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Rejuvenation of Gas Turbine Discs

by

S. M. R. Pishva
B. Eng. (Mechanical)

**A thesis submitted to the Faculty of Graduate
Studies and Research in partial fulfillment
of the requirements for the degree of**

Master of Engineering

**Department of
Mechanical and Aeronautical Engineering
Carleton University
Ottawa, Ontario
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ISBN 0-315-40635-6

The undersigned recommends to the Faculty of Graduate Studies and Research, acceptance of the thesis: *Rejuvenation of Gas Turbine Discs*, submitted by Seyed Mohammad Reza Pishva, in partial fulfillment of the requirements for the Degree of Master of Engineering.

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Abstract

This research project is part of the National Research Council of Canada's investigation into possibilities of extending the usable life of Canadian Armed Forces gas turbine engine components through a damage tolerance approach or component rejuvenation. The concepts of safe-life and damage tolerance are explained as applied to gas turbine discs.

The microstructural damage suffered by Inconel 718 turbine discs during service and their influence on fatigue crack growth rates (FCGR) at 650°C are investigated. The possibilities of rejuvenation of FCGR properties, by reheat treatment of Inconel 718 and A-286 turbine discs from the J85-CAN-15 and J85-CAN-40 engines, respectively, are explored. The effects of service exposure and rejuvenation treatment on dimensional stability of the turbine discs are also investigated.

An electrical direct current potential drop technique was successfully adapted, calibrated and implemented at temperatures up to 650°C in this work, for crack length measurement. Two calibration curves were obtained for compact tension specimens with different thickness and potential probe locations.

Acknowledgements

The majority of the work for this research project was performed at the Structures and Materials Laboratory of the National Aeronautical Establishment, National Research Council of Canada. There were many people who helped in bringing this thesis to fruition. Many have helped in small things that I cannot recall. But I take this opportunity to thank people who made major contributions that I do recall.

I wish to express my sincere gratitude to Dr. A. K. Koul for his constant help and guidance concerning this research. Mr. N. C. Bellinger deserves credit for his helpful contributions in all aspects of the project. I wish to thank the technical staff of the Structures and Materials Laboratory for their kind help and technical assistance. In particular, the following people took extra time on my behalf to assist me in the work performed. These people are: Mr. R. Dainty, Mr. T. Terada, Mr. D. Morphy, Mr. D. Chow, Mr. P. Adams, Mr. P. Au, Ms. S. Taylor, Ms. C. Dionne, Mr. D. Raizenne, Mr. C. Mackenzie, Mr. R. Bracket, Mr. A. Luteyen, Mr. M. Edkins, and Mr. B. Wells. I would also like to thank the staff of the NAE-M.E. Library.

The interest, critical assessment and support shown during the seminars and presentations at NRC and Orenda, by Dr. W. Wallace, the Head of the Structures and Materials Laboratory, Dr. J-P. Immarigeon, the Head of the Materials Section and Dr. R. Thamburaj of the Orenda Division of Hawker-Siddeley Canada, are greatly appreciated.

Thanks are also due to Professor G. L'Espérance of École Polytechnique for the technical assistance in the interpretation of the TEM results. A special note of thanks to my supervisor, Professor M. C. de Malherbe, who provided the opportunity to work in this field. His interest, concern, encouragement, and advice are much appreciated.

I am also forever indebted to my wife, Hope, who understood the demands placed on my time and helped in every way that she could. A final note of thanks to my parents for their encouragement, understanding, and prayers.

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Nomenclature

a	crack length
a/w	nondimensional crack length
A	Ampere, the unit of current strength
B	compact tension specimen thickness
C	Paris law equation coefficient
$^{\circ}C$	degrees Centigrade
hr	hour
hrs	hours
Hz	Hertz
I_i	current Input lead location
K	stress intensity factor
k	kilo ($\times 10^3$)
kN	kilo Newton(s)
kV	kilo Volt(s)
L	Load
lb	Pound ($=4.448$ Newton)
mm	millimetre
MPa	MegaPascal ($10^6 N/m^2$)
n	Paris law equation exponent
N	Number of fatigue cycle
O_i	current Output lead location
P_j	Potential probe locations
R	load or stress Ratio
R_C	hardness in Rockwell 'C' unit
V	Volt(s)
V_{on}	Voltage measured during current on
V_{off}	Voltage measured during current off
w	width of the compact tension specimen as defined by ASTM standard E-647
X_i	DC-PD calibration polynomial coefficients
α	non dimensional crack length, a/w
γ', γ''	strengthening phases
δ	Ni_3Nb phase in Inconel 718
ΔK	stress intensity factor range
η	needle like phase in A-286
μ	micro ($\times 10^{-6}$)
ρ	specific electrical resistivity
σ_{net}	net section stress

Acronym

AC	Alternating Current
AGARD	Advisory Group for Aerospace Research and Development
ASIP	Aircraft Structural Integrity Program
ASTM	American Society for Testing and Materials
APU	Auxiliary Power Unit
BCT	Body Centre Tetragonal
CA	Combined Analysis
CF	Canadian Forces
CIP	Component Improvement Programme
COD	Crack Opening Displacement
CT	Compact Tension
DC	Direct Current
DC-PD	Direct Current Potential Drop
DFM	Deterministic Fracture Mechanics
DND	Department of National Defence
ENSIP	ENGINE Structural Integrity Program
FCC	Face Centre Cubic
FCGR	Fatigue Crack Growth Rate
FCGR's	Fatigue Crack Growth Rates
Fe-Ni-base	Iron-Nickel-base
FOD	Foreign Object Damage
HCP	Hexagonal Close Packed
HIP	Hot Isostatic Press
LCF	Low Cycle Fatigue
LEFM	Linear Elastic Fracture Mechanics
LOC	Life On Condition
LPI	Liquid Penetrant Inspection
MSF	Mission Severity Factor
MTS	Materials Testing System
NAE	National Aeronautical Establishment
NDI	Non Destructive Inspection
NRC	National Research Council
PD	Potential Drop

RFC	Retirement For Cause
RHT	ReHeat Treatment
RPM, rpm	Revolution Per Minute
SEM	Scanning Electron Microscope
S.D.	Standard Deviation
SII	Safe Inspection Interval
TEM	Transmission Electron Microscope
TF	Thermal Fatigue
TSN	Time Since New
TSO	Time Since Overhaul
TTP	Time Temperature Precipitation
USAF	United States Air Force
VR	Voltage Ratio = $[(V_{on} - V_{off})_{sample-specimen}] / [(V_{on} - V_{off})_{reference-specimen}]$

Chapter 1

Introduction

1.1 Background

Today's turbine engines operate at high turbine inlet temperatures and high thrust-to-weight and pressure ratios compared with previous generation engines [1, 2, 3]¹. Therefore, highly stressed rotating parts, such as discs, are manufactured from high strength nickel-base superalloys to obtain these service conditions. To avoid structural failures under these demanding service conditions, significant advances have also been made in analytical methods for predicting component lives. The life of compressor and turbine discs, is limited by the number of low cycle fatigue (LCF) cycles [4, 5, 6, 3, 7, 8]. LCF behaviour is a complex phenomenon which is not only affected by factors such as service conditions, environments, and the microstructure of the materials, but also by the statistical variation in properties of the manufactured components [9, 10, 1, 11]. This causes a scatter in the fatigue performance of materials exposed to a given service environment. Premature crack initiation followed by crack propagation can cause disc failures, which will have serious consequences on the operational safety of the aircraft. This provides strong motivation to quantitatively evaluate and control the risk of various engine components. Hence, the methods used to predict the life of gas turbine engine rotor components have resulted in overly conservative estimates of useful life, and crack free components with residual life remaining in them are unnecessarily scrapped [4, 6, 12, 5, 13, 7, 8, 14, 10]. Figure 1-1 for example, from a study conducted by Pratt and Whitney (Florida) on the F100 first stage turbine discs at retirement, indicates that over 80% of the discs have at least ten lifetimes remaining [7]. This results in increased engine ownership costs by requiring that components be replaced before service exposure has a significant effect on their material properties.

¹ Numbers in brackets designate references.

In order to reduce these costs and also to conserve strategic materials, a great deal of effort is being expended to develop processes and regenerative heat treatments that would allow used components to be refurbished and returned to service [15, 11, 16, 17].

The incentives for extending service lives of these components are considerable. For instance, the total cost of replacing a full row of blades in an industrial gas turbine may be as much as \$100,000 [9]. In military aircraft or marine gas turbines, the replacement costs of hot section components are even higher. The replacement cost of a set of turbine blades of the General Electric F404 engine used in the newly acquired CF-18 aircraft is as high as \$200,000. These examples demonstrate the potential of rejuvenation by cost-effective methods to reduce the user operating costs.

1.2 Safe Life Limit History of the J85-CAN-40/15 Engines

There are three variations of the J85 engine which are included in this study. The J85-5 is a 17,125 N (3850 lbs max) thrust afterburning version of the J85 used in a twin mount in the Northrop T-38 Talon aircraft. This aircraft is used as an advanced trainer in the United States Air Force (USAF) and was also used, until 1983, in the aerobatic flight demonstration team, the *Thunderbirds*. The Canadian version of this engine, the J85-CAN-15, shown in Figure 1-2, has a maximum thrust of 19,125 N (4300 lbs) and is used in the *Northrop CF-5 Freedom Fighter* by the Canadian Armed Forces. It is an afterburning engine mounted in a twin configuration. A schematic view of the turbine section of this engine is shown in Figure 1-3. The final version of interest is the J85-CAN-40 engine. This non-afterburning turbojet has a maximum thrust of 12,675 N (2850 lbs) and is used in the *Canadair CT-114 Tutor* in a single engine configuration. This aircraft is the basic jet trainer used by the Canadian Armed Forces. It is also used by the Canadian flight demonstration team, the *Snowbirds*.

The J85-CAN-40 and CAN-15 engines were first purchased by the Canadian Armed Forces in the 1960's. However, the safe life limit of this engine was not specified by its manufacturer, General Electric Company, until 1970.

From May 1970 to October 1983 continuous efforts were made by the Canadian Department of National Defence (DND) and the American General Electric (GE) to establish

safe life limits for turbine and compressor components in the J85-CAN-40/15 engines. Several revisions were made to estimate life limits as operating experience was gathered from the USAF J85-5 engine and as mission severity factors for the J85-CAN-40/15 engines became better established. However, between September 1979 and November 1981, it became apparent that the mission severity factor for the J85-CAN-40 engines was almost double the value previously assumed [18], see Table 1-1.

A situation arose where many parts had already accumulated operating lives well in excess of the newly revised safe life limits as shown in Table 1-2. Therefore serious logistic problems arose in the procurement of new parts and the replacement of life expired parts. At the request of DND, GE carried out a *damage tolerance analysis* in which the Weibull distribution was used to estimate times to crack initiation, based on a 0.135% probability of "failure", and a similar approach was used to estimate times for cracks to reach a safe dysfunction limit². The two sets of data were used to determine a *safe inspection interval*.

Since the beginning of 1983, the Structures and Materials laboratory of the National Aeronautical Establishment of National Research Council of Canada (NAE/NRC) has been conducting a post-facto investigation to verify the damage tolerance analysis and to assess its extensions and limitations. This study has been performed in co-operation with Carleton University, University of Toronto and Department of National Defence.

In the case of the hot section parts in the J85-CAN-40/15 engines, all the recovered turbine discs had service lives well in excess of manufacturer's safe life limits, as shown in Table 1-2, and parts were retired in a crack free condition. Previous work performed by Carleton University [18], has shown that while these parts were crack free, they were not free from internal damage, and work has been done to characterize this damage. The damage is most simply described as high temperature low cycle fatigue and preliminary work on laboratory specimens [18] has shown that this damage can be annealed out by heat treatment, thus restoring parts to an essentially damage free condition.

² *Dysfunction* refers to the condition where crack growth rate may become rapid, causing eventual fracture, however, it is not necessarily a critical crack length for catastrophic failure.

1.3 Objectives

The objectives of the first phase of the research project were:

- 1) To develop a computerized high temperature fatigue crack growth rate (FCGR) monitoring technique.
- 2) To evaluate the FCGR behaviour of service exposed and reheat treated turbine discs from J85-CAN-40 engines, at elevated and room temperatures.

The second phase of the research project was to use the procedures established in the first phase, to carry out a comprehensive rejuvenation study on the J85-CAN-15 turbine discs. Consequently, the following objectives were set for the second phase of the research project:

- 1) To examine the microstructure of new and reheat treated first and second stage turbine discs.
- 2) To characterize the microstructure of the first and second stage service exposed turbine discs and determine the type or types of damage.
- 3) To evaluate high temperature FCGR of new, service exposed, and reheat treated first stage turbine discs.
- 4) To determine the effect of rejuvenation treatment on the distortion of the J85-CAN-15 turbine discs.

Chapter 2

Literature Review

2.1 Life Prediction of Turbine Discs

Some 30-40 years ago, gas turbine discs were recognized as the most critical components in the engine as pieces cannot be contained if failure occurs, see for example Figure 2-1 [19]. Table 2-1 shows a survey of the accidents due to turbine section component failure for the world's air transport fleet from 1982 to 1986. Note that disc failure in all cases were uncontained, generally, causing engine fire.

Since that time, various criteria have been used to define both the design parameters of the discs, and the lives of such components. Until the end of the 1950's the main factors considered were

- (a) creep for the rims of turbine discs, and
- (b) tensile strength of bore and rim, to give a satisfactory overspeed margin without burst, for both compressor and turbine discs [19].

These criteria changed with the introduction of Iron-Nickel-base (Fe-Ni-base) superalloys in turbine discs. These modern turbine discs experience service temperatures up to 800°C and therefore the combined effects of creep, fatigue, oxidation and their synergistic interactions are of concern [20]. It has also been widely recognized that defects may be present in parts "as-manufactured", and hence an understanding of the growth of small cracks from pre-existing flaws is considered important. Sensitive and reliable non-destructive inspection (NDI) procedures to detect these flaws have emerged. Therefore the methods of treating and analyzing the cyclic durability of discs have passed through several phases of development. Methods based on fracture mechanics concepts have been the most recent to emerge. Although original creep and tensile criteria still need to be observed, it has been found that the most important factor governing component design

and performance is low cycle fatigue (LCF)¹ [19, 2]. Therefore disc manufacturers began to base life predictions on the *LCF crack initiation* behaviour of disc materials [7, 4, 2, 12, 5, 6, 13].

2.1.1 Conventional Life Prediction

2.1.1.1 The Initiation Criterion

The fatigue cracking process for a typical rotor component such as a disc can be visualized as illustrated in Figure 2-2. Total fatigue life consists of a crack initiation phase followed by crack growth and linkup of microcracks. The resulting macrocracks then propagate subcritically until the combination of service load and crack size exceed the material fracture toughness. Catastrophic failure (see Table 2-1 and Figure 2-1) results if the component is not retired from service.

The "crack initiation" life is defined as the number of LCF cycles or the amount of time required to produce a crack of a size that is large enough for detection by currently available NDI techniques. Therefore, LCF crack initiation data is generated on smooth specimens. The number of cycles necessary to produce a barely detectable crack of about 0.8 mm ($\frac{1}{32}$ inch) long (this depends on material type, crack location, accessibility and confidence level of NDI method) is obtained for a given set of loading conditions such as stress/strain, time, and temperature. The data base must be applicable to all discs of a given material and therefore includes test results from many "heats" and "sources". To compensate for the inherent scatter of fatigue data, a statistical analysis is performed to determine a lower bound for the data, as shown in Figure 2-3. This lower bound is defined as minus three times the standard deviation ($-3 \times S.D.$) from the mean and this value is taken as the "safe life" of the component. The final step involves verification through spin-pit testing, where stress-strain distributions that are more typical of flight conditions can be generated. Generally, pilots adapt their flight plans to the particular route and weather conditions. But engine manufacturers consider the flight profile to be of one or several typical flight cycles [21]. For example, commercial engines typically

¹ Generally, in metal fatigue, LCF may be considered to be consequential if less than 10,000 cycles are required to cause failure.

experience quick acceleration to maximum power level at take-off, hold-times at various power levels and another large power level variation during landing (reverse). Missions vary widely for combat aircraft, the general structure of which is shown in Figure 2-4 [22, 23]. Life is then calculated based on the assumption that the aircraft will be used for a given mission profile (for instance, low altitude penetration and interception followed by air combat training). Accelerated mission tests and simulated endurance tests are also conducted to check engine endurance [24]. The high cost of component testing, however, dictates the use of test data from smooth specimens and finite element based component analysis for design [25, 26, 2].

To provide the operator with a more meaningful life limit, a mission spectrum analysis is performed to determine the number of LCF cycles expected per mission flight hour. These analyses allow the safe life of a component to be specified in terms of either mission equivalency cycles or mission flight hours [5].

2.1.1.2 Problems with Initiation Criterion

As shown for the F100 turbine discs Figure 1-1, the *initiation criterion* is an overly conservative method of estimating the *safe life* of gas turbine rotating components [4, 12, 5, 6, 13, 3, 127, 7, 8, 14]. First, this method ignores the inherent *damage tolerance* in these components which can provide the opportunity to safely extend its serviceability beyond the stage of crack initiation [14]. Second, the $\pm 3 \times S.D.$ bounds encompass 99.9% of all data that may span two orders of magnitude in fatigue initiation life as illustrated in Figure 2-3. Therefore, no allowance is made for the vastly superior properties of the majority of the material tested. This results in retiring one thousand discs because one disc may contain a 0.8 mm crack even though the life remaining in the 999 uncracked discs is not used, and the amount of usable life remaining may be significant, Figure 1-1. Obviously, the inadequacy of this approach is associated with a significant cost to the engine user. If component *lives* could be determined based on reliable knowledge of critical flaw sizes, and the time or number of operational cycles required for a flaw to grow from its size at the detection limit to its critical size, then lower operating costs might be expected. The potential cost saving that could be derived from an alternate rejection criterion based on some measure of crack propagation life, for instance, is indicated in Figure 2-5 taken from

a study by General Electric [1].

Many unknowns and uncertainties associated with the *initiation criterion* approach can be eliminated because of the recent advances in fracture mechanics technology. The size at which a propagating crack will become unstable can now be calculated accurately for parts of complex geometry subjected to a wide variety of environmental and service conditions. These data in conjunction with the statistical data, based on engine user experience, can be used to reliably predict the critical crack size of a specific component. These advances have provided the means through which an improved gas turbine life prediction philosophy can be developed without jeopardizing the safety of the aircraft or the crew [2, 4, 12, 5, 13, 8, 14, 28].

2.1.2 Alternatives to The Initiation Criterion

2.1.2.1 Damage Tolerance Based Life Prediction

Recognizing the shortcomings of the initiation criterion, the safe life approach of using full scale or specimen fatigue test results, and a scatter factor to maintain structural safety, is changing. In its place, the USAF is now implementing the Engine Structural Integrity Programme (ENSIP) [29]. This is part of the USAF Aircraft Structural Integrity Programme (ASIP) for design, analysis, testing, and force management, of new USAF systems defined in a Military Standard. It includes a series of related military specifications and a handbook [30, 31]. The ASIP programme has been applied to the development of all new USAF aircraft, including the B-1, F-15 and F-16 [32]. In addition, damage tolerance and durability assessments have been completed and are being performed on a number of other older aircraft to determine their condition with respect to fatigue. These include the F-4, T-38, F-5, SR-71 and C-130 aircraft [32].

The ENSIP durability and damage tolerance based life prediction procedures are currently the subject of much discussion. They are receiving attention both for establishing component design life limits in new engines, and for extending the usable service lives of discs in conventionally "life predicted" engines. The concepts have been considered for designing several new generation engines, including F100-PW-220, F110-GE-100 and F-109-GA models [33]. They have also been considered for extending the usable service lives

of discs in the Canadian GE-J85-CAN-40/15 and the American PW-F100 and GE-TF34 [33]. The damage tolerance approach is also being considered for application for other Canadian Forces engines including the GE-F404 of the new CF-18 fighter aircraft [33]. The ENSIP programme is very wide in scope, covering the design and development of engine components, a comprehensive engine monitoring programme, an improved maintenance methodology, and investigations into rejuvenation of service induced damaged components through conventional heat treatment and hot isostatic pressing [30, 31, 32, 9, 34, 11]. This study deals specifically with the development of an improved maintenance methodology based on the rejuvenation of turbine discs through heat treatment.

2.1.2.2 Application of the Damage Tolerance Concepts

Damage tolerance based life prediction is also known as *fracture mechanics life prediction*, or as *Life On Condition* by Rolls-Royce, and *Retirement For Cause* by Pratt and Whitney. The inherent defects are assumed to be in the fracture critical locations of a component that are just below the detection limit of the NDI technique used to inspect the component [4, 7, 2, 12, 5, 6, 3, 13, 8, 14, 28, 35, 36, 37]. These cracks are then assumed to grow during service in a manner that can be predicted by either linear elastic fracture mechanics (LEFM), or some other acceptable method, to a predetermined dysfunction limit beyond which the risk of failure due to rapid crack growth becomes excessive.

The FCGR's and the dysfunction crack sizes are established analytically based on best estimates of service loads and material properties. The service time, or number of cycles, required to grow the assumed crack to its dysfunction size, is then used to define a Safe Inspection Interval (SII). The SII is usually calculated by dividing the life to dysfunction by a "safety" factor of 2 [38], to account for changes in the aircraft role, and service environments. This life cycle management concept is illustrated schematically in Figure 2-6 which shows that at the end of one SII, crack free components are returned to service for another SII. The procedure is then repeated until a crack is found. In this manner, components are retired on an individual basis when their condition dictates. Therefore, the purpose of damage tolerance life prediction is not to extend the life of a rotor component, but to safely utilize the full life capacity inherent in the component. The above procedure is also known as the deterministic fracture mechanics (DFM) approach.

It is based on the worst case assumptions for all parameters used, and therefore results in conservative estimate for SII. It has been suggested that more realistic estimates of SII's are possible using probabilistic fracture mechanics (PFM) principles. Using these alternative procedures, the probability distributions associated with the occurrence of initial flaw sizes, crack detection, FCGR's and dysfunction crack sizes, are used to establish a SII distribution, from which an inspection interval can be selected, which ensures a sufficiently low but cost effective failure probability [39 18, 38]. The predictions of either approach can be further revised on the basis of field experience and this is known as Combined Analysis (CA) [18, 38].

2.1.2.3 Establishing the Crack Growth Rates

In order to apply damage tolerance maintenance methodology to actual components, considerable information must be available to the engine life cycle manager. One of the most important pieces of information is that of crack growth rates beyond the NDI detection limit. Therefore, the FCGR's must be established under conditions simulating the discs centrifugal and aerodynamic loads resulting from rotation and the gas bending loads, respectively. These forces create a three dimensional stress state ahead of the crack tip that is responsible for crack advancement. Since FCGR's depend upon the conditions ahead of the crack tip, intense efforts have been made to determine a parameter that would uniquely describe these conditions. So far, three main parameters have been developed

- i) the stress intensity factor, K ,
- ii) the J -integral, and
- iii) the J^* -integral.

The stress intensity factor is suitable for use in cases where plastic deformation is limited to a very small zone ahead of the crack tip and where fracture occurs under nominally *brittle* conditions [40]. In cases where significant deformation results, the other two parameters may be more applicable [41]. The J -integral is said to describe the *stress-strain* field ahead of the crack tip while the recently developed J^* -integral is believed to describe the *stress-strain rate* field [42]. However, the difficulties associated with the experimental evaluation of J -integral and J^* -integral, have inhibited the widespread application of

these parameters to FCGR's [43, 2]. These limitations force the use of the stress intensity factors corresponding to appropriate centrifugal and aerodynamic loads to calculate FCGR's at varying crack lengths at operating service temperatures. This type of data can be used to define a critical crack length or a dysfunction limit for a given component. Crack lengths greater than the dysfunction limit results in unstable crack growth or excessive FCGR's under service conditions [2].

A considerable variation in FCGR's is generally observed in a given set of discs that are manufactured from the same heat and subjected to nominally similar operating conditions. These variations result from localized differences in the microstructure, the fabrication history and the time-temperature-stress history of the component. Consequently, the life from crack detection to dysfunction also shows a considerable amount of variation from one part to another. Therefore, a data base which is statistically significant is required to establish the lower limit of the service life causing earliest dysfunction. Using a Weibull plot, the life distribution to dysfunction can be superimposed on the detectable crack distribution plot, Figure 2-7. A $-3 \times S.D.$ crack propagation interval, in terms of number of missions or service hours, can therefore be determined.

Service induced microstructural degradation effects further complicate the problem of crack growth rate variations in turbine discs. The microstructure and mechanical properties of most engine alloys may change significantly at higher service temperatures. It is therefore essential to generate dysfunction limits of the degraded microstructure for accurate prediction of a crack propagation interval. The fracture mechanics based life predictions should generally be verified through spin-pit testing of a few selected service exposed discs. In this study the characterization of the service induced microstructural damage of the J85-CAN-15 turbine discs is specifically addressed.

2.1.2.4 Potential and Limitations of the Damage Tolerance Life Cycle Management Approach

Damage tolerance based life prediction can be employed both at the component design stage, or as an in-service life extension scheme [33]. In order to apply the concept for life extension, the fracture critical areas of the components must be known and inspectable. As well, statistically significant information must be available on:

- i) operational loads, service temperature and simulated service environments to determine the crack growth rates,
- ii) sensitivity and probability of detection of NDI technique, and
- iii) dysfunction crack size.

This is probably one of the major limitations in applying this concept for post design life extension, as the required information is generally not readily available [33].

2.2 Material Considerations for Gas Turbine Discs

2.2.1 Operating Environments in Turbines

A schematic cross section of an early single spool axial flow turbo-jet engine and the gas temperature, pressure and velocity profiles which may be expected, are shown in Figure 2-8 [44]. Gas temperature is highest in the combustion chamber. It then drops gradually as the gas expands across the turbine. In the most advanced engines, gas inlet temperatures may reach as high as 1300°C [45]. Metal temperatures are kept well below this level by internal cooling using air from the compressor. For example, the temperature of the discs does not generally exceed 750°C [46]. The temperature varies across the disc, being highest at the rim and lowest at the bore [47], Figure 2-9. Typical rim temperatures of the J85-CAN-40 and CAN-15 turbine discs are estimated at 550°C and 650°C , respectively, at take off. Coles [1] indicates an operating temperature of $370 - 650^{\circ}\text{C}$ ($700 - 1200^{\circ}\text{F}$) for the turbine discs of the J85 engine powering the American F-5 aircraft, Table 2-2.

The turbine section of an aero engine provides the power to drive the compressor and accessories. It does so by extracting energy from the hot gases released from the combustion chambers, and expanding them to a lower pressure and temperature, see Figure 2-8. Due to centrifugal loading, very high stress levels are involved in this process, Figure 2-9. Turbines may rotate at tip speeds of up to 400 m/sec (for efficient operation). Rotational speeds of the J85 engine can reach as high as 16,500 rpm.

2.2.2 Requirements and Development of Disc Materials

As indicated in Section 2.1, LCF has been recognized as the principal cause of most

disc failures. One area of progress made during the last thirty years, in disc material LCF properties, is illustrated in Figure 2-10. Here, a comparison is made between the properties of conventionally cast and forged materials, and those of the more recently developed powder processed alloys [45]. Note that the 650°C yield strength, which is an indication of the LCF crack initiation property of a disc material, has more than doubled during this period.

After the birth and early development of the jet engine during the World War II in Europe, the first engines to be produced in significant quantities were the Derwent, J-47 and J-48 engines made by Rolls-Royce, General Electric and Pratt and Whitney Aircraft, respectively [48]. The Fe-Ni-base superalloy A-286 and H46 were the most popular alloys during this time period. In the late 1960's, Astroloy became famous after its successful use as a disc material in the J-58 engine powering the SR-71 aircraft.

Because of the Pratt and Whitney's success with Waspaloy and Astroloy, General Electric introduced the Ni-base superalloy René 95 into the marketplace. Meanwhile Inconel 718 evolved into Super-718 about 10 years ago and has recently evolved into DA-718².

To date, these materials have generally met the requirements, but as turbine inlet temperatures continue to increase and stronger blade and vane materials are required, the superiority of the Ni-base superalloy forgings and rapidly solidified powders with respect to LCF and creep properties, make these more costly materials essential for some applications. Recent LCF test requirements have triggered new efforts [49, 50] for cleaner and more controlled processes in gas turbine disc manufacturing, costing more but serving longer in service.

The turbine discs of the J85-CAN-40 and J85-CAN-15 engines are manufactured from austenitic stainless steel alloy A-286 and Ni-base superalloy Inconel 718, respectively. In the following section a brief review of the structure-property relations of A-286 and Inconel 718 turbine disc materials is presented.

² DA stands for *Direct Aged*.

2.2.3 Structure-Property Relations in A-286 and Inconel 718

2.2.3.1 Physical Metallurgy of A-286 (J85-CAN-40 Turbine Discs)

A-286 is an iron-rich alloy which is strengthened by γ' (gamma prime) but also contains η (eta) phase. This alloy has been used at temperatures up to 650°C . The chemical composition of A-286 alloy is given in Table 2-3. Chromium (Cr), molybdenum (Mo) and iron (Fe) are added for solid solution strengthening, whereas aluminum (Al) and titanium (Ti) are added for precipitation strengthening. A small amount of boron (B) is also added for grain boundary strengthening [51]. The most significant contribution to the strength of A-286 results from the precipitation of γ' which has an ordered Face Centred Cubic (FCC) structure. Al and Ti combine with nickel (Ni) to form γ' $[\text{Ni}_3(\text{Al}, \text{Ti})]$ phase, the morphology of which can be either spherical, cuboidal, or plate shaped. An increase in the volume fraction of γ' in A-286 is known to result in enhanced stress rupture properties [51, 52]. On the other hand, an increase in the size of γ' precipitate increases the strength until peak age hardening is reached. After this peak age hardening is reached, further growth in the precipitate size causes a decrease in the strength [53]. The solvus temperature of γ' phase in A-286 is 855°C [53]. It precipitates at temperatures between 650 and 815°C as shown by the Time-Temperature-Precipitation (TTP) diagram of Figure 2-11 [54].

The η phase has a Hexagonal Close Packed (HCP) structure which can form either during processing and heat treatment, or during service. The γ' transforms to the η phase in A-286 after extended exposures to temperatures between 600 to 850°C [53, 55, 56]. It forms as intragranular platelets with a Widmanstätten morphology or as a cellular grain boundary. Cellular η appears in A-286 and other Fe-Ni-alloys, as alternate coherent lamellae of η and γ originating at a grain boundary with a random orientation relationship to the grain into which a cellular zone is growing [53, 56, 57]. Severe degradation of mechanical properties is often caused by coarse interlamellar spacing of the η phase, as well as loss of the more finely distributed γ' in the matrix which is involved with the cellular precipitate formation [53]. The intragranular platelets (the Widmanstätten form of the η phase) have been associated with a loss of strength but not ductility [58]. Exposures to temperatures of 800 to 850°C or higher, may cause the formation of both Widmanstätten

and cellular forms of η phase in A-286 [53, 56, 57, 59, 60]. The formation of platelet η phase can best be avoided by heat treating the alloy above the η solvus temperature of 915°C , and by utilizing the alloy at temperatures below which the γ' to η transformation does not occur [53].

Carbides can form at grain boundaries during processing, heat treatment, or service. $M_{23}C_6$ carbide "films" may form during service in A-286 where M is primarily chromium. A fine cellular form of $M_{23}C_6$ can form due to improper processing and cause stress-rupture notch sensitivity [53, 61, 62].

2.2.3.2 Standard Heat Treatments for A-286

After forging rough shapes, components are solution heat treated to control the grain size, followed by aging heat treatment for optimizing various material properties. The different heat treatments which have been used for this alloy are presented in Figure 2-12. A-286 may be solution annealed at either 899°C or 982°C to produce a fine or coarse grain size, respectively. The alloying elements which are taken into solution during solution annealing are retained in the solution by rapid cooling of the forgings. This rapid cooling also ensures full aging response in all sections of the forging. The morphology of the η phase which is observed after such a process is usually finely distributed intergranular platelets in the vicinity of the grain boundaries. Some cellular η phase may also be observed after such a process [53, 63, 64].

Selection of the appropriate solution annealing temperature depends on the properties which are desired in the hardened component. The lower anneal temperature of 899°C generates a fine grain structure, improved fatigue strength and a better balance of tensile strength and stress-rupture ductility up to service temperatures of 650°C . The 982°C annealing temperature results in better creep ductility and tensile strength but with some sacrifice in the stress-rupture life as shown in Table 2-4. Aging at a temperature of 718°C for 16 hours is used to precipitate an effective strengthening array of γ' phase, Table 2-5.

2.2.3.3 Physical Metallurgy of Inconel 718 (J85-CAN-15 Turbine Discs)

Inconel 718 has been in extensive use since it was first developed by the International Nickel Company [65]. An extensive data base and much literature on the physical metal-

lurgy of this alloy is available [65, 66, 67, 68, 69, 70, 64, 71, 72, 43]. Therefore only a brief review of the physical metallurgy of this alloy is presented in this section [73].

Inconel 718 is a versatile, high strength, precipitation hardenable Ni-base superalloy developed for use at temperatures up to 700°C [74]. Like most superalloys, Inconel 718 is a complex alloy, strengthened by the precipitation of several phases [74, 66, 67, 68, 69, 70]. The chemical composition of this alloy is given in Table 2-6. The major elevated temperature strengtheners are the ordered FCC γ' and ordered Body Centered Tetragonal (BCT) γ'' (gamma double prime) precipitates. The most notable strengthener of Inconel 718, γ'' , is a Niobium (Nb)³ rich Ni_3Nb precipitate. In contrast, most Ni-base superalloys are primarily hardened by Al and Ti rich γ' . Along with these two strengthening phases a number of other minor phases can also form in this alloy, Figure 2-13, depending on the heat treatment used. Both "Laves" phase and an incoherent form of Ni_3Nb precipitate may appear in this alloy. Traces of NbC and TiN are also present. Incoherent Ni_3Nb precipitates (generally known as δ , delta, phase) form upon overaging of γ'' , and the phase appears as needles within the grain interiors and at the grain boundaries. While Ni_3Nb needle precipitation reduces the volume fraction of the major strengthening phase γ'' and may harm strength [74, 66], its precipitation at the grain boundaries in small quantities may lead to improved creep properties. It is believed that the irregular or "serrated" grain boundaries formed by Ni_3Nb needles inhibit grain boundary sliding at elevated temperatures [64, 75]. The sites at which Ni_3Nb needles precipitate are dependent on the heat treatment [64]. Laves phase on the other hand, appears as large, massive particles that are usually found in heavily cored or segregated regions. During solidification, Laves phase forms in segregated interdendritic areas at temperatures of 1150 to 1200°C [67]. Depending upon the heat treatment, Laves phase may appear upon aging at temperatures as low as 780°C [71]. This phase is very harmful to mechanical properties because of its brittle nature. It also decreases strength by depleting the matrix of valuable alloying elements. NbC and TiN particles also form during solidification of the alloy, Figure 2-13. Since NbC removes Nb from the matrix, it reduces the strength of the alloy, although some amount of NbC precipitation is essential for controlling the grain size of the alloy during

³ This is also known as Columbium (Cb).

heat treatments. The solvus temperatures of both major and minor phases of Inconel 718 are listed in Table 2-7.

2.2.3.4 Heat Treatments for Inconel 718

A fine grained microstructure is generally considered desirable for good ductility, tensile strength, resistance to LCF crack initiation, and sonic penetrability [72]. A uniform, fine grained microstructure is also preferred in the disc forging of Inconel 718. However, for reasons of crack propagation and creep resistance, the grain size should not be too fine [43]. Heat treatments for Inconel 718 usually involve a solution treatment, to dissolve precipitates in the matrix, followed by one or more ages to obtain the desired precipitate microstructure. These effects of solutioning and aging have been considered in detail by examining microstructural features produced by a wide variety of heat treatments recommended for this alloy by Doswell et al. [73]. Since the J85-CAN-15 turbine discs were originally heat treated according to the conventional heat treatment, a brief description is given below.

The conventional heat treatment, which is often specified for stress-rupture applications, consists of a solution treatment at 955°C and a double age at 718°C and 621°C . The conditions associated with this heat treatment are:

$955^{\circ}\text{C} / 1\text{hr} / \text{Air Cool} +$

$718^{\circ}\text{C} / 8\text{hr} \Rightarrow \text{Furnace Cool at } 1^{\circ}\text{C}/\text{min} \Rightarrow 621^{\circ}\text{C} / 8\text{hr} / \text{Air Cool}$

These conditions are superimposed on the Inconel 718 TTP diagram and is shown in Figure 2-14. During the solution anneal, stable Ni_3Nb is precipitated. Additional amounts of stable Ni_3Nb will form during the first age along with the hardening constituents γ' and γ'' to increase the strength [66], whereas only γ' and γ'' form during the second age which also increases the strength. The final microstructure will consist of γ' and γ'' uniformly distributed throughout the γ matrix, and Ni_3Nb needles at the grain boundaries. NbC film does not form, since it was not dissolved during annealing. The Ni_3Nb needles at the grain boundary creates an irregular boundary that are believed to be beneficial to the stress-rupture properties, creep, and FCGR's.

2.3 Degradation Modes of Hot Section Components

The lives of various engine components are limited by service induced degeneration effects, such as fatigue, creep, erosion, and a variety of different environmental reactions. The precise form of deterioration, however, depends on the part, its design function, operating environment and temperature, and the material used in its manufacture. The life limiting factors of the major turbine engine sections are summarized in Table 2-8 [33, 2]. A listing of materials, operating temperatures and potential life-limiting failure modes for the J85 engine components [1] are given in Table 2-2. This review will concentrate on the rejuvenation of service-induced damage in the major hot section components, namely turbine discs, and turbine blades. The nature of the damage is first identified, and then discussed in terms of the type and operating conditions of the engines.

Two distinct types of damage are prevalent in hot section components:

- 1) Surface Damage, and
- 2) Internal Microstructural Degradation.

Surface damage, especially in the airfoil of turbine blades, may be due to mechanical impact, wear, fretting, erosion, corrosion, oxidation, or thermal fatigue. Internal degradation, on the other hand, is related to microstructural changes which occur slowly during service, and depends on the alloy type and the service operating conditions, and therefore to some extent, on the type of turbine. In this respect, there are two major differences between operating environments of industrial gas turbines and military aircraft or marine gas turbines. First, the land based power plants tend to operate for long periods of time under relatively stable conditions, and therefore problems due to thermal fatigue are usually less severe. Secondly, industrial turbines are often operated on relatively clean fuel and in a salt free atmosphere, and therefore problems with hot corrosion, oxidation and sulfidation, also tends to be less severe than in marine gas turbines. However, it is difficult to generalize as in reality these problems are not unknown to land based turbines particularly if they run on low quality fuel with a high sulfur content, or when they operate in corrosive marine environments [33]. There are situations where land based turbines are repeatedly started at short intervals, as for instance in peak load generator drives, in

which cases thermal fatigue can prove life limiting for hot parts.

Surface damage can promote crack nucleation at stress concentration sites associated with the damaged areas. These cracks can then propagate under the influence of centrifugal and/or cyclic loads during service, be they mechanical or thermal, and eventually cause component failure. This was the case for a failed J85-CAN-15 turbine disc examined by General Electric. It was determined that a crack had initiated from a heavily oxidized surface which extended between a bolt hole and a cooling hole, see Figure 1-4. The fault was attributed to a forging lap.

This study does not deal with the repair of surface damage, such as corrosion or surface connected fatigue cracks. It deals only with the elimination of internal microstructural damage. It also deals specifically with Fe-Ni-base superalloy components, which account for a large proportion of the hot section of any turbine.

2.3.1 Internal Microstructural Degradation in Fe-Ni-base Superalloy Components

The Fe-Ni-base superalloys are complex alloys, and the changes in microstructure that may occur in service are also complex. Internal damage is related to the microstructural changes which occur slowly during service exposure at a rate that is strongly influenced by service temperature and stresses, and therefore, to a great extent, by user practice. These microstructural changes can be associated partly with plastic damage accumulation, and partly with metallurgical ageing. Depending on the component, alloy composition variation and service conditions, the plastic damage may be due to creep, LCF or creep-fatigue interactions. Creep is associated with cavitation or void formation which represents the initial step in creep failure, which consists of the nucleation and growth of voids on grain boundaries aligned normal to the stress axis.

LCF damage is dominant in bores and bolt hole areas of compressor and turbine discs. It is associated with engine start-up and shut-downs and with RPM excursions during service. Microstructurally, LCF damage manifests itself in the form of intense slip banding, micro-twinning and dislocation activity. LCF damage can also lead to precipitation of embrittling phases such as η phase in A-286 turbine discs, or to coarsening

of strengthening phases such as $M_{23}C_6$ in AM-355 compressor discs [18, 38]. While the driving force for LCF damage accumulation in compressor discs is primarily mechanical, in hot section areas of the engine it may also be associated with thermal cycling. Thermal fatigue (TF) damage in vanes and high performance blades is often a life limiting factor in modern turbines [33].

2.3.2 Deterioration of Mechanical Properties

The aforementioned microstructural changes are detrimental to properties such as tensile strength, ductility and notch sensitivity. As pointed out previously, these changes will occur slowly depending on alloy composition and service operating conditions. It is difficult to predict the rate of deterioration from first principles and attempts are being made to monitor this experimentally [9]. In this study components were removed from service and their FCGR's and tensile properties were compared to new and heat treated discs to predict the effects of microstructural damage. Data obtained in this manner, are not necessarily typical, and there may be considerable scatter from turbine to turbine, thus making it difficult to establish life limits without being unduly conservative.

2.4 Rejuvenation Possibilities of Turbine Components

Once the safe operating life of a turbine disc or a set of turbine blades has been exhausted, they may either be replaced at great cost, providing there is no lack of replacement parts, or they may be rejuvenated and put back in service, consequently reducing the cost of ownership of the engine. This requires an appropriate heat treatment that can regenerate the precipitate structures and eliminate other internal damage. This may be achieved either by conventional heat treatments, or by Hot Isostatic Press (HIP) processing. The application of either process is considered in turn for turbine blades and discs in the following sections. In the case of surface damage, it is sufficient to say that components with severe damage or cracks are scrapped, whereas minor damage can be repaired by blending, dressing, brazing, welding, or machining. For example, Gamble et al. [76] extended the total life of an A-286 first stage turbine disc of an industrial turbine rotor by a factor of three. This was achieved by the machining of the cracked turbine disc.

The repair procedure, illustrated in Figure 2-15, involved removing the cracked turbine disc from service, machining the existing cracks from the blade slots, and replacing the original notch configurations with ones having larger radii. The disc was then put back into service.

Of the types of microstructural, or internal damage discussed earlier, the γ' morphology changes should be reversible by conventional heat treatments involving complete solutioning, followed by controlled ageing. If carbides are formed during solidification of the alloy, their morphology is not reversible [9]. It is also unlikely that cavitation damage would be eliminated by a straight conventional heat treatment. Meanwhile, the precipitation of embrittling minor phases should be reversible.

2.4.1 Rejuvenation of Turbine Blades

It has been reported that reheat treatment of service exposed Nimonic 90 specimens resulted in considerably increased total life, although ductility was reduced [77]. On the other hand, reheat treatment did not fully restore the stress rupture life of forged Inconel X-750 turbine blades [78]. These and other reports [79, 80, 9] on rejuvenation of turbine blades, indicate that damage such as voids due to creep, is introduced during service, and cannot be eliminated by conventional heat treatments. In the alternative HIP treatment, the parts are simultaneously exposed to high temperature and high hydrostatic pressure⁴. Void removal is ensured by HIP processing. Even gross internal shrinkage porosity in IN-738 turbine blade investment castings can be removed using this process [81]. In a study conducted on samples removed from the same Inconel X-750 blades, and subjected to either conventional reheat treatment or HIP treatment, it was clearly shown that the HIP processed material was superior to both commercial and laboratory reheat treated material [80] as shown in Figure 2-16. The laboratory heat treated blades should represent the maximum property improvement attainable by this approach, since ideal temperature control and cooling rates were employed, and the microstructure was fully restored. Therefore, the improvement achieved using the HIP cycle must be due to complete removal of

⁴ The HIP facility at NAE/NRC can reach temperatures and pressures as high as 2,000°C and 207 MPa.

service damage. The comparison of service exposed and HIP reheat treated material revealed that in both cases the grain size was the same. This observation is important since it demonstrates that improved creep resistance in the HIP processed material is due to removal of service damage and was not obtained by increasing the grain size at the expense of tensile and fatigue crack initiation properties. Metallographic examination of airfoil sections showed negligible internal oxidation, nitriding or alloy depletion at the surface in either material condition. It should be noted however, that acceptable microstructural features and satisfactory test results after HIP rejuvenation are not sufficient to provide complete assurance that original properties are in fact fully restored. This is because of possible changes in elemental segregation that might result from service exposure and resolutioning of those parts. Also, the testing conditions can only approximate the service conditions, and therefore test results cannot tell with any certainty whether the part deterioration will not be faster after being returned to service. Such assurance can only be provided in the end through real-time in-service monitoring of rejuvenated parts.

2.4.2 Rejuvenation of Turbine Discs

In principle, the techniques applied to turbine blades may be applied to turbine discs. Aside from the previously mentioned report, where A-286 turbine discs were repaired by machining, there has only been one published study on the rejuvenation of turbine discs [11]. This study focused on cracked turbine discs of the TSC700 Auxiliary Power Unit (APU) used in the DC-10 aircraft. The material was Waspaloy, and the surface connected cracks had to be sealed and healed. Figure 2-17 compares the LCF properties of crack repaired plus HIP processed specimens, with those of uncracked specimens. Both curves are extremely close with the $3 \times S.D.$ estimates showing that the pre-HIP tests exhibit less scatter than the post-HIP tests. Transmission electron microscopy of the HIP processed Waspaloy discs showed a reduction in the dislocation density when compared to pre-HIP samples. It was concluded that retired crack-free Waspaloy turbine discs could be successfully HIP rejuvenated to restore their original tensile strength, microstructure, and uniaxial LCF properties. However, the study did not address the fracture mechanics properties of the discs. Monitoring the fracture mechanics properties of the service exposed and rejuvenated A-286 and Inconel 718 discs is the major thrust of this work.

In the past, crack free service exposed discs have been retired due to excessive blade play in the fir-tree configuration. This condition is apparently caused by creep in the rim of the disc and may be enhanced by wear. Canadian representatives of the Industrial Gas Turbine User Group on Gas Turbine Operations and Maintenance have indicated that creep damage in A-286 turbine discs is of some concern [82]. Unexpected high temperatures in the turbine section of several T56 engines, due to escape of the hot gases, has also caused failure of turbine spacers which has been attributed to stress rupture failure [83]. In such cases, the HIP treatment may be the most successful in the rejuvenation of creep properties of the service exposed components. The metallurgical factors that should be considered prior to the selection of HIP temperatures [84, 85] are: (i) grain coarsening temperature including the time and temperature dependence of grain size, (ii) γ' , $M_{23}C_6$ and MC solvus temperatures, and (iii) extent of primary MC carbide degeneration that occurs during service. The selected HIP temperature should be above the γ' and $M_{23}C_6$ solvus temperatures and below the MC solvus temperature to avoid grain growth [79].

The important points of microstructural features to be considered prior to the selection of post-HIP thermal cycle [84, 85] are: (i) serrated grain boundary formation, (ii) carbide precipitate morphology, and (iii) primary and secondary γ' precipitate size and shape.

All these microstructural features are interrelated and need to be optimized for specific materials by controlling cooling rate and partial solution temperatures during the reheat treatment or the HIP processing [79]. Finally, it should be noted that if the microstructural degradation of a disc during service is only of the LCF induced damage, conventional or specifically designed heat treatments may rejuvenate the original properties of the material. The aim of this study is to investigate such a possibility on the turbine discs of the Canadian Forces J85-CAN-15 and J85-CAN-40 engines.

2.5 Methods of Measuring Crack Growth at Elevated Temperatures

As was described in Section 2.1.2.3, in order to predict critical gas turbine engine component lives based on *damage tolerance* methodologies, accurate FCGR's must be available under simulated service loading (fatigue) conditions at elevated temperatures.

FCGR data at elevated temperatures can either be generated by spin pit testing of actual discs or by testing fracture mechanics compact tension (CT) specimens machined from discs or forged experimental plate stock. The latter apparatus is much cheaper for generating large FCGR data bases. To establish FCGR for CT specimens as a function of stress intensity factor range (ΔK), crack length (a) versus number of fatigue cycles (N) curves must be available. This implies that crack length as a function of N must be measured accurately during fracture mechanics testing of CT specimens. Although many methods are well established for room temperature applications, they are not all suitable for high temperature situations involving crack growths where stability over long periods of time is essential. Many of the measuring devices are incapable of being inserted into a hot furnace.

The three most popular and widely used crack monitoring techniques at elevated temperatures are:

- 1) optical,
- 2) compliance, and
- 3) electrical potential methods.

The optical technique is usually manual but the others can be automated in conjunction with a computer. The following brief review of these methods is undertaken to determine the most suitable method for application in this study.

2.5.1 Optical Method

In an optical system, a specimen is surrounded by a furnace which may be provided with viewing ports to enable the specimen to be seen with appropriate illumination. Otherwise, the test must be stopped, the furnace opened, crack measured, furnace closed, and temperature brought back up to test temperature before the test can continue. A travelling telescope, attached to a vernier scale, is used to focus on the crack tip and follow the progression of its growth. To check whether the crack grows symmetrically for a thick specimen, it is desirable to be able to view the crack from both sides of the specimen, which again may require that the test be stopped and the specimen removed for crack length measurement on both of its sides.

Another complication that arises with optical methods is when surface oxidation occurs. Under such circumstances it may not be possible to detect the fine cracks against the dull background of an oxidized surface. Clearly, optical methods will only give the crack length on the surface and a reliable indication of average crack length can only be obtained if the crack front is relatively straight through the thickness of the specimen. In such cases where extensive creep deformation takes place, crack tunnelling is often obtained which may be pronounced. This leads to inaccurate estimates of the stress intensity factor K .

An advantage of this technique is that crack extension is obtained directly, as no calibration is required. However, it has the disadvantage of being labour intensive and its accuracy depends upon the operator's skill. Even then, only surface readings are possible. The precision of this method depends mainly on the precision of the travelling microscope used.

2.5.2 The Compliance Method

There may be occasions during high temperature testing when parameters such as the J -integral or C^* correlate better with FCGR data than the stress intensity factor (K or ΔK), in which case the values of crack opening displacement (COD) of the test specimen would be needed. In these situations use of the compliance technique is necessary. The compliance technique relies on the change in the effective stiffness of a cracked specimen. As the crack extends, an increase in the crack opening displacement, or COD, results across the specimen notch. The COD is measured during cycling using a clip gauge which is a self contained bridge circuit consisting of strain gauges. At least two of these strain gauges are mounted on arms that are attached to the specimen notch by clipping onto knife edges, which are either machined into or bolted onto the specimen edge. The movement and resultant variation in strain in these arms is indicative of the COD which is compared to a predetermined calibration curve to obtain the actual crack length.

The main advantage of this method of measuring crack extensions is that automatic recording of results is possible. It is also possible to make allowance for curvature of the crack front. Since the displacements are measured remote from the crack tip, an indication of overall specimen response will be obtained and an approximately average crack length

determined.

This method is most satisfactory for those circumstances where load changes are required during a test, such as in thermal fatigue and creep/fatigue investigations, when one of the normal load changes can be used to obtain an instantaneous estimate of compliance.

The use of the compliance technique for determining FCGR's in compact tension specimens is also quite popular, particularly when conducting computer controlled testing [86]. Room temperature precisions of 0.07 mm have been reported [87, 88] for the compliance technique using a calibration curve obtained from an optical tracking technique.

2.5.3 The Electrical Potential Method

The electrical potential method technique relies on the fact that an electric field in a current carrying body is disturbed by the introduction of a crack. As the crack length increases the un-cracked cross-sectional area decreases, its resistance increases and therefore the potential difference across two points on either side of the crack increases, Figure 2-18. By monitoring the potential increase and comparing it with a reference potential, the crack growth can be determined with the aid of an appropriate calibration curve for the particular specimen [25, 89, 90, 91, 92, 93]. The positions of the current supply and potential measuring leads are chosen to optimize the system [89, 94, 95].

This technique has been applied successfully to a wide range of crack growth investigations including fatigue, fast fracture, initiation, stress corrosion and creep studies [25, 95, 96]. The DC version has been used at NAE/NRC for obtaining FCGR data on AM-355 specimens at temperatures up to 200°C [97].

In order to ensure repeatability, it is essential that the current supply is stable and the calibration curve is not unduly sensitive to small errors in the lead positions. A high stability, high gain amplifier capable of measuring potential differences across the crack of the order of μV (microvolts) with a high degree of precision is needed. The size of the test piece is limited by the capacity of the power supply in the DC-PD system. Current densities of about 0.1 A/mm² are needed [98] for acceptable accuracy resulting in currents of a few amps [99] to 100 amps [100]. The potential drop is reported to be very precise, and satisfactory detection of better than 0.05 mm of fatigue crack growths are not uncommon

[94, 101, 102, 95, 103, 104, 105] with careful attention to efficient electrical and magnetic screening and shielding of cables, provision of a good ground, and consideration of thermal electromotive forces (emf's).

Both alternating current (AC) and direct current (DC) versions of the potential drop (PD) technique have been developed. The AC method has been reported to offer higher sensitivity and improved accuracy over the DC method [102, 106], however, the DC version has been the more popular. It has been particularly successful in monitoring crack growth under LCF loading where the extent of crack tip plasticity is small [107].

In the AC-PD version, the current flow is confined to the surface layer because of the "skin effect" which is particularly pronounced in ferritic steels [92]. Frequency is likely to be an important variable affecting the ability to detect and monitor slow stable growth. The DC-PD version enables a material-independent determination of crack length whilst the AC-PD version gives a material dependent crack length. The main reason for this is the difference in the permeability and its variation with the density of the applied current. Also the AC version is much more complex and difficult to implement and requires more elaborate equipment.

Chapter 3

Experimental Materials and Methods

3.1 Materials and Heat Treatments

3.1.1 Turbine Discs of J85-CAN-40 Engine

The A-286 alloy steel forgings are used in solution treated and aged condition for special high temperature applications. The chemical composition requirements of this alloy are given in Table 2-3. A-286 turbine disc material was supplied by Orenda in the form of "as-retired" components. A list of the discs used, along with their Time Since New (TSN) is given in Table 3-1. The 1771.9 and 2210.9 (also referred to as 1772 and 2211, respectively) hrs TSN discs were subjected to a reheat treatment schedule of:

$1000^{\circ}\text{C} / 1\text{hr} / \text{Argon Quench} + 718^{\circ}\text{C} / 16\text{hr} / \text{Argon Quench}$

The 1000°C and 718°C treatments were carried out in an argon environment. In contrast, the heat treatment for virgin forged discs manufactured by GE is

$982^{\circ}\text{C} / 1\text{hr} / \text{Oil Quench} + 704^{\circ}\text{C} / 16\text{hr} / \text{Air Cool}$

The mechanical property specification for heat treated disc forgings is given in Table 3-2. GE does not specify the maximum crack growth rate allowed for disc forgings. However, values taken from the literature for A-286 are given in Table 3-3.

The AM-355 compressor disc material was also supplied by Orenda in "as-retired" components. Specimens machined from these discs were only used for implementation of the elevated temperature DC-PD technique, and therefore it is sufficient to say that it is a precipitation hardened martensitic stainless steel used, presently, for manufacturing the compressor discs of the J85-CAN-40/15 engine.

3.1.2 Turbine Discs of J85-CAN-15 Engine

The J85-CAN-15 turbine discs are forged from Inconel 718 billets with the chemical composition given in Table 2-6. The billets are produced by vacuum induction melting, followed by vacuum consumable electrode remelting. The forgings are then solution treated in the range of 954 to 982°C for 1 hour, at the selected temperature, within $\pm 14^\circ\text{C}$, and air, oil or water quenched. This is followed by ageing at $718 \pm 14^\circ\text{C}$ for 8 hours, furnace cooled to $621 \pm 14^\circ\text{C}$ at a rate of $56 \pm 14^\circ\text{C}$ per hour, held at $621 \pm 14^\circ\text{C}$ for 8 hours, and then air cooled. Tensile specimens prepared from these heat treated forgings must meet the minimum properties given in Table 3-4. A hardness of 346 to 450 Brinell or equivalent (e.g. 34.6 to 45.0 R_C) must also be met. The average grain size of the forging must be ASTM No.4 (average diameter of 90 μm) or finer. Occasional grains as large as ASTM No.2 (average diameter of 180 μm) are allowed. Inconel 718 turbine discs were supplied by Orenda in the form of "as-retired" components.

A list of the specimens used, along with their TSN in hours is given in Table 3-5. The 1523.1 and 1617.4 (also referred to as 1523 and 1617, respectively) hrs TSN discs were subjected to a reheat treatment schedule of:

955°C / 1hr / Argon Cooled +

718°C / 8hr $\Rightarrow$ Furnace Cooled at 1.0°C/min $\Rightarrow$ 621°C / 10hr / Argon Cooled.

All treatments were carried out in an argon environment and a 10^{-6} torrs¹ vacuum, to reduce any oxidation of the surfaces. GE does not specify the maximum FCGR allowed for disc forgings. However, selected FCGR's available from the literature for Inconel 718, with similar fabrication and heat treatment history, is given in Table 3-8.

It should be mentioned that during the manufacturing process the turbine discs are frequently inspected for defects, dimensional accuracies and correctness of assembly. Dimensional inspection consists of measuring specific dimensions to ensure that they are within the limits and tolerances laid down and are known as "fits and clearances". Turbine discs are generally measured at each overhaul because only a small amount of wear or

¹ 1 torr = 1 mm of Hg at 0°C.

distortion is permissible. The detection of defects or cracks that are not normally visible to the naked eye is most important, particularly in discs. Orenda inspects all turbine discs prior to service and also at retirement, by either ultrasonic inspection or production liquid penetrant inspection (LPI).² All A-286 and Inconel 718 turbine discs used in this study were inspected and were found to be free of any cracks at retirement.

3.2 Optimum Crack Measurement Technique for Application in this Study

The purpose of reviewing the various methods of crack growth monitoring at elevated temperature in Section 2.5 was to determine a crack length measurement technique suitable for computer controlled fracture mechanics testing of CT specimens machined from J85-CAN-40/15 turbine discs at temperatures up to 650°C. Consequently the crack growth measuring technique should be simple to use, independent of material type, applicable at room and elevated temperatures and fully automated in conjunction with a computer. These constraints limit the choice to either the potential drop or the compliance techniques. These techniques provide a total measurement of the crack length, inclusive of crack front curvature, and since it does not require visual accessibility, the tests may be conducted in any sealed environment. The output, which is continuous, allows automated data collection and processing together with 24 hour usage of testing machine capacity.

The use of the capacitance gauge, which is based on the compliance technique, was explored. With this technique a 9 mm diameter by 25 mm long Inconel 600 sensor is mounted inside a guard element which is an 18mm x 18mm x 12.5mm block, and this is used to measure capacitance changes with reference to a target surface of a similar Inconel 600 block, Figure 3-1. The sensor and the guard element are separated by a ceramic insulator and both elements are driven at the same electrical potential by a capacitive reactance instrument and are referenced to the earth grounded target. Independent of temperature, the capacitance function, X_C , varies linearly with the distance D between

² Also abbreviated FPI if a fluorescent dye is used.

the sensor and the target surface according to the expression:

$$X_C = B \times D \quad (3.1)$$

where B is a constant and $X_C = \frac{1}{2}\pi fC$, where f is the frequency and C is the charging capacitance.

The sensor-guard element and the target block are screw mounted on either side of the specimen notch as shown in Figure 3-1. The instrument output X_C in volts is calibrated against D in mm at room temperature and finally D is calibrated against crack length a through interrupted elevated temperature tests to obtain da/dN data. The cumbersome nature of the latter part of the calibration between D and a has to be repeated if the microstructure of the test material is changed. On the other hand, in potential drop technique the room temperature calibration curve can be used at elevated temperatures to measure the crack length, irregardless of the material type and microstructure. This method is also capable of detecting small increments in crack length which cannot, otherwise, be resolved optically.

Therefore, it was decided to implement and use the DC-PD technique for FCGR measurement at elevated temperatures. Since the DC-PD technique had already been used at NAE/NRC for crack growth monitoring [108], the elevated temperature capability would compliment the system and would prove to be very useful for any future fatigue crack growth measurement at both room and elevated temperatures.

3.3 Elevated Temperature DC-PD Technique Implementation

3.3.1 Modifications to Improve the Basic Method

The basic method, Figure 2-18, has been modified to increase its accuracy, precision, reliability, and long term stability [89, 97]. A reference voltage specimen, identical to the test specimen, is added in series to the test specimen being tested. The PD across the notches of both specimens is simultaneously monitored and the ratio of the two is taken as the crack length dependent parameter, Figure 3-2. The use of a reference specimen in

series with the test specimen eliminates any effect of temperature variation, because both specimens are maintained at the same temperature and environment. This allows the use of the room temperature calibration curve for elevated temperature applications of similar specimen geometry [101, 95, 103, 94, 109]. Since the specimens are in series, and a ratio of the voltages is used as the crack dependent parameter, any variation in current has no net effect on the results as shown in Figure 3-3. A relatively large current is applied to the system to insure proper induced voltage. However, a pulsed current system is used to eliminate heating of the current leads. The current is turned on for the data acquisition sequence every 50 to 500 cycles, depending on the FCGR. The data acquisition is typically accomplished in a fraction of a second to a few seconds, hence reducing the heat up time of the leads and the junctions appreciably [103, 102, 110].

An automatic voltage zeroing technique is performed at both the reference and test specimens simultaneously [94, 102, 110, 104]. The voltage is measured with the current on then with the current off and the difference of the two readings is used to automatically eliminate the effects of amplifier offset and thermocouple induced voltages at the dissimilar metal junctions and other external interference which may be present in the laboratory.

The DC-PD technique was used in conjunction with a digital computer. A computer programme was written [111] such that the potential drop data is acquired during a maintained maximum load plateau, Figure 3-4, to consistently measure the maximum crack extension and to eliminate any crack closure effects, as well as any short circuit across the cracked surfaces. Furthermore, any magneto-mechanical effect is eliminated by halting the cycling during data acquisition [103, 112]. The voltage measurements can be significantly altered by the magneto-mechanical effect or the "inverse effect of magnetostriction", Figure 3-5 [113], which are the result of the magnetic domains in the material rotating under stress such that the direction of magnetization becomes parallel to the direction of deformation [103]. In fatigue tests, however, the deformation is cyclic in nature thus causing a cyclic variation in the magnetic intensity. This in turn can induce a current in the specimen which affects the measured voltage. This effect is not neutralized just through the use of a reference specimen, since it is not cycled during a test. However, by maintaining the test specimen at a constant load during data acquisition this effect is

eliminated [103].

The computer system used had the capability of measuring the induced potentials at a rate of 10 to 100 times per second. By using this feature, an average voltage across the reference and the test specimens is obtained, and any noise or voltage spikes which may occur is smoothed out. The number of data acquisitions per second depends on the data acquisition time (e.g. 10 or 20 data acquisitions per second for 2 or 1 seconds, respectively). The net effect of these modifications is a temperature, environment, and current independent and dimensionless voltage ratio (VR) given by,

$$VR = \frac{(V_{on} - V_{off})_{\text{sample-specimen}}}{(V_{on} - V_{off})_{\text{reference-specimen}}} \quad (3.2)$$

where V_{on} and V_{off} are the measured voltages with current on and off, respectively.

3.3.2 Instrumentation

The instruments used consisted of a dual voltage recording device, a constant current power supply, and two amplifiers. These instruments were interconnected with copper and stainless steel wires to the test and the reference specimens in accordance with the circuit diagram shown in Figure 3-2. The voltage recording device used in this work was a "Digital" PDP-11/04 series computer. As well as performing the data acquisition for the tests, the computer was also used to turn the power supply on and off as required. This was accomplished by supplying a ten volt ramp to a switching unit. This unit activated a normally open relay to turn the power supply on and off. The amplifiers were used to increase the measured mV (millivolts) to a value more compatible with the 0 to 10 volt range of the analogue to digital converter used to interface with the computer. The amplifiers were "Accudata" model 218 which are bridge type amplifiers. The gain setting for each set of tests at room and elevated temperature is given in Table 3-7. A "Lambda" power supply was used as the constant current source for the system, supplying a current to induce a voltage drop across the crack front of the specimen. Current densities of about $0.1 A/mm^2$ are needed for acceptable accuracy in the DC-PD system [98]. Therefore the amount of current depends on the size and geometry of the specimen, the material properties, and the equipment capabilities. For the compact tension specimens used in this

study, 0.1 A/mm^2 current density translates into a current magnitude of approximately 10 amps at pre-crack and as the crack grows, the current density increases.

In this work, the potential drop technique was calibrated optically, thus necessitating the use of a travelling microscope. Two "Geertner" telemicroscopes were used for this purpose in conjunction with a model 7109-C-46 open type micrometer slide. The slide had a range of 50.00 mm and minimum graduations of 0.01 mm. The tracking system also came with an attachable lamp for illuminating the specimen and crack tip.

Other equipment directly associated with the operation of the potential drop technique include electrical connectors, insulating materials, potential probes and current leads.

3.3.3 Potential Probes and Current Leads

3.3.3.1 Optimum Locations

The position of the current leads on CT specimens is reported to be important, since large differences in sensitivity and reproducibility may result due to small changes in lead positions. High sensitivity, that is large changes in potential drop per unit increase in crack length, is produced with the current leads near the crack plane, Figure 3-6(a). In such a case, the calibration is very sensitive to lead position. By moving the current leads further out, the loss in sensitivity is increased, Figure 3-6(b). This drawback is compensated for by an increase in reproducibility. For the compact tension type specimens, the loading holes also have an influence on the potential field near the crack plane. The most popular location of the current leads appears to be along the side faces, designated as I_3 and O_3 in Figure 3-7.

The potential probes are usually placed adjacent to the crack plane as shown by points, P in Figure 3-7. The position P_1 on the specimen centreline has been most widely used but the diagonal positions, P_2 , are also favoured because of the belief that the effects of crack front irregularity are averaged out [114]. However, locating the probes this close to the corners isolates them from the active current flow and potential distribution.

A general rule regarding the positioning of current and probe leads on CT specimens, is that those lead positions which increase sensitivity also result in the largest errors in crack length predictions due to positioning errors. It was decided to locate the current

leads on the side faces (I_3, O_3) and the potential probes on the specimen centreline (P_1) for the A-286 and Inconel 718 (4.75 mm thick) and the diagonal positions (P_2) for the AM-355 (0.94 mm thick) specimens, Figure 3-8.

3.3.3.2 Current Leads and Probes Material, Size, and Attachment Methods

The preparation, material type, and gauge size of the potential probes and the current leads depend on the specimen size, current density, environment, test temperature, type of material, test duration, and method of connection to the specimen. The type 302 stainless steel wire was selected to reduce the thermocouple effects at potential probe/specimen junctions; this is not critical since any mV induced due to thermocouple effects are compensated for by using a reference specimen. Table 3-8 lists the gauge size of the stainless steel leads used for room and high temperature tests. The gauge size for high temperature tests was selected in a manner such that the amount of surface oxidation did not affect the precision of the method. The leads were covered with insulating fibreglass sleeves to reduce the oxidation effect as much as possible.

The potential wires leading from the specimen potential leads, to the amplifier inputs, were twenty-two gauge copper wires. These wires were connected to the amplifiers through five pin plugs designed for use with the amplifiers. In the room temperature tests, the stainless steel probe wires were soldered to the copper amplifier wires. However, this was impossible in the elevated temperature tests because test temperature was above the melting point of the solder. This problem was rectified by welding longer potential probes to the specimens and making the junction between the two wires at a terminal block. This by far proved to be the superior method since soldering stainless steel is rather difficult at best. The major drawback was the added resistance of the current leads to the circuit, causing a large increase in the voltage requirement of the power supply to maintain a constant current. Therefore the leads were made as short as possible.

Accurate attachment of the probes and the leads to the specimen was accomplished by the use of a laser welding technique which allowed accurate placement of the leads with a minimum weld size and heat affected zone. In order to increase the handling and the durability of the wired A-286 and Inconel 718 specimens, holes with a depth of twice the current lead diameter (2×0.048 inch) were first drilled (with a tungsten carbide twist

drill no. 56 - 0.0465 inch in diameter) and then centre drilled (with a no. 1 centre drill) in I_3 and O_3 positions, Figure 3-7. The current leads were then fitted into the holes and laser welded to prevent any lateral movement of the leads with respect to the specimen. This procedure proved to be very successful in terms of the potential probe and current lead location reproducibility and handling of the specimens. Other methods of attachment such as spot welding and the use of screws were explored but did not seem feasible for this application.

3.3.4 Electrical Grounding and Isolation

3.3.4.1 Amplifier Grounding and Shielding

The DC potential drop system uses two high impedance differential amplifiers to amplify the potential drop across the specimens such that the output signal from the amplifiers would be within the range of the data acquisition system and at the same time increase the sensitivity of the overall system to pick up small changes in crack length. In order to obtain such accuracy and precision of electrical measurements in environments where ground-loop currents are large and electrical or magnetic interference sources abound, proper techniques in system wiring must be employed. Twisted, shielded cable had to be used to connect the amplifier to the signal source, and the shield had to be operated at the same AC common mode voltage as the signal source. Also, wiring and grounding of the amplifier outputs had to be done in a manner such that ground loop or interference problems were not introduced on the output side of the amplifiers.

The ground and shield connections for the DC-PD system, shown in Figure 3-9, set up on the NAE/NRC MTS frame "M-13D", resulted in accurate amplified potential drop measurements with a gain setting between 300 to 1500, Table 3-7. It should be noted that erroneous results are obtained by grounding the shields at the amplifier instead of grounding them at the signal source. The wiring of the outputs in the common ground environment of this system must be done in a manner shown in Figure 3-9 in order to eliminate ground-loops and interference problems in this environment.

3.3.4.2 System Isolation from Load Frame

In order to take advantage of the full sensitivity and current capability of the DC-PD

system, the insulation between the specimen and the load frame (i.e. infinite resistance) must be good. The specimen may be insulated by insulating the loading pins within the clevis, by insulating the clevis from the tie rods or by insulating the tie rods from the frame. At first, during room temperature testing, "bakelite" tie rods were used to insulate the specimen from the frame. One of these tie rods failed at about 10^6 -cycles. The second method involved the use of ceramically sputter coated grip pins. This arrangement is good for temperatures up to 650°C . However, it was not deemed to be the optimum solution in this instance, because of the small load bearing surface of the specimens on the grip pins. It was felt that over the course of a test at elevated temperatures, the coatings could either wear off or break off which would affect the potential drop readings. Therefore it was decided to design and manufacture special adapters to isolate the MAR-M246 tie rods from the frame as shown in Figure 3-10. These adapters were made of AISI 4340 steel and heat treated to yield the highest strength. They were then sprayed with a high temperature non-conducting paint. In addition, "nylon-teflon washers and "bakelite" pipes were used to further isolate the tie rods from the adapters. Since these adapters were situated outside the furnace, this apparatus could be used to very high temperatures.

3.3.5 Correlation between Crack Length and Potential Drop

Clearly a prime source of error in using the potential drop method lies in the accuracy of the calibration which is used to relate potential drop to the crack length. Theoretical calibrations are based on solutions to Laplace's equation for specific boundary conditions of a given test piece geometry. This has not yet been produced for compact tension specimens [90]. Numerical solutions, however, from finite element analysis have been obtained for the CT specimen [91]. Other methods include the use of a thin sheet of conducting paper cut to the shape of the specimen [114], specimen "beach marking" [91], and the use of a travelling microscope [90, 91, 114, 103]. In this study, the latter approach was used where calibration was performed by measuring a series of potential drop and crack length pairs using a calibrated travelling microscope. To aid in the visualization of the crack tip, the region of crack propagation was first polished. A focused light attached to the microscope also aided in discerning the crack tip. Readings of the crack length were taken at approximately 0.10 to 0.25 mm intervals in order to provide a consistent calibration

curve. These optically measured crack lengths were input into the computer after which a corresponding potential drop data acquisition sequence was performed. After completion of the test, the calculated voltage ratio was plotted versus a/w , where " a " is the crack length and " w " is the width of the specimen as defined by ASTM standard E-647 [115]. A polynomial was then fitted to this data. This equation was then put into the main programme for the next test.

The crack length on either side of the specimen must be observed if the thickness to specimen width ratio is greater than 0.15 (see ASTM standard E-647), as was the case for the A-286 and Inconel 718 specimens, Figure 3-11. If crack tunneling occurs in the specimen, the optically measured crack lengths must be corrected accordingly. According to ASTM E-647 [115] and E-399 [116], after completion of testing, the fracture surfaces should be examined, preferably at two locations (for example, at the pre-crack and terminal fatigue crack lengths), to determine the extent of crack tunneling. The crack length should be measured to the nearest 0.5% at the centre of the crack front, and midway between the centre of the crack front and the end of the crack front on each surface of the specimen. The average of these three measurements should be used as the crack length. The difference between the average of crack tunneling length and the corresponding optically measured crack length during the test is defined as the crack curvature correction.

If the crack curvature correction results in a difference in calculated stress intensity factor greater than 5% at any crack length, then the optically measured crack lengths must be adjusted before determining the coefficients for the calibration curve. After such correction, the potential drop crack length measurements automatically take the crack tunneling effect into account thus resulting in accurate crack length determination.

Representative samples of A-286 and Inconel 718 CT specimen fracture surfaces were examined and in all cases the crack curvature correction was lower than the specified 5% limit. Hence, no crack tunneling correction was required for these specimens.

3.4 Microstructural Characterization

3.4.1 Optical Microscopy

Samples were obtained for optical microscopy examination from the Inconel 718 turbine discs. Radial sections were machined from these discs, and cut into smaller portions. These specimens were then prepared to reveal the grain boundaries and carbides. They were (i) mounted in "bakelite", (ii) grounded through to 600 SiC paper, (iii) polished on a flocked twill cloth wheel to a 1 μm diamond paste, (iv) etched by dipping in a solution composed of 2 parts H_2O , 1 part H_2O_2 , and 3 parts HCl for a few seconds and (v) washed with water and methanol.

These specimens were then observed and micrographed using a "Leitz" MM6 metallograph. As the etch solution aged, the etching time had to be increased. The overall shelf life of the etch was about 4 hours. After reheat treatment and dimensional stability investigations, which is described in a later Section, similar procedures were employed for optical microscopy examination of reheat treated and fatigued CT specimens. In all cases grain sizes were measured using the ASTM mean linear intercept method [117].

3.4.2 Transmission Electron Microscopy

Transmission Electron Microscopy (TEM) was used to characterize γ' , γ'' , and δ phases, dislocation structures and the grain boundary and carbide morphology of new, service exposed and rejuvenated Inconel 718 specimens. The TEM microscope used was a "Phillips" model EM-201 with a 100 kV electron beam.

The replica and thin foil techniques are the two major specimen preparation techniques used for TEM investigations. Previous attempts at NAE/NRC to replicate the surface of the A-286 specimen using plastic-carbon replicas had not been successful in resolving the very fine strengthening precipitates of the alloy [118]. Hence, it was decided to prepare Inconel 718 thin foils for TEM examination. Foils also have the added advantages of revealing the nature and distribution of defects, as well as being suitable for electron diffraction microscopy. The procedure used to produce foils from Inconel 718 turbine disc material was as follows:

- i) Radial rods, 3 mm in diameter, were lathe-machined from each of the turbine discs. This type of machining eliminates the depth of damage induced in the rod.
- ii) A "Buehler Isomet" low speed diamond wafering saw, was employed to slice the rod into approximately 0.2 mm thick mini-discs.
- iii) The location of each mini-disc from the bore of the turbine disc was then determined.
- iv) The mini-discs were further ground to 0.1 mm thickness using 320 and 600 grit SiC paper.

These mini-discs were then electropolished and perforated in a commercially available "Tenupol" unit, manufactured by "Struers". The electrolyte used had a composition of 94 parts acetic acid (CH_3COOH) to 6 parts perchloric acid ($CHClO_4$, highly dangerous) by volume. The flow rate of the electrolyte was experimentally determined and set at 2.3. A current density of 50-300 mA/mm² was applied at 45-60 V using a power supply.

3.4.3 Scanning Electron Microscopy

A "Noyascan 30" scanning electron microscope (SEM) with a 15 kV electron beam, was used for examining the fracture surfaces. Fracture surfaces of the CT specimens were ultrasonically cleaned in an acetone bath and then mounted on aluminum holders ensuring that the specimens were electrically conductive. The morphology of crack growth was characterized. In order to avoid any ambiguity in the interpretation of the results, stereo pairs were taken, wherever needed.

3.5 Elevated Temperature FCGR Testing

3.5.1 Preparation, Geometry and Orientation of CT Specimens

The geometry of the compact tension (CT) specimens used in this study is shown in Figure 3-11. All the dimensions conformed to the ASTM Standard E-647 minimum size recommendations [115]. Since the CT specimens were machined from retired components their thicknesses were limited by the minimum thickness of the turbine discs. This corresponded to the 4.75 mm (0.187 inch) dimension in the illustration shown in Figure 3-11. The CT specimen was selected for this application because its configuration allowed the

maximum number of specimens to be machined from each disc [119], while maintaining a constant radial position from the bore, and the appropriate direction of crack propagation. By machining the CT specimens from the disc as close to the bolt hole as possible, the microstructural degradation which occurs around the bolt hole in service is automatically taken into consideration. One of the rejection criterion defined by GE is any indication of a crack at a bolt hole of a disc. To simulate the FCGR behaviour of service induced cracking as closely as possible, the CT specimens were machined in a manner such that the direction of crack propagation through the specimen was radial towards the bore of the disc. This orientation is illustrated in Figure 3-12.

3.5.2 The FCGR Test Programme and Parameters

The FCGR tests were conducted on the Inconel 718 and A-286 specimens using a sawtooth waveform as shown in Figure 3-13. This loading cycle was selected because it was deemed more severe than a comparable sine waveform, and because it more closely matched the trapezoidal waveform [120]. Tests were also conducted on A-286 with trapezoidal waveform. A hold time of 20 seconds was imposed at the peak cyclic load, as shown in Figure 3-13. The 20 seconds hold time was determined from mission profile analysis of J85-CAN-40 engine [121] to be the length of a cruise portion of a specific mission profile. The rate of loading and unloading remained essentially the same for tests using either of the waveforms, Figure 3-13. Consequently any observed hold time effects on FCGR would be attributable to the time spent at peak load (or hold time). As a result, the frequency was 0.05 Hz for the trapezoidal waveform, compared to 1.00 Hz for the sawtooth waveform. For all tests, the load ratio, R , was 0.1 and the maximum load applied was 5.0 kN.

The room temperature tests were performed at a temperature of 24°C while the elevated temperature tests were performed at 550°C for A-286 and 650°C for Inconel 718 CT specimens. The relative humidity varied during the test, and from one test to another, from 50 to 100% whereas the test temperature varied within $\pm 3^\circ\text{C}$.

All CT specimens required a fatigue pre-crack of 6.67 mm (a/w of 0.2625). This crack length corresponds to a stress intensity factor range of $30.3 \text{ MPa}\sqrt{\text{m}}$. The final crack length varied, as the yield and ultimate tensile strength of the specimen varied,

in order to meet the elasticity requirements set by ASTM standard E-647 [115]. These final valid crack lengths and their corresponding stress intensity factors for each type of specimen are given in Table 3-9. The stress intensity factor was calculated from the following equation:

$$\Delta K = \frac{\Delta P(2 + \alpha)}{B\sqrt{w(1 - \alpha)^{1.5}}} [0.886 + 4.640(\alpha) - 13.320(\alpha)^2 + 14.720(\alpha)^3 - 5.600(\alpha)^4] \quad (3.3)$$

where ΔP is the load range in N , B is the specimen thickness in mm , w is the width of the specimen in mm , $\alpha = a/w$, a is the crack length in mm and ΔK is the stress intensity factor range in $MPa\sqrt{m}$.

Most specimens were pre-cracked using the same loading, frequency, and environmental conditions as used in the actual tests. Some of the A-286 room temperature fatigue pre-cracks were carried out at a frequency of 5 Hz to reduce the time for pre-cracking. A few of the Inconel 718 specimens were pre-cracked at room temperature to reduce any effect of ageing during the pre-crack time, which varied from 1 to 4 hours depending on the specimen. The FCGR testing of A-286 CT specimens was performed in three phases, Table 3-1. First, the FCGR's of the reheat treated first stage turbine disc with 1772 hrs TSN were determined at room temperature by testing five specimens. These tests were also used to calibrate the room temperature application of the DC-PD technique. Second, three and two specimens taken from two second stage service exposed turbine discs with 3737.6 and 1953.5 (also referred to as 3738 and 1954, respectively) hrs TSN, respectively, were also tested at room temperature. The third phase of testing involved specimens from the second stage reheat treated 2211 hrs TSN turbine disc, at elevated temperature, using three specimens with sawtooth waveform, and two specimens with a hold time at maximum load.

The FCGR testing of the Inconel 718 CT specimens were also performed in three phases, namely the evaluation of the FCGR properties of new, service exposed, and reheat treated first stage turbine discs, Table 3-5. It is well known that FCGR's show a large amount of scatter due to the inherent scatter in material properties. In order to ensure that the test data obtained was statistically significant and representative of the material, test conditions and environment, four CT specimens were used in each set of condition.

3.5.3 Test Equipment

An MTS 810 automated fatigue test system was used in this study, Figure 3-14. The series 810 system consisted of a 100 kN fatigue rated two column frame, a 25 kN fatigue rated actuator, and a 15 kN rated single bridge load cell, plus the required hydraulic hoses, couplings and fittings. As the Materials Laboratory of NAE/NRC operated several similar systems at the time of acquisition, the new system was connected to the existing hydraulic pump and associated systems.

The system was monitored and controlled by an 810A electronic control package, Figure 3-14. This package consisted of a type 422 controller, a type 417 control panel, a type 413 master control panel and a type 464 data display unit. A type 433 processor interface was also used to allow the system to be computer controlled. Hence the load sequence was generated by a computer software which is explained later.

The system was capable of operation in any one of three modes; load control, strain control, or stroke control. All FCGR tests were conducted in load control. The system provided for several ranges of full scale load cell capability. The ranges offered included 100, 50, 20 and 10% of the load cell rating. In this work, a range of 50% was used giving an effective load cell rating of 7.5 kN. Thus, the required 5 kN load was well within the accurate range of the load cell.

The grips used to hold the CT specimens in the load frame were designed according to ASTM standard E-647 specification [115]. The grip material was Nimonic 105, a Ni-base superalloy which possesses good elevated temperature strength. The pins were manufactured from MAR-M200, also a Ni-base superalloy with excellent elevated temperature properties such as strength, toughness and resistance to creep. This was proven in practice as all tests were conducted with the same set of pins at room and at elevated temperatures. The pins were held on to the grips with "C" type clip retainers on either end. This is not strictly necessary as these fatigue tests were conducted in tension-tension mode of loading. The load train was also equipped with a universal joint to ensure correct alignment of the load train under loading. Extension rods of MAR-M246 were used when the furnace was used at elevated temperatures. The DC-PD adapters used to insulate the

load train from the specimen were designed and manufactured from AISI 4340 stainless steel and heat treated³ for optimum strength.

3.5.4 The Computer Control and the Computer Programme for FCGR Testing

The computer used to control the loading and data acquisition sequences for the FCGR testing was a Digital PDP-11/04, Figure 3-14. Specific functions, such as turning the power supply on and off for the DC-PD crack monitoring system, and controlling the load, were performed through an MTS 433 processor interface. The computer operated two disc drives; one for operating the software and the other for data storage. Access to the computer was provided by either a Tektronix graphics display terminal or a Digital Decwriter III, Figure 3-14. The Decwriter also functioned as a hard copy printer.

The software used in this work was a modified version of the original programme written and implemented at NAE/NRC, and is described elsewhere [111, 122, 89]. It was written in MTS Basic, a modified form of basic employed by the MTS company for their test systems. The software prompts the user for specimen specifications and test parameters, and provides a check list to ensure the correct set up of the hydraulic and DC-PD systems.

In the initial calibration runs, optically measured crack lengths were entered into the computer, and these data were used to determine a calibration curve. The calibration curve was obtained by fitting a polynomial through the data. The expression for the polynomial is of the form:

$$\frac{a}{w} = X_0 + X_1 \times (VR) + X_2 \times (VR)^2 + X_3 \times (VR)^3 + \dots \quad (3.4)$$

where a/w is the nondimensional ratio of crack length and specimen width, VR is the ratio of the differences in the potential drop readings for the sample and the reference specimens, with the power supply on and off as given in Equation (3.2), and $X_0, X_1, X_2, X_3, \dots$ are the coefficients of the fitted curve.

³ 843°C / 1hr / Oil Quench + tempered at 650°C / $\frac{3}{4}$ hr / Air Cooled.

Upon storing the coefficients of the polynomial in the main programme, crack lengths are automatically calculated by the programme by determining VR and substituting it in Equation (3.4) to obtain a/w .

Once the main test has been initiated, the programme performs its first data acquisition to provide the initial potential drop readings and consequently convert them to crack length based on the calibration curve. The maximum and minimum loads and the number of cycles completed are also provided. The computer software then cycles the specimens through its closed loop hydraulic system according to the function generation system specified. This can be modified to allow the use of various loading cycles including sawtooth, sine and trapezoidal waveforms.

The loading sequence is then continued for the specified number of cycles, between DC-PD data acquisition. After completing the number of cycles specified, the function generator is once again stopped at maximum load to allow for data acquisition, Figure 3-4.

The data from each data acquisition cycle is stored on the data disc to free the hardware memory for the next cycle. The programme is written so that the cycling may be stopped by the operator at any time to either change a parameter or input an optical crack length measurement. The test can also be stopped permanently at any time, but it is set up so that it can be resumed with the data contained in the same data file.

The software developed also includes a data reduction programme which takes the data from the data disc and calculates the stress intensity factor from Equation (3.3), and FCGR on the basis of crack length function derived from the crack length measurements, as outlined in the ASTM standard E-647 [115]. This computation involves the calculation of the slope between two adjacent points on the curve of crack length versus the number of cycles. The computer fits a least squares curve through the data points that conform to the minimum crack extension criterion. This crack length interval criterion can be input by the user, and is usually taken as 0.25 to 0.50 mm. The slopes between these points are then determined.

The Paris relationship [123] between the crack propagation rate and the stress intensity

factor,

$$da/dN = C(\Delta K)^n, \quad (3.5)$$

for each test group, is determined in a subroutine using the least squares method of linear regression [124]. Use of this technique to determine the parameters in the Paris relationship for FCGR is quite common [125, 126]. The line determined using this method is referred to as the mean line in this study.

Besuner [127] reports that C varies according to a "log-normal" distribution in FCGR testing. Therefore, da/dN varies "normally" with ΔK , since the data points are plotted on a log-log scale, and the variance is the same for any given ΔK value [128]. In order to account for the scatter of the data, upper bound lines with 99.9% confidence bounds are determined using the following equation:

$$\text{Upper Bound}_{99.9\%} \text{ Line} = \text{Mean Line} \pm 3 \times S_{y/x} \quad (3.6)$$

where 3 is the normal standard variate for a confidence interval of 0.999 [128] and $S_{y/x}$ is the conditional standard variance given by:

$$S_{y/x} = \left[\frac{S_{yy} - \frac{S_{xy}^2}{S_{xx}}}{k-2} \right]^{1/2} \quad (3.7a)$$

where k is the no. of data points and S_{yy} , S_{xy} and S_{xx} are given by:

$$S_{yy} = \Sigma y^2 - \frac{(\Sigma y)^2}{k} \quad (3.7b)$$

$$S_{xy} = \Sigma xy - \frac{\Sigma x \Sigma y}{k} \quad (3.7c)$$

$$S_{xx} = \Sigma x^2 - \frac{(\Sigma x)^2}{k} \quad (3.7d)$$

This assumption results in an upper bound confidence line which is also a log-logistic curve with the same slope value (n) as the mean line from linear regression analysis but with a different intercept (C).

3.6 Tensile Testing

Four sets of A-286 specimens were used in tensile tests at room and elevated temperatures. Miniature tensile specimens were machined from the J85-CAN-40 turbine discs in

the radial direction, Figure 3-12. All specimen dimensions conformed to ASTM standard E-8 [129] as shown in Figure 3-15(a). Tensile specimens were also machined from the J85-CAN-15 turbine discs according to the dimensions shown in Figure 3-15(b), which are in agreement with the ASTM standard E-8.

All tests were performed on an MTS 810 Electro-Hydraulic Test Machine at stress rates of up to 690 MPa/min while in load control, and at strain rates of 3×10^{-3} to $3 \times 10^{-2} \text{ sec}^{-1}$ while in stroke control. The strength properties and ductility of A-286 alloy are reported not to be greatly influenced by strain rate in the range of 3×10^{-6} to 1 sec^{-1} for test temperature from room to 538°C [130]. Strain rate values given here are based on the constant crosshead velocity and the initial specimen gauge length. Two methods were used to produce engineering stress-strain curves. For the room temperature tests, the output of a conventional strain gauge load cell was used for determining specimen load, whereas the specimen extension was measured using both an MTS electronic strain gauge recorder and the MTS crosshead displacement (or the stroke). Load, strain and displacement values were stored on a floppy disc and then plotted using a software programme. The tensile properties of interest were determined from these plots. For the elevated temperature tests, the specimen extension was determined only from the cross head displacement. The MTS electro-hydraulic test machine and the electronic strain gauge recorder provided load and strain measurements by means of a piezoelectric load cell and strain gauge bridge, respectively. The crosshead displacement provided the specimen extension by means of a linear variable differential transformer (LVDT) on the actuator. Voltage outputs from the load cell, strain gauge and LVDT were recorded with a Digital PDP11/04 computer and then translated to load, strain, and displacement data. Ductility values calculated were obtained from pre- and post-test specimen measurements. All tests were conducted in air using recommended ASTM practices.

3.7 Dimensional Stability of the Turbine Discs

To investigate the dimensional stability of the J85-CAN-15 turbine discs, the first and second stage crack-free, service exposed discs were dimensionally inspected at critical locations. These results were later compared to the results obtained from a dimensional

check of the rejuvenated discs and the GE blueprint drawings [131]. The amount of distortion, which may have occurred during the service or rejuvenation treatment was compared in terms of the error in

- i) the flatness of both first and second stage discs,
- ii) cylindricity of the hub in the first stage discs,
- iii) the perpendicularity of the axis of the hub to the plane of the disc also in the first stage discs, and
- iv) firtree slots.

Dimension locations are shown in Figures 3-16 to 3-18. The measurements were made on a Brown and Sharp validator coordinate measuring system using the VAL-MEAS-II software programme. Fourteen points were measured on surface "C" of each disc and averaged to obtain its flatness.

Chapter 4

Results

4.1 Elevated Temperature DC-PD Technique

4.1.1 Power Supply Response

An ideal power supply would be a step shape current source (i.e. zero peak time, no overshoot, and stable output). Three power supplies were tested in order to determine the response time required to give a stable current. Two of these response times have been compared in Figure 4-1. The Lambda power supply was considered to be best suited for the PD technique because of its minimum peak time and good current stability. The Sorensen power supply would introduce inaccuracies due to its rippled current output and slow response time to current stabilization. Although the rippled current output of the Sorensen power supply cannot be eliminated, its slow response time may be overcome by supplying the current to a dummy load first and once the current has peaked, it could then be switched on to the test and reference specimens.

The Lambda power supply has a current output capability up to 13.5 A and a voltage output of up to 20 VDC. A current of 10 A was used for all experiments unless otherwise specified. Using a current of significantly smaller magnitude could introduce inaccuracies and reduce the precision of the technique, while significantly higher currents could create a heating problem in the current leads attached to the specimens. Although it was verified that the magnitude and consistency of the supplied current did not affect the accuracy of the results as shown in Figure 3-3, the magnitude of the current did not vary during a test despite any minor variations in the resistance of the test specimen due to crack growth. This was taken into account by automatic variation of the supplied voltage by the power supply. The current on and off times for data acquisition sequences are given in Table 3-7. The current was first allowed to stabilize and then voltage readings were taken and averaged. The shorter data acquisition times employed in elevated temperature tests

were chosen to minimize the contribution of creep deformation due to hold time effects at maximum load.

4.1.2 Calibration Curves Between Crack Lengths and DC-PD Technique Output

The first set of tests at room temperature was conducted on AM-355 CT specimens by Aker and Terada [97]. Since the specimen thickness was only 0.94 mm (0.037 in), the potential probes were laser welded on the diagonal positions (P_2) of the CT specimens. This allowed excellent reproducibility during their experiments. The equation of the calibration curve was determined to be:

$$a/w = -0.5611 + 0.9884(VR) - 0.7579(VR)^2 + 0.0359(VR)^3 \quad (4.1)$$

Three AM-355 CT specimens of the same geometry, size, thickness and material with identical lead and probe locations were tested at 550°C to examine the validity of using the room temperature calibration curve at elevated temperatures. A current of 10 A was used for test specimen nos. 1 and 3, while only 4 A was used for specimen no. 2. The crack lengths were measured optically by stopping the elevated temperature test and opening the furnace to expose the specimen surface. These results are compared in Figure 4-2 and tabulated in Table 4-1. Excellent agreement is seen between the optical crack length measurements and the calculated crack lengths using the room temperature calibration curve described by Equation (4.1).

A second set of tests was conducted on A-286 CT specimens at room temperature. The CT specimen thickness was 4.75 mm (0.187 in) which allowed the potential probes to be attached to the specimen centreline accurately (position P_1 , in Figure 3-7). Four CT calibration tests were performed. Of these four tests, two calibration tests resulted in erroneous readings since the calibration curves were shifted to the right as shown in Figure 4-3. This inconsistency was later corrected by adjusting the zero and the gain on the amplifiers such that the tests started at a VR value of unity in all cases. The remaining tests were deemed valid and VR was plotted against crack length to specimen width ratio, a/w ; Figure 4-4, and a polynomial was fitted through the data. A nondimensional calibration polynomial equation, independent of temperature, environment, and current

is given by:

$$a/w = -0.471496 + 0.785050(VR) - 0.114153(VR)^2 \quad (4.2)$$

A plot of crack lengths determined with the help of potential drop versus optically measured crack lengths for all subsequent tests is presented in Figure 4-5. These data fit through a 45° straight line thus indicating that the DC-PD technique and optically measured technique show identical crack lengths.

The room temperature calibration curve was then used to obtain the crack lengths at 550°C on A-286 CT specimens. During the test, the crack lengths were also periodically monitored optically by stopping the test, opening the furnace, and measuring the crack length using the travelling microscope. The results are shown in Table 4-2. There is excellent agreement between measured and predicted crack lengths.

4.1.3 Effect of Material Type on the Calibration Curve

To investigate the influence of material variation on the calibration curve, FCGR tests were also conducted on the Inconel 718 CT specimens. These Inconel 718 specimens were of the same geometry and thickness as the A-286 specimens. The current leads and potential probe locations for Inconel 718 specimens were also identical to those used for A-286 CT specimens. The tests were carried out at 650°C and Equation (4.2) was used for PD predicted crack length estimates. Cracks were also monitored optically using the travelling microscope. These results are shown in Table 4-3. Again, excellent correlation between the optical and PD crack lengths is observed. Therefore, it can be concluded that the room temperature calibration curve is indeed valid for elevated temperature testing and unlike the capacitance gauge technique the calibration curve does not change with different materials provided (i) a reference specimen is used in conjunction with the test specimen, and (ii) the specimen geometry is identical to the one used to obtain the calibration curve.

4.1.4 Precision and Accuracy of the DC-PD Technique

A comparison of the CT calibration curve with those obtained for different material and specimen size and thickness is shown in Figure 4-6. The information regarding each

curve is given in Table 4-4. Each data set was passed through a voltage ratio of 1.00 and a/w of 0.20, and a second order polynomial was fitted to the data points. The coefficients for each polynomial are given in Table 4-4.

The term *precision* is used to indicate the closeness with which the predicted crack lengths by DC-PD agree with one another quite independently of any systematic error involved [132]. ASTM standard E-647-83 defines the precision of a crack length measuring system as "the standard deviation on the mean value of crack length determined for a set of replicate measurements". To determine the precision of the adapted DC-PD method, crack lengths were measured at 100 and 500 cycle intervals for elevated and room temperature tests, respectively, before a crack was initiated from the starter notch. For the room temperature tests only data obtained after the first 1500 loading cycles was considered in determining the precision of the technique because of the increase in measured potential resulting from the formation of a plastic zone ahead of the starter notch of the specimen. The standard deviation from the mean was then calculated from eight data sets that were collected on specimens tested at room and elevated temperatures, Table 4-5. The standard deviation was determined from

$$S.D. = \sqrt{\frac{1}{k-1} \left[\sum x_i^2 - \frac{(\sum x_i)^2}{k} \right]} \quad (4.3)$$

where $S.D.$ is the standard deviation, k is the number of data points, and x_i is a data point.

The term *accuracy* refers to the closeness of the crack length predicted by the DC-PD to the "actual" value or the "real" value of the crack length [132]. The standard deviation from the mean was calculated from the deviation in crack lengths between the optical and the potential drop crack length readings, Tables 4-1 to 4-3. In all cases, the accuracy was evaluated during the valid crack length region and are given in Table 4-5.

4.1.5 Probe Location and Thickness Effects on the Calibration Curve

To study the influence of different potential probe locations on the calibration curve, a selected number of AM-355 CT specimens, with potential probes welded at the specimen centreline, were tested at room temperature. The potential probes in the reference

specimens were also located on the specimen centreline. Figure 4-7 compares the optical crack length measurements for these tests with the previously established calibration curve for AM-355 specimens (given by Equation 4.1) where potential probes were located on the diagonal positions of the test and the reference specimens. It is noteworthy that the calibration curve does not vary with the potential probe locations for the thin AM-355 CT specimens at room temperature.

To evaluate the effect of specimen thickness on the calibration curve, Figure 4-8 compares the optical crack length measurements for the above mentioned AM-355 CT specimen tests with the calibration curve for thicker but similar geometry A-286 CT specimens (calibration curve represented by Equation 4.2) where potential probes were located on the centreline positions of the test and the reference specimens. The crack lengths calculated from the calibration curve of the A-286 specimen is noticed to deviate from the optically measured crack lengths of the AM-355 specimens even though the potential probe locations are adjusted proportionally. These findings suggest that the calibration curve is dependent on the CT specimen thickness. However, it should be pointed out that the error between the optical and potential drop crack length measurements is less than 4% which is within the crack lengths variation associated with the DC-PD technique.

A finite element analysis was carried out to obtain a voltage contour map of a CT specimen using the "ASGARD" finite element package available at Carleton University [133]. The results from this study are presented in Figure 4-9: It can be seen that the induced voltage across the notch edge is constant as shown by contour "A" in this figure. This verifies the experimental results obtained by changing the potential probe locations from the diagonal positions to the centreline positions of the 0.94 mm thick CT specimen as shown in Figure 4-7.

4.1.6 Effects of Microstructural Variation

The effect of microstructural variation between the test and reference specimens on the measured crack lengths was investigated. Four test specimens were tested in turn, in conjunction with similar and dissimilar microstructure reference specimens. FCGR's were determined and are shown in Figure 4-10. It can be seen that no significant effect

is observed on the FCGR upon using test and reference specimens with slightly different microstructures.

4.2 Microstructural Characterization of J85-CAN-15 Turbine Discs

4.2.1 The New Turbine Disc

This disc was rejected by Orenda due to a machining error and therefore was never put in service. It is also reported that the disc may have been damaged by being hit by a hammer in the rim area.

A 10 mm thick radial section was cut from the disc and this section was further cut into five smaller samples that were examined by an optical microscope. The new disc revealed a banded grain structure and these banded regions contained large numbers of grain boundary NbC carbides, Figure 4-11. Average grain sizes from bore to rim area were quantitatively determined from the optical micrographs, Figure 4-12, and the grain size values as a function of sample's location are shown by the "dashed line" in Figure 4-13.

TEM examination revealed very fine γ' (about 950 Å in diameter¹) in the γ matrix, Figure 4-14(a). Dislocation pile ups at the grain boundaries, Figure 4-14(b), along with a variety of other dislocations activity within the matrix was also observed. Second phase particles appeared to have acted as dislocations, Figure 4-14(c), and homogeneous dislocation structure within the matrix were also observed, Figure 4-14(d). They are apparently created from movement of edge and screw dislocations and their interactions [134]. Areas showing dipole trails and independent dislocation movement were also identified and these effects are said to result from screw dislocation mobility and their jog heights [134]. Some areas of the disc showed continuous networks of grain boundary carbides, Figure 4-14(e). Electron diffraction analysis of these carbides identified them as $M_{23}C_6$ type carbide. The amount of δ needle phase was negligible in the new disc.

¹ 1 Å = 1×10^{-8} cm.

4.2.2 Service Exposed Turbine Discs

4.2.2.1 Medium Time Turbine Discs

Both the first and second stage turbine discs of an engine with 1520.3 (also referred to as 1520) hrs TSN were examined. As mentioned earlier, all turbine discs, including these, were retired by Orenda in a crack free condition as determined by the industrial LPI technique.

Average grain sizes from bore to rim were quantitatively determined for the first and the second stage discs, and the results are shown as "dash-dot lines" in Figures 4-13 and 4-15, respectively. Optical microscopy indicated an equiaxed grain structure with many anneal twins as shown in Figure 4-16. No obvious signs of microstructural degradation could be seen in these discs by optical microscopy.

Therefore, thin foils were prepared from the bore and bolt hole areas of each disc, and examined by TEM. There were no distinguishable differences in either of the four groups of specimens. Hence all results were grouped together and are referred to as the "medium time disc". Figure 4-17 shows the typical γ' and γ'' precipitation morphology. The average diameter of the γ' precipitates was measured to be approximately 200 Å. A number of *cellular-type dislocation* structure, Figure 4-18(a), were observed near the bore area. Only one specimen from the bolt hole area showed some evidence of slip banding, Figure 4-18(b). This was not a general feature of the medium time service exposed discs. A large amount of δ phase was also observed to be present in the disc bore. Some areas of the disc revealed a unique *grain boundary* δ needle phase, as shown in Figure 4-19(a), that appeared to have resulted in a serrated grain boundary morphology, Figure 4-19(b).

4.2.2.2 High Time Turbine Discs

Both the first and second stage turbine discs of an engine with 2044.9 (also referred to as 2045) hrs TSN were examined. Optical microscopy revealed a necklace-type grain structure for the second stage disc, Figure 4-20(a), whereas the first stage disc contained an equiaxed grain structure, Figure 4-20(b). The latter microstructure essentially consisted of larger warm-worked grains surrounded by a shell of finer recrystallized grains. Considerably fewer annealing twins were observed in these discs. Average grain sizes from

bore to rim area were also measured for the first and the second stage discs and these results are plotted in Figures 4-13 and 4-15, respectively. Again no obvious signs of microstructural degradation could be detected by optical microscopy.

Thin foils were prepared from bore and bolt hole areas of both discs and examined by TEM. As in the case for the medium time disc, there were no apparent differences between either the bore/bolt hole area of the same disc or the first/second stage discs. Therefore they were all grouped together as the "high time disc".

Figure 4-21(a) shows the γ' and γ'' precipitates of the disc. The average diameter of the γ' precipitates was approximately 200 Å. A large number of grain boundary carbides and Ni_3Nb needles were also observed to be present in these discs, Figures 4-21(b) and 4-21(c). A large number of slip lines and *dislocation arrays* were formed in the bore and bolt hole areas of all high time discs, Figure 4-22. Many *low angle boundaries* were observed in the bore region of this high time disc and typical micrographs, showing the sub-structure, are shown in Figure 4-23.

4.2.3 Rejuvenated Turbine Discs

The microstructure of one first and one second stage rejuvenated disc, with 1523 and 1617 hrs TSN, respectively, was examined. The optical micrograph in Figure 4-24, shows the grain structure of the rejuvenated second stage disc. It consists of both small and large grains. The average grain sizes were determined from bore to rim area and these are plotted in Figure 4-15. TEM examination of thin foils taken from the 1523 hrs TSN rejuvenated disc indicated the presence of dislocation activity in the rim area which was usually prevalent in the high time service exposed discs, Figure 4-25. The sub-grain boundary map, shown in Figure 4-26 is typical of the many sub-grain boundaries present in the rim area of the 1617 hrs TSN rejuvenated disc.

4.2.4 Hardness Profiles

To investigate the strength variation in new, service exposed and rejuvenated discs, the hardness profiles from the bore towards the rim of all discs were measured. The surface hardness was also measured to indicate the effect of surface oxidation, if any, on the disc hardness. Figures 4-27, 4-28, and 4-29, show the hardness of the first and second stage

discs, and the surface hardnesses from the bore to the rim area, respectively. Except for the high time disc the surface hardness values of all service exposed and rejuvenated discs were within the material specification, Figure 4-29.

4.3 FCGR of the New Disc

4.3.1 Turbine Discs from J85-CAN-15 Engine

The 650°C FCGR test results for the Inconel 718 compact tension specimens taken from the new disc, are plotted on a log-log scale (with the abscissa as the stress intensity factor, ΔK , in $MPa\sqrt{m}$, and the ordinate as the amount of crack extension per cycle, da/dN , in m/cycle) in Figure 4-30. This figure shows the actual data points and the regressed mean line and the 99.9% upper bound line. The equation for the mean line within the Paris law region is given by:

$$da/dN = 1.44 \times 10^{-9} (\Delta K)^{1.81} \quad (4.4)$$

The C and n values for the upper bound line and the correlation coefficient, "R", which measures how well the curve fits the data, are given in Table 4-6.

Figure 4-31 compares the mean line results with those of James et al.[135]. Their tests were conducted in air at 650°C at a frequency of 0.67 Hz and a load ratio of 0.05. Their specimens were taken from 1.5cm x 4cm bar stock produced by vacuum induction melting and vacuum arc remelting. Note that this is the same heat as the GE turbine disc forgings. This heat produced grain sizes of ASTM No. 11½ with coarse δ precipitation and MC carbides at the grain boundaries. The FCGR results of the present investigation are lower than those reported by James, Figure 4-31.

4.3.2 Turbine Discs from J85-CAN-40 Engine

There was no J85-CAN-40 new disc available at NAE/NRC for FCGR testing. Figure 4-32 shows the FCGR data obtained by James [136] for the virgin A-286 material. He used two material forms; 12.7 mm thick strip material and 38.1 mm diameter bar, where two different crack orientations were tested in each product form. He reported that in

general, no significant differences in crack growth behaviour were noted between product forms or between different crack orientations.

4.4 Mechanical Properties of Service Exposed Discs

4.4.1 Turbine Discs from J85-CAN-15 Engine

Figure 4-33 shows the FCGR results of the two first stage service exposed discs, with 1520 and 2045 hrs TSN values. The solid line represents the mean line whereas the dashed line represents the 99.9% upper bound line. The equations of the mean line within the Paris region are given by:

$$da/dN = 1.17 \times 10^{-9} (\Delta K)^{1.86} \quad (4.5)$$

$$da/dN = 5.70 \times 10^{-11} (\Delta K)^{2.88} \quad (4.6)$$

for the 1520 and 2045 hrs TSN discs, respectively. The C and n values for the 99.9% upper bound line and the correlation coefficients are given in Table 4-6. It should be pointed out that these discs come from two separate J85-CAN-15 engines. The FCGR of the high time disc was approximately two to five times higher than the medium time disc. Figure 4-34 compares the fitted mean lines to the mean line obtained for the new disc including the results obtained by James [135]. Relative to the new disc, the medium time disc showed a slight reduction of FCGR whereas the high time disc FCGR's were considerably higher. However, relative to the result reported by James, the FCGR's of both service exposed discs are lower over a wide range of ΔK values.

4.4.2 Turbine Discs from J85-CAN-40 Engine

4.4.2.1 Fatigue Crack Growth Rates

The room temperature FCGR results for the A-286 compact tension specimens are shown in Figure 4-35. The points represented by a "x" sign are from three CT specimens taken from a second stage disc with 1954 hrs TSN, whereas the point represented by "open

circle" are from two CT specimens taken from a second stage disc with 3738 hrs TSN. The equations of the mean lines are:

$$da/dN = 2.86 \times 10^{-12}(\Delta K)^{2.99} \quad (4.7)$$

$$da/dN = 1.13 \times 10^{-12}(\Delta K)^{3.19} \quad (4.8)$$

for the 1954 and 3738 hrs TSN discs, respectively. The C and n values for the 99.9% upper bound line and the correlation coefficients are given in Table 4-7. It appears that increased service exposure decreases the room temperature FCGR of A-286 discs.

4.4.2.2 Tensile and Hardness Properties

The results of the hardness and room and 550°C temperature tensile tests are given in Table 4-8 for 1954 and 3738 hrs TSN service exposed and 1772 and 3738 hrs TSN rejuvenated discs. Increasing TSN appears to reduce the room temperature yield strength and increase the % elongation. Rejuvenation treatment, in the 2211 hrs TSN, decreases the room temperature yield strength and increases the tensile ductility. The yield strength, modulus of elasticity, and tensile ductility in the 2211 hrs rejuvenated disc decreases with increasing test temperature, Table 4-8. Typical room temperature and 550°C engineering stress-strain plots for the 2211 hrs TSN disc are shown in Figures 4-36 and 4-37.

4.5 Mechanical Properties of Rejuvenated Discs

4.5.1 Turbine Discs from J85-CAN-15 Engine

Figures 4-38(a) and 4-38(b) show the FCGR data for CT specimens machined from first stage turbine discs, with 1523 and 1617 hrs TSN, respectively. The equations of the mean lines are:

$$da/dN = 9.86 \times 10^{-11}(\Delta K)^{2.76} \quad (4.9)$$

$$da/dN = 1.22 \times 10^{-11}(\Delta K)^{3.33} \quad (4.10)$$

for the 1523 and 1617 hrs TSN discs, respectively. The C and n values for the 99.9% upper bound line and the correlation coefficients are given in Table 4-8. These discs were rejuvenated by using the conventional heat treatment for Inconel 718, see Section 2.2.3.4.

It should be mentioned that the information about the mission profiles experienced by the discs were not available. Figure 4-39 compares the FCGR mean lines of the rejuvenated disc with those of the new disc tested in the present investigation and the results of James [135]. The FCGR of the rejuvenated discs are very similar to those reported for the high time service exposed disc.

4.5.2 Turbine Discs from J85-CAN-40 Engine

Five CT specimens taken from a 1772 hrs TSN first stage disc were tested at room temperature. Figure 4-40 compares the mean lines of the service exposed discs described in the previous section, the results obtained by James [136], and this rejuvenated disc. The equation of the mean line for the rejuvenated disc is given by:

$$da/dN = 1.57 \times 10^{-12} (\Delta K)^{3.15} \quad (4.11)$$

The C and n values for the 99.9% upper bound line and the correlation coefficient are given in Table 4-7. The FCGR of this rejuvenated disc is very similar to the 1954 hrs TSN service exposed disc.

The 550°C FCGR test were conducted on three CT specimens taken from a second stage reheat treated disc with 2211 hrs TSN. The data points and the mean and 99.9% upper bound lines are plotted in Figure 4-41 and are compared to the results obtained by James (the solid line) [136]. It can be seen that in comparison with the published results, the 550°C FCGR's of the 2211 hrs TSN are lower over a wide range of ΔK values. The equation for the fitted mean line for the 2211 hrs TSN disc, shown in this figure is given by:

$$da/dN = 1.28 \times 10^{-11} (\Delta K)^{2.90} \quad (4.12)$$

The C and n values for the 99.9% upper bound line and the correlation coefficient are given in Table 4-7.

Figure 4-42 shows the FCGR results of the 2211 hrs TSN disc at 550°C compared to the FCGR results of the 1772 hrs TSN disc at 24°C. As was expected, the FCGR's were increased with an increase in the test temperature. Figure 4-43 shows typical SEM fractographs of the compact tension specimen fracture surfaces of FCGR tests from the

room temperature (Figure 4-43(a)) indicating a mostly transgranular fracture and the 550°C temperature (Figure 4-43(b)) indicating a mostly intergranular fracture.

Two CT specimens from the 2211 hrs TSN disc, were tested at 550°C using a trapezoidal waveform, with 20 seconds hold time, Figure 3-12. Contrary to what was expected, the fatigue/creep crack growth rates were slower than those tested using the sawtooth waveform, as shown in Figure 4-44. The equation of the mean line with hold time in the Paris region is given by:

$$da/dN = 7.94 \times 10^{-11} (\Delta K)^{2.34} \quad (4.13)$$

However, if the hold time is extended to the region where the creep effect predominates over the fatigue effect, da/dN begins to lose significance. Then, crack growth per unit time (da/dt) might be a more meaningful measure of crack growth. Therefore, the data were analyzed in terms of da/dt as well as da/dN and were plotted as a function of ΔK using logarithmic scales as shown in Figure 4-45. The "solid circles" in this figure represent the data from the sawtooth wave, whereas the "open circles" represent the data from the hold time test. It can be seen that the 550°C FCGR's in terms of da/dt are much lower for the trapezoidal waveform as compared to the sawtooth waveform tests. The equation of the mean line within the Paris region of the sawtooth and the trapezoidal waveform tests are given by:

$$da/dt = 3.93 \times 10^{-8} (\Delta K)^{2.95} \quad (4.14)$$

$$da/dt = 1.16 \times 10^{-8} (\Delta K)^{2.39} \quad (4.15)$$

respectively, where da/dt is in m/hrs and ΔK is in $MPa\sqrt{m}$. The C and n values for the 99.9% upper bound lines and the correlation coefficients are given in Table 4-7.

Figure 4-46 shows the results from the sawtooth (solid circles) and trapezoidal (open circles) waveforms plotted using a log-log scale, with the abscissa as the net section stress in MPa , and the ordinate as da/dN in m/cycle. The net section stress in MPa was calculated from

$$\sigma_{net} = \frac{P}{B(w-a)} \quad (4.16)$$

where P is the load in N , B is the specimen thickness in mm , and $(w-a)$ is the uncracked ligament of the CT specimen in mm . The FCGR as a function σ_{net} for the sawtooth

waveform are again higher than the values obtained for the trapezoidal waveform tests.

Typical SEM fractographs of the compact tension specimen, tested with a 20 second hold time, are shown in Figure 4-47. These fractographs indicate a transgranular and intergranular mixed mode of fracture.

The number of cycles required for each specimen to fail is given in Table 4-9. Failure, in this context, occurs either when the specimen fractures or when the crack length has reached the elastic limit as given in Table 3-9.

4.6 Disc Distortion

The critical dimensions for the first and second stage turbine discs, before and after rejuvenation treatment, are given in Tables 4-10 and 4-11. Table 4-10 gives the dimensions for the Inconel 718 discs given the conventional heat treatment. All dimensions are within the given tolerances except the perpendicularity of the plane of the 1617 hrs TSN disc. This out of tolerance dimension is, however, removed after the rejuvenation treatment. Table 4-11 gives the dimensions of discs given the modified heat treatment proposed for Inconel 718 disc material (see Section 5.3). All dimensions are within the given tolerances except the perpendicularity of the plane of the 1544 hrs TSN disc. Again, this out of tolerance dimension is, however, removed after the rejuvenation treatment.

Chapter 5

Discussion

5.1 Structure-Property Relationships in New and Service Exposed Turbine Discs

5.1.1 Turbine Discs from J85-CAN-15 Engine

A total of 12 first and second stage J85-CAN-15 turbine discs were received from Orenda. The total service time for each disc varied from 1520.3 hrs to 2044.9 hrs TSN. None of these discs had any reported occurrences of in-service cracking, as determined by the industrial LPI technique at Orenda. There was no apparent surface damage on any of the discs due to erosion, corrosion, oxidation, wear, fretting, or foreign object damage (FOD). However, surface hardness measurements indicated that the maximum R_C hardness level of the 2045 hrs TSN first stage disc, was approximately 47 which exceeds the maximum 45 R_C hardness specified by GE, the manufacturer of the discs. Optical microscopy indicated that the 2044.9 and 1617.4 hrs TSN second stage discs had a necklace-type structure. This is not unusual in wrought and forged material, where warm working can preferentially recrystallize the material along the grain boundaries. Therefore, the microstructure essentially consists of large warm-worked grains surrounded by a shell of fine recrystallized grains. This necklace structure was also found near the rim of the disc, but it was less pronounced at the rim locations than at the mid-section of the disc. Therefore the average grain size in these discs appeared to decrease from the bore towards the web section, and then increase towards the rim, where the grains were found to be coarsest, Figure 4-15. This observation can be attributed exclusively to the forging process where pancakes are deformed more in the web section compared to the rim and the bore area of the necklace structure describe above.

In addition to grain size variations within a disc, differences in grain sizes were also prevalent from one disc to another, Figures 4-13 and 4-15. This may be due to non

uniformity of the deformation and temperature variation during the forging and solution treatment of the discs prior to being put in service. On the other hand, it may simply be a reflection of the scatter in the material grain size variation from one heat to another. In all cases, the average grain size met the GE specification of ASTM No. 4 (i.e. $90\text{ }\mu\text{m}$) or finer. There was no indication of any grain coarsening in service exposed discs.

The new disc showed somewhat banded structures, Figure 4-11, which may be attributed to the forging process. It is not reported whether the banded structures have any harmful effects on the mechanical properties of Inconel 718. However, these banded or necklace regions may be a preferred path for crack propagation, at least along its boundary, under the right stress conditions [137]. Jeal [49], however, indicates that necklace structures in turbine discs provides the best balance of properties between a coarser grain in the rim, where creep deformation is predominant, and finer grains in the bore, to give better LCF and tensile properties.

Transmission electron microscopy clearly indicated severe plastic deformation had occurred in the 2045 hrs TSN first and second stage turbine discs. An excessive number of intergranular dislocation walls (Figure 4-22) and low angle boundaries (Figure 4-23) were present in their bolt hole and bore areas, respectively. This damage is clearly caused by the stress cycles associated with engine start up/shut down, and RPM excursions during service, and is said to be characteristic of high strain low cycle fatigue damage [138]. In contrast the dislocation arrangements in the 1520 hrs TSN disc was of the cellular-type and there was no indication of any low angle boundary formation either. This observation indicates that these discs may have been subjected to a low plastic strain amplitude [138].

Other TEM observations show that there are no essential differences between the γ' and γ'' precipitates of the new and service exposed discs, Figures 4-14(a), 4-17(a), and 4-21(a). These findings indicate that elevated temperature exposure by itself, does not seem to create any significant variation in the microstructure. These results are consistent with the hardness measurements taken along the radial section of the discs from the bore towards the rim, for both the first stage (Figure 4-27) and the second stage (Figure 4-28) turbine discs. The hardness values tended to remain constant over the entire radial distance in all discs. This indicates that matrix precipitate coarsening or averaging at

the rim, is quite minimal for discs subjected to service temperatures up to 650°C.

The large number of annealing twins observed in the new and the 1520 hrs TSN discs are the result of their processing history where the growth of recrystallized grains in highly deformed areas occurs in low stacking fault material during forming or solution or heat treatment. This characteristic is common to both forged and annealed Inconel 718 and is harmless as the annealing twins simply remain in the material and do not degrade its mechanical properties. The annealing twin frequency is usually used to qualify the forged material by the engine manufacturers. Too high a frequency is considered to be indicative of excessive grain growth.

There are many factors in addition to the microstructure which effect FCGR's. These include test parameters such as temperature, cyclic frequency, waveform, load ratio, environment, and crack orientation, which were all carefully controlled. Factors which could not be controlled included manufacturing variations and service conditions, such as heat to heat differences, chemical composition, and mission profiles experienced by the individual components. It was however assumed that all turbine discs were subjected to similar manufacturing and service conditions. Therefore, the changes with FCGR's have been interpreted on the basis of the microstructural differences between various discs.

It was found that compared to the new disc, service exposure had both beneficial (in the case of the 1520 hrs TSN disc) and detrimental (in the case of the 2045 hrs TSN disc) effects on the FCGR's, Figure 4-34. Literature contains very little information concerning the effects of service exposure on FCGR's in Ni-base superalloys. These superalloys contain complex microstructure thus making it very difficult to analyze the service exposure effects systematically. Some investigations have been carried out on the effects of pre-straining, or plastic deformation by application of tensile loads, on FCGR's in different materials. Increases in FCGR following pre-strain, have been observed in Ni-base turbine disc alloy N901 [139], low carbon steels [140], and aluminum alloys [141]. In contrast, FCGR's have been observed to reduce by pre-straining stainless and mild steels [142], dual-phase steels [143], copper [144], and 90/10 α -brass. In 70/30 α -brass pre-straining does not affect the FCGR's [144, 145]. The literature suggests that pre-straining does not affect the final failure mechanism for any one of the materials [146, 139]. In the present

investigation, all the Inconel 718 CT specimens revealed a mixed intergranular and transgranular mode of fracture, Figures 5-1. Therefore, it is suggested that the pre-straining or service exposure has little effect on the final failure mechanism.

There is however, evidence that the nature of the dislocation structures plays a significant role in FCGR behaviour of different materials. It has been reported that cellular dislocation arrays produced in copper and 90/10 α -brass during pre-straining lowered the FCGR's relative to the unstrained materials [144, 146]. In Ni-base alloy N901, on the other hand, the introduction of intense planar slip bands have been shown to increase the FCGR's [139]. In the present study therefore, much higher FCGR's observed in the 2045 hrs TSN disc relative to the new disc can be attributed to the presence of service induced *dislocation wall and low angle boundary* formation, Figure 4-22. In contrast, *cellular dislocations* observed in the 1520 hrs TSN disc (Figure 4-18(a)) cause a reduction in the FCGR's relative to the new disc (Figure 4-34). It has been suggested that pre-existing cellular dislocation-arrays may hinder the development of intense slip bands and thus reduce the amount of crack extension per fatigue cycle [147]. Conversely, prior localization of slip into dislocation walls, provides a favourable path for fatigue crack growth, and leads to an increase in the FCGR's [139].

5.1.2 Turbine Discs from J85-CAN-40 Engine

The microstructural examination of the J85-CAN-40 turbine discs was carried out by Thamburaj et al. [18] and a summary of their findings is given here for completeness sake. The study was carried out on two second stage discs which had seen 1954 and 3738 hours of service. The 3738 hrs TSN disc had some surface discolouration. TEM foils taken near the rim indicated that very little plastic deformation had occurred at this location. However, only a small distance away from the rim towards the bore, evidence of significant plastic deformation was found. In the bore regions too severe plastic deformation had manifested itself in the form of intense planar slip. An important observation from these foils was that in both discs, in regions that had undergone significant plastic deformation needle-like precipitates, η , were always present. These precipitates appeared to occur along slip bands, and in areas of intense dislocation activity near the bore. This may be due to

transformation of γ' to η by slipping of the closed-packed planes to rearrange the FCC γ' into HCP η . This transformation can be attributed to the presence of higher stress levels in the bore region of the disc (see Figure 2-19).

There were no essential differences between the γ' precipitates in the rim sections of the 1954 and 3738 hrs TSN discs. These findings indicate that microstructural degradation in the turbine discs of the J85-CAN-40 engine is primarily caused by high strain low cycle fatigue damage. The hardness values tended to remain constant over the entire radial distance in both discs, indicating that matrix precipitate dissolution or coarsening at the rim is negligible at the 550°C service temperature.

The increase in the FCGR's with increasing service exposure in A-286 discs appears contradictory to the results obtained from the Inconel 718 discs. The average grain sizes of the 1954 hrs and 3738 hrs TSN discs were found to be approximately equal (18 and 15 μm , respectively) as were their average hardness values (35 and 37 R_C , respectively). The observed difference in the FCGR's could not be correlated directly to these minor differences in the grain size. In any case, James [136] reported that grain size does not play a major role in the fatigue crack growth behaviour of austenitic stainless steels at room temperature. Postans [148] has also reported that grain structure does not influence the crack propagation rate in Fe-Ni-base alloys at room temperature.

Since the size of the plastic zone ahead of the propagating crack is proportional to the inverse of the square of the yield strength, lower yield stress values obtained for the 3738 hrs TSN disc as compared to the 1954 hrs TSN disc, Table 4-8, results in a larger plastic zone size. This larger plastic zone size results in slower FCGR's for the 3738 hrs TSN disc compared to the 1954 hrs TSN disc. Another contributing factor between the lower FCGR's in the high TSN disc, may result from the effect of ageing (number of service hours) on the microstructure at elevated temperatures. Prolonged exposure of austenitic stainless steels can result in precipitation of various phases such as carbides and intermetallics [149, 150]. Precipitation of η phase was observed in both service exposed discs as explained in the previous section on the effect of service exposure on microstructure. It has been postulated [151] that as a propagating crack encounters various second phase particles it becomes partially blunted, tending to decrease its rate of growth.

However, from literature it also appears that long term thermal exposures can prove detrimental to crack growth in annealed austenitic stainless steel at room temperature [136].

5.2 Structure-Properties Relationships in Rejuvenated Turbine Discs

5.2.1 Turbine Discs from J85-CAN-15 Engine

Having characterized the type of microstructural degradation that occurs in service exposed J85-CAN-15 turbine discs, attempts were made to restore the original microstructure by applying the conventional heat treatment. It was expected that the reheat treatment would lead to uniform precipitation of δ phase at the grain boundaries and of γ' and γ'' within the matrix. It was also expected that the rejuvenation would considerably reduce the intensity of the service exposed dislocation sub-structures, or at least redistribute the dislocation networks found in the high time disc. Ralph [152] reports that during the anneal (or recovery) most point defects anneal out by thermally activated movement under the influence of strain fields. He indicates that the deformation on a microscopic level, consists of "geometrically necessary" dislocations, which are formed as a result of strain gradients across the grains, and "statistically stored" dislocations, which are the additional dislocations in the deformed material resulting from random interactions. Comparatively, the geometrical dislocations are easily removed by diffusion assisted processes, during the material annealing. Statistical dislocations, on the other hand, are less easily removed, but local rearrangement may occur to form low angle grain boundaries [153].

The reheat treatment was carried out in an argon environment and 10^{-5} torrs vacuum at a commercial heat treatment facility. Comparison of the service exposed and rejuvenated discs showed negligible amounts of oxidation or alloy depletion at the surface of the rejuvenated discs. This observation indicates that surface protection measures i.e. argon environment, during reheat treatment were adequate in the present case. It is clear that rejuvenation treatment did not induce any appreciable amount of grain growth in the service exposed material because of the low solution treatment temperature used. TEM examination indicated that the rejuvenation did not have any deleterious effects on

the intra-granular microstructural features such as the size, shape and distribution of the strengthening precipitates. However, the most obvious microstructural characteristic was the presence of a considerable amount of prior plastic deformation as shown in Figures 4-25 and 4-26. The existence of these low angle boundaries indicate that the rejuvenation treatment for Inconel 718 (with a solution anneal temperature of 955°C) was not successful in removing the microstructural damage induced during service. This damage may be reversed by treating the material at higher solution anneal temperatures. This would, of course, be associated with a trade off in tensile strength and grain growth, which in turn would be expected to affect the LCF crack initiation resistance of the material.

As mentioned previously, rejuvenation treatment did not restore the FCGR's of the service exposed disc to the as new disc condition, Figure 4-39. In fact the FCGR behaviour of both 1520 hrs and 1617 hrs TSN discs were similar to that of the 2045 hrs TSN service exposed disc. This was not totally unexpected, since service induced low angle boundaries were retained in the case of all these discs after rejuvenation, Figures 4-23 and 4-26. Again very little has been published on the effects of low angle boundaries on FCGR's in Inconel 718. What is clearly evident, is that the conventional reheat treatment is ineffective in restoring the FCGR's of service exposed Inconel 718 turbine discs. It should however be pointed out, that the rejection criterion of the J85-CAN-15 turbine discs are based on crack initiation, and a rejuvenation treatment that is ineffective in improving crack propagation, may not necessarily be ineffective in improving the crack initiation properties. Therefore, further work should include investigation into crack initiation properties of the new, service exposed and rejuvenated discs.

5.2.2 Turbine Discs from J85-CAN-40

Specimens taken from the bore and rim areas of the 1954 hrs and 3738 hrs TSN second stage discs were subjected to:

1000°C /1hr /Oil Quench + 718°C /16hr /Air Cool

heat treatment by Thamuraj et al.[18]. Examination of the TEM foils showed that the needle-like phase, and the intense dislocation activity present near the rim regions were virtually eliminated in both discs. However, some remnants of prior plastic deformation

were still present in the bore regions. It may be postulated that as the reheat treatment may have removed the bulk of the needle like phase, η , a crack may propagate onward encountering no second phase particles to blunt it, and slow its growth rate. The effects of rejuvenation treatment on FCGR at 550°C are inconclusive as no service exposed specimens were tested at this temperature due to the lack of specimens. However the FCGR's of these rejuvenated discs are superior to those of published values for A-286 virgin material, Figure 4-41.

It should be pointed out that the reheat treated compact tension specimens were machined from turbine discs which were subjected to a

1000°C / 1hr / Argon Quench + 718°C / 16hr / Argon Quench

reheat treatment schedule. Note that the discs were not oil quenched, after the 1000°C anneal temperature, in order to reduce any dimensional distortion. Obviously the cooling rate is appreciably slower in argon quenching compared to oil or water quenching, resulting in a much softer material as indicated by the hardness and the tensile properties, Table 4-8.

The elastic moduli (E) of the discs, obtained from tensile specimens, are given in Table 4-8. These results suggest that the higher FCGR for the 1772 and the 1954 hrs TSN discs may also be due to the reduction in the modulus of elasticity.

5.3 Improved Rejuvenation Scheme for the J85-CAN-15 Turbine Discs

The conventional heat treatment used for Inconel 718 was not originally developed to produce effective LCF crack initiation and crack growth properties for turbine discs. In fact, Thamburaj [154] has reported that the conventional heat treatment produced the highest FCGR's at 650°C when compared to several other available heat treatments for Inconel 718. In order to anneal out any plastic deformation induced during service, and to optimize mechanical properties in terms of elevated temperature creep and LCF crack initiation and propagation, a specific heat treatment for Inconel 718 turbine discs must be designed. As grain size, grain boundary Ni_3Nb precipitates, and the size of γ'' affect the mechanical properties, they are important parameters to consider. For example a fine

grain size is desired to develop high strength and excellent resistance to crack initiation. On the other hand, a fine grain size may not be desirable for the balance of creep and crack propagation properties. Therefore, the best structure would be a coarser grain in the rim where creep is required, with finer grain in the bore to give better cyclic initiation life. However this is not easily achieved and a grain size balance must be obtained. Ni_3Nb precipitation should be sufficient to produce a serrated grain boundary morphology. Very coarse Ni_3Nb needles may benefit creep properties and improve ductility and FCGR's [43].

A novel heat treatment developed at NAE/NRC by Koul and co-workers [73, 155], seems to be promising for the combination of properties needed in gas turbine discs. A solution treatment is carried out at $1032^{\circ}C$ to anneal out any dislocation network and to recrystallize the microstructure and produce a relatively fine grain size. This is followed by a direct age to the first intermediate ageing temperature of $843^{\circ}C$. This ageing temperature precipitates and coarsens the grain boundary Ni_3Nb needles, and produces serrated grain boundaries. A partial solution treatment at $926^{\circ}C$, solutions any coarser γ'' precipitates particles that form within the grains during the direct age. A final double age at 718 and $621^{\circ}C$ is used to precipitate γ' and γ'' intragranularly in an optimum morphology. The conditions associated with this heat treatment are shown on the Inconel 718 TTP diagram, Figure 5-2, and also given below,

$1032^{\circ}C / 1hr \Rightarrow$ Furnace Cool at $1^{\circ}C/min \Rightarrow 843^{\circ}C / 4hr /$ Air Cool +
 $926^{\circ}C / 1hr /$ Air Cool +
 $718^{\circ}C / 8hr \Rightarrow$ Furnace Cool at $1^{\circ}C/min \Rightarrow 621^{\circ}C / 8hr /$ Air Cool.

A comprehensive and detailed programme should be conducted to determine the tensile, creep, stress-rupture, fatigue crack initiation and propagation properties of Inconel 718 turbine disc material based on this heat treatment.

5.4 Other Considerations for J85-CAN-40 Turbine Disc Rejuvenation

5.4.1 Effect of Waveshape on FCGR's

The shape of the loading waveform used for FCGR testing greatly depends on the type of mission (stress, temperature, time profile) experienced by the aircraft. However, each mission in a military aircraft is associated with manouvers which might be simulated with sinusoidal or sawtooth waveforms, while those associated with engine start up and shut down might involve hold times at, or near, the maximum stress. Although the specific mission profiles for these engines were not readily available, a representative mission profile is shown in Figure 5-3 [121, 47].

Since in these studies, relative comparisons of FCGR's were involved, the thermal mechanical cycle of Figure 5-3 can be simplified considerably for ease of analysis while still adequately assessing material response [47]. The mission simplification is illustrated in Figure 5-4, where the overall stress cycle is comprised of two parts:

- 1) The take off, cruise, and landing portion (engine start up, hold time, and shut down),
- 2) Military manouvering portions (major throttle excursions).

The first portion may be simplified to a trapezoidal waveform with a 20 second hold time at maximum tensile load, while the second portion may be considered as a sawtooth waveform where approximately similar cycles of the same stress range are experienced.

Since only relative comparisons were being made, and for the sake of simplicity, only the sawtooth waveform was used in assessing the FCGR's of the CT specimens. However, in order to investigate what effect, if any, the hold time might have on FCGR behaviour and also to assess the suitability of the DC-PD technique for the trapezoidal waveform, two CT specimens from a J85-CAN-40 turbine disc with 2211 hrs TSN were tested with a 20 second hold time. Since changes in the deformation mode are more sensitive to strain rates than frequency [138], and they can be associated with large changes in FCGR performance, the loading rate was chosen to be the same as that of the sawtooth waveform, as shown in Figure 3-12. It can be seen from the FCGR results given in Figure 4-44, that

tests with hold time produced somewhat lower FCGR's than did the sawtooth waveform. There is also a decrease in the slope of the curve for the tests with hold time; that is, the two curves are diverging from each other with increasing stress intensity factors. One possible mechanism might be the crack tip blunting each time the crack tip encounters a second-phase particle during the hold time, thereby requiring the crack to partially reinitiate itself before proceeding [151]. These findings are similar to those of James [156] and Shahinian [157], but contrary to studies by Popp et al. [158], Michel et al. [159], and Clavel et al. [138]. James reported a lower rate of crack growth in type 304 stainless steel at 538°C for square wave loading (effectively hold time) as opposed to a sawtooth waveform. This effect was observed at a frequency of 0.0014 Hz but not at two higher frequencies. Shahinian reported that, at 593°C, increases in the hold time caused higher FCGR's for a given stress intensity factor range, for type 304 steel and 20% cold worked type 316 steel. On the other hand, Popp et al., Michel et al. and Clavel et al. reported that the introduction of a tensile load period during cyclic loading of Inconel 718 at 538°C and 650°C significantly increased FCGR's, when compared to continuous cycling.

In terms of da/dt , the hold time waveform also decreased the growth rates substantially, as shown in Figure 4-45, thus indicating that at 550°C, cyclic loading is damaging in A-286, but static loading or creep, is not. Apparently very little creep occurs, Figure 4-47, and contributes to crack growth. The inclusion of the 20 second hold time also caused a reduction in the number of cycles to failure, but in terms of total times to failure, life increased, Table 4-9. This is in agreement with other investigators results who reported that hold times introduced at the maximum stress level tend to decrease the number of cycles to failure and increase the total time to failure [160, 157, 156]. The plot of data points for tests with and without hold time, Figure 4-44, shows that both sets of FCGR data points approach the same critical stress intensity factor of approximately $70 \text{ MPa}\sqrt{m}$. This signifies that the material has not undergone a metallurgical change at the 550°C test temperature during the long test periods of up to 90 hours, Table 4-9.

The mode of failure was affected by hold time period, shifting from mostly transgranular to mostly intergranular fracture, as shown by the SEM micrographs given in Figure 4-47. Solomon and Coffin [161] who studied the effects of frequency and environment on

FCGR in A-286 at 593°C, indicated that the mode of failure had a pronounced frequency dependence when tests were run in air, Figure 5-5. These frequency dependencies are mainly due to environmental effects.

It should be remembered that under generally elastic conditions, the stress intensity factor range, ΔK , rather than the simple parameters of either stress or strain, is the rate-controlling parameter in the propagation of fatigue cracks [162]. The methods employed for the analysis of the FCGR results in this study were those of LEFM. These methods require that the crack propagation be independent of load at a constant value of stress intensity factor [159]. Although the experimental results generally show good agreement on the basis of these methods, it is reasonable to expect that fracture mechanics parameters may not apply under conditions where time dependent deformation processes may become active, such as during hold time tests. In these instances, the propagation of fatigue cracks under plastic conditions, requires a compound parameter of strain range multiplied by the crack length, or by the square root of the crack length, to adequately describe the propagation of fatigue cracks [156]. Therefore, the J -integral may be more representative of the FCGR under hold time loading. In order to calculate the J -integral, the COD must be determined, but the potential drop technique only measures the crack length and does not provide the COD. Hence, this type of analysis could not be carried out using this method. Future hold time tests on this material should include COD measurements by utilizing a suitable COD measurement technique, such as the capacitance gauge technique, in order to provide the J -integral calculation which accounts for plastic deformation at the crack tip.

Koteraazawa, [163] in an examination of the applicability of fracture mechanics parameters to crack propagation under creep condition with 304 stainless steel, concluded that the creep crack propagation rate could not be described in terms of elastic stress intensity factors, and that the net section stress could be used instead of the J -integral. Therefore, the FCGR's for the tests with hold times and those with the sawtooth waveforms are presented as a function of net stress, in Figure 4-46. The net stress does not, however, appear to be a better correlating factor. Again this may be due to significant crack opening and considerable plastic deformation around the crack tip in the tests with hold times.

5.4.2 Effect of Temperature on FCGR's

Knowledge of the FCGR behaviour of structural material such as turbine discs, at typical elevated temperatures, is of significant importance. A typical mission for these engines, shown in Figure 5-3, illustrates the temperature variation in a turbine disc, during a mission, as "dashed line". Since relative comparisons of the FCGR's were involved, the thermal cycling portion of the mission was reduced to a constant temperature of 550°C and 650°C for the J85-CAN-40 and the J85-CAN-15 engines, respectively. In the case of A-286, room temperature tests were also conducted. A comparison of these results, to those obtained at 550°C , Figure 4-42, indicates a dramatic increase in FCGR's as the temperature increases. This behaviour is in agreement with many similar observations when other alloy systems are tested in air at elevated temperatures and more specifically with those of James [164] who investigated the effect of temperature on FCGR behaviour of A-286 steel, Figure 4-32. James reported that, in general, the FCGR's increased with increasing temperatures, while no significant differences in crack behaviour were noted between product forms or the different crack orientations. In this study, examination of the fracture surfaces with a scanning electron microscope revealed features that varied with changes in temperature. A mostly transgranular mode of fracture was indicated at room temperature, whereas at 550°C a mostly intergranular mode of fracture had taken place, Figure 4-43. It is suspected that the major reason for the higher FCGR at elevated temperature is the reduction of the modulus of elasticity of the material [165], Table 4-8.

5.5 Disc Distortion

Of all the measured dimensions, it appears that only one first stage disc (S/N 11J774-67) with 1617 hrs TSN exceeds blueprint limits on perpendicularity of hub to the plane of the disc by 0.094 mm^1 before the heat treatment, as given in Table 4-10. Incidentally this is the disc with the highest FCGR results, Table 4-6. This distortion was, however, corrected after it was heat treated, Table 4-10. This rectification may be due to the relaxation of residual stresses produced in the original manufactured disc. Since the hub runs at low

¹ 0.00371 inch ($0.00471 - 0.001$).

temperatures during service, the stresses produced during manufacturing would not be relieved in service. The forgings can contain high residual stresses due to inconsistencies in cooling during the initial heat treatment. Another possible explanation is the development of residual stresses in service. Similar results, but to a lesser extent (0.023 mm or 0.00089 inch), were obtained for the first stage disc (S/N 11J778-76) which was subjected to the proposed modified heat treatment.

All other dimensions including the firtree slots and rabbet radii were well within the specified tolerances [131] and the extent of distortion is quite limited in each case. Therefore, it may be concluded that neither the conventional nor the proposed modified heat treatments affects the critical dimensions of the turbine discs of the J85-CAN-15 engine.

5.6 Elevated Temperature DC-PD Technique

5.6.1 Calibration, Accuracy and Precision

5.6.1.1 The Calibration Curve

The calibration curve was obtained by the use of a travelling microscope. The potential probes were attached to the centreline positions. There were no published calibration curves for this condition and specimen size. Hence no true comparison of the calibration curves was possible. However, the calibration curve is in agreement with those determined by Hicks and Pickard [166] and Aker [97], who had identically placed their current leads but with potential probes in the diagonal positions. Aker used a travelling microscope for calibration of 0.94 mm thick $\times$ 25.4 mm wide CT specimens. Hicks and Pickard used finite element programmes and a conducting paper method for calibration. The calibration curve is in good agreement with that determined by Andrews [167], who had identically placed potential probes and current leads. He used a travelling microscope for calibration of 12.5 mm thick $\times$ 50.8 mm wide CT specimens.

— For a given voltage ratio, higher crack lengths are predicted by using the calibration curves of those specimens with the potential probes in the diagonal positions. This may indicate the loss in sensitivity of the system by placing the potential probes away from

the active voltage field near the centreline. In any case the maximum deviation in crack length from one curve to another is 0.65 mm, Figure 4-6.

5.6.1.2 Accuracy and Precision

In FCGR testing, it is important to accurately measure the crack extension from one point to another point consistently, as incremental crack extensions are extremely significant in evaluating the FCGR, da/dN . The DC-PD technique employed is very effective in this respect. In fact, the accuracy and the precision of the technique was determined to be better than 0.100 mm and 0.075 mm, respectively, Table 4-5. It should be noted that the accuracy of the DC-PD technique can not exceed that of the travelling microscope used in determining the crack lengths for the calibration curve. The precision is well within the 0.1 mm minimum required precision, specified in the ASTM standard E-647-83 [115].

5.6.2 Advantages and Disadvantages

5.6.2.1 Advantages of the DC-PD Technique

The most obvious advantage of the DC-PD technique, is the use of the room temperature calibration curve at elevated temperatures, regardless of the material type or microstructure. This eliminates the need for exposing the surfaces of the specimen for crack length measurements from furnaces or any closed aggressive environments. Another important advantage is the requirement for minimal operator intervention once a test is underway. In this study, most of the tests were performed at night with no supervision. This results in full utilization of expensive equipment. The computer that was used to control the tests and acquire the data also performed the data reduction, Figure 3-2, and plotted output graphs of crack lengths versus the number of cycles, and da/dN versus the stress intensity factor. Further utilization of the various MTS machines is also possible by using a central computer to control many load frames and their respective DC-PD systems for measuring FCGR in compact tension, corner crack, or notch specimens. This is effectively used at the Rolls-Royce engine laboratory in England.

5.6.2.2 Disadvantages of the DC-PD Technique

Perhaps the biggest disadvantage of the use of this technique in this study, was electromagnetic interference (EMI) and grounding problems, as very sensitive amplifiers were used to detect the small changes in the potential drops. Grounding of all equipment including the load frame and the computer to a common earth ground point, Figure 3-9, eliminated the voltage fluctuations. However, the EMI problems caused by the operation of a 750 kHz magnetic induction heater on another MTS load frame within the laboratory persisted, even after shielding the power supply, amplifiers, and cables. Therefore only one of the load frames could be operated at a time. Since the DC-PD system could be operated without supervision, most FCGR tests were carried out at night. Another disadvantage of the technique, as a fracture mechanics tool, is its inability to provide crack opening displacements for determining parameters such as the J -integral.

5.6.3 Precautionary Measures for High Temperature Application of the DC-PD Technique

By the addition of a reference specimen, the calibration curve obtained at room temperature may be used at any temperature, providing the reference specimen is subjected to the same environment as the test specimen and that the calibration curve is plotted in non-dimensional form. This is based on the argument that only material resistivity, ρ , varies as a function of temperature, and since the calibration parameter is the ratio of the test and reference specimen voltage drops, V_R , Equation (3.2), the net effect on the calculated ratio under a given set of testing conditions will not change. It is not even necessary to know the resistivity of the material being tested. However, the voltage ratio is affected by the change in temperature between the reference and the test specimens. For the AM-355 CT specimens tested at 550°C, it was found that the accuracy of the system was not affected if the temperature difference between the reference and the test specimens did not exceed 10°C, Figure 5-6. On the other hand, however, it should be noted that the effect of test and reference specimen temperature difference, on the accuracy of the technique, is dependent on the magnitude of material resistivity, ρ , which will vary from

one material to another. The electrical resistivity varies linearly with temperature, i.e.

$$\rho(T) = \rho_0(1 + b \times T) \quad (5.1)$$

where $\rho(T)$ is the specific electrical resistivity at a given temperature, ρ_0 is the residual resistivity at T_0 , and b is the slope of the line [109]. For highly alloyed materials like austenitic stainless steels, Fe-base superalloys (e.g. A-286 alloy) and Ni-base superalloys (e.g. Inconel 718), the residual resistivity is reached at room temperature. It should be noted that b can become negative if precipitates are formed or phase transformations occur. In such a case, the correction factor described below may be employed [109].

The specimen is pre-fatigued and the crack is measured by means of an optical microscope. From the potential drop measurement at room temperature, the crack is corrected using the calibration curve. The furnace is then turned on, and after reaching the test temperature, the specimen is subjected to a small preload and the potential drop is measured again. Hence, the correction factor is given by:

$$C_T = \frac{V(T, L \approx 0)}{V(RT)} \quad (5.2)$$

where $V(T, L \approx 0)$ is the potential drop at the test temperature under a preload and $V(RT)$ is the corresponding potential drop from the calibration curve at room temperature [109].

Another concern may be the formation of a large plastic zone ahead of the crack tip after the maximum load is reached at high temperature. This may be corrected by,

$$C_P = \frac{V(T, L)}{V(T, L \approx 0)} \quad (5.3)$$

where $V(T, L)$ is the potential drop under load at the test temperature [109]. For ferritic (low alloyed) materials, this correction is large, while for highly alloyed materials, such as those used in this study, the correction may be small as the potential drop increases by only a small amount. Furthermore, the influence of the increase in plastic zone size on the potential drop may be neglected if the specimen thickness is sufficient [109]. The measurement of the initial rates of fatigue crack growth made using the potential drop technique are also likely to be influenced by the build up of the plastic zone at the notch tip,

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particularly at small growth rates and low temperatures. This is because of the resistivity change of the material in the plastic zone. The effect is most pronounced for the first few cycles, going from a starter notch with no plastic zone, to a small crack plus a plastic zone. However, by using standard pre-cracking procedures and decreasing stress intensity, this effect is not significant.

There is also concern about the conduction path between the crack faces of the test specimen resulting from surface oxide formation and magnetic effects at elevated temperatures [96]. This problem is more pronounced for low-alloyed ferromagnetic materials like ferritic steel. However, by keeping the load at a maximum level during data acquisition, Figure 3-4, this problem is all but eliminated as the crack length is increased. In this study, it was found that the potential drops for the AM-355 CT specimens at 550°C must be measured at loads greater than 50% of the maximum load, as shown in Figure 5-7. This graph was obtained by plotting the potential drop crack lengths at various load levels, versus the ratio of the load level to the maximum applied load. It can be seen that the potential drop crack length measurement approaches the optically measured crack length at a ratio greater than 0.5. Therefore, there should be little concern regarding the contribution of creep crack growth during FCGR testing as the load may be kept at half the maximum load for a few seconds for data acquisition.

When little creep deformation accompanies fracture, the crack opening remains small. Some variability in the potential drop is observed due to this small COD [98]. This variability has been attributed to crack shortening due to surface roughness (particularly when the fracture is intergranular), uncracked ligaments left behind the crack front, voiding around the crack tip, and oxide bridging across the crack faces. Since the extent of these features may be rate sensitive, different calibration curves may be obtained for different loading conditions. Therefore under these circumstances, a unique calibration curve cannot be adapted, and it is recommended [98] that periodic optical checks of crack lengths be made, or, if this is not possible, the test is stopped prior to final fracture so that the final crack length can be measured and the calibration curve forced through the known end points. This is especially important if the voltage ratio at the beginning of the test is not set to unity.

Chapter 6

Conclusions and Recommendations

6.1 Conclusions

The major results, observations, and conclusions of this research project for each alloy and the DC-PD technique are summarized below:

6.1.1 Inconel 718 Turbine Discs from the J85-CAN-15 Engine

- One of the most significant factors in determining whether service exposure is beneficial or detrimental to FCGR properties is the nature of the dislocation substructure introduced during service.
- The FCGR's in the high time disc were two to three times higher than in the new disc. The FCGR's of the medium time disc were slightly lower than those of the new disc.
- SEM studies of the fracture surfaces of the CT specimens tested at 650°C indicated, in all cases, a mixed mode of intergranular and transgranular fracture. Using standard metallographic techniques, creep cavities were not detected in the service exposed discs, even at very high magnifications. Service exposure did not cause any grain coarsening.
- Rejuvenation treatment based on the Inconel 718 conventional heat treatment did not completely remove the low angle boundaries or the dislocation arrays. The FCGR's in the rejuvenated discs were comparable to the high time disc. A modified heat treatment, with a higher solution treatment temperature and an extra ageing treatment, for the rejuvenation of service exposed discs is outlined.
- Dimensional analysis indicated that the discs did not undergo significant creep or warping due to service. The discs were dimensionally stable during both the conventional and the proposed rejuvenation treatments.

6.1.2 A-286 Turbine Discs from the J85-CAN-40 Engine

- As expected FCGR's increased with an increase in test temperature. Therefore, FCGR tests for life predictions, must be conducted at the best estimated operating temperature of the component.
- The FCGR's at room temperature were lower for the high time disc than the medium time disc. Both room and elevated temperature FCGR's were lower than the published values for this alloy.
- Rejuvenation treatment did not improve the room temperature FCGR properties.
- Inclusion of a hold time during cyclic loading decreased the FCGR's (da/dN and da/dt) in terms of ΔK . Therefore, the most suitable waveform must be established for use in specimen testing and data generation for life predictions.

6.1.3 DC-PD Technique

- The direct current potential drop technique for crack length measurement at temperatures up to $650^{\circ}C$ was successfully adapted, calibrated, verified, and implemented in this work. The precision of the technique was better than 0.075 mm.

6.2 Recommendations for Future Work

- One of the most vital areas requiring future work, is the development of a more reliable life prediction system for the critical components of an engine. This would be a method of continuously tracking the rate at which LCF cycles are consumed in each critical component, and retiring each one upon the accumulation of a certain number of LCF cycles, determined as the safe life of the component. This approach would be different from the present approach of retiring life limited parts after a certain number of "hours" of service exposure, which in reality may not be applicable since parts tend to experience completely different levels of severity when operating under a variety of missions. It is suggested that for future rejuvenation treatment investigations, the discs be divided in two and the treatment be applied to one half, the other half being

used for comparative purposes. This eliminates uncertainties about different mission profiles experienced by different discs.

- A second area of interest is the development of advanced nondestructive techniques to determine the extent of metallurgical damage, which will have a life limiting impact on gas turbine engine components, particularly discs. These techniques should include detection and quantification of metallurgical damage due to creep, low cycle fatigue, and high cycle fatigue, and consequently would allow reliable component residual life prediction. In the mean time, more work is needed to investigate the effect of different pre-existing dislocation substructures on FCGR's of turbine and compressor disc materials, specifically Inconel 718 and A-286 alloys.
 - Thirdly, the turbine discs of the J85-CAN-15 engine rejection criterion is based on crack initiation. Therefore, appropriate LCF tests must be conducted, at the operating temperature and a realistic waveform, on hour-glass specimens removed from each half of service exposed discs, one half of which has been rejuvenated. Since it is not possible to machine a whole hour-glass specimen from a disc, a specimen can be fabricated with the gage section taken from the disc with ends welded on as shown schematically in Figure 6-1. The weld and its heat affected zone, should be located well outside the gage section of the specimen.
 - Finally, appropriate methods should be developed which would provide the COD to enable the determination of FCGR's using the J -integral or the J^* -integral.
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Table 1-1: The J85-CAN-40 Engine Mission Severity Factors
Supplied by General Electric Company

MSF as supplied by General Electric		e.g. Safe Life of the 5th Stage Compressor Disc	
Date	MSF	USAF J85-5 (hrs)	CAF J85-CAN40 (hrs)
May 1970	1.00	3,000	3,000
June 1978	0.94	4,500	4,787
Sept. 1, 1979	0.94	4,000	4,255
Sept. 17, 1979	0.94	4,400	4,680
Sept. 3, 1980	1.60	4,400?	2,750?
Sept. 24, 1981	1.80	3,200	1,777
Oct. 12, 1983	1.40	3,200	2,285

MSF = Mission Severity Factor

USAF = United States Air Force

CAF = Canadian Armed Forces

$$(J85-CAN-40 \text{ Life}) = \frac{(J85-5 \text{ Life})}{MSF}$$

Table 1-2(a): J85-CAN-40 Turbine Discs Received by NRC/NAE

Stage	Serial No.	TSN (hrs)
First	11J56-204	1453.9
First	11J56-230	2492.6
First	11J56-248	1771.9
First	P02675-23	3737.6
Second	11J96-205	1953.5
Second	11J96-207	3653.4
Second	11J96-223	2210.6
Second	11J96-280	2700.7
Second	P10323-7	3737.6
Second	P22096-34	3791.6

Table 1-2(b): Statistical Data on the Retired J85-CAN-40 Turbine Discs

Stage	Safe Life Limit (hrs)	Safe Inspection Interval (hrs)	Longest Life Disc (hrs)	Mean Life (hrs)	Average Life Extension (hrs)
First	1111	555	3737.6	2364	1253
Second	1111	555	3791.6	2971	1860

Table 2-1: Some Airline Flight Accidents Due to Turbine Section Component Failure

(Cont'd on the next page)

Passenger Flights	Cicumstances and Remarks
<p>Date : March 17, 1982 Carrier : Air France Aircraft: A300B4 (F-BVGK) Location: Sana'a, North Yemen Injuries: 2 Total Occupantes: 124 Phase : Take-Off</p>	<p>Stage one high pressure turbine disc in No. 2 engine broke up during take-off run. Aircraft aborted take-off. Serious fire started as a result of turbine fragments penetrating fuel tanks and lines, but all on board were evacuated. Aircraft a write-off. Engine: P&W JT9D? [Reference, FLIGHT 22-01-83]</p>
<p>Date : March 22, 1984 Carrier : Pacific Western A/C : 737-200(C-GQPW) Location: Calgary, Alberta Injuries: - Total Occupants: 118 Phase : Take-Off</p>	<p>13th stage compressor disc failed at 75kt during take-off run. Debris penetrated wing. Major fire. Engine: P&W JT8D-9 [Reference, FLIGHT 26-01-85]</p>
<p>Date : June 1, 1984 Aircraft: MD-80 Injuries: - Total Occupants: 85 Phase : Climb</p>	<p>Uncontained turbine failure in No. 2 engine. Damage to hydraulics and APU Engine: P&W JT8D-207 ? [Reference, FLIGHT 26-01-85]</p>
<p>Date : June 13, 1984 Carrier : Austrian Aircraft: DC-9 Location: Vienna Injuries: - Total Occupants: ? Phase : Take-Off</p>	<p>Uncontained engine failure. APU set on fire, hydraulic lines severed. Gear had to be lowered manually. Engine: P&W JT8D [Reference, FLIGHT 26-01-85]</p>
<p>Date : April 20, 1984 Carrier : Trans-European Aircraft: 707 Location: Brussels Injuries: - Total Occupants: 160 Phase : Take-Off</p>	<p>Broken turbine blade caused engine fire. Take-off aborted. Engine: P&W JT3D-7 [Reference, FLIGHT 26-01-85]</p>

Table 2-1 Cont'd: Some Airline Flight Accidents Due to Turbine Section Component Failure

Passenger Flights	Cicumstances and Remarks
<p>Date : January 5, 1984 Aircraft: 720 Location: Philadelphia Injuries: - Phase : Climb</p>	<p>Turbine blade failure in No. 3 engine caused fire and smoke entered cabin. Safe landing. Engine: P&W JT3D [Reference, FLIGHT 26-01-85]</p>
<p>Date : April 9, 1985 Carrier : British Airways Aircraft: 757 (G-B1KA) Location: Near Heathrow Injuries: - Total Occupants: 127 Phase : Climb</p>	<p>At 8,000 ft, No. 1 engine high pressure compressor disc (first stage) failed and penetrated the cowling. Engine safely shut down. Engine: RR RB.211-535C [Reference, FLIGHT 25-01-86]</p>
<p>Date : May 11, 1985 Carrier : Saudia Aircraft: 737 Location: Doha Injuries: - Total Occupants: 104 Phase : Take-Off</p>	<p>Uncontained failure of 2nd stage low pressure turbine disc. Engine fire. Fuel tank holed. Engine: P&W JT8D [Reference, FLIGHT 25-01-86]</p>
<p>Date : Sept. 8, 1985 Carrier : Jet America Aircraft: DC-9 Location: Oakland Int. Injuries: - Phase : Take-Off</p>	<p>Take-off aborted following turbine failure in one engine. Engine: P&W JT8D [Reference, FLIGHT 25-01-86]</p>
<p>Date : Sept. 9, 1985 Carrier : TWA Aircraft: 747 Location: London Heathrow Injuries: - Total Occupants: 94+ Phase : En route</p>	<p>Turbine failure in No. 1 engine. Landed back. Engine: P&W JT9D [Reference, FLIGHT 25-01-86]</p>
<p>Date : April 7, 1986 Carrier : WDL Flugdienst Aircraft: Metro (D-1BCF) Phase : Climb</p>	<p>Fuselage holed by uncontained failure of 3rd stage turbine. No injury. Engine: Garrett 331 [Reference, FLIGHT 24-01-87]</p>

Table 2-2: Material Related Life Considerations for
the J85 Engine Components [Coles, 79]

Component	Material	Temperature (deg. C)	Potential Concern
COMPRESSOR			
Blades	TI-6AL-4AV	< 370	FOD, HCF; HCF/LCF
Vanes	TI-6-4, INCONEL 718	< 150 < 370	
Spools	TI-6AL-4V, TI-6-2-4-2-	< 205, < 370	LCF, Critical Speeds
Disc*	INCONEL 718	400	Burst, LCF
Casing	GE AM355	400	LCF
SHAFT	A-286	205	LCF
COMBUSTOR			
Liner	HS188	985	Thermal Fatigue
Casings	INCO-W	540	
TURBINE			
Blades	RENE' 80	980	T-M-F, HCF, S.R.
Discs	INCONEL 718	370-650	Burst, LCF, Creep

*The Canadian J85-CAN-40/15 Compressor Disc Material is AM355

FOD = Foreign Object Damage
HCF = High Cycle Fatigue
LCF = Low Cycle Fatigue
TF = Thermal Fatigue
T-M-F = Thermo-Mechanical-Fatigue
S.R. = Stress Rupture

Table 2-3: Chemical Composition of A-286 from Literature
and as Required by General Electric Company

(Weight %)	AMS 5735 Bars and Forgings		General Electric Disc Forgings	
	Range	Typical	Range	Typical
Carbon (C)	0.08 max	0.045	0.08 max	
Manganese (Mn)	1.00-2.00	1.35	1.00-2.00	1.50
Silicon (Si)	0.40-1.00	0.95	0.40-1.00	
Chromium (Cr)	13.50-16.00	15.52	13.50-1600	14.75
Nickel (Ni)	24.00-28.00	26.06	24.00-27.00	26.00
Molybdenum (Mo)	1.00-1.50	1.25	1.00-1.50	1.25
Titanium (Ti)	1.75-2.25	1.95	1.75-2.25	2.00
Vanadium (V)	0.10-0.50	0.32	0.10-0.50	0.25
Aluminum (Al)	0.35 max	0.20	0.35 max	low
Iron (Fe)	Balance		Balance	
Sulfur (S)	0.04 max	0.018	0.04 max	
Phosphorus (P)	0.04 max	0.210	0.04 max	
Boron (B)	0.001 min	0.005	-	

**Table 2-4: Effect of Heat Treatment on the Properties
of A-286 Alloy [Muzyka, 72]**

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* (A) 980 C /1hr /Oil Quench + 720 C /16hr /Air Cool * (B) 900 C /2hr /Oil Quench + 720 C /16hr /Air Cool		Heat Treatment*	
		(A)	(B)
Tensile Properties at 21°C			
Ultimate Tensile Strength	(MPa)	1103	1076
0.2% Yield Strength	(MPa)	745	690
Elongation to Fracture	(%)	25	24
Reduction of Area	(%)	46	46
Stress Rupture at 650°C and 465 MPa			
Life	(hrs)	64	85
Elongation	(%)	15	10
Reduction of Area	(%)	20	15

**Table 2-5: Mechanical Properties of A-286
During Heat Treatment (A)**

Property	As Solution Annealed at 980°C	Solution Annealed + Precipitation Hardened at 718°C
UTS (MPa)	627	1007
0.2% YS (MPa)	255	690
Elongation in 2.25 inch (%)	47.5	25.0

**Table 2-6: Chemical Composition of Inconel 718 from Literature
and as Required by General Electric Company**

(Weight %)	Literature		GE, Forgings
	Range	Typical	
Carbon (C)	0.01 max	--	0.02-0.08
Manganese (Mn)	0.50 max		0.35 max
Silicon (Si)	0.75 max		0.35 max
Sulfur (S)			0.015 max
Phosphorus (P)			0.015 max
Chromium (Cr)	17.00-21.00	19.00	17.00-21.00
Iron (Fe)	15.00-21.00	18.50	15.00-21.00 (Balance)
Cobalt (Co)			1.00 max
Molybdenum (Mo)	2.80-3.30	3.00	2.80-3.30
Niobium (Nb) + Tantalum (Ta)	4.50-5.75 -	5.10	see below 4.75-5.50
Titanium (Ti)	0.30-1.30	0.90	0.75-1.15
Aluminum (Al)	0.20-1.00	0.50	0.30-0.70
Boron (B)	0.002-0.006	0.005	0.006 max
Copper (Cu)	0.15 max		0.30 max
Nickel (Ni)	50.00-55.00	25.50	50.00-55.00

Table 2-7: Solvus Temperature of Various Phases

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in Inconel 718

Phase	Solvus Temperature (°C)	Reference
Gamma Prime	843-871	[Paulonis, 69]
Gamma Double Prime	900	[Paulonis, 69]
Delta (Ni ₃ Nb)	982-1037	[Hall, 67]
NbC	1040-1093	[Hall, 67]
Laves	1148-1200	[Eiselstein, 65]

Solvus Temperature: The temperature at which the phase dissolves into solution.

**Table 2-8: Degradation Modes and Life Limiting Properties
for Turbine Engine Components [Koul, 87]**

Component	Land Based Turbines	Aero Engine Turbines
COMPRESSOR		
Blades	ER - COR, HCF	ER - COR, HCF
Vanes	ER - COR	ER - COR
Discs	LCF	LCF
TURBINE		
Blades	C, HC	TF, C, HC
Vanes	HC, C	TF, HC, C
Discs	C, LCF	LCF, C

ER = Erosion
HCF = High Cycle Fatigue
TF = Thermal Fatigue
HC = Hot Corrosion

COR = Corrosion
LCF = Low Cycle Fatigue
C = Creep

Table 3-1: A List of the A-286 Compact Tension Specimens
from the J85-CAN-40 Turbine Discs

Stage	Serial No.	TSN (hrs)	Condition	No. of CT Specimens	Test Temp. (°C)	Waveform
First	11J56-248	1771.9	RHT	5	23	Sawtooth
Second	P10323-7	3737.6	S.E.	3	23	Sawtooth
Second	11J96-205	1953.5	S.E.	2	23	Sawtooth
Second	11J96-223	2210.6	RHT	3	550	Sawtooth
"	"	"	"	2	"	Trapezoidal

TSN = Time Since New
RHT = ReHeat Treated
S.E. = Service Exposed
CT = Compact Tension

Table 3-2: General Electric Specification for Heat
Treated A-286 Disc Forgings

<u>Tensile Properties at 21°C</u>	
Ultimate Tensile Strength	896-1034 MPa
0.2% Yield Strength	483-620 MPa
Elongation in 2" or 4D	12-25 %
Reduction of Area	20-35 %
Stress Rupture Life at 650°C and 414 MPa	
	24 hrs

Table 3-3: Fatigue Crack Growth Rate Constants from Literature for A-286 Alloy 102

T	f	R	C	n	e.g. da/dN	Reference
23	3.000	0.05	1.8E-12	3.22	0.5	James, 76
23	24.00	0.10	9.7E-12	2.72	0.4	Reed, 77
538	0.067	0.05	1.1E-10	2.47	1.7	James, 76
650	0.500	-1.0	7.0E-10	1.95	1.4	Kitamura, 85

T = Test Temperature in deg. C

f = Test Frequency in Hz

R = Load Ratio

n = Paris law equation exponent

C = Paris law equation coefficient in (m/cycle)/(MPa $\sqrt{m}$)ⁿ

e.g. da/dN = An Example of the Crack Growth Rate in $\mu\text{m}/\text{cycle}$
at a Stress Intensity Factor of 50 MPa $\sqrt{m}$

Table 3-4: General Electric Specification for Heat Treated Inconel 718 Disc Forgings

Tensile Properties	Temperature (°C)	
	21	650
Ultimate Tensile Strength (MPa)	1276	1000
0.2% Yield Strength (MPa)	1034	862
Elongation in 2" or 4D (%)	12	12
Reduction of Area (%)	15	18
Stress Rupture Properties at 650°C and 690 MPa		
Minimum Life	25 hrs	
Elongation measured at RT	5% in 4D	

Table 3-5: A List of the Inconel 718 Compact Tension Specimens from the J85-CAN-15 Turbine Discs

Serial No.	TSO (hrs)	TSN (hrs)	Condition
WYOR-0123	0.0	0.0	New
11J774-54	1379.0	2044.9	Service Exposed
11J774-88	N/A	1520.3	Service Exposed
11J774-43	691.0	1523.1	Reheat Treated
11J774-67	1156.7	1617.4	Reheat Treated

4 Compact Tension Specimens Tested in each Group with

Test Temperature = 650 C

Test Frequency = 1 Hz

Load Ratio = 0.1

Max. Load = 5.0 kN

Environment = Lab Air

Waveform = Sawtooth

TSO = Time Since Overhaul

TSN = Time Since New

Retired Discs Part No. = 12A00288P01

Table 3-6: Fatigue Crack Growth Rate Constants for Inconel 718 [James, 79]

Constant	Mean Line	Upper Bound Line
C	5.7E-10	2.32
n	9.2E-10	2.32

Test Parameters

grain size = ASTM No. 11.5

melt heat = VIN-VAR

stock = 1.5 X 4 cm bar

temperature = 650 C

frequency = 0.67 Hz

load ratio = 0.05

Table 3-7: Summary of DC-PD Parameters

Parameter	AM355		A-286		Inconel 718
	RT	HT	RT	HT	HT
Current (Amp)	10	10	10	10 & 4	10 & 9
Amplifier Gains	500	500	1000	300-800	500-1500
Current On (Sec)	2.75	1.0	4.0	1.0	1.0
DAQ Time (Sec)	1.0	0.5	2.0	0.5	0.5
DAQ Rate (1/Sec)	20	20	10	30	20

RT = Room Temperature

HT = High Temperature

DAQ = Data Acquisition

Table 3-8: Type 302 Stainless Steel Gauge Sizes of
the Current Leads and Potential Probes

	Room Temp. (inch)	High Temp. (inch)
Current Leads	0.035	0.048
Potential Probes	0.008	0.018

Table 3-9: Valid Final Crack Lengths and the Corresponding Stress Intensity Factors for the A-286 and Inconel 718 CT Specimens

TSN (hrs)	T °C	Yield Stress (MPa)	UTS (MPa)	Flow Stress (MPa)	a (mm)	a/w final	Delta K	K _{max}	K _{maxL}
A-286 =====									
1772 RHT	23	705	1036	870.5	14.48	0.570	72.5	80.5	80.6
3738 SE	23	840	1122	981.0	15.09	0.594	79.3	88.1	88.3
1954 SE	23	896	1125	-	14.61	0.575	73.8	82.0	82.5
2211 RHT	550	616	907	761.5	13.72	0.540	65.3	72.5	72.9
Inconel 718 =====									
all	650	862	1000	-	14.43	0.568	72.0	80.0	80.0

$$\Delta K = \frac{\Delta P(2 + \alpha)}{B\sqrt{w(1 - \alpha)^{1.5}}} [0.886 + 4.640(\alpha) - 13.320(\alpha)^2 + 14.720(\alpha)^3 - 5.600(\alpha)^4]$$

where ΔK is the stress intensity factor range in $MPa\sqrt{m}$, ΔP is the load range (and is $= (1 - R)P_{max} = (1 - 0.1) \times 5000 = 4500N$), B is the thickness ($4.75 \times 10^{-3} m$), w is the width of the specimen ($25.4 \times 10^{-3} m$), $\alpha = \frac{a}{w}$, and a is the crack length.

$$\begin{cases} K_{maxL} = \sqrt{\frac{\pi}{4}(w - a)\sigma_{ys}} & \text{for } \frac{UTS}{\sigma_{ys}} < 1.3 \\ K_{maxL} = \sqrt{\frac{\pi}{4}(w - a)\sigma_{fs}} & \text{for } \frac{UTS}{\sigma_{ys}} \geq 1.3 \\ K_{max} = \frac{\Delta K}{1 - R}, \quad \sigma_{fs} = \frac{1}{2}(UTS + \sigma_{ys}) \end{cases}$$

ELASTIC LIMIT REQUIREMENT: $K_{max} \leq K_{maxL}$

Pre-crack Length $a = 6.67 mm$, $a/w = 0.2625$, $\Delta K = 30.3$ and $K_{max} = 33.7 MPa\sqrt{m}$

Table 4-1: a/w Comparison of Potential Drop (PD) versus Optical (OP) Crack Lengths at 550C for AH-355 CT Specimens

(a) Test Specimen No. 1

PD	0.2367	0.2748	0.3032	0.3300	0.3461	0.3772	0.4246	0.4778
OP	0.2303	0.2767	0.3033	0.3258	0.3461	0.3779	0.4280	0.4781
Dev. (mm)	+0.163	-0.048	-0.003	+0.107	0.000	-0.018	-0.086	-0.008

(b) Test Specimen No. 2

PD	0.2277	0.2600	0.2984	0.3362	0.3671	0.4069	0.4474	0.5066
OP	0.2280	0.2636	0.2988	0.3358	0.3791	0.4074	0.4551	0.5040
Dev. (mm)	-0.008	-0.091	-0.010	+0.010	-0.305	-0.013	-0.196	+0.066

(c) Test Specimen No. 3

PD	0.2445	0.6024
OP	0.2461	0.5868
Deviation (mm)	-0.041	+0.396

Dev. = Deviation of potential drop crack length from the optically measured value in mm

Table 4-2: a/w Comparison of Potential Drop vs Optical Crack Lengths at 550C for A-286 CT Specimens

Potential Drop	0.3623	0.4187	0.5687
Optical	0.3629	0.4198	0.5764
Deviation (mm)	-0.015	-0.028	-0.196

Table 4-3: a/w Comparison of Potential Drop vs Optical Crack Lengths at 650C for Inconel 718 CT Specimens

Potential Drop	0.2005	0.2478	0.2668	0.5824
Optical	0.2000	0.2461	0.2669	0.5827
Deviation (mm)	+0.013	+0.043	-0.040	-0.050

Table 4-4: A Comparison of the Various DC-PD Calibration Curves Available for CT Specimens

Reference	Material	CT Specimen Size		Potential Probe Locations	X ₁	X ₂	X ₃
Width	Thickness						
[Hicks and Pickard, 82]	-	26.0 mm	13.0 mm	Diagonal	-0.545516	+0.884318	-0.138778
[Aker, 86]	AM-355	25.4 mm	0.94 mm	Diagonal	-0.482249	+0.785741	-0.103492
[Andrews, 88]	Inconel 718	50.8 mm	12.5 mm	Centreline	-0.469594	+0.770866	-0.102825
This Study	AM-355	25.4 mm	0.94 mm	Centreline	-0.471496	+0.785050	-0.114153
	A-286	25.4 mm	4.75 mm	"	"	"	"
	Inconel 718	25.4 mm	4.75 mm	"	"	"	"

X₁, X₂, and X₃ are coefficient factors of the fitted second order polynomial

Table 4-5: Accuracy and Precision of the DC-PD System

Material Type	Standard Deviation from the Mean	
	Precision (mm)	Accuracy (mm)
AM-355	0.072	0.080
A-286	0.065	0.101
Inconel 718	0.023	0.023

- Accuracy refers to the closeness of the measurement to the actual value or the real value of the crack length.
- Precision indicates the closeness with which the measurements agree with one another quite independently of any systematic error involved.

Table 4-6: Fatigue Crack Growth Rate Constants from the Inconel 718 CT Specimens

TSN (hrs)	Condition	da/dN (m/cycle)			R	da/dN example at 50 MPa $\sqrt{m}$ (micrometer/cycle)
		C _{mean}	n	C _{upper}		
0.0	new	1.44E-09	1.81	1.63E-09	0.98601	1.94
2044.9	Service Exposed High Time	5.70E-11	2.88	6.49E-11	0.98581	5.07
1520.3	Service Exposed Medium Time	1.17E-09	1.86	1.26E-09	0.99292	1.82
1523.1	Reheat Treated	9.86E-11	2.76	1.20E-10	0.95834	5.87
1617.4	Reheat Treated	1.22E-11	3.33	1.41E-11	0.98758	6.41
Results from James, 79		5.72E-10	2.32	9.20E-10		8.04

Four 25.4mm CT specimens with 4.75mm thickness were tested for each group

Test Temperature = 650°C

Test Frequency = 1.0 Hz

Load Ratio R = 0.1

Max. Load = 5.0 kN

Environment = Lab Air

Waveform = Sawtooth

R = Correlation coefficient of the mean line

'James' Tests were performed at 650°C, 0.067 Hz, and R=0.05

Delta(K) unit is in MPa $\sqrt{m}$ for all results

da/dN values are calculated from upper bound C and n values

obtained from the CT specimens for each group and

a hypothetical stress intensity factor range

Table 4-7: Fatigue Crack Growth Rate Constants from the A-286 CT Specimens

TSN (hrs)	Condition	da/dN (m/cycle)			Corr. Coeff. (R)	da/dN example at 50 MPa $\sqrt{m}$ (μm /cycle)
		C _{mean}	n	C _{upper}		
1771.9	23C, 1.00Hz, RHT	1.57E-12	3.15	2.27E-12	0.96615	0.51
3737.6	23C, 1.00Hz, SE	1.13E-12	3.19	1.32E-12	0.98323	0.35
1953.5	23C, 1.00Hz, SE	2.86E-12	2.99	3.30E-12	0.98323	0.40
2210.6	550C, 1.00Hz, RHT	1.28E-11	2.90	1.52E-11	0.99052	1.28
2210.6	550C, 0.05Hz, RHT	7.94E-11	2.34	8.53E-11	0.99079	0.81

RHT : Reheat Treated

SE : Service Exposed

CT specimens: W=25.4mm, B=4.75mm

Load Ratio R: 0.1

Max. Load : 5.0 kN

Environment : Lab Air

Waveform : Sawtooth for 1.00Hz tests and

Trapezoidal for 0.05Hz tests with 20 sec hold time at max load

Delta(K) : the unit is in MPa $\sqrt{m}$ for all results

C and A : Paris law coefficient in $\{(m/cycle)/(MPa \sqrt{m})^n\}$

da/dt	hold time	A	m
Sawtooth	-	3.93E-08	2.95
Trapezoidal	20 sec	1.16E-08	2.39

da/dN values in the last column are calculated from upper bound C and n values obtained from the CT specimens for each group and a hypothetical stress intensity factor range

Table 4-8: A-286 Tensile Test Results

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TSN (hrs)	Test Temp (C)	0.2% YS (MPa)	E (GPa)	min. UTS (MPa)	El. (%)	RA (%)	Hardness (R _C)	da/dN example at 50MPa $\sqrt{m}$ (micrometer/cycle)
1771.9 RHT	23	705	180	1036	28.0	48.0	15-29-34	0.36
3737.6 SE	23	840	189	1122	17.6	42.9	33-37-40	0.29
1953.5 SE	23	896	183	1125	15.5	42.5	27-35-39	0.36
2210.6 RHT	23	702	187	1060	25.4	50.3	31-34-36	-
2210.6 RHT	550	616	173	907	16.9	46.8	--	1.10

TSN = Time Since New

YS = Yield Strength

E = Elastic Modulus

UTS = Ultimate Tensile Strength

El. = Elongation to Fracture

RA = Reduction of Area

R_C = Hardness in Rockwell C, given values are: min-average-max

da/dN values are calculated from the C and n values obtained from the CT specimens for each group and a hypothetical stress intensity factor range

Table 4-9: No. of Cycles to Failure for the A-286

Compact Tension Specimens

Specimen No.	TSN (hrs)	Test Temp. (°C)	Waveform [Hold Time] (Frequency) (Hz)	Failure	
				Cycles (X 1000)	Time (hr:min)
248A C D E F Average	1771.9 RHT	23	Sawtooth [0 sec] (1.00)	96 78 76 88 87 85	6:09 9:14 9:37 11:37 10:37 9:26
-7A B C Average	3737.6 SE	23	Sawtooth [0 sec] (1.00)	106 100 97 101	12:48 13:12 12:26 12:48
205A B Average	1953.5 SE	23	Sawtooth [0 sec] (1.00)	96 87 91	10:38 9:48 10:13
223A D E Average	2210.6 RHT	550	Sawtooth [0 sec] (1.00)	26 22 21 23	3:35 3:01 2:57 3:11
223B C Average	2210.6 RHT	550	Trapezoidal [20 sec] (0.05)	14 16 15	80:00 91:00 85:30

Table 4-10: Dimensions Measured on the J85-CAN15 Turbine Discs
Prior and After to Conventional Reheat Treatment

ALL DIMENSIONS IN INCHES	STAGE DISC S/N	1	1	2
	TSN (hrs)	1523.1	1617.4	1617.4
FLATNESS OF SURFACE °	measured	0.00028	0.00030	0.00077
	deviation	-	-	-
AFTER	measured	0.00045	0.00071	0.00075
	deviation	-	-	-
BEFORE	diameter	3.14976	3.14976	-
	measured	-	-	-
	deviation	-	-	-
AFTER	diameter	3.14979	3.14978	-
	measured	0.00006	0.00011	-
	deviation	-	-	-
BEFORE	measured	0.00066	0.00470	-
	deviation	-	(0.00371)	-
AFTER	measured	0.00021	0.00016	-
	deviation	-	-	-
BEFORE	distance	-	-	-
	measured	<0.001	<0.001	<0.001
	deviation	-	-	-
AFTER	diameter	0.0017	0.0015	0.1271
	measured	0.00015	0.00035	0.00075
	deviation	-	-	-

Table 4-11: Dimensions Measured on the J85-CAN15 Turbine Discs
Prior and After to Proposed Modified Reheat Treatment

ALL DIMENSIONS IN INCHES	STAGE DISC S/N	1	1	2
	TSN (hrs)	1714.8	1544.3	1727.5
FLATNESS OF SURFACE °	measured	0.00054	-	0.00036
	deviation	-	-	-
AFTER	measured	0.00068	0.00030	0.00036
	deviation	-	-	-
BEFORE	diameter	-	-	-
	measured	0.00007	0.00005	-
	deviation	-	-	-
AFTER	diameter	3.14962	3.14970	-
	measured	0.00021	0.00030	-
	deviation	-	-	-
BEFORE	measured	0.00100	0.00189	-
	deviation	-	(0.00089)	-
AFTER	measured	0.00087	0.00078	-
	deviation	-	-	-
BEFORE	distance	-	-	-
	measured	<0.001	<0.001	<0.001
	deviation	-	-	-
AFTER	diameter	0.0005	0.0005	0.1265/0.1274
	measured	<0.001	<0.001	<0.001
	deviation	-	-	-

Figure 1-1

Distribution of Calculated Residual Lives for Retired F100 Rotor Showing that the Majority of Discs Have Useful Life After Retirement When a Conventional Crack Initiation Life Prediction Criterion is Used to Life the Components

Figure 1-2 A Schematic of the J85-CAN-15 Gas Turbine Engine

OUTER
SEAL

FIRST STAGE
DISC

TORQUE
RING

SECOND STAGE
DISC

Figure 1-3

A Schematic View of the J85-CAN-15 Turbine Section Components

Figure 2-1 Effects of a Fan Disc Failure

Figure 2-2

Total Fatigue Life Segmented into Stages of Crack Development, Subcritical Growth and Final Fracture

Figure 2-3 Disc Material Data Scatter Based on LCF Crack Initiation Approach,

Figure 2-4 The General Structure of a Flight

Figure 2-5 Schematic Field Life Distribution Showing the Potential Benefits of Using a Rejection Criterion Based on Some Measure of Crack Propagation Rather than Conventional Crack Initiation

Figure 2-6 Schematic Representation of the Damage Tolerance Based Life Cycle Management of a Component where Fracture Mechanics is Used to Establish a Safer Return to Service Interval until a Dysfunction Crack is Found and the Part is Retired

Figure 2-7

The Use of Weibull Probability Plots Showing the Number of Service Hours Required to Propagate a Pre-Existing Flaw to a Detectable Size (N_i) and to a Dysfunction Limit (N_d) in Order to Determine a Safe Inspection Interval

Figure 2-8

A Schematic Cross Section of a Typical Single Spool Axial Flow Turbo-Jet Engine and the Gas Temperature, Pressure and Velocity Profiles which may be Expected

Figure 2-9 Stress and Temperature Distribution for Advanced Engine High Pressure Turbine Disc

Figure 2-10 Trends in Disc Alloy Developments

Figure 2-11 Time-Temperature-Precipitation Diagram of Vacuumelt ATS 28 which is Similar to A-286

Figure 2-12 Standard Heat Treatments for A-286 Alloy

Figure 2-13. Expected Phases in Inconel 718 Ni-Base Superalloy

Figure 2-14

Time-Temperature-Precipitation Diagram of Inconel 718 Ni-Base Super-alloy Showing the Conventional Heat Treatment Schedule

Figure 2-15 Repair Scheme to Remove Fatigue Cracks from Turbine Disc Blade Slots Involves Removing the Cracked Turbine Disc from Service, Machining the Existing Cracks from the Blade Slots, and Replacing the Original Notch Configurations with Ones having Larger Radii

Figure 2-16 Comparison of Stress Rupture Life at 345 MPa and 735°C in Service Exposed, Conventionally Reheat Treated, Laboratory Reheat Treated and HIP Reheat Treated Used Inconel X-750 Turbine Blades

Figure 2-17 UnHIPped versus HIPped Load Controlled LCF Properties of Waspaloy Turbine Discs at 538°C

Figure 2-18 Principle of DC-PD Method where Electric Current is Passed through the Specimen via 'A' and 'D' and the Voltage Drop Across 'B-C' is measured and Related to Crack Size

Figure 3-1 Capacitance Gauge Technique Assembly on a CT Specimen

Figure 3-2 DC-PD Test Setup and Measuring System

Figure 3-3. Stability of Voltage Ratio with Respect to Current Variation

Figure 3-4

Principle of Pulsed DC-PD and Synchronization of Potential Drop Measurement with Specimen Loading

Figure 3-5 The Magneto-Mechanical Effect on Voltage Ratio at a Frequency of 1.0 Hz

(a) Non-uniform Current Configuration.

(b) Uniform Current Configuration.

Figure 3-6 Potential Field Distribution in Single Edge Notch Specimens

Figure 3-7 Alternative Attachment Points for Current Leads and Potential Probes for Compact Tension Specimens

Figure 3-8 Laser Weld Position of Current Leads and Potential Probes on CT specimen

Figure 3-9 DC-PD System Grounding and Shielding on MTS Frame 'M-13D'

Figure 3-10

Specially Designed Adapters for Insulating the CT Specimens from the Load Frame

NOTES:

1. DIMENSIONS ARE IN INCHES

2. A-SURFACE SHALL BE PERPENDICULAR AND PARALLEL AS APPLICABLE TO WITHIN 0.002 TIR

3. THE INTERSECTION OF THE TIPS OF MACHINED NOTCH (X) WITH THE SPECIMEN FACES SHALL BE EQUALLY DISTANT FROM THE TOP AND BOTTOM EDGES OF THE SPECIMEN TO WITHIN 0.005.

LOADING HOLE Φ 2 X $\Phi .250 \pm .003$

Figure 3-11 ASTM Standard Compact Tension Specimen Dimensions

Figure 3-12 Location and Direction of the CT and Tensile Specimens Machined from the J85-CAN-40/15 Turbine Discs

Figure 3-13 Sawtooth and Trapezoidal Waveform Shapes for FCGR Testing

Figure 3-14 A Photograph of the Equipment Used for FCGR Testing

(A) A-286, J85-CAN-40 TURBINE DISCS

(B) INCONEL 718, J85-CAN-15 TURBINE DISCS

NOTE 1: IN THE REDUCED SECTION MACHINE A GRADUAL TAPER FROM THE ENDS TOWARD THE CENTRE, WITH ENDS NOT MORE THAN 1% LARGER IN DIAMETER THAN THE CENTRE (CONTROLLING DIA.).

NOTE 2: SURFACE FINISH TO BE SMOOTH AND FREE FROM SCRATCHES.

NOTE 3: NO UNDERCUT AT FILLET ENDS.

NOTE 4: ALL DIMENSIONS IN INCHES.

Figure 3-15

Tensile Specimen Geometry and Dimensions

J85-CAN-15
Turbine Disc, Stage 1
CF Drng. No. 501175

9.6000 inch Dia.

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Figure 3-16 Dimensions of the Critical Locations Investigated for Dimensional Stability of the First Stage J85-CAN-15 Turbine Discs

Figure 3-17 Dimensions of the Critical Locations Investigated for Dimensional Stability of the Second Stage J85-CAN-15 Turbine Discs

Figure 3-18

Firtree Slot Dimensions of the First and Second Stage J85-CAN-15 Turbine Discs

Figure 4-1 Response Time Comparison of Two Power Supplies

Figure 4-2 Comparison of Optical and Potential Drop Crack Lengths of AM-355 at 550°C

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Figure 4-3 Shifted Calibration Curves of A-286 CT Specimens. Adjust Zero and Gain at the Beginning of Test to Get a Voltage Ratio of 1.00

Figure 4-4 Calibration Curve Obtained from the A-286 CT Specimens

Figure 4-5 Comparison of Optical and Potential Drop Crack Lengths of A-286 at Room Temperature

Figure 4-6 Comparison of the CT Calibration Curve with those Obtained for Different Material, Potential Probe Locations, and Specimen Size and Thickness

Figure 4-7

Comparison of Room Temperature Optical Crack Lengths of AM-355 CT Specimens with Potential Probes at Centreline and Potential Drop Crack Lengths Calculated from AM-355 CT Specimens with Potential Probes at Diagonal Positions

Figure 4-8

Comparison of Room Temperature Optical Crack Lengths of AM-355 CT Specimens with Potential Probes at Centreline and Potential Drop Crack Lengths Calculated Thicker A-286 CT Specimens with Potential Probes at Centreline as Well

Figure 4-9 Finite Element Analysis of CT Specimen Simulating the Voltage Contours

Figure 4-10

The Effect of Microstructural Variation Between the Test and Reference Specimens on Crack Lengths Measured by the DC-PD Technique

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Figure 4-11 The Banded Regions Seen in the New Disc

Figure 4-12 Typical Optical Micrographs of the New Disc

Figure 4-13 Average Grain Size of the Examined First Stage J85-CAN-15 Turbine Discs

A) The $\gamma(\gamma')$ Precipitates

B) Dislocation Pileups

C) Dislocation Source at Particles

Figure 4-14 Transmission Electron Micrographs Obtained from the New Disc.

D) Dislocation Interactions

E) Grain Boundary Carbides

Figure 4-14 Cont'd Transmission Electron Micrographs Obtained from the New Disc

Figure 4-15

Average Grain Size of the Examined Second Stage J85-CAN-15 Turbine Discs

Figure 4-16 Typical Optical Micrographs of the 1520 hrs TSN Service Exposed Turbine Discs

Figure 4-17 Transmission Electron Micrographs Obtained from the 1520 hrs TSN Service Exposed Turbine Discs Showing the γ' - γ'' Morphology and Size

Figure 4-18 (A) Typical Cellular-Type Dislocations and (B) Only Slip Banding Observed in the 1520 hrs TSN Service Exposed Turbine Discs

Figure 4-19 Typical TEM Micrographs Showing (A) Grain-Boundary δ and (B) Serrated Grain Boundary structure in the 1520 hrs TSN Service Exposed Turbine Discs

A) Necklace-type Grain Structure

B) Equiaxed Grain Structure

Figure 4-20 Typical Optical Micrographs of the 2045 hrs TSN Service Exposed Turbine Discs

Figure 4-21 Transmission Electron Micrographs Obtained from the 2045 hrs TSN Service Exposed Turbine Discs Showing (A) the γ - γ' Precipitates, (B) Grain Boundary Carbides and (C) Ni_3Nb Needles at Grain Boundaries

Figure 4-22 Typical Dislocation Walls Observed in the 2045 hrs TSN Service Exposed Turbine Discs

Figure 4-23 Typical Low Angle Boundaries Observed in the 2045 hrs TSN Service Exposed Turbine Discs

Figure 4-24 Typical Optical Micrographs of the Second Stage Reheat Treated 1617 hrs TSN Turbine Disc

Figure 4-25 Typical Dislocations Observed in the 1523 hrs TSN First Stage Reheat Treated Turbine Disc

Figure 4-26

Typical Low Angle Boundaries Observed
in the 1617 hrs TSN Second
Stage Rejuvenated Turbine Disc

Figure 4-27 Average Hardness of the Examined First Stage J85-CAN-15 Turbine Discs

Figure 4-28 Average Hardness of the Examined Second Stage J85-CAN-15 Turbine Discs

Figure 4-29 Average Surface Hardness of the Examined First and Second Stage J85-CAN-15 Turbine Discs

Figure 4-30

FCGR Data and the Mean and 99.9% Upper Bound Lines Obtained from the New Inconel 718 Disc

Figure 4-31

Comparison of the FCGR Mean Lines of the New Inconel 718 Disc and the Literature

Figure 4-32 FCGR Mean Lines for A-286 Obtained by "James, 79"

Figure 4-33

FCGR Data and the Mean and Upper Bound Lines Obtained from the 1520 and 2045 hrs TSN Service Exposed Inconel 718 Discs

Figure 4-34

Comparison of the Inconel 718 FCGR Mean Lines of the 1520 and 2045 hrs TSN Service Exposed Discs, the New Disc, and the Literature

Figure 4-35 [†] FCGR Data and the Mean Lines Obtained from the 1954 and 3738 hrs TSN Service Exposed A-286 Discs.

TENSILE TEST IN LOAD AND STROKE CONTROL

SPECIMEN # 2238T

160

INIT AREA = 12.7779 SQ MM

AL = 2.58445

ML = 2.45 CM

TEST # 871015.165

25 DEG C.

TENSILE TEST IN LOAD AND STROKE CONTROL

SPECIMEN # 2238T

INIT AREA = 12.7779 SQ MM

AL = 2.58445

ML = 2.45 CM

TEST # 871015.165

25 DEG C.

Figure 4-36

Room Temperature Stress-Strain Curve Obtained from the 2211 hrs TSN Rejuvenated A-286 Turbine Disc

TENSILE TEST IN LOAD AND STROKE CONTROL SPECIMEN # 223CT
 INIT. AREA = 12.8101 SQ MM L = 3.095 CM TEST # 871103.150
 549 DEG C. 'M' = .15 'B' = -6.10000E-04

TENSILE TEST IN LOAD AND STROKE CONTROL SPECIMEN # 223CT
 INIT. AREA = 12.8101 SQ MM L = 3.095 CM TEST # 871103.150
 549 DEG C. 'M' = .15 'B' = -6.10000E-04

Figure 4-37

550°C Temperature Stress-Strain Curve Obtained from the 2211 hrs TSN
 Rejuvenated A-286 Turbine Disc

Figure 4-38 FCGR Data and the Mean and Upper Bound Lines Obtained from the (A) 1523 and (B) 1617 hrs TSN Reheat Treated Inconel 718 Discs

Figure 4-39

Comparison of the Inconel 718 FCGR Mean Lines from the 1523 and 1617 hrs TSN Reheat Treated Discs, the New Disc, and the Literature

Figure 4-40

Comparison of the A-286 Room Temperature FCGR Mean Lines from the 1772, 1954, and 3738 hrs TSN Discs, and the Literature

Figure 4-41

Comparison of the A-286 High Temperature FCGR Mean Lines from the 2211 hrs TSN Reheat Treated Discs and the Literature

Figure 4-42

Comparison of the A-286 Room and High Temperature FCGR Data and the Mean and 99.9% Upper Bound Lines

Figure.4-43 Typical SEM Fractographs of the Compact Tension Specimen Fracture Surfaces from (A & B) Room and (C & D) Elevated Temperature FCGR Tests

Figure 4-44

Comparison of the A-286 FCGR's in Terms of da/dN for Tests with Sawtooth and Trapezoidal Waveforms

Figure 4-45

Comparison of the A-286 FCGR's in Terms of da/dt for Tests with Sawtooth and Trapezoidal Waveforms

FCGR Curves for J85-CAN40 Turbine Discs With and Without Hold Time at 550C

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Figure 4-46

Comparison of the A-286 FCGR's in Terms of "Net Section Stress" for Tests with Sawtooth and Trapezoidal Waveforms

Figure 4-47 Typical SEM Fractographs of the Compact Tension Specimen Fracture Surfaces from the FCGR Tests with (A & B) Sawtooth and (C & D) Trapezoidal Waveforms

A) The New Disc

B) The 1520 hrs TSN Service Exposed Disc

C) The 2045 hrs TSN Service Exposed Disc

Figure 5-1 Typical SEM Fractographs of the Compact Tension Specimen Fracture Surfaces from J85-CAN-15 Turbine Discs at a Crack Length of 7 mm

D) The 1523 hrs TSN Rejuvenated Disc

E) The 1617 hrs TSN Rejuvenated Disc

Figure 5-1 Cont'd Typical SEM Fractographs of the Compact Tension Specimen Fracture Surfaces from J85-CAN-15 Turbine Discs at a Crack Length of 7 mm

Figure 5-2

Proposed Modified Heat Treatment Superimposed on the Inconel 718 Ni-base Superalloy Time-Temperature-Precipitation Diagram

Figure 5-3 Typical Stress, Temperature and Time Mission Profile for an Advanced Engine Operation (The Profile is for the Disc Rim Region)

Figure 5-4 Illustration Showing the Simplification Process for the Advanced Engine Mission

Figure 5-5 Effect of Frequency on FCGR's of A-286 Alloy

Figure 5-6

Allowable Temperature Difference Between AM-355 Sample and Reference Specimens at 550°C During FCGR Testing

Figure 5-7

Static Load Effect on Potential Drop Crack Length Measurement of AM-355 CT Specimens at 550°C During FCGR Testing

Schematic of Recommendation for Low Cycle Fatigue Test Specimens Preparation from Service Exposed Turbine Discs where Complete Standard Specimen Sizes are not Possible to be Machined

Figure 6-1

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