than the macroscopic da/dN , measured on the surface which indicates that surface crack growth is only an average of local rates in the general directions of Stage II growth³⁷. While Stage I cracking is closely related to the microstructure and the slip character, Stage II is governed more by the modulus, the ultimate tensile strength and the strain hardening rate. ~~Slower growth rates would be obtained~~ by raising the modulus, the ultimate tensile strength and the strain hardening rate.¹³

Striations are formed during Stage II crack growth as a result of the operation of local crack tip slip line fields during the opening and closing of the crack. During the tensile part of the load cycle, large plastic strains at the crack tip cause localized slip and blunting of the crack. Upon reversal of the load, the crack faces are

forced together and the new surface created in tension is folded partly at the crack tip as shown in Fig. 1.14.

After completion of the compressive half cycle, the crack tip is resharpened and the sequence may be repeated all over again. This process is termed the "plastic blunting" mechanism³⁵ and said to account for the fracture surface topography observed in ductile materials as a result of cyclic loading. Another mechanism given by Tomkins³ and called the "Shear Decohesion Mechanism" is shown in Fig. 1.15.

Striations are observed in Stage II growth in high strain low cycle fatigue tests as well as with low stress crack growth tests with precracked specimens. In the low stress crack growth rate test, under plane strain conditions the crack tip yield zone is small. Localised plastic flow occurs ahead of the crack at $\pm 45^\circ$, but this localized flow field is still within the yield zone (Fig. 1.16). In fully yielded, high strain low cycle fatigue tests this localized field is within the plastic matrix (Fig. 1.16). Thus, at the crack tip, a similar slip line field is generated by strain intensification for both cases.

leading to the same Stage II crack growth mechanisms in low stress and high strain fatigue.

At small crack lengths (and hence small crack openings and low crack tip stress intensities) the symmetrical shearing process at $\pm 45^\circ$ shown in Fig. 1.16 cannot operate³ and cracking occurs along a specific shear band, as shown in Fig. 1.17. This essentially is Stage I crack growth. New crack surface is still created by shear

decohesion, but the process is much more reversible, relying on fresh surface-environment interactions to determine the degree of reversibility. The similarities of the Stage I and Stage II processes guarantee a smooth transition in the crack growth curve.

According to Grinberg^{82,83} Stage II can be divided into Stages IIA and IIB. In Stage IIA crack extension occurs only once for every dN cycles. During the rest of the $dN-1$ cycles, the microcrack remains closed and does not advance. Over this period the accumulation of plastic strain in the microscopic zone ahead of the crack tip leads to the energy essential for opening and advancing the crack.

In other words, the crack tip generates dislocations which form stable three-dimensional configurations of dislocation cells. The process of cell formation lasts over a certain number of cycles, the number of which, dN , is determined both by the energy required for formation of a minimum cell substructure. This substructure formation occurs in a small area adjacent to the crack, i.e. in the cyclic plastic zone. Within this zone, each of the substructure parameters (the cell size, the total cell disorientation and the degree of crystal lattice distortion) is a maximum^{84,85} and remains unchanged or varies only slightly throughout the whole stage. The number of cycles which is necessary to form the substructure diminishes as the crack length (or the stress intensity factor) increases. In the remaining macroscopic part of the plastic zone, the substructure is less distorted. The substructure formation in the whole plastic zone and the formation of the most distorted structure in the area adjacent to

the microcrack produces the crack opening. This occurs at the moment when the potential for stress relaxation due to the plastic strain is exhausted and a subsequent increase in the stress leads to the breakdown of the bonding forces between atoms in the plane of maximum normal stresses. The microcrack increases in step by the cell size because the cell boundary, which has a high density of dislocations, is a barrier to crack propagation and the density of dislocations in the bulk cell is ~~several~~ orders of magnitude lower⁸⁶. When crack extension occurs, a striation is formed by the crack tip blunting mechanism mentioned earlier.

The macroscopic mechanism of crack growth is the same for Stage IIb which follows Stage IIa, with the principal difference that the crack opens and advances in each loading cycle. Thus, the presence of fatigue striations in Stage IIa and IIb is associated with the microplastic strain in a grain, at the crack tip. The formation of striationless facets of cleavage or quasi-cleavage does not imply that the fatigue fracture is caused by some other mechanism, but rather that the local conditions of the microplastic strain at the crack tip in a given material were not exactly those required for striation formation. If conditions were changed, for instance, the temperature of cyclic strain, striations could perhaps be produced^{87,88}.

1.9 Factors Influencing Fatigue Crack Propagation³

The influence of a variety of parameters on Stage I cracking has been discussed already. After the transition to Stage II growth,

the net plastic strain in each cycle determines the crack extension in that cycle. The variables which affect Stage II crack growth manifest themselves through their effect on the plastic strain per cycle. Since the elastic energy of plastic deformation is directly proportional to the elastic modulus of the material, any variation of the modulus with temperature affects the growth rates. The elastic modulus has been the most significant variable affecting crack growth rate in vacuum at temperatures below the creep range and other factors which influence the net plastic work per cycle are considered to be relatively unimportant. Speidel⁸⁹ has shown that for several metals and alloys the Stage II crack growth rates in vacuum are virtually the same if the data is normalised, taking into account variations in elastic moduli. However, this does not apply to alloys that become unstable and go through phase changes with temperature or for those that show a ductile-brittle transition with temperature.

An air environment can apparently affect crack growth behaviour very significantly even at temperatures well below the creep range by diffusing into the material and alternating the stress-strain response and slip character of the material close to the crack tip⁷⁸.

Studies on a number of metals and alloys (e.g. iron⁸⁹, Cu-7.5% Al⁹⁰ and magnesium alloy MA12⁹¹) have shown that the character of plastic deformation in the vicinity of a fatigue crack is similar to that at the surface of a given alloy during the initial period of deformation under higher load (strain) amplitudes. In the case of cycling under vacuum, the plastic strain distribution over this zone differs from that in atmospheric conditions.

In an optical study⁷⁸ of the plastic zone in a Cu-7.5% Al alloy, the main crack length was the same in air and vacuum but numerous uniformly distributed thin slip bands could be seen in vacuum and widely spaced intense bands in air. This leads to some changes in the rate and mechanism of crack growth since in air, secondary microcracks are initiated in intense slip bands that subsequently become attached to the main crack. In vacuum, such microcracks are not formed, due to plastic strain homogenization, and hence crack growth proceeds solely by extension of the main crack.

Similar changes in the plastic strain distribution in the vicinity of a crack in vacuum, have been observed in Al-4% Cu alloy⁹². Bouchet et al.⁹² noted that with the same stress intensity factors the slip band density is higher in vacuum than in air. A similar effect in vacuum has been observed for a Al-3% Mg alloy⁹³, as well.

A vacuum environment may influence not only the plastic strain distribution near the crack tip, but also the size of the plastic zone. In a number of metals and alloys, for the same crack length and loading conditions, the size of the plastic zone has been found to be larger in vacuum than in air. Figure 1.18, shows the plastic zone depth as a function of crack length for vacuum tested and air-tested specimens of iron for the high and low strain amplitude ranges⁹⁴. In both ranges, the plastic zone can be seen to be larger in vacuum than in air. Similar results have been obtained for a Al-4% Cu alloy⁹⁵. In X-ray diffraction analyses of fractured surfaces and subsurfaces in magnesium alloy MA12⁷⁸, it has been found that in both

air and vacuum environments a plastic zone composed of subgrains is formed (Fig. 1.19), its depth being larger in vacuum than in air. It is also believed⁷⁸ that in vacuum, a greater number of grains are involved in plastic deformation.

When the extent of yielding at the crack tip is limited, it is usually the case that larger plastic zone sizes correlate with higher fatigue crack growth rates⁹⁴. However, the larger plastic zone formed in vacuum is generally found to correlate with lower crack growth rates. This indicates that it is not only the size of the plastic zone which controls crack growth but also the character of plastic deformation within this zone.

Verkin and Grinberg⁷⁸ point out that the decrease in fatigue crack growth rate in vacuum, in the case of a deeper plastic zone may be the result of a lesser degree of strain concentration (better homogenization of slip) in its local volume.

It is also typical of a plastic zone formed under vacuum that its microhardness is higher than that of a plastic zone formed in air⁷⁸. The higher microhardness of the plastic zone in vacuum is in agreement with the behaviour of the subsurface layers mentioned earlier. This is good support for the hypothesis that the effect of vacuum on plastic strain distributions is the same in the subsurface layer of a smooth specimen and in a near crack area of a precracked specimen.

The most prominent differences between fractures in vacuum-tested and air-tested specimens have been observed by

microfractography of aluminum and titanium alloys⁷⁸. In atmospheric conditions, distinct regular striations indicating the position of the propagating fatigue crack have been observed. In vacuum, such striations were either not observed at all⁹⁶⁻⁹⁹ or were poorly resolved and not regularly ordered. This has suggested the idea that crack growth mechanisms are different in vacuum and atmospheric conditions.

Since the plastic zone formed in vacuum penetrates deeper and extends over a larger number of grains than the zone formed in air, the degree of elastic constraint of grains in it (and hence the level of plastic relaxation) should be less than that in an air-formed zone for the same crack length. This may be one of the reasons why striations are irregular and faint in vacuum. When the plastic zone exceeds a critical size, striations are not formed at all¹⁰⁰.

The deepening of the plastic zone in vacuum does not occur with all materials. In Fe-3.7%Si alloy, the size of the plastic zone is practically the same for air and vacuum¹⁰¹. Slip bands at the surface of this alloy are sparse, straight and weakly pronounced over the whole strain range after cyclic loading. It was difficult to identify any differences in the number and distribution of slip bands in specimens strained in air and vacuum. Microhardness measurements on this alloy proved that some surface hardening takes place up to failure, the microhardness being the same in air and vacuum at the same number of cycles. Detailed X-ray analysis of the specimen surface revealed no signs of the subgrain structure typical of a wavy slip mechanism. This is not unexpected since cross slip is known to be

hindered in this alloy by the low SFE resulting from silicon addition^{102,103}. However, the characteristics of the planar slip mechanism were not very evident either since microcracks initiated at grain boundaries rather than at PSBs. Thus, it appears that the effects due to a vacuum environment mentioned earlier, i.e., homogenization of slip, deepening of the crack tip plastic zone and subsurface hardening are pronounced only in material which exhibit distinctively, the characteristics of planar, heterogeneous slip behaviour (small amount of cyclic hardening, low internal stresses, considerable cumulative plastic strain, bands of dislocation pile ups, formation of PSBs, absence of cell structure).

1.10 Fracture Mechanics Characterization of Fatigue Crack Growth at Room Temperature

Fracture Mechanics essentially involves characterization of the local stress and deformation fields ahead of the crack tip. Under linear elastic conditions, the local crack tip stresses (σ_{ij}) at a distance r ahead of a nominally stationary crack subjected to tensile (Mode I) opening can be characterized in terms of a parameter called the "stress intensity factor" (K_I) defined¹⁰⁴ as follows:

$$\sigma_{ij}(r, \theta) \rightarrow \frac{K_I}{\sqrt{2\pi r}} f_{ij}(\theta), \text{ as } r \rightarrow 0$$

where K_I is the Mode I stress intensity factor,

θ is the polar angle measured from the crack plane,

and f_{ij} a dimensionless function of θ .

The use of K_I to uniquely characterize the elastic stress field at the crack tip is valid only when small-scale yielding conditions exist. Thus, the plastic zone size (r_y) at the crack tip given by,

$$r_y \approx \frac{1}{2\pi} \left(\frac{K_I}{\sigma_o} \right)^2$$

where σ_o is the yield strength of the material is required to be small compared to the in-plane dimensions of the body, namely crack length (a) and remaining ligament depth (b)¹⁰⁵.

Specifically, a small scale crack tip yielding situation is said to exist¹⁰⁶ when,

$$r_y \ll a, b, \quad \text{typically } \ll \frac{1}{15} (a, b)$$

Under cyclic loading, the range of stress intensity ΔK , given by the difference between the maximum (K_{max}) and minimum (K_{min}) stress intensities in the cycle, is commonly used to correlate crack extension, through power-laws of the form¹⁰⁷,

$$\frac{da}{dN} = C \Delta K^m$$

where C and m are scaling constants. Here, the extent of local yielding is defined in terms of maximum (monotonic) and cyclic plastic zones (r_{max} and r_{Δ}) respectively, given approximately by¹⁰⁶,

$$r_{max} \approx \frac{1}{2\pi} \left(\frac{K_{max}}{\sigma_o} \right)^2$$

and,

$$r_{\Delta} \approx \frac{1}{2\pi} \left(\frac{\Delta K}{2\sigma_o} \right)^2$$

Figure 1.20 shows the variation of fatigue crack growth rate with ΔK and the stress ratio R given by $\frac{\sigma_{\min}}{\sigma_{\max}}$. As stated earlier, three distinct regimes can be identified in this sigmoidal curve. Stage II is the least sensitive to ΔK with the exponent 'm' generally in the range 2-4 for most alloys. Stages I and III may be viewed as transient regimes, the first due to the existence of a threshold stress intensity ΔK_{th} , and the second due to the fast fracture upon approaching the fracture toughness K_{IC} .

1.10.1 Influence of Stress Ratio (R)

Figure 1.20, clearly illustrates the effect of the stress ratio upon fatigue crack growth behaviour. In general, R has a large effect in Stage I and Stage III and the least effect in Stage II. In Stage I, crack growth rates are low and therefore the time of exposure of the crack tip to the surrounding damaging environment is high. The observed increase of growth rate with increasing stress ratio is likely to be due to stress-assisted environmental effects^{108,109}. In Stage III, not only cyclic crack growth is occurring, but also subcritical crack growth due to monotonically increasing stress, giving rise to R effects on da/dN vs. ΔK plots. According to Forman¹¹⁰, these effects can be taken into consideration by a modified power-law relationship given as follows:

$$da/dN = A(\Delta K)^m / [(1-R)K_C - \Delta K]$$

Pearson¹¹¹ showed that in a number of cases, a further modification of the above equation, as follows:

$$da/dN = A(\Delta K)^m / \sqrt{[(1-R)K_C^2 - \Delta K]}$$

provides better correlation of crack growth rate data. However, both these modified equations have the drawback that they require prior knowledge of the fracture toughness parameter K_C , which may vary during service exposure.

Elber¹¹² has argued that the so called stress ratio effects are actually due to crack closure effects which may occur for example, due to compressive residual stresses created at the crack tip during cyclic loading. As a result of these residual compressive stresses, the crack closes prematurely during unloading and therefore during subsequent loading the crack is open only during part of the ramp rather than during the complete ramp. By this argument, an effective ΔK (ΔK_{eff}) needs to be computed from the load at which crack opening occurs in each cycle rather than from the minimum load. Based on experimental results, an empirical relation was deduced for U , the ratio of ΔK_{eff} to applied ΔK , as

$$U = \Delta K_{eff} / \Delta K = 0.5 + 0.4R \quad \text{for} \quad -1 < R < 0.7$$

It should be noted that closure can arise from a number of sources, such as the compressive residual stress effect mentioned earlier (commonly called plasticity induced closure¹¹³), the presence of corrosion debris within the crack (oxide-induced closure¹¹⁴⁻¹¹⁶) and the contact at discrete points between faceted or rough fracture

surfaces (roughness induced closure)¹¹⁷⁻¹¹⁸. These various mechanisms are illustrated schematically in Fig. 1.21.

In spite of numerous studies, the crack closure concept is not fully understood. There appears to be a consensus, however, that crack closure does occur mainly at the surface before complete unloading but the crack opening load may not be just a function of the stress ratio R as depicted by the equations mentioned earlier but rather a complete function of K_{max} , specimen thickness and crack length as well.

Thus, Hadrarh et al.^{121,122}, have combined the crack closure concept with superimposed monotonic loading effects and derived the following expression for ΔK ,

$$\overline{\Delta K} = \frac{\Delta K_{eff}}{(1 - \frac{K_{max}}{K_f})}$$

and maintain that this provides a better correlation than ΔK_{eff} . In this expression, K_f is a fracture parameter related to K_{IC} and requires experimental determination.

Sadananda and Shahinian¹²¹ have proposed yet another expression for an effective ΔK by defining an effective crack length as $L+(l-dl)$ where L is the elastic crack length, l is the plastic zone size corresponding to K_{max} and dl corresponds to relaxation of this zone during unloading.

It may be shown¹²¹ that l and dl respectively are given by,

$$l = A \left(\frac{K_{max}}{\sigma_y} \right)^2 \quad \text{and} \quad dl = A \left(\frac{\Delta K}{\sigma_{yc}} \right)^2$$

where $A = 1.06 (1-\nu)/(1+\pi/2)$ and σ_y and σ_{yc} are the yield stresses under monotonic and cyclic loads respectively. Then, an effective ΔK was deduced in terms of applied ΔK as,

$$\Delta K_{eff} = \Delta K \sqrt{1 + A \left(\frac{\sigma_m}{\sigma_y} \right)^2 \left[1 - (1-R)^2 \left(\frac{\sigma_y}{\sigma_{yc}} \right)^2 \right]}$$

where σ_m is the maximum applied stress. This expression has not been justified by experimental results as yet, but points out that stress ratio effects are not the same for all materials but depend upon the monotonic and cyclic flow properties as well. It is claimed that this equation predicts lower R effects for higher strength materials, which is consistent with experimental observations.

Purely empirical relationships are also available to account for observed stress ratio effects. Walker¹²³ has proposed a simple relationship of the form,

$$\Delta K_{eff} = K_{max} (1-R)^m$$

where m has to be determined empirically for a particular material and temperature by correlating known crack growth data¹²⁴. Expressions of this nature tend to be of limited value since the determination of the empirical constants is usually cumbersome. Furthermore, these constants may not have any systematic variation with temperature and other test conditions.

1.10.2 Threshold Behaviour

It was mentioned earlier that below a "threshold" stress intensity factor crack growth does not occur, or is too slow to be detectable. The characteristics of crack propagation at near-threshold

rates are often surprisingly different from behaviour at higher growth rates¹²⁵. Ritchie¹⁰⁹, as well as Dickson et al.¹²⁶, have reviewed the mechanical, microstructural and environmental factors influencing near-threshold growth and the principal aspects mentioned therein are summarized as follows.

Fracture Morphology: While striated fracture surfaces are often considered the principal characteristic of fatigue crack propagation, for near-threshold crack growth rates strongly crystallographic fracture surfaces are observed, which suggest microstructure sensitive fracture modes. These features have been described in terms such as quasi-cleavage, crystallographic decohesion and decohesion on slip planes. Intergranular fracture facets have also been observed in ferritic steels near the threshold limit¹²⁷. Intergranular facets as well as cleavage facets are suppressed in vacuum. This and the observation of several investigators¹²⁸⁻¹³² that corrosion products are formed on the fracture surfaces near ΔK_{TH} for tests carried out in air strongly suggests that the environment plays a significant role in the near-threshold region.

Influence of Extrinsic Parameters: A number of studies have shown that testing in a vacuum better than 10^{-4} - 10^{-5} Torr gives higher ΔK_{TH} than in air and the near-threshold propagation rates are generally lower. However, in most studies of the influence of various types of gaseous environments, ΔK_{TH} values for air appear to be almost the same as for dry hydrogen¹³³ (A517 steel), chlorine¹³⁴ (A36 steel) dry and humid argon^{125,135} (D6ac steel and Ti-6Al-4V). However, this is not an

universal result and depending upon the type of material and the nature of the gaseous environment (e.g. dry or moist) ΔK_{TH} may or may not vary from the value for air.

The load ratio has a very strong influence upon ΔK_{TH} and can be described by a general equation.

$$\Delta K_{TH} = \Delta K_0 \cdot f(R)$$

where ΔK_{TH} is the threshold for a given R-ratio and ΔK_0 is the threshold for $R=0$.

Klesnil and Lukas¹³⁶ have proposed that,

$$f(R) = (1-R)\gamma,$$

where $\gamma \approx 0.7$. In many alloys γ varies from 0.5-0.8 in laboratory air, although cases of $\gamma=1$ are reported for some aluminum alloys. The environment can considerably influence the value of γ and in vacuum, the value of γ can be zero.

A variety of other relationships¹²⁶ have been proposed relating ΔK_{TH} and ΔK_0 and each appears to hold good only for a certain material tested at a certain set of experimental conditions. Several of these relationships involve forms of $f(R)$. According to Dickson et al.¹²⁶, this diversity of experience is due to an inadequate understanding of the influence of the load ratio upon ΔK_{TH} . The most conservative prediction is believed to be given by the simple relationship,

$$\Delta K_{TH} = \Delta K_0 (1-R)$$

Influence of Intrinsic Parameters: An increase in grain size can result in a significant increase in ΔK_{TH} . There is generally

a marked effect in mild and low alloy steels¹³⁷⁻¹³⁹ but little or no effect for aluminum alloys¹⁴⁰. Robinson and Beevers¹⁴¹ report an order of magnitude decrease in near-threshold growth rates in alpha titanium after coarsening the grain size from 20 μm to 200 μm . Similar effects have been observed with Ti-6Al-4V.

The influence of grain size on ΔK_{TH} is contrary to its well known effects on the yield stress (as given by the Hall-Petch relationship) and the endurance limit. Thus, grain size refinement is beneficial for the endurance limit, but detrimental to ΔK_{TH} . This can be better understood by noting that ΔK_{TH} is related to the arrest of a pre-existing macro-crack in the absence of overloading effects and does not concern nucleation, while the endurance limit is related to the probability of crack nucleation-followed by crack growth to complete failure.

Several explanations have been proposed to explain this grain size effect, none of which have been entirely satisfactory. It has been suggested¹⁴² that near threshold growth is confined to specific crystallographic planes such that, in coarser structures, greater deviations of the crack path may be required from the plane of maximum tensile stress. It is also speculated¹⁴³ that, since in high strength steels plastic zones are often confined to a single grain during near-threshold growth, the probability that hydrogen atoms, generated by chemical reactions with the environment at the crack tip can be swept into the grain boundaries by dislocation motion is merely smaller if the grain size is large compared to the size of the plastic zone.

However, more recent experimental results tend to contradict this explanation since not all high strength steels exhibit the grain size effect¹²⁶.

It has been pointed out by Dickson et al.¹²⁶ that in order to be applicable to a number of materials, these models should not only take into account microstructural parameters but environmental factors as well. Based on the previous discussions upon the effect of microstructure and environment on deformation modes (in particular, the slip character) it is felt that more work is required to relate the observed differences in threshold behaviour to the fundamental differences in aspects of deformation in various materials.

The presence of inclusions has not been found to significantly affect near-threshold propagation rates¹⁴⁴. This is likely to be due to the smallness of the plastic zone relative to the average distance between inclusions normally present in commercial alloys.

Impurity induced embrittlement has also been found to increase¹⁴⁵ near-threshold propagation rates. Segregation of impurities such as P, S and Sb have been shown to cause a decrease of ΔK_{TH} in a high strength martensitic steel.

In summary, the initial propagation of macrocracks at near-threshold levels is extremely sensitive to mean-stress (R value) and microstructural factors. Also, the environment appears to play a significant role. Better resistance to such propagation may be achieved by using materials of a coarse grained microstructure, low

impurity levels and good resistance to environmentally induced cracking. Dickson et al.¹¹⁵ point out however, that such microstructures do not always guarantee similar optimum resistance to crack nucleation.

1.10.3 Short Crack Behaviour

One definition of a "short" fatigue crack involves cracks which are comparable in size to the scale of microstructural features. Several recent experimental studies^{125,146-151} on the initiation and growth of cracks in a wide range of materials have revealed that short cracks, initiated near regions of surface roughening or at inclusions and grain boundaries propagate at rates which are different from those of equivalent long cracks, when characterized in terms of the conventional fracture mechanics concepts mentioned above. This is brought out by the work of Lankford¹⁵², for example, on T075-T6 aluminum (Fig. 1.22). It is apparent from this figure that the growth rates associated with the short cracks are up to two orders of magnitude faster than those of the long cracks, and further than such accelerated short crack growth occurs at stress intensities well below the so-called fatigue threshold stress intensity range (ΔK_{TH}), below which long cracks remain dormant or grow at experimentally undetectable rates¹⁵³. The initially higher growth rates of the short cracks can be seen to progressively decelerate (and even arrest in some cases) before merging with the long crack data at stress intensities close to ΔK_{TH} ¹⁵⁴⁻¹⁵⁶. Tanaka et al.^{155,156} have attributed such short crack behaviour to a slowing down of short crack advance through interaction

with grain boundaries. According to the model developed by Morris et al.¹⁵⁴, crack propagation of a short crack into a neighbouring grain ceases when a critical plastic zone size is reached at which crack closure stresses drastically retard crack growth rates. Tanaka et al. opined that the growth of microstructurally short cracks is impeded due to pinning of slip bands emanating from the crack tip, by the grain boundaries. Lankford et al.¹⁵² found that the crack growth rate minimum corresponds to a crack length roughly equal to the smallest grain dimension. Thus, as observed by Ritchie and Suresh¹⁵⁷ a consensus from these studies is that despite their accelerated propagation rates, compared to long crack data, short cracks are apparently impeded by the presence of grain boundaries, which in general, would be unlikely to significantly affect the local propagation of long cracks.

It has also been observed that the fatigue crack propagation threshold stress intensities for long and short cracks tend to be very different. Although conventional fracture mechanics predicts that the threshold stress intensity range for short cracks (ΔK_{TH}^*), for a particular material should be independent of crack length, Kitagawa and Takahashi¹⁵⁸ have shown that the threshold stress intensity factor for short cracks actually decreased with decreasing crack length, whereas the threshold stress $\Delta \sigma_{TH}^*$ approached that of the smooth bar fatigue limit $\Delta \sigma_e$ at very short crack lengths. It has been shown^{137,138} that this critical crack size (below which ΔK_{TH}^* is no longer constant with crack length) is dependent upon microstructural and mechanical

factors, and that it is approximately given by $1/\pi(\Delta K_{TH}/\sigma_e)^2$, where ΔK_{TH} is the long crack threshold. Values of this dimension, which represent the limiting crack size for valid LEFM (Linear Elastic Fracture Mechanics) analysis range from typically 1-10 μm in ultrahigh strength materials (yield strength ~ 2000 MPa) to 0.1-1 mm in low strength materials (yield strength ~ 200 MPa).

According to El-Haddad et al.¹⁵⁹, the stress intensity expression for a short crack of length "a" may be written as,

$$\Delta K = \Delta \sigma \sqrt{\pi(a+a_0)}$$

where a_0 is a material constant and is defined as,

$$a_0 = (\Delta K_{TH}/\Delta \sigma_e)^2 \cdot 1/\pi$$

where ΔK_{TH} is the limiting threshold corresponding to a long crack.

The threshold stress intensity for any crack length is then given by,

$$\Delta K_{TH} = \Delta \sigma_{TH} \sqrt{\pi(a+a_0)}$$

"Short crack behaviour" can also occur when the size of the crack is small compared to the scale of local plasticity, either generated by the crack itself or by the presence of a larger stress concentrator like a notch. This has led to attempts to correlate short crack growth behaviour using elastic-plastic fracture mechanics parameters like the J-integral¹⁶⁰. (The concept of the J-integral and other elastic-plastic fracture mechanics parameters is explained in a later section.) However, even with the more appropriate characterization afforded by elastic-plastic fracture mechanics, significant scatter is still observed.

"Short crack behaviour" may be observed when the crack is long compared to both the scale of microstructure and the scale of local plasticity, yet simply physically small, i.e., typically less than 0.5-1 mm in length. This is attributed to crack closure effects, and as mentioned earlier crack closure can be plasticity induced, oxide induced or roughness induced¹⁵⁷. Suresh and Ritchie¹⁵⁷ have stated that the major reason for the faster growth of short cracks and the fact that they can propagate below the long crack threshold is associated with a decreasing rate of closure at decreasing crack sizes. Some experimental evidence has been presented¹⁶¹ to support this view.

In summary, the lack of similitude between short and long cracks can arise due to a number of reasons, such as,

- differences in the nature of interaction with grain boundaries, hard second phases and inclusions.
- interactions of a short crack with other microcracks.
- differences in the elastic restraint surrounding the crack.
- crack closure effects.
- differences in the nature of interaction with the environment.

1.11 Methods of Life Prediction for Fatigue at Room Temperature

In the case of critical components for aircraft engines the low cycle fatigue "life" of the part is usually determined using crack initiation as the failure criterion. A number of empirical relationships have been developed for the purpose of predicting the life to initiation for cyclic loading conditions and it is the purpose

of this section to briefly review the principal methods available. Time-dependent fatigue is discussed in a later section and hence the following discussion is limited to cycle dependent fatigue.

Conventional stress amplitude-cycles to failure plots (S-N curves) do not produce accurate results in the case of low cycle fatigue which is a plastic strain controlled damage process. Coffin and Manson¹⁶² were among the first to develop a "strain-life relationship" of the form shown below.

$$\frac{\Delta \epsilon}{2} = \underbrace{\frac{\sigma'_f}{E} (2N_f)^b}_{\text{Elastic}} + \underbrace{\epsilon'_f (2N_f)^c}_{\text{Plastic}}$$

where, $\Delta \epsilon$ = total strain amplitude

$2N_f$ = reversals failure

σ'_f = fatigue strength coefficient

ϵ'_f = fatigue ductility coefficient

b = fatigue strength exponent

c = fatigue ductility exponent

This equation may be graphically represented as in Fig. 1.23 and essentially, this illustrates that at short lives, lesser than $2N_t$ (the transition fatigue life when $\Delta \epsilon_p = \Delta \epsilon_e$), plastic strain predominates and ductility will control performance. At longer lives, greater than $2N_t$, the elastic strain is more dominant and strength will control performance. The former is more likely for LCF controlled failure whereas the latter is more relevant in a HCF situation.

In order to apply this relationship to a practical situation, it would be necessary to determine the strain amplitudes occurring at

notches and other strain concentrations in the component under service loading conditions - through a detailed elasto-plastic analysis. The various coefficients and exponents would have to be determined through laboratory testing of smooth LCF specimens.

Experimental data⁴ suggests that the exponent c ranges from about -0.5 to -0.7 with -0.6 as a representative value. The exponent b ranges from about -0.6 to -0.14 with 0.1 as a representative value. The term ϵ'_f may be approximated by the true fracture strain ϵ_f in a monotonic tensile test and the coefficient σ'_f may be taken as a close approximation to the true fracture stress σ_f in a monotonic tensile test. Based on this information Manson simplified the strain-life relationship to the form,

$$\Delta\epsilon = 3.5 \frac{\sigma_u}{E} (N_f)^{-0.12} + \epsilon_f^{0.6} (N_f)^{-0.6}$$

where σ_u (the ultimate tensile strength), E (the Young's Modulus) and ϵ_f can all be obtained from a monotonic tensile test. This relationship, which is an approximation to the earlier one, is commonly referred to as the universal slopes equation.

At temperatures in which time-dependent effects such as creep and environmental damage are minimal these empirical relationships have been found to be quite useful in predicting fatigue life within a factor of 2. However, at elevated temperatures the situation becomes very complex and more sophisticated relationships are required to effectively predict life. This aspect will be dealt with in a later section.

Chapter 1

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1.1 Hysteresis Loop Obtained in LCF Testing².

(A) CONTROL CONDITION

(B) RESPONSE VARIABLE

1.2 Cyclic Hardening During LCF Testing²⁾

(A) CONTROL CONDITION

(B) RESPONSE VARIABLE

1.3 Cyclic Softening During LCF Testing²⁾

1.4 Various Cyclic Responses Possible in a Strain Controlled Test²

1.5 Characteristics of Planar and Wavy Slip

1.6 Influence of Environment on Microhardness of
Cu-7.5% Al Specimens Tested Under Cyclic
Loading¹⁶

1.7

Microhardness vs the Depth of the Removed
Layer for Specimens of Cu-7.5% Al Strained in
Air and Vacuum for Various Periods¹⁶

1.8 Stages in Fatigue Crack Growth³

1.9 Persistent Slip Bands at Surface³

1.10 Influence of Grain Size on LCF Life³

1.11 Effect of Grain Size on LCF Life in a High Purity 7075 Alloy²

STRESS CYCLE

VACUUM

AIR

DISSOLVED O_2

INCIPIENT CRACK

1.12

Environment Assisted Crack Nucleation Through
the Prevention of Crack Re-Welding²

STRESS CYCLE

VACUUM

AIR

OXIDE FILM

HIGHLY STRESSED
REGION

VOIDS

1.13

Environment Assisted Crack Nucleation Through
Interaction with Surface Oxide Scale²

1.14

Plastic Blunting Mechanism for Stage II Fatigue Crack Growth²¹

1.15 Shear Decohesion Mechanism for Stage II Fatigue Crack Growth³

1.16 Comparison of Growth Mechanisms in Low Stress and High Strain Fatigue³

1.17 Mechanisms Operative in the Three Stages of Fatigue Crack Growth³

1.18 Plastic Zone Depth vs. Fatigue Crack Length for Armco Iron Cycled in Air (full symbols) and in Vacuum (open symbols)¹⁶

1.19 Influence of Air and Vacuum Environments Upon
the Depth of the Plastic Zone in Magnesium Alloy
MA12¹⁶

1.20 Influence of Stress Ratio on Fatigue Crack Growth Behaviour³

1.21 Mechanisms of Crack Closure¹¹⁹

1.22 Short Crack Behaviour in 7075-T6 Aluminum¹⁵²

1.23 Graphical Representation of Manson-Coffin Relationship and "Universal Slopes" Equation ¹⁶²

Chapter 2

ELEVATED TEMPERATURE CREEP DAMAGE AND FAILURE

UNDER UNIAXIAL LOADING

2.1 The Phenomenon of Creep

A schematic representation of the typical strain vs. time relationship which develops during elevated temperature creep deformation is shown in Fig. 2.1. After the instantaneous strain ϵ_0 , a decelerating strain rate stage (primary creep) leads to a steady creep rate (secondary creep) which is finally followed by an accelerating stage of deformation (tertiary creep) which ends in fracture. The transition from the primary creep to the secondary creep stage is thought to be due to the gradual development of a balance between the strain hardening of the metal caused by deformation on the one hand and thermal softening or recovery on the other. The transition to the tertiary stage takes place when sufficient creep damage has occurred within the material in the form of cavities or cracks. The minimum creep rate $\dot{\epsilon}_{min}$ which occurs during secondary creep is one of the most important parameters for fundamental and engineering studies. It is usually found that $\dot{\epsilon}_{min}$ is inversely proportional to the rupture life, t_r :

$$\dot{\epsilon}_{min} \cdot t_r = \text{constant}$$

2.2 Deformation Regimes

Several mechanisms can influence slow, creep deformation of materials at high temperature and the type of mechanism which dominates is primarily a function of stress and temperature. It is possible to construct deformation mechanism maps showing the regimes for each

mechanism in a stress-temperature space, as shown in Figs. 2.2 and 2.3. It is evident that two basic mechanisms, exist, namely, dislocation creep and diffusion creep. Creep flow will, at a given strain rate, occur by diffusional flow at low stresses and sufficiently high temperatures, while dislocation creep will predominate at relatively lower temperatures and higher stresses². These two mechanisms are discussed further in the following sections.

2.2.1 Diffusional Creep

In the presence of stress, a diffusional flux of matter could be created, through or around the surfaces of grains in a polycrystal². This trans- or circumgranular flux leads to strain, provided it is coupled with sliding displacements in the plane of the boundaries themselves. If both lattice and grain boundary diffusion are permitted, a simplified rate equation of the form given below can be derived²

$$\dot{\epsilon} = \frac{14\sigma\Omega}{kT} \cdot \frac{1}{d^2} D_V \left(1 + \frac{\pi A_B D_B}{d D_V} \right) \quad (2.1)$$

where, $\dot{\epsilon}$ is the secondary creep strain rate, σ is the stress, Ω the atomic volume, d the grain size, D_V the bulk diffusion coefficient, D_B the grain boundary diffusion coefficient and A_B the effective cross-section of a grain boundary for diffusional transport.

For a very coarse grained material, the term $\pi A_B D_B / (d \cdot D_V)$ in eqn. 2.1 can be much smaller than unity, and under these conditions the influence of grain boundary diffusion becomes unimportant. The transgranular process of diffusion creep, also called "Nabarro Creep"

is the dominant mechanism. At high temperatures, the ratio of D_B/D_V decreases (because the activation energy for grain boundary diffusion is only about half that of volume diffusion) and this also promotes the Nabarro creep mechanism. On the other hand, when creep occurs primarily as a result of grain boundary diffusion, deformation is said to be controlled by "coble creep".

2.2.2 Dislocation Creep

At temperatures above $0.3 T_M$, dislocations can climb as well as glide and hence are no longer constrained to move only in their slip planes. As a result of this, dislocation tangles and networks created by plastic deformation will gradually be released by a recovery process. The released dislocation will glide until it reaches the next obstacle and this glide step is responsible for the observed strain.

Thus, dislocation creep is controlled by a climb-plus-glide sequence. The vacancies enabling climb can diffuse either by pipe diffusion (along dislocations) at intermediate temperatures or by bulk diffusion at elevated temperature. In contrast with diffusional creep discussed earlier, in which creep strain is directly proportional to stress, dislocation creep exhibits a power law stress dependence, i.e. $\dot{\epsilon}_{min} \propto \sigma^n$. Precipitation hardened alloys and Fe-Ni-Ti-Al alloys have been shown to exhibit stress exponents in the range $5-15^3$.

Dispersion strengthened materials such as TD-NiCr may exhibit even higher exponents. An Arrhenius-type equation can be satisfactorily used to describe the thermal dependence of $\dot{\epsilon}_{min}$. At high temperatures, especially for materials of high stress sensitivity, the stress

normalised by the elastic modulus yields a better description of experimental data:

$$\dot{\epsilon}_{\min} = \exp\left(\frac{-Q}{kT}\right) \left(\sinh \frac{c\sigma}{TE}\right)^n \quad (2.2)$$

where Q is the apparent activation energy, k is Boltzmann's constant, c is a structure dependent constant and E is Young's Modulus.

2.3 Stages in Creep Fracture

It is generally accepted that creep fracture occurs in three basic steps:

- (i) nucleation of cracks
- (ii) stable growth of cracks
- (iii) unstable crack growth leading to failure

It is important to note that the nucleation and the stable growth stages are time-dependent and can occur at a constant stress. Creep fracture occurs typically by intergranular cracking, which can usually be classified into two categories:

- (i) intergranular cavitation (r-type voids)
- (ii) intergranular wedge cracks (w-type cracks)

r-type voids appear as discrete microvoids strung out along the grain boundaries while w-type cracks concentrate at grain boundary triple points. Cavitation is generally favoured at low stresses, whereas w-type cracking is preferred at relatively higher stresses⁴. McLean⁵ observed that the transition from r-type to w-type cracks occurs quite sharply in several superalloys, occurring at a different stress for each alloy. However, this finding has met with criticism from

Weaver⁶, who in a study of Nimonic 80A determined that this transition was significantly influenced by microstructural variables and hence not distinct for a particular alloy. The mechanisms responsible for the nucleation and growth of r-type and w-type voids are discussed separately as follows.

2.3.1 r-type Voids (Cavitation)

A number of investigations have clearly demonstrated that grain boundary sliding is a prerequisite for the nucleation of r-type cavities during creep deformation². The sliding of grains sets up stress concentrations at various irregularities along the boundaries. These stresses promote local decohesion and cavities are initiated. Irregularities that are considered to be essential for cavity nucleation are ledges and steps in the grain boundaries^{2,4} (Fig. 2.4). As will be discussed later, particles located at the grain boundaries can also play a very significant role in promoting cavity nucleation⁴.

The transition to the growth stage is generally felt to be dependent on the stability of the growing nucleus. Below a certain critical size which is a function of the surface energy of the material in question, the nucleated void may collapse, while above that size, stable growth may occur⁴. As cavity growth progresses, the cavities on a single grain boundary facet link up together. Many grain facets may become completely cavitated this way, further growth being limited by the change of direction required at grain boundary triple points.

As noted by Perry⁴ in a comprehensive review on creep cavitation, there is considerable debate over which process controls

the stable growth of nucleated cavities. One school favours the idea that it occurs by vacancy condensation on cavities and that the diffusion of vacancies along the grain boundaries is the controlling step of the growth. In contrast, another school advocates that not only the initiation but also the growth is governed by grain boundary sliding. The growth of cavities under the combined action of grain boundary sliding and intragranular slip³ is shown schematically in Fig. 2.5. Since grain boundary sliding is intimately connected with intragranular creep deformation, it follows that this type of growth will be strain controlled².

While there is no clear-cut answer as to which mechanism dominates, the weight of evidence available so far indicates that a strain controlled cavity growth mechanism is more likely. For instance, detailed microscopy of cavities has shown that they frequently have non-spherical shapes^{7,8}; alloying does not change the diffusion coefficient but does change the creep rupture life considerably^{9,10}; void growth has also been observed to be rather temperature independent¹¹; creep strain and void area have been found to be linearly related⁴. Thus, while a grain boundary sliding mechanism may play the dominant role in most cases, it is also likely that no one mechanism operates individually. The situation is more likely to be one in which several mechanisms such as tearing between voids, crack propagation from void apices, grain boundary sliding and vacancy condensation are all possible, of which one or more may dominate depending upon microstructural variables and testing conditions.

In the final, unstable stage of fracture, the cavitation damage in the material proceeds to an extent sufficient to cause fracture. Attempts have been made to derive a fracture criterion using parameters such as critical crack length¹², critical void spacing¹³ and critical void volume fraction¹⁴. It is unlikely that an universally acceptable fracture criterion will be developed since the fracture criteria and the final fracture mode can be different for different microstructural conditions even in the same alloy⁴.

2.3.2 w-Type Cracks

Both the initiation and growth of wedge cracks are brought about by grain boundary sliding. Typically, the cracks form in grain boundary triple points as shown in Fig. 2.6, at an approximate 45° angle to the tension direction. Then they grow as wedges in the grain boundary forming approximately a 90° angle to the tension direction.

It is often assumed that the nucleation of w-type cracks can be modelled by brittle crack models as the one due to Stroh¹⁵:

$$\sigma_s^2 = \frac{3\pi\gamma G}{8(1-\nu)d} \quad (2.3)$$

where σ_s is the shear stress on the sliding boundary, γ the effective surface energy per unit of fracture surface, G the shear modulus, d the grain size and ν the Poisson's ratio. However, this simple model ignores the stress relaxation of stress concentrations caused by creep deformation.

Regarding final fracture due to w-type cracks, support has been found for a fracture criterion based on a critical stress intensity factor $\sigma\sqrt{a}$ (σ = normal stress and a = length of crack)^{16,17}.

The various stages described so far are normal processes which lead to creep fracture. The rate at which creep damage is accumulated in the material depends on how easy or difficult it is for cavities to nucleate, grow and interlink. In this connection, the following factors are of importance.

2.3.3 Intrinsic Factors

Grain Size. Generally, an increase of grain size causes creep rate and ductility to decrease (Fig. 2.7), particularly when the grain size is small. This grain size effect on creep rate is likely to be due to larger grains leading to longer diffusion paths and smaller grain boundary area per unit volume¹⁸. Also, it is well known that the extent of grain boundary sliding will be minimized at coarse grain sizes.

The observed decrease of creep ductility with decreased grain size is likely to be due to a decrease in the number of grain boundary triple points which can effectively inhibit cavity linkage. The critical crack length required for fracture may therefore be reached only after a greater elongation⁴.

Recent studies indicate that the observed grain size effect on creep rate is, in part, an environmental effect. This will be discussed in detail, in a later section.

Grain Shape and Boundary Morphology. Grain aspect, orientation of elongated grains and texture can all have significant influences upon high temperature creep resistance¹⁹. Irregular, "wavy" or "serrated" grain boundaries are particularly useful in improving creep rupture

properties²⁰⁻²³. The rupture life and ductility (particularly the latter) have been found to be enhanced considerably. The increase in life is likely to be due to a decrease in the creep rate resulting from marked inhibition of grain boundary sliding by grain boundary irregularities or serrations. The increase in rupture ductility may be a direct consequence of the difficulty of cavity linkage along a serrated grain boundary.

In the case of Ni-based superalloys strengthened by the coherent gamma prime ($\text{Ni}_3(\text{Al}, \text{Ti})$) precipitate, grain boundary serrations are created by a solution anneal above the γ' solvus temperature followed by a slow cool through the γ' solvus temperature range²⁰⁻²³. The serrations are well rounded (Fig. 2.8) and their spacing and amplitude are important parameters which determine the extent of their influence on stress rupture behaviour. In Ni-based alloys such as Inconel 718, which are strengthened by γ'' precipitation the mechanism of formation as well as the appearance of serrations can be quite different (Fig. 2.9). Nevertheless, a beneficial effect on creep properties can be expected as a result of such serrations as well²⁴.

Inclusions and Second Phase Particles at the Grain Boundary. The role of grain boundary particles in nucleating voids was stated earlier. According to a theoretical model presented by Smith and Barnby²⁵, void nucleation, followed by growth to a stable size will be unlikely at particles which are either too small or too large. There is believed to be an intermediate range of precipitate sizes for a given alloy such that both nucleation and growth can readily occur. According to

another theoretical study by Raj²⁶, small grain boundary particles that are closely spaced would offer the best resistance to cavity nucleation.

In Ni-based superalloys, the presence of chromium carbides as discrete particles at the grain boundaries is generally considered beneficial to creep properties. However, the formation of chromium carbide at the grain boundaries can, in some instances be accompanied by the development of a precipitate free zone (PFZ) along the grain boundaries²⁷. Heslop²⁸ has observed that if such denuded zones are too wide or too weak, excessive local deformation can occur in these regions and result in premature failure.

Matrix Precipitate. In precipitation hardened high temperature alloys, the volume fraction of strengthening precipitate is obviously an important parameter. Figure 2.7 also shows an example of creep resistance variations as a result of differences in the volume fraction f of the hardening phase γ' , in four Ni-Cr-based alloys differing mainly in the precipitation fraction content²⁹. At constant grain size and particle size the stress plots are as follows:

$$\sigma_s = \sigma_{ss} + \frac{C}{\lambda}$$

where σ_{ss} is the solid solution strength, C is a constant and

$\lambda = \rho(f^{-1} - 1)$ is the interparticle distance; ρ is the average particle size. An example of the influence of grain size and precipitate volume fraction on creep rate is shown in Fig. 2.10³⁰.

For a given volume fraction of the strengthening precipitate, the creep behaviour can also vary as the particle size varies.

In-service exposure can lead to a gradual coarsening of the strengthening precipitate and to account for this effect, relationships as the one shown below for Inconel 700³¹ may be developed.

$$\log \dot{\epsilon} = -8.88 + 17.81 d - 40.55d^2$$

where $\dot{\epsilon}$ is in s^{-1} , and d is in μm .

2.3.4 Extrinsic Factors

Temperature. An increase in temperature increases the contribution of grain boundary sliding to elongation. This, in turn will tend to increase the probability of cavity nucleation. But, at very high temperatures diffusional accommodation around the obstacles in the boundary may cause the stress concentrations induced by sliding to drop, hence lowering the probability of nucleation²⁶.

Strain Rate. The calculations of Raj predict that there will be a minimum in the case of nucleation with respect to strain rate. This is because, at high strain rates the contribution of grain boundary sliding to deformation decreases, leading to a low probability of nucleation. On the other hand, at very low strain rates, diffusional creep becomes rate controlling and the magnitude of stress concentrations set up at irregularities in the grain boundaries decays to a low value, causing a low probability of cavity nucleation.

Service Environment. During high temperature creep of Ni-base superalloys, interaction with an air environment leads to several competitive reactions. "Air strengthening" and "Air weakening" can occur simultaneously and depending upon which effect dominates, an alloy may show an increase or decrease in creep strength in air, as compared to that in vacuum³².

Thus, Shahinian and Achter³³ in a study of the creep rupture properties of nickel in air and vacuum found a reversal in relative strengths with variations in stress. At low stresses, the properties were better in air while at high stresses they were better in vacuum.

In Ni-base alloys, at low stresses, oxidation is believed to have increased the rupture life of Inconel³⁴, Hastelloy³⁵, Inconel X³⁶, Ni/20Cr alloys³⁷, and single crystals of Mar-M200³⁸. Oxidation was also found to reduce the secondary creep rate and enhance the ductility of Inconel^{39,40}. When oxides formed in surface cracks nucleated during secondary creep, a stage of low creep rate after tertiary creep has been noted in a Ni-19Cr alloy⁴¹, particularly at low stresses, giving rise to a strengthening effect. As in the case of nickel, mentioned earlier, this effect is reversed at high stresses.

The various mechanisms by which a surface oxide scale can contribute to a strengthening effect have been reviewed by Davidson et al.⁴² as follows:

Production of dislocations: It has been shown that single ended dislocations are capable of generating new dislocations by a Frank-Read mechanism at half the stress necessary to activate internal double ended dislocation sources. It has been further hypothesized⁴³ that single ended sources should predominate at free and clean surfaces, but that these sources are rendered ineffective when an oxide scale covers the free surface⁴⁴. Hence when an alloy surface is heavily oxidized, dislocation generation must take place at the internal double ended sources at twice the stress necessary to activate the surface sources

prevalent when the surface is clean and free. That free surfaces are potent sources for dislocations and coated surfaces are not has been confirmed by studies on the plastic flow behaviour of silver coated versus non-coated copper⁴⁵ and oxidized versus unoxidized or polished aluminum single crystals.

Mobility of dislocations: It has also been suggested that adherent, integral scales can provide strength by inhibiting the passage, or egress, of dislocations to the alloy surface^{46,47}. Experimental support for this concept has been found in ambient temperature studies on copper plated zinc crystals⁴⁸, oxidized aluminum crystals⁴⁹, oxidized cadmium crystals⁵⁰ and polycrystals⁵¹. This effect may also exist with respect to grain boundary sliding in that such sliding or the concomitant grain rotation, might also result in surface steps and thus be sensitive to the presence of oxide scales⁴².

It has been reported⁴² that in Ni-base alloys, air strengthening does not involve a drastic change in the thermally activated process of dislocation movement, for the activation energies for creep deformation in air and vacuum are about the same and of the magnitude for vacancy self diffusion. Hence, regardless of the surface effects, vacancy self diffusion apparently controls dislocation movement in both and vacuum crept samples. The strengthening is manifested through significant variations in the stress sensitivity exponent n . For example, in Udimet 700 creep tested at 760°C, the value of n in air is 7.07 while in vacuum, n is 4.03⁴².

It is generally accepted that n is related to ' m ', the stress sensitivity of the mobile dislocation density and to ' l ', the stress sensitivity of the mobile dislocation velocity such that $n = m + l$. Although it is not clear whether the differences in n (in air and vacuum) are due to changes in m and/or l it seems reasonable that air strengthening behaviour at elevated temperatures is a reflection of the inhibition of dislocation generation and/or movement due to the presence of a surface oxide scale.

Internal oxidation strengthening: Oxidation strengthening can involve external effects, such as solid solution or precipitation strengthening from the diffusive penetration of environmental elements into the alloy substrate. Interstitial oxygen has also been noted to improve the creep resistance of various metals^{52,53} and alloys⁵⁴ including stainless steels⁵⁵.

A reaction which is highly likely in the sub-scale region is the combination of diffusive gaseous elements such as oxygen and nitrogen with various highly oxidizable alloying elements to form internal oxides or other corrosion products. Such internally formed particles, can increase alloy hardness⁵⁵, as well as tensile properties⁵⁶ and have been generally known to improve the creep resistance of metals and alloys also^{57,58}. These internal strengthening effects are particularly important at the grain boundaries where creep deformation due to grain boundary sliding can be inhibited via solid-solution strengthening of the grain boundaries or through internal oxides precipitating at the grain boundaries^{59,60}.

Mechanical crack bridging effects: Creep fracture can be delayed by blunting or bridging of surface crack tips by the formation of oxides or nitrides⁶¹. The effect of these oxides is basically to reduce the stress concentration at the crack tip. The extent to which such strengthening occurs will depend upon the oxide scale being sufficiently adherent and strong, but as noted earlier, is likely to be of influence only at low stresses.

While air strengthening is possible under certain conditions, there have been numerous reports in the literature about air weakening as well. Just as intruding environmental species might solid solution strengthen or lead to precipitation strengthening, the loss of alloying elements, e.g. due to selective oxidation or volatilization, might lead to a loss of solid solution strengthening or, ultimately, a dissolution of strengthening precipitates⁴². It has also been hypothesized that the oxidation process itself might induce solid solution weakening by the injection of vacancies into the alloy substrate⁶² which, in turn, could enhance both the diffusional and dislocation climb mechanisms, or lead to the formation of deleterious voids⁶³. Early crack initiation due to stress concentrations caused by surface oxide wedges has been suggested as a reason for premature failure of superalloy turbine blades⁶⁴.

Thus, oxidation plays a dualistic role in creep crack initiation and growth which is important, if not totally life determining, in elevated temperature creep deformation. The overall effect (i.e. strengthening or weakening) that is observed

experimentally is the result of a complex variety of competing reactions.

While considerable work remains to be carried out to identify under what conditions air strengthening or air weakening will occur, a few basic parameters are clearly of crucial importance. The transition related to stress levels was mentioned already. It is possible that at low stresses creep lives are longer, allowing for thicker oxide formation and possibly a greater level of strengthening⁴². It is also possible that at low stresses, the oxide scales have some load bearing capacity while at high stresses they may readily crack and fail to contribute to any load bearing or crack tip blunting effects. Certainly, at very high stresses the air strengthening effect may become unimportant because of the predominance of tertiary creep and the accompanied shortening of the duration of secondary creep⁴².

The effect of temperature upon the transition from air strengthening to weakening is not clear. According to Davidson et al.⁴², at higher temperatures, one would expect thicker oxide scales and hence greater air strengthening of the specimen surface. At lower temperatures, and especially in the high aluminum or chromium oxidation resistant alloys less adherent oxide scale growth will occur and accordingly, less air strengthening. However, in spite of these differences in the oxidation kinetics the transition from strengthening to weakening seems to occur at about the same stress level for various temperatures in fine grained nickel (Fig. 2.11). It may be argued that the differences in properties between air and vacuum portrayed in

Fig. 2.11 are not sufficiently large to draw concrete conclusions from. But, the recent findings of Woodford⁶⁵⁻⁶⁸ throw some light on this question.

Woodford⁶⁵⁻⁶⁸ determined that at a high temperature, although damage penetration might be significant, the sensitivity to environmental embrittlement is low (Figs 2.12 and 2.13). On the other hand, at a low temperature, although the sensitivity to embrittlement is high, the oxidation kinetics would be such that, in the period of a laboratory creep test, damage penetration would be small. Thus, it may be quite difficult to ascertain the effect of temperature on environmental damage through the parameters measured in a laboratory creep test.

The type of gaseous environment plays a crucial role and even materials which exhibit an air strengthening effect at low stresses may be weakened in an atmosphere of high oxidizing potential, such as pure oxygen. Stegman et al.⁶⁹ creep tested nickel in the air-weakening regime at various oxygen partial pressures and found that rupture life decreased with increasing oxygen partial pressures as shown in Fig. 2.14, until a pressure of about 13 mN/m^2 when it became insensitive to further increase in oxygen partial pressure. The observed air weakening was attributed to a lowering of grain boundary surface energy due to gaseous adsorption. However, this explanation has met with criticism due to the fact that when oxygen is admitted to a specimen in vacuum during steady state creep, the new creep rate stabilises only after several hours. If gaseous adsorption were the rate controlling factor, this stabilisation would be expected to occur

much more rapidly⁷⁰. Thus, it appears that other factors such as sub-surface oxygen content and mechanisms such as the dislocation "sweep-in" of oxygen atoms⁷¹ would have to be considered to explain this observed effect.

Perhaps the most important microstructural feature controlling the extent of environmental embrittlement in a given material is the grain size. Figure 2.15 shows that effect of grain size upon creep rates of Udimet 700 in air and vacuum. The grain size transition from air strengthening to air weakening as the grain size gets finer is not surprising considering the fact that at these temperatures and stresses, grain boundary sliding becomes significantly enhanced with finer grained materials⁷². Further, oxidation or gaseous oxygen could enhance grain boundary sliding through, for instance, a reduction in the cohesive boundary forces^{73,74}. Similar results have been reported by Davidson and Tien⁷⁵ for Nimonic 115.

There is a limited amount of evidence⁷³ that an "irregular" grain boundary morphology may give rise to a lower level of environmental embrittlement during creep deformation and further work is required to explore this important aspect further. Virtually no information is available on the effect of grain boundary particles except that in Nimonic 115, it has been observed⁷⁵ that for a given grain size and specimen size, air strengthening occurs only if the grain boundaries are strengthened by carbides. The role of matrix precipitates is even less well understood, but it is likely that when the matrix precipitate is very fine, causing planar deformation, the

stress concentrations caused by dislocation pile-ups at the grain boundaries can intensify environmental embrittlement⁷⁶.

In air and vacuum, thick specimens are generally observed to be more creep resistant than thin specimens⁷⁵. In the air tests, this phenomenon might be attributed to the air weakening of grain boundaries, since a greater fraction of grains intersect the free surface of thin specimens and are therefore open to environmental penetration. The fact that a similar effect is observed in vacuum, however, suggests that even in the absence of significant environmental reaction, near surface layers might be intrinsically weaker than the internal bulk material due to the lack of geometrical constraints at free surfaces⁷⁵.

Chapter 2

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2.1 Typical Strain vs Time Relationship in Uniaxial Creep

2.2 Deformation Mechanism Map for Pure Nickel³

2.3 Deformation Mechanism Maps for Mar-M200³: Influence of Grain Size on Deformation Mechanisms

2.4 Crack Nucleation at Grain Boundary Ledges⁴

2.5 Cavity Growth Under the Combined Action of Grain Boundary Sliding and Intragranular Slip³

2.6 Mechanism of Wedge-Crack Nucleation³

2.7 Influence of Grain Size on Creep Rate and Ductility of Four Ni-Cr Based Alloys Differing in the Volume Fraction (f) of the Hardening Phase³

2.8 Serrated Grain Boundary in IN-738

2.9 Grain Boundary Serration in Inconel 718

2.10 Influence of Grain Size and Precipitate Volume Fraction on Creep Rate⁹

2.11 Transitions from Air Strengthening at Low Stresses
to Air Weakening at High Stresses in Fine Grained
Nickel⁴²

2.12 Effect of Air Exposure on Rupture Life of IN-738. Individual Lives are Indicated in Hours at 700°C, 800°C, 900°C and 1000°C⁶⁵⁻⁶⁸

2.13 Influence of Time of Exposure to an Embrittling Atmosphere Upon the Rupture Life of IN-738⁶⁵⁻⁶⁸

2.14 Influence of Oxygen Partial Pressure Upon Rupture Life of Nickel⁶⁹

2.15 Effect of Grain Size Upon Creep Rates of Udimet 700 in Air and Vacuum⁷²

CHAPTER 3

CREEP CRACK GROWTH FROM PRE-EXISTING FLAWS

Uniaxial creep test data are of value to the engine designer to determine data such as the maximum creep extension that can occur in a turbine blade over a certain number of hours of service exposure. However, this type of data is not useful for making life predictions from a damage tolerance point of view. In order to know the probability that an undetected, pre-existing flaw will propagate to a size sufficient to cause catastrophic failure during a certain service life interval, it is necessary to generate crack growth rate data for conditions representative of the service conditions. In this section, the application of fracture mechanisms to generate creep crack growth (CCG) rate data is discussed in detail. The various parameters that have been developed to uniquely characterize creep crack growth behaviour have been described and the limits of their applicability have been pointed out.

It should be emphasized here that while the damage processes occurring in uniaxial creep tests and creep crack growth tests using precracked fracture mechanics specimens, are similar materials which have creep strength do not necessarily exhibit good resistance to creep crack propagation,¹ perhaps because the scale over which plastic deformation occurs in these two types of tests is entirely different. For this reason, the microstructural and environmental factors influencing creep crack growth resistance are discussed separately here, so that any confusion with uniaxial creep behaviour is avoided.

3.1. Slow Crack Growth Behaviour

In experiments performed using fatigue pre-cracked fracture mechanics specimens (tested under constant load conditions), crack growth does not start immediately on loading. There is usually an incubation period before the material ahead of the crack tip accumulates sufficient creep damage for intergranular propagation to occur². The failure made in the room temperature fatigue pre-crack is essentially transgranular. The incubation time was found to be of the order of 5 to 10 pct. of the total failure time. It is to be noted that such behaviour is quite different from that in a uniaxial creep test, where initially, the rate of straining is quite rapid, and no "incubation" time is involved.

In compact tension specimens, crack propagation occurs at a continuously increasing rate until fast, overload fracture takes place. When the crack length vs. time data is converted to crack growth rate vs stress intensity, the general type of curve shown in Fig. 3.1 is obtained. Three stages of cracking, labelled I, II and III respectively, emerge from this representation of which, Stage I occurs only during the initial onset of crack growth from the fatigue crack and the Stage III region develops only during the final few minutes of the test.

The time to failure in these tests is controlled primarily by the Stage II crack growth rate. It has been found that the following relationship generally holds good, i.e.:

$$a_{II} t_f = \text{constant}$$

where a_{II} is the second stage crack growth rate and t_f is the time to failure.

Metallographic examination of specimens tested in this manner has revealed that creep crack propagation does not necessarily take place by the extension of a single crack. Several microcracks have been found to initiate ahead of the main crack and interlink to it. Thus, creep crack growth may take place in relatively slow steps involving microcrack formation on a grain boundary normal to the tensile axis slightly ahead of the main crack, followed by a rapid shear fracture of the inclined boundary linking the microcrack and the main crack. Thus, at a microscopic level, the material ahead of the crack tip goes through what would be called the primary, secondary and tertiary stages of creep in a uniaxial test. The length of the main crack vs time would be a staircase type of plot showing zero crack growth while the microcrack formed and then a rapid jump when the main crack links up with the microcrack. However, macroscopically, the averaging together of all these local events gives a smooth curve of crack length vs time.

3.2. Fracture Mechanics Characterization of Creep Crack Growth

Several parameters have been proposed for characterizing creep crack growth behaviour of metals. These include: stress intensity factor,³ K , the J-contour integral,⁴ crack tip opening

displacement,^{5,6} net section⁷ or reference stress⁸ and energy rate line integral.^{9,10} The applicability of these parameters is discussed below.

3.2.1 Stress Intensity Factor, K and the J-Contour Integral

The relationship between creep crack growth rate, da/dt and K was investigated by Siverns and Price³ and also by Harrison and Sandor.⁷ In these studies, even though only one specimen geometry was used, the success of K in describing da/dt was at best moderate. For more than one specimen geometry, the correlation is expected to be worse.¹¹ However, for Udimet 700 at 843°C, Sadananda and Shahinian¹² have reported good correlation between da/dt and K . Similarly, for Inconel 718 at 650°C, the stress intensity factor has been found to be an effective correlating parameter.^{13,14} The reasons for these apparent contradictions are related to the extent of creep relaxation occurring ahead of the crack tip and to the extent of sensitivity of the alloy to environmental embrittlement, as explained below.

~~When a crack is subjected to load in the creep regime, a~~
plastic zone is formed instantaneously upon loading, at the crack tip. If this plastic zone is limited to the crack tip and is well embedded over the zone in which elastic stress and deformation fields apply, the stress intensity factor will generally be a good correlating parameter. This is shown schematically in Fig. 3.2, where the "K-zone" is the region over which elastic stress and deformation fields apply.

For high temperature materials such as Udimet 700, even at temperatures as high as 843°C, the extent of crack tip plasticity may be limited to a small region and hence a good correlation of da/dt with K may be observed. Also, materials such as Inconel 718 are highly sensitive to environmental embrittlement by oxygen at temperatures in the vicinity of 650°C.¹⁵ Thus, crack propagation may occur in a nominally brittle manner by an oxygen assisted grain boundary embrittlement mechanism. Here again, the stress intensity factor would be a useful parameter to correlate crack growth rate behaviour.

Under conditions where the plastic zone size becomes large, comparable to the size of the K -zone, the crack tip stresses and strains can be characterized by the J -contour integral. Defined by Rice,¹⁶ this parameter is given by:

$$J = \int_T \left\{ W dy - T_J \frac{\partial u_j}{\partial x} ds \right\}$$

W is the strain energy density given by,

$$W = \int_0^{\epsilon_{jk}} \sigma_{jk} d\epsilon_{jk}$$

T_J is $\sigma_{jk} n_k$, the traction normal to the contour T and u_j is the displacement vector. J can be determined for the elastic/plastic case through a numerical finite element calculation. It has been shown that this contour integral is path independent as long as the contour is sufficiently away from the crack tip.

It has also been shown¹⁶ that J is also a measure of the energy release rate per unit thickness:

$$J = - \frac{du}{da}$$

J can be determined empirically from a non-linear diagram of load P versus load-line displacement V (Fig. 3.3).¹⁶

While the J-integral may be of some use in correlating creep crack growth, this parameter also loses value when significant stress relaxation due to creep occurs.

3.2.2 Net Section Stress and Reference Stress

When yielding occurs over the larger part of the reduced section (of a fracture mechanics specimen) there is some possibility that crack instability will occur for a critical value of the net section stress σ_{net} . However, experimental data concerning this parameter have not conclusively demonstrated the usefulness of this parameter. Harrison and Sandor⁷ reported modest correlation between da/dt and net section stress in Cr-Mo-V steel using center cracked specimens only. Other studies¹⁷ involving various specimen geometries, including some with non-uniform stress fields, reported lack of correlation of da/dt with net section stress. Saxena¹¹ has observed that since this parameter does not describe the magnitude of stress, strain or strain rate in the crack tip region, it is not expected to correlate very well with crack growth rate.

For a crack specimen loaded partly in bending, the net section stress is not very meaningful and hence it has been attempted to use the reference stress σ_{ref} defined as,

$$\sigma_{ref} = \frac{\sigma_g}{m} = \frac{P}{mBW}$$

where σ_g is the gross section stress, P is the applied load, B and W are the specimen dimensions and m is the ratio of the load to produce general yield in a cracked body to the load to yield an uncracked body of the same overall dimensions. Thus, the reference stress is the stress that when applied to a uniaxial creep specimen will produce the same displacement rate. In general, σ_{ref} has been found to give better correlations than σ_{net} .¹⁸

3.2.3 Crack Opening Displacement Rate

A different approach is one based on crack opening displacement rate or crack tip opening displacement. The COD (v) can be measured in the load line, or at other distances from the crack tip, using, e.g. clip gages or extensometers from which the CTOD, δ can be derived.

For a cracked fracture mechanics specimen, the displacement in the load line, v is related to load P and compliance C as follows,

$$v = C \cdot P$$

Differentiation with respect to time gives

$$\frac{dv}{dt} = C \cdot \frac{dP}{dt} + P \cdot \frac{dC}{da} \cdot \frac{da}{dt}$$

For constant load,

$$\frac{dv}{dt} = P \cdot \frac{dC}{da} \cdot \frac{da}{dt}$$

For a tapered double cantilever beam (constant- K) specimen, $\frac{dC}{da}$ is constant. It follows then that only for a constant- K specimen in

the elastic case at constant load will there be a straight proportionality between $\frac{da}{dt}$ and $\frac{dv}{dt}$. Under other conditions, a more general relationship of the form

$$\frac{da}{dt} = H \cdot \left(\frac{dv}{dt}\right)^m$$

may still hold. However, Van Leeuwen has observed that, since the time dependence of crack length and COD can be quite different, (Fig. 3.4) such relationships may have only limited validity.

In a mechanistic sense, the crack tip opening displacement is a more appropriate parameter. However δ is much more difficult to determine than v or $\frac{dv}{dt}$. The usual method of determining δ is to assume that there is a plastic hinge at a certain spot in the unbroken ligament and that δ will be a certain fraction of v . A typical relationship developed by Haigh¹⁹ is given below.

$$\frac{da}{dt} = 2 \cdot \left(\frac{d\delta}{dt}\right)^{0.8}$$

3.2.4 C* (or J*) Line Integral

The C* parameter is a path independent energy rate line integral defined by analogy to J.¹⁹ It is given by

$$C^* = \int_{\Gamma} W^* dy - T_i \left(\frac{\partial u_i}{\partial x} \right) ds$$

where,

$$W^* = \int_0^{\dot{\epsilon}_{mn}} \sigma_{ij} d\dot{\epsilon}_{ij}$$

As illustrated in Fig. 3.5, Γ is the line contour taken from the lower crack surface to the upper crack surface. W^* is the strain

energy rate density associated with a point stress, σ_{ij} and strain rate $\dot{\epsilon}_{ij}$. T_i is the traction vector, defined by the outward normal, n_j , along Γ . $T_i = \sigma_{ij} \cdot n_j$. The displacement vector is denoted by u_i and s is the arc length along the contour.

The energy rate interpretation of C^* was given by Landes and Begley⁹ also, by analogy to the J-integral. They defined C^* as the power difference between two identically loaded bodies having incrementally differing crack lengths, Fig. 3.6.

$$C^* = - \frac{du^*}{da}$$

u^* is the power or energy rate per unit thickness defined for a load P and displacement rate $\dot{u}$. Based on an analogy to McClintock's work²⁰ on the J-integral, describing the nature of the singularity in the crack tip region, Goldman and Hutchinson²¹ have shown that C^* characterizes the stress and strain rate in the vicinity of the crack tip in the secondary creep regime as follows.

$$\sigma_{ij} \cdot \dot{\epsilon}_{ij} = \frac{C^*}{r}$$

For materials undergoing secondary creep deformation (steady state) the stress and strain rate are related by,

$$\frac{\dot{\epsilon}}{\dot{\epsilon}_0} = \alpha \left(\frac{\sigma}{\sigma_0} \right)^n$$

where $\dot{\epsilon}_0$ and σ_0 are reference strain and stress respectively and α and n are material constants.

Combining the above two relations, it can be shown that,

$$\sigma_{ij} \propto \left(\frac{C^*}{r} \right)^{1/(1+n)}$$

and, $\dot{\epsilon}_{1j} \propto \left(\frac{C^*}{r}\right)^{n/n+1}$

where r is the distance to the crack tip. Thus, C^* characterizes the crack tip stress and strain rate field under conditions of local secondary creep behaviour. This has been one of the most successful parameters developed for correlating creep crack growth behaviour and its applicability to various structural alloys has been reviewed in detail by Sadananda and Shahinian.¹⁷ Excerpts from this review are given below.

Al alloys: Shown in Fig. 3.7 are crack growth rates for Al 2219,²² RR58²³ and Alloy 6061,²⁴ at various temperatures, plotted as a function of both linear and non-linear parameters. The data for RR58 covers several different specimen geometries and crack lengths. Though in Fig. 3.7 it appears that both K and C^* (also called J^*) have the same amount of scatter, in the latter case the data was clustered at the center of the band with only a few data points at extremes.

The superiority of the C^* parameter is clearer in the case of the 6061 alloy which was tested at a higher temperature than the other two.

Stainless steels: Extensive work on stainless steels indicates that C^* correlates the data much better than other parameters, at least at 650°C. Crack growth rate data for a Type 304 stainless steel in terms of C^* is represented in Fig. 3.8, where data from different investigations²⁵⁻²⁸ are represented. A wide spectrum of specimen geometries were used and in spite of this, the data fall in a somewhat narrow band with crack growth rates for a given J^* value falling within a reasonable amount of scatter. In contrast, crack growth rates

plotted in terms of the stress factor varied by two orders of magnitude for a given stress intensity value.

Superalloys: The applicability of the C^* integral has been studied in Discaloy, Alloy 718, Udimet 700* and IN-100.¹⁷ The crack growth rate results are shown in Fig. 3.9, plotted in terms of C^* parameter. In Alloy 718 and Udimet 700, crack growth has been studied using only compact tension specimens but in the different loads. The data for IN-100 correspond to one specimen geometry. The work on Discaloy involved both compact tension and centre-notched specimens.

It may be noted that the crack growth rates in terms of C^* in Discaloy and Udimet 700 are nearly the same in spite of the differences in temperatures, but the growth rates in Alloy 718 and IN-100 differ considerably. Unlike the case of the stainless steel data in Fig. 3.8, the C^* parameter does not normalise the temperature dependence of growth rates in Alloy 718. This may be related to the fact that Alloy 718 is strongly attacked by the environment and this effect is not taken into account by C^* .

When all the data available for the crack growth behaviour of various structural alloys in terms of the J^* (or C^*) is plotted in one graph (Fig. 3.10) it is interesting to note that except for Alloy 718, IN-100 and RR58, all of the other alloys fall more or less within the scatter band for stainless steel. This may be taken to imply the crack growth process is essentially the same. The behaviour of Alloy 718, IN-100 and RR58 alloy can be interpreted as being due to the presence of an additional parameter, probably environmental, contributing to crack growth.

Sadananda and Shahinian¹⁷ have concluded that there is a temperature range for each alloy where J^* is a better parameter to characterize creep crack growth provided that the alloy is not very sensitive to the environment. While environmental sensitivity provides one limiting case, extensive creep ductility could also make C^* invalid. For the latter case, the reference stress may provide a convenient means of predicting life. For significantly environment sensitive materials K can adequately describe crack growth rates. It has also been observed by Saxena¹¹ that the crack opening displacement rate and C^* may be uniquely related as K and the crack opening displacement are in the linear elastic case. Indeed, Krishnan and McEviley²⁴ found an even better correlation with the crack opening displacement rate parameter than with C^* in the case of Alloy 6061 crack growth rate data.

3.3 Parameters Influencing Creep Crack Growth from a Pre-existing Flaw

Substantial variations can result in creep crack growth rates due to the influence of parameters such as temperature, environment and microstructure. The manner in which these extrinsic and intrinsic factors play a role in the crack growth process is described below.

3.3.1 Temperature

The influence of temperature on cracking susceptibility of Inconel 718, as reported by Floreen² is shown in Fig. 3.11. Decreasing

the test temperature considerably lengthens the time to failure, but only slightly increases the threshold stress intensities. The large decrease in crack growth resistance in going from 538°C to 650°C is perhaps due to environmental effects playing a dominant role in the crack propagation process in this alloy at 650°C and a much lesser role at 538°C. Similar effects of temperature have been observed with other alloys.²

Floreen² also illustrated the influence of temperature upon Stage II crack growth rates in Inconel 718 at three different stress intensities, as an Arrhenius plot (Fig. 3.12). This illustration indicates that the temperature dependence of the cracking rate is independent of stress intensity and has an apparent activation energy 197 kJ/mole (47 kcal/mole). This value is well below the value expected for self-diffusion in a nickel-base alloy. However, Sadananda and Shahinian^{13,17} have observed that this value is quite close to the activation energy for grain boundary diffusion in Monel (41 kcal/mole) which is also a nickel-base alloy. This, together with the fact that creep fracture in Inconel 718 is generally intergranular led them to conclude that creep crack growth in this alloy is essentially controlled by diffusion of vacancies along grain boundaries.

If the crack growth process is purely diffusion controlled, then, one would expect even larger crack growth rates at 760°C than at 650°C. Since Fig. 3.12 indicates that crack growth rates are only slightly larger than at 650°C, it was agreed that the severe plastic flow which occurs as a result of lowering of the yield stress due to

the increase in temperature as well as due to a continuous change in microstructure is responsible for the retardation of crack growth. Thus, the creep crack growth process is visualized as a balance of two competing processes - a diffusion process which aids the crack growth and creep deformation which retards crack growth. Sadananda and Shahinian¹³ have pointed out that extensive plastic deformation could interfere with the diffusion process contributing to crack growth in two ways. First, because of their stress field, the dislocations could provide alternate sinks to the point defects. This, however may involve some bulk diffusion which may be significant at high temperatures. Secondly, the stress field of the crack could be relaxed to such an extent that the crack becomes a less efficient sink for the point defects.

It was also pointed out by these investigators that though creep deformation may retard the crack growth process, this may not always be the case. If a material undergoes extensive deformation, the deformation field could itself provide a source rather than a sink for point defects. In such cases the deformation may even enhance the diffusion controlled crack growth process. Thus, the plastic deformation at the crack tip can either retard or enhance the crack growth process depending on the creep ductility of the material, test temperature and loading conditions.

Service temperature exposure for an extended period of time can lead to a loss in crack growth resistance due to microstructural degradation. In Inconel 718, a reduction of creep crack growth

resistance of as much as sixty per cent has been reported as a result of a one week exposure at 538°C. However, this would have to be verified since it is generally known that this alloy is microstructurally stable up to temperatures around 650°C for periods up to 10,000h.

3.3.2 Environment

It is generally accepted that an oxygen containing environment has a very strong effect on creep crack growth rates in elevated temperature alloys. In conventionally aged Inconel 718, the growth rate in air is about 100 times higher than in helium.⁵¹

Figure 3.13, given by Floreen²⁹ illustrates the creep crack growth rates for Inconel 718 at various temperatures at stress intensities of 30-50 Mpa $\sqrt{m}$. Also shown is an estimate of the depth of grain boundary diffusion per unit time, $\sqrt{Dt}$, where D is the grain boundary diffusivity and 't' is the time. The value of D has been taken from the work of Hwang and Baluffi³⁰ and is very approximate.

~~The effect of stress upon diffusion rate has been neglected.~~ Floreen²⁹ observed that although the two curves have different slopes, over much of the temperature range of interest the diffusion distances and the crack growth distances per unit time are of comparable magnitude. Thus, it was concluded that it was possible for grain boundary diffusion of some embrittling species such as oxygen to keep pace with creep crack growth. However, it should be noted that this is an oversimplified analysis and that the diffusion rates are strongly

enhanced by the presence of a stress field. A more rigorous analysis is needed to define the manner in which environmental contributions occur during creep crack growth.

While an oxygen containing environment can cause considerable acceleration in creep crack growth rates, sulphidizing environments, particularly H_2S can be even more severe.²⁹ CH_4 , H_2 and CO_2 containing environments could be slightly aggressive.²⁹

3.3.3 Stress, Loading Effects

The importance of considering the effect of stress on environmental contributions was mentioned earlier. Environmental effects are found to be more severe at low stress intensities.² At intermediate temperatures, some minimum stress or stress intensity often appears necessary to produce an environmental effect. Thus Floreen² reports that unstressed specimens exposed to fairly severe oxidizing or sulfidizing environments showed no signs of attack and exposures of unstressed specimens to an aggressive environment had no effect on the subsequent crack growth behaviour. As a result, elevated temperature corrosion tests may be of little value in predicting service behaviour under mechanical loading. However, in reading these statements one should note that the temperature of exposure is a crucial factor. Depending upon the environment, an alloy may be oxidized or corroded above a certain temperature (even in the unstressed condition) to a level that can lead to substantial decreases in creep crack growth resistance.

Creep crack growth data in air often display a threshold stress intensity below which no cracking takes place. This observation is consistent with the hypothesis of Sadananda and Shahinian mentioned earlier that creep crack growth occurs in the presence of two competing reactions. At the threshold, stress relaxation at the crack tip and the resultant blunting of the crack may be more effective than the processes contributing to crack growth. This threshold is sensitive to the environment, and more aggressive environments lower the threshold.¹³

The relationship between the threshold stress intensity factor and level of aggressiveness of the environment may be explained as the basis of a critical stress or strain rate at the crack tip necessary to disrupt protective surface films and allow the environment to react with the alloy.²⁹ However, there is no experimental evidence that such a mechanism is operative.

3.3.4 Microstructure

It has been conclusively demonstrated that the grain size has a very important role in creep crack growth behaviour, with coarser grain sizes exhibiting improved crack growth resistance. This is most likely to be due to the reduced grain boundary sliding which occurs at higher grain size values. There is perhaps a synergistic effect between the rate of environmental attack and the extent or rate of grain boundary sliding as well.

Floreen² has attempted to correlate the creep crack growth behaviour of several Ni-base superalloys simply on the basis of grain size, and surprisingly enough, a reasonable correlation was achieved. However, it would be too simplistic to assume that this is the only influential factor. The extent or rate of grain boundary sliding would be affected by the grain boundary morphology as well, in which case a serrated grain boundary structure would be beneficial.

Sadananda and Shahinian performed creep crack growth rate tests on Inconel 718 subjected to various heat treatments summarized in Table 3.1. These heat treatments produced significantly different microstructures, covering a wide range of precipitate sizes, grain sizes and grain boundary morphologies. The crack growth rates measured are illustrated in Fig. 3.14.

As expected the "conventional" heat treatment, which produces a relative fine grain size led to fairly rapid crack growth rates. However, the "Wilson" heat treatment while producing the same grain size also produced copious amounts of Ni_3Cb needles (δ -phase) at the grain boundaries. This perhaps led to considerable grain boundary strengthening, since this heat treatment was found to result in creep crack growth rates which were an order of magnitude lower than the others. The matrix precipitates were also overaged in the case of the Wilson heat treatment which might have minimized planar slip character, the development of stress concentrations at grain boundaries due to slip band impingement and consequently the rate of diffusion of embrittling environmental species into the material.³²

It is important to note, however that while grain coarsening or overaging may be very beneficial to creep crack growth resistance, heat treatments which achieve this may also result in substantial decreases in high strain LCF crack initiation life, tensile and yield strengths. In gas turbine disc applications, heat treatment which produce the best overall balance of these properties is required rather than those which maximise a specific property. One of the tasks of this study is to develop such a heat treatment for Inconel 718.

Chapter 3

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Table 3.1

Heat Treatments Employed

Heat Treatment	Designation
1. Anneal at 955°C for 1 hour - air cool Age at 720°C for 8 hours, furnace cool to 620°C. Hold at 625°C for total aging time of 18 hours. (Conventional heat treatment.)	A
2. Anneal at 1050°C for 1 hour - air cool Age at Age at 760°C. Hold at 650°C for total aging time of 20 hours. (Conventional heat treatment.)	B
3. Anneal at 1035°C for 20 min, furnace cool to 915° Intermediate anneal at 915°C for 10 hours, air cool Age at 720°C for 8 hours, furnace cool to 620°C. Hold at 620°C for total aging time of 18 hours. (Muller and Donachie heat treatment.)	C
4. Anneal at 1093°C for 1 hour - cool at 38°C/hour to 720°C - Age at 720°C for 4 hours, cool at 38°/hour to 620°C - Hold at 620°C for 16 hours (INEL heat treatment.)	D
5. Anneal at 926°C for 10 hours, furnace cool Age at 734°C for 48 hours (Wilson heat treatment.)	E

3.1 Stages of Cracking Observed in Static Load, Creep Growth Tests

3.2 Schematic Representation of the Deformation Zones Ahead of the Crack Tip at Elevated Temperatures

3.3. Experimental Procedure for Determining the J-Integral and J* Integral¹⁷

3.4 Time Dependence of Crack Length and COD in Single-
Edge Notched Specimens¹⁶

3.5 Definition of the C^* (J^*) Integral¹⁶

3.6 Experimental Determination of the C^* Integral

3.7 Creep Crack Growth Rates for Aluminum Alloys at Various Temperatures¹⁷

3.8 Creep Crack Growth Rates for Stainless Steel (AISI 304) Plotted in Terms of C*¹⁷

3.9 Applicability of the C^* Integral for Correlating Crack Growth Rates in Superalloys¹⁷

3.10 Creep Crack Growth Rates for a Variety of Alloys Plotted in Terms of the C^* Integral¹⁷

3.11 Influence of Temperature on Creep Crack Growth Rates in Inconel 718²

3.12 Temperature Dependence of Cracking Rate in Inconel 718, for Three Different Stress Intensity Factors^{2A}

.3.13 Creep Crack Growth Rates in Inconel 718 Compared With Diffusion Rates of Oxygen ²⁹

3.14 Influence of Heat Treatment Upon Creep Crack Growth Rates in Inconel 718³¹

CHAPTER 4

LOW CYCLE FATIGUE AT ELEVATED TEMPERATURE

With increasing temperature, several additional factors must be considered in describing fatigue behaviour and predicting lifetimes. This is primarily because,

- (a) creep cavitation damage is possible during the tensile loading portion of the LCF cycle.
- (b) metallurgical instability as well as embrittling reactions with the environment can occur more readily.

In this section, the various intrinsic and extrinsic factors which influence fatigue behaviour by promoting one or both of the above interactions have been described. As in Chapter 5, published data obtained using high strain LCF tests and those determined by fracture mechanics tests have been reviewed in separate sections.

4.1 Fatigue Crack Initiation

4.1.1 Transgranular Crack Initiation

Though intergranular crack initiation is encountered most often at elevated temperature, the incidence of transgranular cracking is still significant and is affected by the following factors:

Temperature and Time: These parameters affect the transgranular initiation process by modifying the concentration and scale of slip displacements. Overaging or reversion of precipitates under certain temperature-exposure time conditions concentrates local slip activity and can hasten crack initiation.¹ On the other hand, strain aging or

strain hardening which can occur at certain temperatures can disperse slip and prolong the initiation life. However, excessive dispersal can have adverse consequences, as observed in the case of Udimer 700.² This is because the precipitation of particles accompanying cyclic strain aging can reduce the cohesive strength of slip bands.

Environment: Though rigorous evidence is not available at this time, it is possible that adsorption of environmental gases plays a critical role in slip band crack initiation through progressive passivation of fresh surface exposed during cycling.¹

Microstructural Defects: The presence of intragranular microstructural defects such as pores or brittle phases is likely to promote crack initiation as in the case of room temperature fatigue, but the effect is likely to be felt much less in alloys which undergo wavy slip at elevated temperature.¹

Bashir et al.³ found that pores and ceramic inclusions in AS-HIP and HIP + Forged René 95 promoted Stage I crack initiation at 650°C. In both materials, a marked tendency for planar deformation was observed. Jablonski⁴ found in the case of Astroloy, that ceramic inclusions do not decrease LCF life at 25°C, but substantially reduce it at 500°C. This difference was attributed to the observation that at a ceramic inclusion cracks at 500°C, while it debonds from the matrix at 25°C. The initial starter crack size was found to be much larger for the cracked inclusion and also, propagation from the cracked inclusion occurred more rapidly since it represents a theoretically sharp crack.

Very little further information is available in the literature regarding factors that influence transgranular crack initiation. However, as noted earlier, the influence of various other microstructural or environmental factors can be rationalized on the basis of their influence upon slip character. It is important to note that transgranular crack initiation need not necessarily occur in the Stage I mode. In materials which deform predominantly through wavy slip at elevated temperature a variety of other crack initiation mechanisms can occur which lead to non-crystallographic (Stage II) crack initiation. Examples of these are the development of a notch-peak topography by intrusion and extrusion, pore formation in zones of distortion and surface instability, all of which are consequences of the irreversibility of wavy slip.¹ But, when the deformation of complex polycrystalline alloys becomes homogeneous under creep conditions, the crack initiation mechanism usually changes to an intergranular mode so that the other mechanisms are really of very little practical interest.¹

4.1.2 : Intergranular Crack Initiation

The temperature at which the transition from transgranular to intergranular crack initiation occurs is a function of the specific test conditions, type of alloy and its microstructure. Intergranular crack initiation generally occurs more rapidly than transgranular crack initiation and can thus have a profound effect on total life.

Conditions which promote intergranular crack initiation are generally those which contribute to creep cavitation and/or to environmental embrittlement. The microstructural factors which promote creep cavity nucleation were reviewed in detail in an earlier chapter and need not be repeated here. The environmental contribution is however, mechanistically different from the case of the creep contribution described earlier and therefore, the possible mechanisms leading to this effect and the factors influencing it are reviewed below.

Low cycle fatigue crack initiation in the presence of a strongly oxidizing environment is thought to occur by a repeated sequence of oxide rupture by deformation leading to more oxidation and a concentration of deformation - eventually leading to the formation of a notch, from which a crack proceeds.^{5,6} Due to the repeated rupturing of the oxide film due to plastic strain, the environmental contribution to crack growth is usually more pronounced in cyclic loading conditions compared to static loading.

Only very limited information is available regarding the role of microstructure in promoting the above mentioned environmental effect. In Inconel 718, the carbides present at the grain boundaries oxidize easily and fatigue crack nucleation occurs preferentially at these locations, particularly when the grain size is small.⁷ The locally oxidized region then acts as a notch to further concentrate the strain until a fatigue crack is initiated. Since grain boundary sliding can be expected to amplify the film rupture and local oxidation,

process, the level of embrittlement is more pronounced in a relatively fine grained alloy.⁸ A similar observation, regarding preferred crack initiation from oxidized primary MC carbides has been reported by Reuchet and Remy⁹ who studied a cast cobalt-based superalloy.

Virtually no work has been carried out so far to illustrate whether there is any connection between matrix precipitate size (in precipitation hardening alloys) and the sensitivity to environmental embrittlement. The only detailed work which can be referred is that of Prager and Sines¹⁰ and that of Thamburaj et al.¹¹ who showed that the tensile fracture ductility of superalloys is substantially reduced in the presence of hyperfine precipitates, as a result of enhanced environmental embrittlement. A limited amount of work has been carried out by Woodford and Coffin¹² on A-286 which confirms the correlation between γ' particle size and sensitivity to environmental embrittlement and hence, LCF life at 593°C. In Fig. 4.1, heat treatment (HT) 2 and 3 had coarser γ' particles than HT-1. It is also seen that the effect of heat treatment is more obvious at lower frequencies. A fine matrix precipitate size can enhance coarse, planar slip causing intense stress concentrations at the grain boundaries due to dislocation pile-ups. Under these conditions grain boundary cracking can readily occur even if the grain boundary strength has been only slightly reduced due to oxidation. Also, it is possible that these high stresses somehow aid in the inward diffusion of oxygen into the grain boundaries. Though the precise mechanism by which a planar slip character enhances environmental embrittlement at elevated temperatures

is not known, there appears to be a clear correlation between these phenomena. Further work is needed to fully understand this very important relationship.

While microstructural effects are of considerable importance as described above, the macroscopic surface finish resulting from the manufacturing techniques used to prepare a component can also influence crack initiation. There is some evidence that elevated temperature intergranular fatigue crack initiation in a nickel-base superalloy is accelerated by surface working and this is believed to be related to a surface topography effect.¹

Foreign object damage may also shorten life by increasing the ease of initiation. In a study of iron and nickel-base superalloys subjected to thermal fatigue in a fluidized bed, the number of cycles to the first observed crack was reduced drastically when specimens were subjected to impact.¹³

4.2 Fatigue Crack Propagation: LCF Tests

4.2.1 Transgranular Propagation

As in the case of room temperature fatigue, polycrystalline alloys usually exhibit a transition from Stage I to Stage II transgranular fracture with increasing crack length and stress intensity. In single crystals, a Stage I to Stage II transition is never observed at a given temperature.¹ A change from Stage I to Stage II propagation can, however, occur with increase of temperature or lowering of frequency and this has been identified as being associated with a change in slip character from planar to wavy.¹⁴ It has also

been found in single crystals of carbide-free Mar M-200 that the proportion of Stage I fracture increases with increase in frequency.¹⁵ Similar results have been noted with polycrystalline Udimet 700.¹⁶ Wells et al.¹ have suggested that these results are because the mode of slip is strain rate dependent, becoming more planar with increased frequency and crack length.

It is interesting to note that in some cases, a crack propagating in Stage II mode can undergo a reversed transition to Stage I. This is believed to be once again a result of the increase of planar slip with increased crack length and therefore, strain rate. This has been observed in both single crystals and polycrystals of superalloys at elevated temperature.¹

This somewhat confusing situation can be clarified if it is understood that these transitions occur as a result of two basic competing phenomena. For example, in a polycrystalline alloy a cross-over from Stage I to Stage II occurs at a given temperature due to the misorientation of slip planes across grain boundaries which makes continuous slip band crack propagation difficult. On the other hand for given materials, beyond a certain critical crack tip strain rate planar slip is the favoured deformation mode. Therefore, though the misorientation of slip planes across grains may make slip band crack propagation difficult, the slip character becomes much less homogeneous and a reversed transition to Stage I propagation may be observed.

It has already been explained in the section on room temperature fatigue that Stage I fatigue failure is sensitive to

microstructure whereas Stage II propagation is dependent only on the temperature and the continuum properties (primarily the elastic modulus) of the material. The manner in which microstructural parameters affect Stage I propagation have already been explained and need not be repeated here.

4.2.2 Intergranular Propagation

Grain Size Effects: Wells et al.¹ proposed an early model to predict the effect of grain size on fatigue life. A schematic drawing of this model is shown in Fig. 4.21. It was argued by these investigators that at low stress ranges, a single intergranular crack is initiated and propagates to failure with little or no crack linking. It can be seen in Fig. 4.2a and 4.2c that a crack on a transverse boundary in a fine grained material is much smaller before it reaches a triple point than a crack in a coarse grained material. At the triple point, the cracks must undergo the more difficult processes of crack re-initiation, branching and propagation on a boundary not normal to the stress axis if it continues intergranular (Fig. 4.2a), or must propagate in a much slower transgranular mode (Fig. 4.2c). Under these circumstances, it was postulated that fatigue life should increase with reduced grain size. In the second case, likely to be found at high strain ranges, cracks initiate on transverse grain boundaries with the number of cracks increasing with reduced grain size. If linking of cracks separated by a grain diameter is the critical step in the

fracture process, whether it be intergranular (Fig. 4.2b) or transgranular (Fig 4.2d), the rate of link up should be inversely proportional to grain size. Here, the fatigue life would increase with grain size.

Recent results indicate that this model is far from being generally applicable. For example, in a fine grained material in which propagation occurs intergranularly, crack link up may be difficult, but at elevated temperature, extensive grain boundary sliding can occur to promote creep cavity nucleation and growth. Further, as will be pointed out later on this study, environmental effects can be much more pronounced in a relatively fine grained material. Thus, in practice, it is found that a fine grained material exhibits a relatively poor crack growth resistance due to creep-fatigue-environment interactions.

The fatigue life at elevated temperature is generally improved in directionally solidified alloys and in single crystals. As shown in Fig. 4.3, the fatigue life of both columnar grained and single crystal Mar-M200 is approximately one hundred times greater than that of conventionally cast material at 926°C.^{17,18} At these temperatures, cracks initiate on transverse grain boundaries in the polycrystalline material and propagate rapidly to failure along entirely intergranular paths. In both the columnar grained and single crystal materials, most of the life is spent in the more difficult process of transgranular propagation resulting in much higher lives. Single crystals of Udimet 700 show an increase in fatigue life compared to wrought polycrystalline material at 760°C, although the improvement was not so large as that for Mar-M200 at either 760°C or 926°C.¹⁷

4.3 Factors Influencing Fatigue Crack Propagation

4.3.1 Frequency, Hold-Time, Mean Stress Effects*

All three of these factors influence fatigue crack propagation at elevated temperature by modifying the creep and environmental contributions.

Coffin has presented a model describing the effect of frequency which is shown schematically in Fig. 4.4. It is suggested that three frequency regimes exist in air which can be identified with different damage processes. At high frequencies ($\nu > \nu_e$) the damage process is independent of frequency and failure is by the normal transgranular fatigue mechanism. In the intermediate frequency range ($\nu_m < \nu < \nu_e$) the air environment is capable of interacting with the cracks and there is change of fracture mode from transgranular to intergranular with decreasing frequency. At lower frequencies, ($\nu < \nu_m$) microstructural instabilities may add to the environmental effects and once again, intergranular fracture may result. Most experimental investigations support this model strongly. The effect of frequency upon the LCF life of A-286 has already been referred to earlier in Fig. 4.1.

The effect of increasing the hold or dwell period or increasing the mean stress has been found to have the same effect as decreasing the frequency.^{19,20}

*These parameters affect crack initiation as well as propagation life.

Strain hold-time tests on austenitic stainless steels and materials of similar strength and ductility have indicated that tensile strain holds are the most damaging mode for equivalent periods²¹ (Fig. 4.5). On the other hand, for cast nickel-based superalloys, compressive strain holds are the most damaging²² (Fig. 4.6).

Coffin⁵ has attempted to explain why tensile strain hold periods are less damaging than compressive strain hold periods in high strength nickel-base superalloys using Fig. 4.7. Considering the initial loading to be largely elastic, upon reaching the strain hold point 1, inelastic stress relaxation $(\epsilon_{in}^r)_1$, occurs, such that the unloading results in a compressive stress at 0. Again, assuming nearly elastic loading, further stress relaxation at 1 causes further compression at 0. Continued cycling causes further lowering of the stress-strain loop. Concurrently, inelastic plastic strain $(\epsilon_{in}^p)_0$ begins to develop as the unloading path approaches 0. The magnitude of this plastic strain increases with increasing loop shift. The loop stabilizes when $(\epsilon_{in}^r)_1 = (\epsilon_{in}^p)_0$ and a compressive mean stress is so introduced. For compressive strain hold periods, the reverse action develops and a tensile mean stress results. Since tensile mean stresses generally reduce fatigue life the effect of compressive hold times can be more detrimental than that of tensile hold-times.

Another explanation given by Teranishi and McEvily,²³ who studied 2½ Cr-1Mo alloy steel is that this behaviour is related to the interaction of the oxide formation and spallation with surface deformation. In compression after a tension hold, the oxide film

spalls and produces a fresh surface removing surface damage and inhibiting crack formation, while in tension after a compression hold the oxide cracks rather than spalling, creating localized stress concentrations to accelerate localized oxidation and crack nucleation.

The results of Sanders et al.²⁴ shown in Fig. 4.8 regarding the effect of compressive and tensile hold-times upon life of Inconel 718 at 649°C confirm the above findings.

4.3.2 Wave Shape Effects*

Coffin⁵ made the very important observation that the wave shape in cyclic loading can have a marked effect upon the resistance to fatigue crack initiation and propagation at elevated temperatures. This finding was made by testing A-286 at 593°C with plastic strain ranges of 0.01 and 0.005. Tests were run at the same overall frequency, using fast-slow, symmetrical and slow-fast loading sequences as shown in Fig. 4.9. The results of these tests are shown in Fig. 4.10, and it can be noted that slow-fast cycles are more damaging and fast-slow cycles less damaging than equal tension and compression-going frequencies for both strain ranges.

Sidey and Coffin²⁵ using test results on OFHC copper at 400°C and 304 stainless steel at 650°C developed an explanation for such behaviour as summarized in Table 4.1.

*This parameter affects life to crack initiation as well as propagation life.

The slow-fast type of cycle introduces ~~deep~~ creep damage in addition to environmental damage. Cavity growth can occur in the forward tensile portion of the cycle, the growth rate depending upon the temperature and strain rate. Growth cannot occur during the fast, compressive portion of the cycle and in fact, this latter reversal of stress would tend to heal some of the creep damage caused by the tensile stress.²⁶ However, since the strain rate in the compressive portion of the cycle is not equal to that in the tensile portion, the cavitation damage is not completely healed and with each cycle a steady growth of cavities can be expected depending upon the relative forward and reverse strain rates. A schematic view of the progressive grain boundary cavity growth is described in Fig. 4.14 for each phase of the slow-fast loading.

When tests are conducted in air, the environmental contribution to crack growth may also be expected to be larger in the slow-fast type of cycle than in the fast-slow cycle. It is of interest to note that the difference between the slow-fast and fast-slow wave shapes persists even when tests are conducted in vacuum, showing that the creep contribution is much more pronounced with the slow-fast wave form, as explained above.

Failure in specimens tested under the fast-slow type of hysteresis loop occurred by a propagation of a single transgranular crack. In this type of cycle, the tensile-going strain rate is high, so that the crack is only exposed to the environment for a short period of time and this likely to lead to a transgranular fatigue crack.

Since the strain rate during the compressive part of the cycle is low there is time for any intergranular damage introduced during the tensile ramp to be reversed. A schematic view of this process of cavity collapse at the grain boundaries is shown in Fig. 4.12. Since no creep damage is expected with this type of cycle in air or in vacuum, environmental damage in air depends upon forward strain rates only.

Equal tensile and compressive ramp cycles produce a transgranular or intergranular main crack depending on the frequency. Intergranular surface cracks and secondary intergranular interior cracks near the fracture have been observed with OFHC copper, but no bulk damage. As shown in the schematic in Fig. 4.13, no cavity growth is anticipated on grain boundaries since the processes of growth and collapse during each phase of the cycle are balanced and no net changes are anticipated. In this case, intergranular initiation of cracks is promoted by cracks and further growth takes place along the grain boundaries which are the easiest diffusion paths for oxygen. As shown in Table 4.1, the damage processes in equal ramp rate regimes are identified as environmental and time-independent fatigue in air, while in vacuum, damage is only due to cycle-dependent (time-independent) fatigue.

While these explanations appear reasonable, it should also be noted from Fig. 4.8, that the mean stress for the slow-fast cycle is compressive and it is tensile for the fast-slow cycle. It is generally believed that tensile mean stresses are more damaging. Yet, the slow-fast cycle contributes to less damage than the latter. It appears

therefore, that the effect of mean stress has not been fully understood and hence any explanations for observed behaviour on the basis of mean stress (as that of Coffin mentioned earlier regarding the effect of compressive and tensile hold-times) must be viewed with caution.

4.3.3 Plastic Strain Range*

This is a primary variable influencing the level of oxidation damage. An increase in the severity of oxidation with increase in the plastic strain range has been observed by Antolovich and Rosa²⁷ in the case of René 77. As mentioned earlier there are very few theories at present which account for the interactions between the environment and plastic deformation behaviour. Antolovich and Rosa²⁷ also pointed out that higher plastic strains tend to accelerate coarsening of strengthening precipitates which would, up to a certain extent enhance ductility in some situations. In such cases, the life is then determined by a trade-off between the beneficial internal changes and the detrimental environmental reaction due to the higher levels of deformation.

An interesting effect of the plastic strain range has been observed by Sanders et al.,²⁴ in Inconel 718. Their results are shown in Fig. 4.14 which indicates that for temperatures up to 427°C, the dependence of N_1 (life to initiation) on plastic strain range $\Delta\epsilon_p$ remains essentially unchanged. Above 427°C, the material response changes dramatically. When N_1 exceeds about 10^4 cycles, lifetime at a

*This parameter affects life to initiation as well as propagation life.

given value of $\Delta \epsilon_p$ is greater for the two higher temperatures than for the lower temperatures.

Sanders et al. attributed these results to a change in deformation mode with temperature. At 204°C, 316°C and 427°C deformation was thought to occur predominantly by deformation twinning and hence resulted in the same strain-life response. At 538°C and 649°C the dominant mode of deformation was slip which though still planar, represented a relatively more homogeneous mode of deformation. Hence, specimens tested at these temperatures exhibited a different strain-life response.

It is felt that the results of Sanders²⁴ can be explained also on the basis of a cross-over in the level of environmental contribution to crack growth. At high plastic strain ranges and elevated temperatures (>538°C) stress-assisted oxidation occurs rapidly and give rise to a lower life than at lower temperatures (Fig. 4.14). At low plastic strain ranges, the environmental contribution to crack growth at 538°C and above could decrease markedly and the relatively homogeneous deformation could give rise to higher lives than at lower temperatures.

Apparently conflicting results have been obtained by Fournier and Pineau²⁸ with the same alloy. In their results (Fig. 4.15) depicting the strain life relationships at room temperature and 550°C, no cross-over was observed, as reported by Sanders.²⁴ However, it should be noted that the test conditions of these investigators are not the same. Fournier and Pineau²⁸ used a frequency of 3 cpm while Sanders et al. used 20 cpm. It is quite possible that in the results of the former, the lower frequency contributes to a greater

environmental effect even at low plastic strain ranges, thus making the elevated temperature life always lower than that at room temperatures. Further, at low frequencies, Fournier and Pineau²⁸ found that the incidence of twinning was worse at elevated temperature than at room temperature, which is the opposite of what Sanders et al.²⁴ found at 20 cpm. The deformation modes in these two conditions are quite different.

4.4 Fatigue Crack Propagation: Fracture Mechanics Tests

Clavel and Pineau²⁹ have compared the deformation modes obtained in Waspaloy and Inconel 718 during LCF tests and in fracture mechanics tests and their results are summarized in Fig. 4.16. In most cases, a very close correspondence exists between the deformation modes observed in LCF specimens (tested at a large fatigue life) and in fatigue crack growth rate (FCGR) or fracture mechanics tests in the near threshold region. A limit can be drawn, corresponding to $N_f = (5-10) \times 10^3$ cycles and $da/dN = 10^{-8}$ m/cycle. On the right hand side of this limit correspondence clearly exists. On the left hand side, the situation is more complex. In particular, it is not possible to relate N_f in an LCF test to da/dN in a FCGR test. It was emphasized by these investigators that abundant twinning typical of very high strain LCF tests was never observed in FCGR specimens. It was also stated that in their experiments, the cyclic plastic strain amplitude was always less than 1% within the plastic zone around the fatigue crack tip.

In the following section, elevated temperature fatigue crack growth behaviour under fracture mechanics test conditions has been reviewed.

4.4.1 Fracture Mechanics Parameters

Linear Elastic Fracture Mechanics techniques have been used successfully to predict fatigue crack growth rates in several superalloys at room temperature, due to the fact that a single parameter, namely, the stress intensity factor can adequately describe the stress field around the crack tip. However, at elevated temperature, crack tip plasticity effects become appreciable and non-linear parameters, such as the J-integral and J^* will have a broader range of applicability.

The concept of the J-integral has been explained already. Strictly, the J-integral concept is valid only for monotonic deformation since irreversible deformation during unloading should invalidate its applicability for elevated temperature fatigue. In spite of this, initial attempts^{30,31} to use the J-integral for fatigue were somewhat successful. Sadananda and Shahinian³² explain this as being due to a material developing a characteristic cyclic stress-strain behaviour under fatigue which is different from that under monotonic deformation and yet unique for a given set of experimental conditions such as temperature, environment, etc. At a microlevel, this means that the material ahead of the crack tip acquires a characteristic plastic zone that is different from that under monotonic deformation.

and yet unique for a given crack length and loading condition. This uniqueness of the cyclic plastic zone implies that strain energy for a crack increment is unique for a given crack length and load. Therefore, the J-integral concept may be applicable under fatigue although its value at a given crack length differs from that under monotonic deformation.

Figure 4.17 shows the applicability of the J-integral for load controlled fatigue with reference to high temperature fatigue crack growth in Udimet 700 at 850°C. The crack growth rate data are specifically represented in terms of $\sqrt{\Delta J \cdot E}$ (where E is the Young's Modulus) thus normalizing the data with respect to the Young's Modulus. The extent of spread in the data is less in terms of ΔJ than in terms of ΔK . Dowling and Begley³³ have also shown that the J-integral has a broader range of applicability in terms of specimen geometry.

The superiority of the J-integral over ΔK in correlating crack growth behaviour becomes apparent when one considers cases where the metallurgical structure becomes unstable during the test. For example, if any time-temperature dependent transformations occur at test temperatures, the microstructure of the material ahead of the crack tip changes to a different extent in each specimen depending upon the duration of the test. This in turn, may cause a large spread in the da/dN versus ΔK plot for the same specimen geometry. Here the inapplicability of the linear elastic parameter is due to its inability to take into consideration the changes in the flow properties ahead of the crack tip. On the other hand, since the J-integral is determined

using load-displacement curves, which reflect the in-situ changes in the flow properties of the materials, it could be a better parameter to use to characterise crack growth in such materials. Thus, the J-integral as determined from the load-displacement curves is a kinematic parameter that takes into consideration dynamic changes in flow properties of the material ahead of the crack tip.³²

When crack growth under cyclic load becomes increasingly time-dependent both ΔK and ΔJ parameters may not correlate crack growth rates.³² In the extreme case of purely time-dependent growth, as in the case of Alloy 718 at 650°C, the linear parameter K_{max} or its non-linear counterpart, J_{max} may be better in correlating crack growth rates.^{34,35} This is illustrated in Figs. 4.18 and 4.19(a), where crack growth rates in Alloy 718 are represented in terms of ΔK and ΔJ as well as K_{max} and J_{max} . Although the ΔJ parameter is slightly better than ΔK , the spread in the data in terms of both parameters for different ΔK and ΔJ values is more than an order of magnitude. This spread is significantly less when plotted in terms of K_{max} or J_{max} although crack growth rates for continuous cycling and for 1-min hold fall in separate bands. That the crack growth rate in this alloy is predominately time dependent at 650°C for these test frequencies can be seen clearly in Fig. 4.19(b), where da/dt (da/dN times frequency) is represented in terms of K_{max} and J_{max} . It is interesting to note that K_{max} and J_{max} values for a certain value of da/dt are nearly the same, and this is because crack growth in this alloy occurs essentially under elastic

conditions. This is related to the high tensile and creep strengths of the alloy and to its poor resistance to environmental interactions.

At this time, it is not clear which fracture mechanics parameter is suitable when both time-dependent and cycle-dependent processes occur simultaneously, particularly when there are synergistic effects. Figure 4.20 illustrates a case where crack growth in Udimet 700 is represented in terms in ΔJ for various hold times.³⁶ The crack growth rates decrease and then increase with increase in ΔJ . A similar effect was noticed in terms of ΔK as well. Such crack retardation has been observed in several alloys^{32,36} particularly when the stress intensities are less than the threshold stress intensity for creep crack growth. The retardation occurs at low stress intensity factors due to bulk creep deformation which causes crack tip blunting, thereby contributing to stress relaxation at the crack tip. Since such relaxation effects are not simulated during the evaluation of the J-integral, ΔJ may not be useful in predicting crack growth behaviour. On the other hand, if crack growth occurs continuously by a combination of time and cycle-dependent processes a complex combination of both J_{max} and ΔJ -integral parameters may be useful. The procedures become more complex when there are synergistic effects.

4.4.2 Microstructural Effects

The effects of microstructure on fatigue crack propagation have been considered by several investigators. Due to the very large amount of information available, discussion of this effect will be limited in this section to the superalloys only.

Merrick and Floreen³⁷ studied the effects of γ' particle size in Waspaloy at 650°C and found relatively little effect at a test frequency of 1.0 Hz. The heat treatments employed produced uniform γ' sizes of 0.016 to 0.018 μm in one case and 0.020 to 0.023 μm in another case. In a third heat treatment, a bimodal γ' structure was produced in which the size of the fine γ' precipitate was 0.01 μm and the size of the coarse γ' precipitate was 0.08 μm . The results of their tests are shown in Fig. 4.21. The crack propagation rates seem to be independent of γ' particle size.

This result may seem surprising at first, since it was pointed out several times in earlier sections that γ' particle size and environmental embrittlement at elevated temperature are related. However, in Waspaloy the extent of planar deformation can be quite limited at 650°C. Clavel and Pineau²⁹ have observed very homogeneous cyclic deformation in this alloy at 650°C. Hence, the mechanism of Prager and Sines¹⁰ which explains environmental attack on the basis of particle cutting and the planar movement of dislocations does not apply in this case. Since the deformation apparently proceeds in a homogeneous fashion over the small range of γ' particle sizes studied, no difference in the level of environmental contribution to crack growth occurs and the observed FCGRs seem to be approximately the same.

Merrick and Floreen³⁷ also studied the effect of grain size on cyclic crack growth in Astroloy at 650°C. Heat treatments were carried out in which the γ' particle size was not well controlled but in which the grain size varied from ASTM 8 (fine) to ASTM 5 (coarse).

The fine grained specimens had a bimodal γ' distribution of 0.05 and 0.30 μm . Finally a coarse grained specimen was produced by heat treatment in which the grain boundaries were serrated and there was a bimodal γ' distribution with coarse γ' both in the grain boundaries and in the matrix along with fine γ' (0.065 μm) produced by aging. FCGR testing showed that the propagation rates tended to be grouped solely in terms of grain size independent of γ' and grain boundary morphology (Fig. 4.22).

Here again, the lack of sensitivity of FCGR to γ' particle size and grain boundary morphology can be explained on the basis of limited sensitivity to environmental embrittlement. Thamburaj et al.³⁸ have shown that Astroloy is not sensitive to stress-assisted environmental attack at 650°C, while it is very susceptible at 760°C, in the presence of hyperfine γ' . Also, the effect of grain boundary morphology would not affect FCGR at 650°C substantially in this alloy since grain boundary sliding becomes a dominant mode of deformation only in the vicinity of 760°C in this material.^{11,38} Actually, the fracture mode was found to be transgranular for all the heat treatments studied by Merrick and Floreen³⁷ at 650°C. This being the case it is surprising that some effect of grain size was observed at all. This could be due to differences in the proportion of Stage I fractures produced by the various microstructures.

Bartos and Antolovich³⁹ conducted a more comprehensive study on René 95, in which both γ' particle size and grain size were well controlled. FCGR testing was carried out at 540°C. The lowest FCP

rates were associated with fine γ' ($0.08\mu\text{m}$) and coarse grains ($\sim 50\mu\text{m}$ in diameter) while the highest FCP rates occurred when the γ' size was $0.5\mu\text{m}$ and the grain size was $10\mu\text{m}$. The results are shown in Fig. 4.23 which indicates that above about $0.2\mu\text{m}$, the γ' particle size has little effect on residual life while below this size residual life increases rapidly with decreasing γ' particle size, especially for the coarse grained material. Thus, it was concluded that planar slip which occurs with very fine γ' particle sizes and coarse grain sizes is beneficial to FCGR at 540°C in this alloy.

While it is clear from this study and others⁴⁰ that planar slip is beneficial to FCGR, it is important to note that the temperatures at which it is beneficial are below the ranges at which environmental interactions are dominant. At higher temperatures, when oxygen embrittlement can occur, a planar slip character can produce exactly the opposite effect, namely, lead to the rapid oxidation of grain boundaries and poor resistance to crack growth.

Mills and James⁴¹ have studied the effects of product form, heat-to-heat variation and heat treatment upon FCGR in Inconel 718. For Heat I, the product was bar-stock melted by VIM-VAR (Vacuum Induction Melted and Vacuum Arc Remelted) while for Heat II the product was plate, melted by VIM-EFR (Vacuum Induction Melted and Electroflux Remelted). The compositions were within limits specified for Inconel 718. Each heat was given a "Conventional" and "Modified" heat treatment - the latter designed to increase the fracture toughness.

The schedules for the "Conventional" and "Modified" heat treatments are as follows:

Conventional heat treatment:

$$955^{\circ}\text{C}/1\text{h}/\text{AC} + 718^{\circ}\text{C}/8\text{h} \xrightarrow[1^{\circ}\text{C}/\text{min}]{\text{FC}} 620^{\circ}\text{C}/18\text{h}/\text{AC}.$$

Modified heat treatment:

$$1093^{\circ}\text{C}/1\text{h} \xrightarrow[1^{\circ}\text{C}/\text{min}]{\text{FC}} 718^{\circ}\text{C}/4\text{h} \xrightarrow[1^{\circ}\text{C}/\text{min}]{\text{FC}} 620^{\circ}\text{C}/16\text{h}/\text{AC}$$

The "Conventional" heat treatment employs a 955°C solution anneal which produces a fine grain size (ASTM 11½ in Heat I and ASTM 5 in Heat II) and some coarse δ-phase precipitates at the grain boundaries. In the "Modified" heat treatment a 1093°C solution anneal is employed which being well above the γ and δ-phase solvus temperatures for this alloy leads to a coarse grain size (ASTM 4 in Heat I and ASTM 2½ in Heat II) and significantly reduced the amount of δ-phase as well as Laves phase present. The microstructures described here are shown in Figs. 4.24 and 4.25.

It may be seen that conventionally heat treated Heat II material contained banded regions in which smaller grains and significant grain boundary δ-phase were present. This difference in microstructure between the two heats was attributed by Mills and James⁴¹ to differences in melting practice.

FCGR tests results for these materials are shown in Figs. 4.26 and 4.27. Clearly, for the case of the conventional heat treatment, the Heat II material which has a coarser average grain size

exhibits a lower FCGR. In response to the Modified heat treatment, the FCGR of each heat increased with respect to the conventional and the properties of each heat tended to approach one another. The better FCGR resistance due to the Modified heat treatment is obviously due to the coarser grain size, as well. The lack of heat-to-heat variation in materials given the Modified heat treatment was attributed by these investigators to a more homogeneous microstructure, though no rigorous explanation was given. In this study, no detailed observations on either precipitate morphology or nature of plastic deformation were made.

Pedron and Pineau⁴² investigated three microstructures of Inconel 718, namely, a coarse grained microstructure (ASTM 3-4), a fine grained microstructure (ASTM 6-8) and a necklace structure (ASTM 3-4 and 8-10). Their results for high frequency fatigue at 20 Hz and also for a 5 min hold time are shown in Fig. 4.28.

It is clear that at high frequency, due to the minimization of environmental effects very little difference in crack growth rates is observed between the various microstructures. However, when a 5 min hold time is introduced, dramatic differences are observed particularly at low values of ΔK . The material of the necklace structure shows the best performance while the worst situation is encountered with the fine grained material.

In Fig. 4.28, S-L refers to specimens with the notch parallel to the rolling direction. Quite a difference in behaviour is noticed in crack growth directions parallel (S-L) and perpendicular (L-S) to

the rolling direction. In particular, it is important to note that there is very little difference between the FCGRs of the coarse grained and fine grained materials, when in both cases the crack growth direction is parallel to the rolling direction.

Sadananda and Shahinian⁴³ investigated the "Conventional" and "Modified" heat treatments mentioned earlier for Inconel 718 and another treatment called the Wilson heat treatment, as well. This latter treatment leads to an overaged microstructure with coarse γ and significant amount of δ -phase needles at the grain boundaries. The grain size is fine, and is similar to that of the conventional heat treatment.

It was found that the overaged microstructure of the Wilson heat treatment led to dramatic improvements in FCGR resistance at 650°C for long hold periods of 10 min or more (Fig. 4.29). More than two orders of magnitude of difference in FCGR were noticed between the Wilson heat treatment and the other two, though the fracture mode was intergranular in all three cases. The Wilson heat treatment proved superior for shorter hold times of 1 min as well at 650°C (Fig. 4.30). However, at 540°C, the Wilson heat treatment gave rise to higher crack growth rates than for the conventional heat treatment when 1 min hold periods were used (Fig. 4.31). At zero hold very little difference was noticed between the two heat treatments.

These results can be explained once again on the basis of concepts mentioned earlier. The overaged microstructure of the Wilson heat treatment leads to minimal sensitivity to environmental

embrittlement since in this material dislocations loop around coarse precipitates in homogenous fashion rather than cutting the particles and being confined to move in planar arrays. Further, the significant amount of σ -phase present at the grain boundaries gives rise to an irregular or "serrated" grain boundary morphology and minimizes the extent of grain boundary sliding. Hence, the superior FCGR resistance of material given the Wilson heat treatment at 650°C.

At the lower temperature of 540°C, the level of environmental embrittlement is much lower and so is the contribution of grain boundary sliding to deformation. At this temperature better FCGR resistance is obtained with planar slip which is pronounced in the case of the conventional heat treatment due to the very fine γ " precipitate particles present. The creep strength of the material given the conventional heat treatment is also likely to be higher. As a result, the Wilson heat treatment may actually be somewhat inferior to the Conventional treatment at these lower temperatures.

The superior FCGR resistance exhibited by the Wilson heat treatment at 650°C with respect to the conventionally heat treated material has been confirmed by Floreen and Kane⁴⁴ over a frequency range of 0.01 Hz to 1.0 Hz.

4.4.3 Environmental Effects

Oxidizing Environments: Figure 4.32, taken from the work of Pedron and Pineau⁴² shows the results of FCGR tests carried out in air and under vacuum for fine grained Inconel 718 at 650°C. At high frequencies, the increase in FCGR observed in air does not occur under vacuum. When a 5

min hold time is included, the FCGR is increased under vacuum, but this increase is much smaller than that observed in the air tests. These results clearly demonstrate the damaging effect of the air environmental on the FCGR of Inconel 718 at 650°C.

Sadananda and Shahinian⁴⁵ have also attempted to determine the relative contributions of creep and environment to the time dependent component of fatigue crack growth in Inconel 718 at 650°C. For this purpose, the increase in da/dN due to creep is taken to be the ratio of the hold time da/dN to the continuous cycling da/dN in vacuum. The increase in da/dN due to the environment is taken to be the ratio of the da/dN in air to the da/dN in vacuum. It is assumed here that no significant creep effect is present under continuous cycling, which appears to be the case in most cases for this alloy. The variation of the relative contributions of the two effects is shown in Fig. 4.33. The contribution from creep increases with increase in ΔK and the slope of the curve is higher for the longer hold time. At the low ΔK end, an increase in the hold period from 1 min to 10 min increased the creep component by a factor of 10 while at high ΔK values, the increase in the creep component was even greater and this is thought to be due to a synergistic effect between creep and fatigue.

While the effect due to creep rose with increase in ΔK , that due to environment steeply declined. At the low ΔK end (30 MPa $\sqrt{m}$) the contribution to crack growth from the effect of environment was appreciably larger than from the creep effect but this effect was reduced with increase in hold time and ΔK . In fact, a cross-over is

present such that creep becomes the major contributor to the time dependent component at high ΔK levels. This cross-over point at which the creep effect becomes larger than the environmental effect shifts to a lower ΔK as hold time is increased.

Another important observation here is that environmental effects are more pronounced at low ΔK values than at high ΔK levels. At high ΔK levels, the crack propagation rates could be quite rapid and these may not be sufficient time for environmental embrittlement to occur ahead of the crack tip.

In an air environment, oxygen has clearly been shown to be the main aggressive species (Fig. 4.34) responsible for the increased crack growth rates.⁴⁵ However, the mechanism by which oxygen enhances fatigue crack growth rates is a matter of conjecture at this time.

The supply of a reactive gas to the crack tip is limited by molecular transport or surface diffusion along the crack. The rate of supply can be insufficient to affect crack tip fracture mechanisms, if the rate of formation of fresh surfaces is too rapid for formation of a passivating film or for adsorption and diffusion effects to take place. The crack retardation effects appear to be plugging of the crack by the reaction products, formation of a ductile film at the crack tip, or blunting of the crack tip, either through uniform chemical attack or by restricting crack-tip closure and resharping through volumetric expansion of the reaction product.^{46,47}

In the transgranular mode of fracture, an oxygen containing environment can exert a significant deleterious effect by diffusing along slip bands and causing slip-band decohesion.⁴⁶ In the intergranular fracture modes the inability to form a protective film would allow deep penetration of the grain boundaries by the aggressive species and lead to embrittlement. Oxidation of grain boundary particles to produce a loosely compacted reaction product may occur.

In order to understand the fundamental mechanism of environmental embrittlement, it appears necessary to study the formation and properties of films at the crack tip.^{46,47} Clearly, if a protective film can form at the crack tip, the effects of an environment, in the crack tip plastic zone are minimized. Diffusion by any reactive species will decrease with time and crack growth will occur by other mechanisms. The film will be subjected to high strain in either creep or fatigue and thus, its ductility as a function of thickness, structure, temperature and strain rate will determine whether (and how frequently) the film will be fractured, thereby repeating the process of film formation.

While the modelling of these effects is important, it should be noted that crack growth rates in oxidizing environments cannot be predicted using thermodynamic calculations and corrosion data on unstressed specimens. The type of environmental damage being discussed here is a stress-assisted phenomenon and is much more severe than general oxidation or corrosion behaviour.

Carburizing Environments: While oxygen containing environments can substantially increase FCGR with respect to vacuum, carburizing environments can be harmful as well. The effect of carburizing environments on the FCGR in Nimonic 115 at intermediate temperatures has been assessed by Floreen and White.⁴⁸ Tests were carried out at temperatures ranging from 650 to 870°C and frequencies of 0.01 to 1.0 Hz. Several He + CH₄ and H₂ + CH₄ environments were also studied.

Carburizing environments were found to increase crack growth rates as in the case of an oxidizing environment. However much higher concentrations of methane (than oxygen) were required to produce a signified change in FCGR. For example, a number of earlier studies⁴⁹⁻⁵¹ of nickel and stainless steel have demonstrated that oxygen pressures as low as 10⁻² to 10⁻³ Torr produced fatigue behaviour comparable to tests at atmospheric pressures. In Inconel 718, addition of 100 ppm of oxygen to a high purity helium environmental was found to produce crack growth rates close to that obtained in air. The results of Floreen and White,⁴⁸ however, show that several pct methane or a partial pressure of the order of 10 Torr was needed to substantially increase the crack growth rate.

Fatigue crack growth rates at 0.1 Hz at 760°C in several He + CH₄ and H₂ + CH₄ environments are shown in Fig. 4.35. At intermediate methane concentrations, e.g. He + 2 pct CH₄ and H₂ + 1 pct CH₄ or H₂ + 2 pct CH₄, the environmental effects were more noticeable at higher stress intensities. This is the opposite of the behaviour noticed in oxygen environments, since as mentioned earlier the crack

growth rates in Inconel 718 were found to have a greater environmental contribution at low stress intensity factors.

The crack paths in pure hydrogen and in all the $H_2 + CH_4$ environments were transgranular. In the $He + CH_4$ environments the crack paths were intergranular, particularly at 870°C. Samples tested in the $He + CH_4$ environments had a deposit of carbon soot on their surfaces after testing while this was not observed in the $H_2 + CH_4$ test samples.

Figure 4.36 shows the effect of methane content of the gas on the crack growth rates at 33 MPa $\sqrt{m}$. The crack growth rates in pure hydrogen were significantly greater than in pure helium and the data shown in Fig. 4.36, suggests that about half the increase in crack growth rates seen in the $H_2 + 4$ pct CH_4 environment can be attributed to hydrogen. In terms of the effectiveness of the methane per se, the helium methane environments appeared to produce a more potent effect than the hydrogen methane mixtures.

The effect of cyclic frequency in helium and $H_2 + 4$ pct CH_4 environments 760°C are shown in Fig. 4.37. As usual, slower frequencies significantly increased crack growth rates. The crack paths were transgranular in both environments at 0.1 and 1.0 Hz, and intergranular in both environments at 0.01 Hz. This data was also cross plotted as shown in Fig. 4.38, for a ΔK of 33 MPa $\sqrt{m}$. It was concluded that the largest environmental contribution occurred at 0.1 Hz. Interestingly enough, at a lower frequency of 0.01 Hz the environmental contribution appeared to become quite small. This

contrary to the usual experience in oxidizing environments of increasing sensitivity with decreasing frequency.

No explanation has been offered by this anomalous result of a maximum in environmental sensitivity at an intermediate frequency. However, it is felt that such interpretations need to be viewed with some caution. On a logarithmic plot, the difference in da/dN between say, $1.0 \cdot 10^{-3}$ m/cycle and $1.1 \cdot 10^{-3}$ m/cycle may appear quite small, but in terms of the absolute values of the crack lengths involved this may be just as damaging an effect was the difference between say, $1.0 \cdot 10^{-4}$ m/cycle and $2.0 \cdot 10^{-4}$ m/cycle. The separation between the latter two values will appear to be much greater on a logarithmic plot and this may be misleading.

The mechanisms by which H_2 , CH_4 or combinations of these gases lead to an increase in crack growth rates (with respect to vacuum) is not clear. These investigators have not made a detailed analysis of this aspect except noting that there was not any detectable carburization damage ahead of the crack tips.

Sulphidizing Environments: These environments, H_2S in particular have been shown to have a very damaging effect on FCGR resistance, even more damaging than oxygen. The results of Floreen and Kane⁵² for Inconel 718, at a test temperature of 650°C are shown in Fig. 4.39. Helium + 0.5 pct H_2S and helium + 5 pct SO_2 environments have been considered.

The crack growth data for the same alloy and temperature in a air + 5 pct SO_2 environment are shown in Fig. 4.40. At low stress intensity ranges, it may be seen that the FCGR in air + 5 pct SO_2 is

significantly less than in He + 5 pct SO₂ Fig. 4.41. Thus, air appears to help retard the effect of the more aggressive SO₂ environment.

The severity of attack in these sulphidizing environments led to the conclusion that a nickel-sulfur eutectic formed during these tests. The published Ni-Ni₃S₂ eutectic temperature of 647°C is slightly below the test temperature used.⁵³ The mitigating effect of air was thought to be due to oxidation inhibiting sulphide formation. Such protection from sulphur bearing species should be most effective at lower stress intensity ranges where mechanical damage to the surface or grain boundary regions would be less severe. At higher ΔK ranges, mechanical disruption would minimize the protection effects of oxygen as shown in Fig. 4.41.

Figure 4.41 also shows the effect of the R-ratio upon crack growth behaviour in Inconel 718 in a sulphidizing environment. At low stress intensity ranges, the formation of a hard, dense sulphide scale at the crack tip led to crack tip blunting and led to a considerable decrease in the effective ΔK such that crack arrest occurred after a while. Increasing the R-ratio from 0.1 to 0.2 so that the minimum load is increased led to extension of the useful stress intensity range, confirming that crack closure was responsible for crack arrest at lower ΔK values.

Floreen⁵⁴ compared the behaviour of Inconel X-750 with various grain boundary carbide microstructures in He, Air, He + 4 pct CH₄ and He + 3 pct SO₂ environments, the results of which are given in Table 4.2. It may be seen from these results that the treatments which

gave rise to the lowest crack growth rates in He and He + 4 pct CH₄ had the highest crack growth rates in air and in He + 3 pct SO₂. It was therefore concluded that the optimum grain boundary carbide morphology changes with the type of environment. This result also indicates that the mode of embrittlement in each one of these environments may be quite different, giving rise to a variety of observed effects.

Water Vapour Containing Environments⁵²: The effect of helium + 5 pct water vapour upon the FCGR of Inconel 718 at 650°C is shown in Fig. 4.42. This environment produces crack growth rates intermediate between air and helium environments.

In Fig. 4.43, the results for H₂ + 5 pct water vapour and air + 5 pct water vapour are shown. The hydrogen + 5 pct water vapour results were once again intermediate between the values obtained for air and helium. The air + 5 pct water vapour data appears identical to that for air. At higher temperatures, however, the presence of water vapour can accelerate the oxidation of Ni-base superalloys.

The increased crack growth rates in the presence of water vapour may be due to the reaction of water vapour with the alloy to liberate oxygen which then causes embrittlement. This is only a matter of speculation at this time and further work is needed before any specific conclusions are drawn in this report.

In summary, the FCGR of Ni-base superalloys is accelerated in the presence of a reactive environmental species, some of which may produce a more dramatic effect than others. Based on the above mentioned results, three categories of environmental aggressions can be distinguished as shown below.⁵⁴

Very Aggressive	Slightly Aggressive	Non-Aggressive
H ₂ S	CH ₄	N ₂
SO ₂	H ₂	He
O ₂	CO ₂	
H ₂ O		

There is clear evidence that the processes leading to environmental embrittlement are not unique and can very significantly depending upon microstructure, alloy composition, temperature and loading conditions. Very little is known at present, particularly in the area of microstructure-environment interactions and further work is encouraged in this direction.

4.4.4 Frequency, Hold-Time and Wave Shape Effects

The effects of frequency and hold-time have been referred to several times earlier, in connection with microstructural and environmental effects. In summary, the effect of lowering frequency is generally an increase in FCGR, primarily due to environmental effects at low ΔK values and primarily due to creep effects at high ΔK values. The effect of hold-time is clearly to increase FCGR and the extent to which this increase occurs will be proportional to the duration of the dwell-period. Short hold-times may introduce primarily environmental contributions while longer dwell-periods may allow sufficient time for creep effects to be significant. It is important to note that hold-times can in some cases be beneficial, if the hold period is at a ΔK value, just below the threshold for creep crack growth, crack tip

blunting can occur due to stress relaxation ahead of the crack tip, and a retardation in crack growth rate may be observed.

These results have been explained in detail in earlier sections and hence the following discussion given by Antolovich and Campbell⁴⁰ pertains mainly to the effects of waveshape.

The most significant study of the effect of waveform on FCGR has been carried out by Clavel and Pineau⁵⁵ who used triangular and square waveshapes as shown in Fig. 4.44a. The alloy was Inconel 718 and the test temperature was 550°C. The data obtained using the triangular wave at 2 Hz were the same as the data obtained in other tests using the sine wave at the same frequency. These two waveforms produced the lowest FCGR in this investigation. The effect of loading at the same rate as the triangular wave at 2 Hz but imposing a 10-second hold-time at maximum load was to increase FCGR only slightly as shown in Fig. 4.44b. The deformation mode was similar in both cases, consisting of homogeneously distributed faults in the γ phase. Thus, the small increase in FCGR appears to be clearly associated with a creep effect. For the low-frequency triangular wave, the amount of intergranular fracture increased and there were bands of intense deformation which were identified as twins. In addition, there were planar faults between the intense deformation bands, the density of which decreased with decreasing frequency. The FCGR for the low frequency triangular wave seemed to be significantly higher than for the case of the 10-second hold period.

The conclusions drawn from this were that the change in deformation mode is more sensitive to strain rate than to frequency per se, and that changes in deformation mode can be associated with large changes in FCGR. The major difference between the effects of the square wave and the low frequency triangular wave (which have the same frequency) was therefore associated with a change in deformation mode and creep was not thought to be a major contributor. The greater amount of twinning which occurs at lower strain rates was thought to promote environmental embrittlement since the fracture mode changed to intergranular.

While it appears to be clear from the above study that the strain rate has an important influence on the deformation mode in Inconel 718, it should also be noted from Fig. 4.44a that the low frequency triangular wave and the square wave (with a 10-s hold time) used by Clavel and Pineau⁵⁵ are not entirely equivalent in terms of the time for which the material is subjected to a significant tensile load. The square wave gave rise to a tensile stress exposure for approximately 10 seconds while the triangular wave exposes the material to a significant tensile load for nearly 20 seconds, though not at the same value always. This being the case it is difficult to categorically assume that the higher FCGR for the case of the low frequency triangular wave is not due to an increased creep contribution or due to an increased period of exposure to environmental attack - but mainly due to a strain rate effect.

4.5 Life Prediction Under Elevated Temperature Fatigue

Components in aircraft engines operate under a combination of complex loading and temperature cycling which makes it extremely difficult to predict a safe life. This is especially difficult because the fatigue and creep contributions may not be additive, even if it proved possible to distinguish between these simultaneously occurring effects.

This problem can be approached in two ways. The first approach is to calculate component life based on available low temperature fatigue life relations. Time and temperature effects are incorporated in those relations to account for operation at elevated temperatures or to account for thermal cycling conditions. The other approach is based on available creep life equations which are adopted to take into account the effect of cyclic loading conditions. The fatigue approach works best in fatigue dominated conditions where time dependent effects are less significant. The creep approach is most appropriate for creep dominated conditions where life is strongly dependent on time at temperature and failure occurs by intergranular void formation and linkage.

4.5.1 Extension of Low Temperature Fatigue Life Relations

As mentioned in Chapter 1, the first life prediction method in the area of low temperature LCF was the Manson-Coffin law relating the inelastic strain range ($\Delta \epsilon_{in}$) to the number of cycles to failure,

N_f

$$\Delta \epsilon_{in} = C \cdot N_f$$

(4.1)

In this equation, C is a constant directly related to the tensile ductility of the material.

Later on, Manson⁵⁶ proposed a more general equation called the Method of Universal Slopes. This technique was also described earlier in Chapter 1. Manson and Halford⁵⁶ have tried to modify the universal slopes equation to extend its use to elevated temperatures. A "10 percent rule" was proposed according to which the number of cycles predicted by the universal slopes equation (calculated for the actual temperature) should be reduced by a factor of 10 to account for creep and oxidation effects. Although this approximation was shown to be effective over a range of temperatures, it overestimated life especially at very high temperatures and also at low frequency or when hold time effects were considered.

Harrison and Cockcroft⁵⁷ introduced a further modification based on creep rupture data as follows:

$$N'_f = \frac{N_f}{1 + \frac{K}{FA}} \quad (4.2)$$

where, N'_f = revised number of cycles in relation to 10 percent rule

K = fraction of cycle time at peak load

F = frequency

A = time to rupture on a stress rupture test.

Coffin, later extended the Manson-Coffin relation (eqn. 4.1) into an elevated temperature form by introducing a frequency term, resulting in what is generally known as the "Frequency Modified Fatigue Relation"⁵⁸ written as follows:

$$\Delta \epsilon_t = \Delta \epsilon_p + \Delta \epsilon_e = C_2 (N_f \cdot v^{k-1})^{-\beta} + \frac{AC_2^n}{E} \cdot N_f^{-\beta n} \cdot v^{k_1} \quad (4.3)$$

where, v = frequency

β = slope of plastic curve, no hold time

n = cyclic strain hardening exponent

C_2 = constant in Manson-Coffin Law

A, k, k_1, k'_1 = material constants ($k'_1 = -\beta n(k-1) + k_1$)

This equation transforms into the universal slopes equation if the frequency effects are eliminated, i.e. if $k = 1$, $k'_1 = 0$.

The Coffin frequency-modified relation can be written in a "creep" form⁵⁹ by replacing $\frac{N_f}{v}$ by the time to failure t_f .

$$\Delta \epsilon_t = C_2 \cdot t_f^{-\beta} \cdot v^{-k\beta} + \frac{AC_2^n}{E} \cdot t_f^{-\beta n} \cdot v^{k'_1 - \beta n} \quad (4.4)$$

For hold-time testing the frequency can be replaced by the inverse of the hold-time τ , as follows,

$$\Delta \epsilon_t = C_2 \cdot t_f^{-\beta} \cdot \tau^{k\beta} + \frac{AC_2^n}{E} \cdot t_f^{-\beta n} \cdot \tau^{\beta n - k'_1} \quad (4.5)$$

The above-mentioned frequency modified relationship have found considerable application especially under those conditions which resemble those for which the constants were calculated. However, a considerable amount of input data is required, since for each temperature, at least six coefficients should be determined.

4.5.2 Linear Damage Summation

Also called time and cycle fraction summation, this is another well known method of life prediction. Originally proposed by Taira,⁶⁰ it is written in the following form:

$$\sum_{j=1}^p \frac{n_j}{N_j} + \sum_{k=1}^q \frac{t_k}{T_k} \leq D \quad (4.6)$$

where, n_j = number of cycles at load level j ; p = number of load levels

N_j = number of cycles to failure at load level j

t_k = time duration at load level k

T_k = creep rupture time at load level k

D = total damage factor.

In this approach it is assumed that the damage fraction occurring in fatigue (the cycle fraction) and the damage fraction occurring in creep (time fraction) may be simply added up. However, experience indicates that this is not necessarily true. Further, there are serious limitations with regard to compressive loading⁶¹ and hold-times.⁶² For example, the method does not take into account the stress relaxation that can occur during a hold-time period.

4.5.3 Non-Linear Damage Summation (Ductility Exhaustion)

Polhemus⁶³ developed a model in which the total initial available ductility is gradually exhausted due to the accumulated damage contributions of cyclic loading and creep. In this model, the damage occurring in every cycle is converted into an equivalent creep time. This equivalent creep time is added to the actual creep time in the cycle. This procedure therefore calculates at any cycle the position along the static creep curve. This stepwise progression is followed until failure occurs, which is the case when the remaining ductility is zero - that is, when the total strain equals the fracture strain of the creep rupture curve.

4.5.4 Damage Rate Method

This is a strain based damage rate approach developed by Majumdar and Maiya.^{64,65} In its simplest form, failure was said to occur when pre-existing cracks of length a_0 grew to a critical length a_f , one or more of these cracks linking to form a macrocrack. The growth of each microcrack was assumed to be governed by the equations:

$$\frac{da}{dt} = a \left[\frac{T}{C} \right] [\epsilon_p]^m [\dot{\epsilon}_p]^k \quad (4.7)$$

where T , C , m and k are material parameters which are functions of temperature, environment and the metallurgical state of the material. T and C are included to account for differences that can occur in growth rates under tension and compression.

For continuous cycling over a plastic strain range $\Delta\epsilon_p$ at constant plastic strain rate $\dot{\epsilon}_p$, eqn. 4.7 can be integrated to give the cycles to failure, viz:

$$N_f = \left[\frac{(m+1)}{4A} \right] \left[\frac{\Delta\epsilon_p}{2} \right]^{-(m+1)} [\dot{\epsilon}_p]^{1-k} \quad (4.8)$$

where,

$$A = \frac{(T+C)}{2} \ln(a_f/a_0)$$

An extension of the original approach was subsequently proposed by Majumdar and Maiya⁶⁶ in which damage was considered to be made up of crack damage (fatigue) and cavity damage (creep) which accumulate independently. The rate of fatigue damage is still given by eqn. 4.7, but in addition, cavity growth rates are assumed to be given by:

$$\frac{1}{c} \frac{dc}{dt} = \left(\frac{G}{-G} \right) |\epsilon_p|^m |\dot{\epsilon}_p|^{k_c} \quad (4.9)$$

where c is the cavity size at time t , and G and k_c are material parameters. The parameter G is given the appropriate sign for tensile or compressive stress regimes. Final failure is calculated as the reciprocal of the sum of the crack and cavity damage.

More recently, an interactive damage equation has been proposed by Majumdar and Malya in which fatigue crack growth rate is assumed to be increased by the presence of cavities, the cracks and cavities are assumed to grow according to the following interactive damage equations⁶⁷:

$$\frac{1}{a} \frac{da}{dt} = \left(\frac{T}{C} \right) \left[1 + \alpha \ln \frac{c}{c_0} \right] |\epsilon_p|^m |\dot{\epsilon}_p|^k \quad (4.10)$$

$$\frac{1}{c} \frac{dc}{dt} = \left(\frac{G}{-G} \right) |\epsilon_p|^m |\dot{\epsilon}_p|^{k_c} \quad (4.11)$$

where the material parameters T , C , m , K , G and K_c are functions of temperature, environment and microstructure of the material. Similarly α is a material parameter which is zero for continuous cycling or loadings with compressive imbalance, when cavities are assumed not to initiate.

This approach is attractive in that it can model the "healing" effect of a compressive dwell on a previous tensile dwell. However, values for many coefficients in the lifetime equations must be obtained before the method can be used. Indeed, evaluation of some of these coefficients depends upon the results of complex cyclic-hold tests and so their generation is both expensive and time-consuming.⁶⁷

4.5.5 Strain-range Partitioning (SRP)

This approach was developed by Manson and co-workers^{68,69} in an attempt to take account of the damage caused by different types of strain. It was proposed that the inelastic strain accumulated during one high temperature cycle could be partitioned into four possible kinds of strain range, depending upon the direction of straining (tension or compression) and the type of inelastic strain accumulated (creep or time-independent plasticity). Thus, any reversed inelastic strain range could be partitioned into a combination of the following components: $\Delta\epsilon_{pp}$, $\Delta\epsilon_{cp}$, $\Delta\epsilon_{pc}$ and $\Delta\epsilon_{cc}$. Here, for example, $\Delta\epsilon_{cp}$ represents tensile creep strain reversed by compressive plastic strain.

Figure 4.45 shows the idealized hysteresis loops for the four generic types of strain ranges. For these, the basic life relationships can be written as,

$$\begin{aligned} N_{pp} &= A(\Delta\epsilon_{pp})^{\alpha_1} \\ N_{cp} &= A(\Delta\epsilon_{cp})^{\alpha_2} \\ N_{pc} &= A(\Delta\epsilon_{pc})^{\alpha_3} \\ N_{cc} &= A(\Delta\epsilon_{cc})^{\alpha_4} \end{aligned} \quad (4.12)$$

where the coefficients A and exponents α are experimentally determined material constants.

For actual cycles, a cumulative damage rule is required to account for the different types of strain occurring. Assuming that the inelastic strain in the loop is $\Delta\epsilon_{in}$, the component fractions are:

$$\begin{aligned} F_{pp} &= \frac{\Delta \epsilon_{pp}}{\Delta \epsilon_{in}}, & F_{cp} &= \frac{\Delta \epsilon_{cp}}{\Delta \epsilon_{in}} \\ F_{cc} &= \frac{\Delta \epsilon_{cc}}{\Delta \epsilon_{in}}, & F_{pc} &= \frac{\Delta \epsilon_{pc}}{\Delta \epsilon_{in}} \end{aligned} \quad (4.13)$$

The predicted lives $N_{f(pred)}$ can then be obtained from the following interactive damage rule

$$\frac{1}{N_{f(pred)}} = \frac{F_{pp}}{N_{pp}} + \frac{F_{cc}}{N_{cc}} + \frac{F_{cp}}{N_{cp}} \quad \left[\text{or } \frac{F_{pc}}{N_{pc}} \right] \quad (4.14)$$

Though much data have been presented which apply the SRP approach to a range of materials, the approach has yielded conflicting evidence for and against the method.⁶⁷ Problems with the SRP approach are associated with the complex cyclic-hold tests that must be performed to generate the base-line data. Also, the SRP method, in its present formulation is not suitable to predict the effect of mean stress on life time or to be used to correlate crack propagation. Thus, the applicability of the method varies markedly with the material and test conditions.

Several investigators have found it difficult to partition inelastic strain. This is especially so with primary creep on load reversal and very small time-dependent strain components. The interpolation or extrapolation of test results requires either a time-dependent, cyclic constitutive model for the material or an experimental simulation of the structural response with a test specimen. At the strain ranges encountered in turbine discs, the inelastic components are very small of the order of 10^{-4} and so, and hence accurate computation or measurement is very difficult. The SRP

method is very sensitive to the time-dependent strain computation and this is regarded as a serious impediment to its application to turbine discs.

In summary, a variety of methods are available to predict life under creep-fatigue conditions. Mechanistic approaches such as the damage rate method seem promising in terms of accuracy but are difficult to apply. Simpler methods are required which can be used with an acceptable level of accuracy.

However, progress is being made in this direction. In the past few years, the original picture of creep-fatigue, which was based on interaction between a growing fatigue crack and creep damage has now changed. The initial picture evolved due to tests being conducted at short hold-times and high strain ranges on creep ductile materials. It is now becoming clear that the creep-fatigue interaction regime is only a narrow zone of behaviour, dividing the creep and fatigue dominated regimes. This realization is leading to simpler predictive relationships based on a ductility exhaustion concept.⁶⁷

It is felt that there is much knowledge to be tapped from mechanistic studies of creep-fatigue cracking towards developing simple but efficient predictive relationships. More emphasis is needed towards taking microstructural parameters into account. For example, grain size has been shown to be a primary variable affecting elevated temperature crack growth and its influence shown to dominate over many other microstructural variables. A relationship combining this parameter with creep ductility (which could account for environmental

as well as grain boundary morphology effects) and strain rate (which could account for frequency, hold-time and waveshape effects) may prove to be quite powerful as a predictive tool. Further work is required towards developing relationships such as this, which approach the problem not from a purely phenomenological sense, but with a clear understanding of the metallurgical processes associated with failure in elevated temperature fatigue.

Chapter 4

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4.1 Influence of Frequency and Heat Treatment Upon LCF Life of A286 at Elevated Temperature

4.2 Model of Grain Size Effects on Intergranular Crack Propagation

4.3 Influence of Directional Solidification or Single Crystal Fabrication Upon LCF Life of Mar-M200

4.4 Influence of Frequency Upon Fatigue Life: Schematic⁵

4.5 Influence of Tensile Hold Time on Fatigue Life of AISI 304 Stainless Steel at 650°C²¹

4.6 Influence of Hold-Periods Upon Fatigue Life of Cast René 80²²

4.7 Explanation of the Effect of Tensile and Compressive Hold-Periods⁵

4.8 Influence of Compressive and Tensile Hold Times Upon Fatigue Life of Inconel 718²⁴

4.9

Wave Shapes Investigated by Coffin⁵

4.10

Influence of Wave Shape Upon LCF Life in A-286

at 593°C⁵

4.11

Mechanism of Cavity Growth Under the Slow-Fast

Mode⁵

4.12 Mechanism of Cavity Collapse Under the Fast-Slow Mode⁵

4.13 Absence of Cavity Growth Under the Equal Ramp Rate Mode⁵

4.14 Effect of Plastic Strain Range Upon LCF Life in Inconel 718 at Various Temperatures²⁴

4.15 Influence of Strain Amplitude Upon LCF Life in Inconel 718²⁸

4.16 Deformation Modes in Inconel 718 and Waspaloy During LCF Tests and Fracture Mechanics Tests²⁹

4.17 Applicability of the J-Integral for Correlating
Fatigue Crack Growth Rates in Udimet 700 at 850°C

ALLOY 718
850 C
0.17 Hz

ALLOY 718
850 C
0.17 Hz

4.18 Fatigue Crack Growth Rates in Inconel 718 Plotted as a Function of ΔK and ΔJ ⁴⁵

4.19(a) Fatigue Crack Growth Rates in Inconel 718 Plotted as a Function of K_{max} and J_{max} ⁴⁵

4.19(b) Fatigue Crack Growth Rates in Inconel 718:

da/dt vs K_{max} and J_{max}

4.20 Fatigue Crack Growth Rates in Udimet 700

Represented in Terms of ΔJ for Various Hold

Times³⁶

4.21 Effect of γ' Particle Size on FCGR in Waspaloy
at 650°C³⁷

4.22 Influence of Grain Size on FCGR in Astroloy³⁷

$\Delta K = 30$ to $60 \text{ MPa}\cdot\text{m}^{1/2}$ (27 to $60 \text{ ksi}\cdot\text{in.}^{1/2}$). Temperature, 540°C (1000°F). Frequency, 0.3 Hz .

4.23 Influence of Grain Size and γ' Particle Size Upon FCGR in René 95³⁹

200 μm

(a)

200 μm

(b)

9 μm

(c)

9 μm

(d)

4.24 Microstructures of Inconel 718 (Heat 1)
 (a) and (b) after Conventional Heat Treatment
 (c) and (d) after Modified Heat Treatment

200 μm

(a)

9 μm

(b)

200 μm

(c)

5 μm

(d)

- 4.26- Influence of Heat to Heat Variations Upon FCGR in
- 4.27 Inconel 718 Given the Conventional Heat Treatment
and Modified Heat Treatment, Respectively ⁴⁰

4.28 Influence of Microstructure Upon FCGR in Inconel

718⁴²

4.29

Superior FCGR Resistance Exhibited in Inconel 718
Due to the Wilson Heat Treatment at 650°C⁴³

4.30 and Comparison of the FCGR Resistance in Incoel 718

4.31 Due to the Conventional and Wilson Heat Treatments
at 650°C and 540°C, Respectively ⁴³

4.32 Influence of an Oxygen Containing Environment Upon FCGR in Inconel 718 at 650°C⁴²

4.33 Variations in the Creep and Environmental Contributions to FCGR with Stress Intensity Level⁴⁵

4.34 Comparison of the Effects of Oxygen and Nitrogen
Upon FCGR in Inconel 718 at 650°C³¹

4.35 FCGR in Nimonic 115 at Various He + CH₄ and H₂ + CH₄ Environments ^{48;52.}

4.36 Effect of Methane Content of the Environment Upon
FCGR in Nimonic 115 at $\Delta K = 33 \text{ MPa } \sqrt{\text{m}}$ ^{48,52}

4.37 Effect of Cyclic Frequency Upon FCGR in Nimonic
115 in Helium and $\text{H}_2 + 4 \text{ pct. CH}_4$ Environments
^{48,52}
at 760°C.

4.38 Comparison of FCGR at a Stress Intensity Range of 33 MPa $\sqrt{\text{m}}$ vs Frequency ^{48,52}

4.39 Comparison of FCGRs in Inconel 718 in Helium +
0.5 pct. H₂S and Helium + 5 pct. SO₂ Environments
at a Temperature of 650°C.^{44,52}

4.40 Comparison of FCGRs in Inconel 718 at 650°C in Air and Air + 5 pct. SO₂ Environments ^{40,52}

4.41 Influence of R-Ratio Upon Crack Growth Behaviour in Inconel 718 in Sulphidizing Environments ^{45,52}

4.42 The Effect of Helium + 5 pct. H₂O Upon the FCGR in Inconel 718 at 650°C ^{45,52}

4.43

Comparison of the Effect of Air + 5 pct. H_2O
 and He + 5 pct. H_2O Upon FCGR in Inconel 718 at
 650°C^{48, 52}

(a)

(b)

4.44 The Influence of Waveform Upon FCGR in Inconel 718
at -550°C⁴²

(a) Idealized hysteresis loops for the four basic types of inelastic strain range.

(b) Hysteresis loop containing $\Delta \epsilon_{pp}$, $\Delta \epsilon_{cc}$, $\Delta \epsilon_{cp}$

4.45

Idealized Hysteresis Loops for Four Basic Types of Strain Ranges

END

FIN

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