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**High Temperature Fatigue Crack Initiation
and Propagation Behaviour in
Inconel 718 Turbine Discs**

by
Robert G. Andrews
B. Eng.

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

Master of Engineering

Department of
Mechanical and Aerospace Engineering
Carleton University
Ottawa, Ontario
Canada

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Andrews, 1991

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**HIGH TEMPERATURE FATIGUE CRACK INITIATION
AND PROPAGATION BEHAVIOUR OF
INCONEL 718 TURBINE DISCS**

submitted by Robert G. Andrews, B.Eng.
in partial fulfilment of the requirements for
the degree of Master of Mechanical Engineering

Thesis Supervisor

for Chair, Department of Mechanical and Aerospace Engineering

Carleton University

June 20, 1991

Abstract

This thesis is a part of an investigation that is being conducted by the Institute for Aerospace Research (IAR) to look into the possibilities of extending the useable lives of Canadian Forces gas turbine engine components through a damage tolerance based life cycle management approach or component rejuvenation.

The effects of the conventional and an IAR developed damage tolerant heat treatment on the microstructure and elevated temperature mechanical properties of Inconel 713 wrought bar and service exposed J85-CAN15 turbine discs have been investigated. These mechanical properties include tensile, creep, fatigue crack initiation and short and long fatigue crack propagation (FCP) behaviour.

A direct current potential drop (DCPD) technique has been calibrated for use with a double edge notch (DEN) specimen for crack initiation, short and Paris region FCP at elevated temperatures. The sensitivity of the DCPD technique for initiation was $50 \pm 20 \mu\text{m}$ and the precision for FCP was $20 \mu\text{m}$. The DEN crack initiation and propagation results were substantiated by smooth low cycle fatigue and compact tension specimen testing, respectively.

Acknowledgements

I wish to express my sincere gratitude to my friend and colleague, Nick Bellinger for his constant help and guidance throughout the duration of this thesis. I would also like to thank Dr. Ashok Koul for his supervision and assistance in the development, progress and documentation of this work.

I would like to thank the entire staff of the Structures and Materials Laboratory, IAR, NRC for their kind help and assistance. In particular, the following people took extra time on my behalf: Dave Morphy and his phone, Tak Terada, Ray Dainty and Phil Adams. I would also like to thank Peter Au for his useful discussions and his assistance in the testing laboratory.

I wish to thank my thesis supervisor Professor M.C. de Malherbe for his encouragement and for providing the opportunity to work in this field.

I would like to thank Jane for her support and understanding throughout my Masters program as well as for her assistance in the preparation of the manuscript.

A final note of thanks to my parents for their constant encouragement, understanding and support throughout the development of my career.

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Nomenclature

A	Cross-sectional area of current path
a	Crack length or depth
AGARD	Advisory Group for Aerospace Research and Development
AMS	American Materials Society
ASTM	American Society for Testing and Materials
ATS	Applied test system
B	Specimen thickness
b	Width of narrowest section
C	Paris law constant - intercept
c	Half the surface crack length
C_{LM}	Larson Miller parameter constant
D	Grain size
d	Plate thickness
da/dN	Crack growth rate
DC	Direct current
E	Modulus of elasticity
DCPD	Direct current potential drop
F_{sn}	Short Crack growth geometric factor
FCGR	Fatigue crack growth rate
FCP	Fatigue crack propagation
IAR	Institute for Aerospace Research
K	Stress intensity factor
K_c	Critical stress intensity factor
K_{eff}	Adjusted stress intensity factor
K_f	Fatigue stress concentration factor
K_I	Mode I Stress intensity factor
K_{IC}	Fracture toughness for mode I crack growth
K_{II}	Mode II Stress intensity factor

$K_{\max}$	Maximum cyclic stress intensity factor
K_t	Stress concentration factor
K_1	Short crack stress intensity factor
K_2	Long crack stress intensity factor
k	Number of data points
L	Length of voltage path
LCF	Low cycle fatigue
LVDT	Linear variable displacement transducer
MTS	Material test system
m	Walker parameter to adjust for stress ratio
n	Paris law constant - slope
$P_{\max}$	Maximum load
p	Average stress at narrowest section
Q	Shape factor
q	Relative notch sensitivity
R	Stress ratio
r	Notch radius
R_c	Rockwell hardness number
RA	Reduction of area
r_p	Plastic zone size
$S_{y/x}$	Two variable standard deviation
T	Temperature
t	Half the specimen thickness
t_n	Depth of notch
t_R	Rupture life
UTS	Ultimate tensile strength
V_{off}	Voltage measured while current is off
V_{on}	Voltage measured while current is on
W	Specimen width
W_s	Adjusted single edge notch specimen width

x_i	Single variable data point
x,y	Two variable data point
Y	Linear elastic fracture mechanics geometric factor
Y_{CT}	Linear elastic fracture mechanics compact tension geometric factor
Y_{SEN}	Linear elastic fracture mechanics single edge notch geometric factor
α	Nondimensional crack length parameter (a/W)
β	Beta parameter - nondimensional voltage ratio
γ	Material matrix
γ'	Gamma prime strengthening precipitate
γ''	Gamma double prime strengthening precipitate
ΔK	Stress intensity factor range
ΔK_{th}	Threshold stress intensity factor range
ΔP	Load range
ΔR	Change in resistance
ΔV	Voltage range
$\Delta \sigma$	Stress range
δ	Ni ₃ Nb Precipitate
ϵ_f	Failure strain
ϵ_o	Instantaneous strain
ρ	Material resistivity
ρ_r	Radius of curvature of a notch
σ	Standard deviation
σ_a	Applied stress
σ_m	Mean stress
σ_{max}	Maximum stress
σ_{min}	Minimum stress
σ_{nom}	Remote applied stress
σ_y	Notch root stress
σ_{ys}	Two percent offset yield strength
ϕ	Parametric angle of the notch

Chapter 1

Introduction

In the late 1960s, low cycle fatigue (LCF) was recognized as a major damage mode for rotating components such as discs and spacers of gas turbine engines. By 1975, LCF performance had become the life limiting factor for over 75 percent of the major rotating components of advanced military engines such as the F100 [1,2]¹. In order to avoid catastrophic failures, engine manufacturers began to predict disc and spacer design lives on the basis of LCF crack initiation behaviour. In this design method, known as the safe life approach, a set of components is retired after a specified cyclic life, which is based on the number of cycles required to initiate a LCF crack in 1 in 1000 components. A major disadvantage of using this approach is that 999 components are retired in a crack free condition and the number of cycles required to propagate a crack to a subcritical length is not taken into account. Depending on the material and service stresses, crack propagation could constitute a large fraction of the component life. Recently, an alternative approach known as retirement-for-cause (RFC), is receiving more attention because it allows a component to remain in service until a crack is detected.

In the present RFC methodology, the fracture critical locations of a component are assumed to contain defects of a size corresponding to the detection limit of the nondestructive inspection technique used to inspect the component. These defects are then assumed to act as propagating cracks and their rates of propagation are established on the basis of fracture mechanics principles, using experimental crack growth rate data, until a predetermined crack length limit is reached beyond which the risk of failure becomes excessive[3]. This limiting crack length is usually determined by the engine manufacturer, by dividing the critical crack length after which failure of the component would occur, by a safety factor and this

¹Numbers in brackets designate references

is known as the dysfunction limit. The number of cycles required to grow these inherent cracks to the dysfunction limit is known as the crack propagation interval which is then divided by a safety factor to establish a safe inspection interval. This inspection interval is used as the basis on which discs are repeatedly returned to service until a crack is eventually detected, Figure 1.1 [4]. If the crack propagation rates are excessively high, the safe inspection interval may prove too short to be economically viable.

Retirement-for-cause is a form of the damage tolerance maintenance approach that has been used successfully in aircraft structural maintenance [3,5,6]. The success of this approach is heavily dependent upon the sensitivity and reliability of non-destructive inspection techniques and the accurate prediction of crack propagation lives of components using fracture mechanics data [1,3,5,6]. Ideally this data would be obtained from a statistically significant number of specimens which would be tested under conditions that simulate the engine operating environment. The specimen loading would also simulate a typical mission profile that an engine would experience during flight which includes various throttle excursions; engine start up, take-off and climb, cruising, manoeuvres, landing and shut down [2]. The economic benefits of the damage tolerance maintenance approach have been the driving force for much of the fatigue crack propagation (FCP) and threshold data generated for nickel-base alloys and for the development of turbine disc alloys with improved crack propagation resistance [1,3,5,6]. Damage tolerance concepts have already been used for designing several new generation engines, including F100-PW-220, F100-GE-100 and F109-GA-100 engines [7]. They have also been applied to existing engines such as PW F100, GE TF34 in the United States and GE J85-CAN40/15 engines in Canada for extending their usable service lives [7].

1.1 Background

Forged Inconel 718 turbine discs are used in the Canadian Forces Northrop CF-5 Freedom Fighter J85-CAN15 gas turbine engines. In 1970, life limits were

placed on these discs by General Electric, the engine manufacturer, based on a safe life approach where 1 in 1000 components would have developed a detectable crack (0.8 mm) at retirement. These safe life limits were believed to have been determined using the flying experience of the United States Air Force J85-5 engines, and assuming that the US flight missions were comparable to the Canadian missions [8]. Continuous efforts were made to update the safe life limits for the turbine and compressor discs of the J85-CAN15 engines from 1970 to 1981 by the Canadian Department of National Defence and General Electric. In 1981, due to premature LCF cracking encountered in the compressor discs and spacers, the mission severity factor which is an indication of the severity of the stresses encountered in these discs relative to the J85-5 engines, was increased to 1.8. The resulting reduction in the safe life limits created logistic problems in the availability of new parts and replacement of expired parts. In order to allow time for the purchase, delivery and installation of new parts, a damage tolerance based maintenance methodology was developed for these engines [8,9]. Considerable uncertainty still surrounds the safe lives and damage tolerance based safe inspection interval predicted for the turbine and compressor discs and spacers of this engine by the engine manufacturer.

This thesis forms a part of an ongoing investigation that is concerned with life prediction of J85-CAN15 gas turbine engine components. This investigation was started in 1983 by the Structures and Materials Laboratory of the Institute for Aerospace Research (IAR) in co-operation with Carleton University and the Department of National Defence. The compressor section of the J85-CAN15 engine has been extensively investigated with respect to both destructive testing (tensile and crack growth rate) and nondestructive inspection (liquid penetrant, eddy current and ultrasonic leaky wave), but no LCF crack initiation tests and only a limited number of crack growth rate tests have been conducted on the hot section disc material. The safe life and damage tolerance based inspection interval predictions demand that extensive data bases be generated for the material. The LCF crack initiation life can be significantly influenced by the notch sensitivity of the disc or spacer materials.

There are no established standards for generating LCF crack initiation data that would simulate the influence of component discontinuities such as bolt holes on the LCF life. Therefore, the development of experimental techniques to obtain data for the initiation of fatigue cracks from stress concentrations, such as bolt holes, is considered important [10]. In addition, advanced damage tolerance based life prediction algorithms require that short crack growth behaviour of disc materials be characterized and that data bases be generated at disc operating temperatures for safe inspection interval calculation. Again, no standard guidelines exist for generating these data bases and the short crack growth rate behaviour of J85-CAN15 disc material has not been characterized to date.

1.2 J85-CAN15 Turbine Disc Material and Heat Treatments

The J85-CAN15 turbine discs are manufactured out of Inconel 718, a Ni-Fe base superalloy, forgings. The forgings are heat treated using a conventional heat treatment developed for this alloy prior to machining the discs. The IAR has recently developed a *damage tolerant* heat treatment for this alloy which produces a microstructure that would ideally reduce creep and fatigue crack growth rates (FCGRs) without sacrificing LCF initiation life, relative to the conventional heat treatment. This damage tolerant heat treatment is being considered for two separate applications, first as an initial heat treatment for new Inconel 718 discs replacing the conventional heat treatment and secondly as a rejuvenation reheat treatment for safe life expired discs.

Once the safe life of a turbine disc has been exhausted, it may either be replaced by a new disc or rejuvenated and put back into service. Rejuvenation would reduce the cost of ownership of the engine [9,11]. The rejuvenation of turbine discs for eliminating service induced high strain LCF damage through reheat treatments is still a new technology [4,9,12,13]. High strain LCF damage prior to cracking is in the form of dislocation substructure formation which can reduce the materials resistance to fatigue crack initiation and propagation [9,13,14,15].

1.3 Objectives

The three main objectives of this thesis are:

- 1) To determine the properties of service exposed Inconel 718 J85-CAN15 turbine discs and to evaluate the effectiveness of conventional and damage tolerant heat treatments in rejuvenating the microstructure and mechanical properties of service exposed discs.
- 2) To determine the influence of notch sensitivity on the fatigue crack initiation of Inconel 718 material subjected to conventional and damage tolerant heat treatments.
- 3) To expand the material data bases required for the damage tolerance based life prediction of J85-CAN15 turbine discs. The data bases of interest are fatigue crack initiation and both short and Paris region fatigue crack propagation rates.

Chapter 2

Literature Review

2.1 Inconel 718

Inconel 718 is the most widely used nickel-base superalloy currently reaching 45 percent of the total production of the nickel-base superalloys in wrought form and 25 percent in cast form [16]. It possess high strength, moderate corrosion resistance, good resistance to elevated temperature LCF, good creep properties and excellent weldability [4,11,17]. Both wrought and cast forms of Inconel 718 are used extensively throughout a gas turbine engine, including: engine frames, casings, exhaust nozzles, shafts, compressor and turbine discs, seals, spacers, compressor blades and vanes and fasteners. Turbine engines which contain Inconel 718 in major amounts are: TF39, CF6, LM2500, 5000, F101, F110, F404, TF34, T700, CT7, J85 and CFM56 [18].

2.1.1 Physical Metallurgy of Inconel 718

Extensive literature is available on the physical metallurgy of Inconel 718 and therefore only a brief review will be presented here [4,11,17,19-30]. Inconel 718 is a versatile, high strength, precipitation hardenable nickel-base superalloy developed for use at temperatures up to $\sim 700^{\circ}\text{C}$ [19,28]. The nominal composition of Inconel 718 is: 18Fe, 18Cr, 3Mo, 5Nb, 1Ti, 0.5Al, .04C, and .004B [30]. Like most superalloys, Inconel 718 is a complex alloy, containing several phases: NbC, TiC, TiN, δ , Laves, γ' and γ'' [17,19,22,31], as shown in Figure 2.1 [17]. The major elevated temperature strengtheners are ordered body centred tetragonal, gamma double prime (γ'') precipitates and ordered face centred cubic, gamma prime (γ') which produce coherency strains in the γ matrix [30,32]. In contrast to most Ni-base superalloys which are primarily hardened by Al and Ti rich γ' , a niobium rich Ni_3Nb precipitate,

γ'' , is the primary strengthener of Inconel 718 . Along with these two strengthening phases a number of other minor phases can also form in this alloy. Depending on the heat treatment used incoherent Ni_3Nb precipitates (δ) can form upon overaging of γ'' and appear as needles within the grain interiors and at the grain boundaries [4,19]. Laves phase 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 the order of 1150°C to 1200°C [9], but depending on the heat treatment, they may appear at aging temperatures as low as 780°C [19]. This phase is very harmful to mechanical properties because of its brittle nature and it also decreases the alloy strength by depleting the matrix of valuable strengthening elements [17]. Similar to the Laves phase, NbC and TiN particles also form during solidification of the alloy. Although some amount of NbC precipitate is essential for controlling the grain size of the alloy during heat treatment it reduces the strength of the alloy by removing γ'' forming Nb from the matrix [4]. The solvus temperatures of both major and minor phases of Inconel 718 are given in Table 2.1 and a time-temperature-precipitation diagram is shown in Figure 2.2.

2.1.2 Heat Treatments for Inconel 718

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. The schedules for the two heat treatments considered in this thesis, the conventional heat treatment, described in AMS 5596C and ASTM B637-84a and the IAR damage tolerant heat treatment [4,19,33,34], are given in Table 2.2.

A fine grained microstructure is generally considered desirable for good ductility, tensile strength and resistance to LCF crack initiation [24]. On the other hand, a larger grain size tends to reduce elevated temperature crack growth rates and improves creep resistance [4,20,35]. The grain coarsening behaviour of Inconel 718 is shown in Figure 2.3 [4,19]. It should be noted that rapid grain growth commences

between 1040°C and 1050°C because the primary NbC precipitates begin to dissolve in this temperature range [4]. Rapid grain growth is not expected to occur at the conventional solution treatment temperature of 955°C because this temperature is below the NbC and Ni₃Nb precipitate solvus temperatures that restrict grain coarsening [4]. Therefore, the original grain size is maintained, but the strengthening precipitates are fully solutioned. A solution treatment temperature of 1032°C for the damage tolerant heat treatment dissolves the Ni₃Nb precipitates without causing rapid grain growth although the grain size is coarser than that obtained during the conventional heat treatment [4]. A solution time of one hour is adequate for homogenizing the microstructure. Longer solution treatment times could increase the grain size through Ostwald ripening of NbC precipitates even if the solution treatment is kept below the NbC solvus temperature [4].

In Inconel 718, the precipitation of the grain boundary Ni₃Nb needles may, under certain conditions, create a *serrated* grain boundary structure [4,36]. The serrations arise from the cellular precipitation of Ni₃Nb following the Tu-Turnbull mechanism and are angular in nature [4,36]. While the Ni₃Nb needles reduce the volume fraction of the major strengthening phase γ'' which may decrease the intragranular strength of the alloy [19,28,31], its precipitation at the grain boundaries, in small quantities, may lead to improved fatigue crack propagation resistance and creep properties. The serrated grain boundaries resist relative grain boundary motion by pinning the grain boundaries and reduce fatigue crack growth rates by suppressing intergranular cracking which forces the crack to follow a transgranular path and leads to slow crack growth [4,11,34,36]. Serrated grain boundaries also improve the creep properties by restricting both grain boundary diffusion and grain boundary sliding which are the two dominant deformation mechanisms operative during service [36-38]. Limited information is available on direct aging conditions where the material is slow cooled from the solution temperature to an intermediate aging temperature and held at this temperature to produce grain boundary Ni₃Nb needles and serrations [4]. For solution treatment temperatures of 1032°C, 1066°C and 1080°C, the time-temperature-precipitation curves for a range of direct aging

conditions are presented in Figure 2.4 [4]. The nose of the Ni_3Nb time-temperature-precipitation curves lies in the vicinity of 843°C in all cases, Figure 2.4. A direct aging treatment of 843°C for four hours has been shown to produce profuse serrations while longer aging times led to intragranular Ni_3Nb precipitation that results in severe loss in notch ductility [19,26]. Therefore, to produce a serrated grain boundary structure the damage tolerant heat treatment uses a direct aging treatment of 843°C for four hours. Adding a one hour reage at 926°C re-solutions any overaged intragranular γ' and γ'' precipitates that form during the direct aging [19]. These overaged γ' and γ'' precipitates can result in a loss of strength if they are not removed [19,36]. Some coarsening of the remaining Ni_3Nb precipitates, resulting from the direct age (843°C), will however occur during the reaging treatment at 926°C [19].

A single age in the temperature range 704 to 760°C for 10 to 20 hours produces fine γ' and γ'' precipitates within the alloy matrix to optimize tensile strength [19]. Tensile strength can be increased by 70 to 90 MPa without decreasing ductility [28] by using a double age with the second aging temperature 100°C below the first age [19]. Both the conventional and the damage tolerant heat treatments use a standard double aging treatment: 718°C for 8 hours, furnace cooled to 621°C for 8 hours, air cooled to room temperature [26,28]. The γ' and γ'' precipitate sizes which result from the double age treatment are given in Table 2.3 [11,22,30,34]. Transmission electron microscopy and dark field electron diffraction techniques, can be used to differentiate between the γ' and γ'' precipitates by orienting the grain in the $[001]$ direction (or crystallographically equivalent), using the (010) reflection to show γ' and γ'' and using the $(1\frac{1}{2}0)$ reflection to show just γ'' [21,22,25,39]. The average hardness is expected to increase from 24 to 44 Rc when using the double aging sequence [11,33,34], however, by lowering the aging temperature, higher hardness values can be reached, but longer times are required [19].

2.1.3 Component Distortion

The damage tolerant and conventional heat treatments have also been considered for use as reheat treatments in the rejuvenation of service exposed turbine discs. The possibility of component distortion during reheat treatment must be considered prior to applying the rejuvenation techniques to actual components. Tests have already been carried out on actual discs taken from the J85-CAN15 engine to determine the amount of distortion that results from the conventional and the damage tolerant reheat treatments [11]. All dimensions of the service exposed discs, prior to rejuvenation, were within the engine manufacture's tolerances, except the perpendicularity of the plane of the 1617 hour and 1544 hour time since new (TSN) discs. After the rejuvenation reheat treatments all dimensions were found to be within the engine manufactures tolerances, including the two out of tolerances which were removed by the reheat treatments.

2.2 Tensile Properties

Tensile tests provide material strength and ductility data in terms of load versus displacement from which basic material properties including: yield strength (σ_{ys}), modulus of elasticity (E), ultimate tensile strength (UTS), failure strain (ϵ_f), elongation and reduction of area (RA) can be determined. The American Society for Testing and Materials (ASTM) has developed accepted standards for performing high temperature tensile tests (E21-79) including standards for the calculation of the material properties [40]. The yield stress can be determined using the intercept given by the 0.2 percent offset strain, while the slope of the offset can be used as an estimate of the modulus of elasticity. The ultimate tensile strength is defined as the maximum stress achieved for each specimen and can be found by dividing the maximum load by the initial cross-sectional area. The specimen elongation and reduction of area can both be calculated as a percentage of their original values,

using the gauge length and cross-sectional area measured from the specimens before and after testing.

The material parameters which have the greatest effect on the tensile strength of Inconel 718 at elevated temperature are grain size, γ' and γ'' precipitate sizes and volume fraction. The tensile strength of Inconel 718 decreases with increasing temperature, with a significant loss of strength above 700°C [41,42]. The maximum operating temperature for the turbine discs in the J85-CAN15 engine has been estimated to be 650°C [3,11,20,34] and therefore the tensile tests must be performed at this temperature in order to simulate the engine operating environment. As mentioned previously, a fine grain size results in the highest tensile strength, as shown in Figure 2.5 [21]. The effects of aging on the tensile properties of Inconel 718 are also shown in Figure 2.5. Table 2.4 summarizes high temperature tensile properties of Inconel 718 for a variety of microstructures.

2.3 Fatigue Crack Initiation

Design defects or metallurgical flaws already exist in a component and these act as stress raisers which can drastically shorten, or completely eliminate the fatigue crack initiation stage. This results in a reduction in the potential life of a component [43]. Stress raisers can either be inherent in the component design (bolt holes and fillets) or form during the manufacturing process (surface scratches and machining marks), or exist as material flaws (inclusions, gas pores and dislocation structures). They can also develop during cyclic straining by producing sharp peaks (extrusions) and troughs (intrusions) in the material due to nonreversible slip which is considered to be the initial stage in microcrack formation [38,44].

Estimation of the fatigue crack initiation life is the basis for the safe life design approach used for the J85-CAN15 turbine engine discs. The definition of crack initiation is a controversial topic in fracture mechanics literature, for instance, at what point does a defect develop into a crack [45]. Many authors define crack initiation by a specific criterion for instance, when the crack reaches a specific length

(1 cm, 1 mm, or 1 μm) or when the crack attains a length equal to some multiple of a characteristic microstructural unit (three grain sizes) [38]. In general, no precise definition for crack initiation has been or perhaps can be identified because the sensitivity and reliability of the crack detection techniques also influence the results while generating the crack initiation data bases [38,45].

Turbine discs are subjected to both radial and circumferential stresses due to rotational forces. The critical locations in the J85-CAN15 turbine engine discs, according to the engine manufacturer, are the bolt holes and fillets, with the latter considered more critical [11]. Fatigue cracks that initiate and propagate from the bolt holes are assumed to grow radially inward due to the circumferential stresses, which is similar to cracks found in the compressor discs of the same engine [9]. Fatigue tests that are carried out using tension-tension axial loading simulate engine operating conditions for either radial or circumferential stresses at the bolt hole. The actual mission profiles which an engine encounters during a flight involve a sequence of cycles that differ in strain range and mean stress [9]. In order to determine the initiation life using these mission profiles, certain cycles that produce negligible damage may be ignored and a cumulative effect of more severe cycles is usually considered [9]. A trapezoidal waveform is the ideal load spectrum to simulate these cycles, but, this waveform is the most severe of all waveforms (lowest number of cycles to failure) and the hold times required to simulate an actual mission profile would require extremely long test periods. A practical alternative to the trapezoidal waveform is the triangular waveform, Figure 2.6, which results in shorter test periods and larger cycle counts [4,11,34]. The standard definitions of key load or stress variables are shown in Figure 2.6 and defined as follows [38]

$$\Delta\sigma = \sigma_{\max} - \sigma_{\min} \quad (2.1)$$

$$\sigma_m = \frac{\sigma_{\max} + \sigma_{\min}}{2} \quad (2.2)$$

$$R = \frac{\sigma_{\min}}{\sigma_{\max}} \quad (2.3)$$

where $\Delta\sigma$ is the stress range, σ_m is the mean stress and R is the stress (load) ratio.

An example of load amplitude fatigue data for 7075-T6 aluminum alloy notched specimens is shown in Figure 2.7. Considerable scatter in the fatigue life is shown in this figure, with less scatter evident at the higher stress levels which is believed to result from the shorter initiation period [38]. Since scatter is common in fatigue tests, the statistical character of the data must be considered when determining life limits [9,38,46]. The scatter originates from variations in testing environment, specimen surface preparation, alignment of the test machine, and a number of metallurgical variables [38]. High temperature fatigue crack initiation results for Inconel 718 are given in Table 2.5.

Smooth cylindrical specimens have been used extensively for both stress and strain controlled fatigue initiation tests because the influence of surface or near surface defects can be simulated for large surface areas [9,33,47]. These specimens assume that the fatigue crack initiation life of a notched component is equal to the fatigue life of a smooth specimen if the notch tip of the component and the smooth specimen both experience the same strain history, which is also referred to as the local strain approach [48]. Therefore, the smooth specimen should simulate the notch root strain behaviour of the material at a discontinuity in a disc, Figure 2.8 [9,48]. ASTM has developed standards for carrying out high temperature fatigue tests (E606-80) using smooth cylindrical specimens and a load drop technique for monitoring initiation [47]. If the smooth specimen results are to be compared to other tests carried out on different specimens, then the loading conditions must be the same for both. Since local strain can be difficult to measure in areas where initiation is expected to occur, some initiation tests are carried out in load control [48]. For load controlled testing, the initiation criterion used for the smooth specimens is commonly specimen failure. Specimen failure is an acceptable approximation in this situation, since the main purpose of the smooth specimen

testing is to compare the influence of microstructural and surface preparation variables on fatigue life. Also, if a small net section is used it can be assumed that the propagation interval will be only a small percentage of the specimen life.

2.3.1 Application of The Direct Current Potential Drop Technique to Fatigue Crack Initiation Testing

In addition to the load drop technique, the direct current potential drop (DCPD) technique has also been used to monitor crack initiation during fatigue. Traditionally, the DCPD technique has been used to monitor crack growth in terms of the change in crack length in fatigue test specimens. This technique supplies a constant DC current to a specimen while the voltage is measured across the expected crack path. As the crack grows the resistance of the specimen increases and subsequently the voltage increases in accordance with Hooke's law. The increase in voltage can then be calibrated against crack length by analytical or experimental techniques [6,34,49-51]. The theoretical change in material resistance can be represented by the differential equation:

$$\Delta R = \frac{L}{A}d\rho + \frac{\rho}{A}dl - \frac{\rho L}{A^2}dA \quad (2.4)$$

where ΔR is the change in resistance, L is the length of the voltage path, A is the cross-sectional area of the current path and ρ is the material resistivity.

The sensitivity of the DCPD technique is limited by a number of factors, such as the electrical stability and resolution of the system, crack front curvature, electrical contact between crack surfaces, and changes in electrical resistivity with plastic deformation and/or temperature variations [11,49,51-53]. The use of a reference specimen can reduce the effect that temperature variations and electrical stability has on the sensitivity of the technique [11,34,49]. The ratio between the test specimen potential and the reference potential results in a nondimensional DCPD parameter, β :

$$\beta = \frac{(V_{on} - V_{off})_{test\ specimen}}{(V_{on} - V_{off})_{reference}} \quad (2.5)$$

where V_{on} is the potential measured with the DC current on and V_{off} is the potential measured with the DC current off. The use of a reference specimen partially accounts for drift in the potential measurements caused by the amplifiers since both the specimen and reference potentials should drift the same amount [11,34,49]. The change in resistivity of the specimen due to a change in temperature is taken into account by the reference as long as it is placed in the furnace along with the specimen [11,34,49]. A pulsed DC current can improve the electrical stability, and also eliminate heating effects in the current leads and at the connection points [11,34,49].

The DCPD technique is based on the measurement of the change in area, equation 2.4, which can create difficulties in determining the number of cycles to crack initiation since different crack front characteristics such as multiple initiation and crack front curvature can occur [54,55]. Figure 2.9 shows examples of some common crack front shapes at initiation, including: (a) straight front, (b) small curvature, (c) multiple initiation and (d) thumbnail (large curvature) cracks. Under most circumstances, the DCPD technique assumes an equivalent straight fronted crack based on the crack area during the calibration between voltage and crack length [56]. It can be seen from Figure 2.9 that for a thumbnail crack, the depth is much larger than for a straight fronted crack of the same crack area, while for multiple initiation and small curvature cases, there is only a small difference in the crack depth [57]. Therefore, by using an equivalent straight fronted crack calibration, the DCPD technique can accurately measure initiated cracks. However, for crack fronts with large curvature (thumbnail), the DCPD crack length may not be as accurate as for other types of cracks. Minimum detectable crack lengths or DCPD sensitivity that were found in the literature are given in Table 2.6.

ASTM standard E647-83 [58] defines the precision of a crack length measuring system as: *the standard deviation (σ) on the mean value of crack length*

determined for a set of replicate measurements. Since the initiation period of a test normally provides a single constant measurement which is taken repeatedly these values can be used to determine the precision of the DCPD technique. For a more conservative value which includes a larger percentage of the data, the precision can be defined by using three standard deviations from the mean precision value (3σ):

$$3\sigma = 3 \sqrt{\frac{1}{k-1} \left[\sum x_i^2 - \frac{(\sum x_i)^2}{k} \right]} \quad (2.5)$$

where k is the number of data points and x_i is a data point.

The DCPD technique has been applied successfully using a one percent voltage increase as the definition of crack initiation in references [59-64]. The DCPD technique was selected because other workers have found it useful for crack initiation studies and because it had already been used at IAR for monitoring crack growth rates and, therefore, only required some modifications to be used for monitoring fatigue crack initiation. The technique used is described in more detail elsewhere [11,49,50].

2.3.2 Fatigue Crack Initiation Double Edge Notch Specimens

The double edge notch (DEN) specimen is normally used for fracture toughness testing, but can also be used for crack initiation testing as proposed in the AGARD engine disc cooperative test programme [60]. Notched specimens are more representative of the stress concentrations that can be present in actual components, such as at bolt holes, than in smooth specimens [38]. The notch sensitivity of a material is not accounted for in smooth specimen tests although the specimen surface is representative of the notch root strain [26,65]. The ideal specimen to simulate the stress concentration of a bolt hole is a plate specimen with a centre-hole, however, there are difficulties in applying the DCPD technique to such a specimen. If the voltage probes are connected above and below the hole, the

shortest current path will always exist on the uncracked side of the hole. However, if a pair of voltage probes are located on each side of the hole, the additional stress concentration resulting from the attachment of the leads (heat affected zone) could result in premature crack initiation at these probes. The DCPD voltage probes can be attached to a DEN specimen above and below each of the notches without causing additional stress concentrations within the notch region. These probes can then monitor each notch separately and detect small voltage increases which correspond to small changes in the voltage path resulting from the formation of a crack.

A stress concentration factor (K_t) is normally used to account for the reduction in tensile strength resulting from the notch geometry in these specimens which can be determined by a number of methods [38,66,67]. For standard notch geometries, graphs based on Neuber's approach, similar to the one shown in Figure 2.10 [66] can provide an accurate value of K_t . For geometries which are not standard, formulas are available, or can be derived, based on Neuber's approach. For example, the equation for shallow external notches on two sides of a bar subjected to axial load is given by [68]:

$$\frac{\sigma_{\max}}{p} = 3 \sqrt{\frac{t_n}{2\rho_r}} - 1 + \frac{4}{2 + \sqrt{\frac{t_n}{2\rho_r}}} \quad (2.6)$$

where ρ_r is the radius of curvature at the base of the notch, t_n is the depth of the notch, $\sigma_{\max}$ is the maximum stress and p is the average stress at the narrowest section given by:

$$p = \frac{P_{\max}}{bd} \quad (2.6a)$$

where d is the thickness, b is the width of the narrowest section and $P_{\max}$ is the maximum load.

The fatigue strength is reduced by a factor K_f , which represents the effective stress concentration factor as affected by plastic flow, notch root surface area and volume considerations [38]. The fatigue stress concentration factor, K_f , is determined as a function of K_t and the relative notch sensitivity factor, q :

$$K_f = 1 + q(K_t - 1) \quad (2.7)$$

where the magnitude of q depends on the material and the notch geometry and can be found from literature, as shown in Figure 2.11 [38,69].

The DCPD technique and DEN specimen make it possible to distinguish between the periods of initiation and propagation when a specimen is cycled to failure. The duration of the fatigue crack initiation stage relative to the propagation stage depends on the notch root radius, Figure 2.12. When the radius is small for a given stress level, the total fatigue life is dominated by the propagation stage. On the other hand, when the notch root radius is large, most of the total fatigue life involves initiation of the crack at the notch root.

2.4 Fatigue Crack Propagation

Component fatigue failure begins with the initiation of a crack from a flaw or stress raiser, which then propagates until a critical length is reached at which point failure of the component occurs. The rate at which a crack propagates depends on the applied stress and the component material.

The rate at which a fatigue crack grows depends on the stress, microstructure and environment ahead of the crack tip. Efforts have been made to determine a parameter that would uniquely describe these conditions and the three main parameters that have been developed are: (1) the stress intensity factor, K , (2) the J -integral and (3) the J^* -integral. The stress intensity factor uniquely describes the stress-state ahead of the crack tip and 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 [9,11,70]. For a given material, stress ratio, and test environment, a specific applied stress intensity factor range (ΔK) produces the same growth rate, in the *Paris* region (section 2.4.3), regardless of the applied stress and crack length [9,70]. In cases where significant deformation results, the other two parameters may be more applicable. The J-integral is said to describe the stress-strain field of the crack tip while the J^* -integral describes the stress-strain rate field [9]. In Inconel 718, ΔK has been found adequate for describing the crack growth behaviour in air and other aggressive environments.

The growth of a fatigue crack can be divided into a number of different regimes including small, short and long crack growth. Small crack growth is described as growth which occurs when the crack is microstructurally small, for example less than one grain size in length. However, this definition has also been suggested as a possible criterion for initiation, see section 2.3. The definition of a short crack is debatable, with one working definition being *any crack whose growth cannot be predicted using fracture mechanics technology* [10], specifically linear elastic fracture mechanics. Mechanically long cracks are those which can be predicted by linear elastic fracture mechanics. Fatigue crack propagation data, in the form of crack growth rate per cycle (da/dN), as shown in Figure 2.13, versus alternating stress intensity range (ΔK) curves, for nickel-base alloys usually fall into three regimes, (1) Near Threshold, (2) Paris and (3) Final Failure, as shown in Figure 2.14 [1,38,71].

2.4.1 Threshold Region

In the near threshold region I, linear elastic fracture mechanics predicts a rapid decrease in crack growth rate with decreasing ΔK , approaching a lower limit which is known as the threshold stress intensity factor range (ΔK_{th}). The FCGR in the threshold region in nickel-base alloys, has been found to be sensitive to changes in both microstructure and load ratio with faceted growth the dominate

micromechanism [1,72,73]. Au et al. [33] and Clavel et al. [74] observed cyclic softening in Inconel 718 for a number of different heat treatments with fine disc shaped γ'' precipitates. Coarse grains promote intragranular deformation by suppressing grain boundary sliding [1]. In Inconel 718 for ΔK -values less than about $20 \text{ MPa} \sqrt{\text{m}}$, James and Mills [75], found lower growth rates for a modified heat treatment with a grain size of 200 to 300 μm than for a conventional heat treatment with a grain size of 10 to 20 μm . With increasing temperature both the elastic modulus and the yield stress of Inconel 718 fall, which should decrease ΔK_{th} and increase growth rates. With increasing temperature, the fracture surface appearance can change from a highly angular one, to a flatter, smoother one, although in the near-threshold region the grain size facets are still present [1]. Faceted transgranular growth persists at higher temperatures in Inconel 718 with little change in the slip distribution and fracture path, while intergranular crack propagation starts to predominate at temperatures above 550°C [1,74]. In real life, however, the fatigue cracks in Nickel-base superalloy components invariably grow after initiation unless severe environmental attack ahead of the crack tip forms large amounts of debris which blunts the crack tip. This has led to the development of short crack growth models that operate below the apparent threshold ΔK_{th} exhibited during standard linear elastic fracture mechanics testing.

2.4.2 Short Crack Region

Short crack growth, which occurs in the near threshold region, can exhibit characteristics as shown in Figure 2.15 [76,77]. Surface microcracks can grow below the long crack threshold, where the threshold is obtained by carrying out tests on long through cracks using linear elastic fracture mechanics specimens [10,38,76-80]. Short surface cracks tend to exhibit a range of behaviour including a considerable amount of scatter in the crack growth rate data [10,77,78,80-82]. As shown in Figure 2.15, some cracks can arrest (A) after growing to a depth of an average surface grain [77]. The growth of some cracks can form a cusp (B-D), where there is an initial

reduction in the rate of growth until a certain depth is reached and beyond this the growth rate increases rapidly [77]. Finally, some cracks may never show a decrease in the growth rate (E) [77,80]. Eventually, all cracks, except those that have been arrested, will follow the long crack curve at some particular short crack length [77]. Since most short crack studies have been conducted at a stress level just below the yield stress, the limiting short crack length can also be expressed as a limiting ΔK value [77]. The validity of ΔK as a description of the crack tip conditions under high applied elastic stress is questionable [77]. However, it is desirable to push the stress intensity concept as far as possible, since it currently provides the most convenient parameter for describing crack growth [70,77].

The crack initiation site can have a significant effect on the growth rate. For cracks that initiate at slip bands, the orientation of the grains as well as the density of the grain boundaries in the vicinity of the crack will all have an influence on its growth [78]. Cracks which initiate at a subsurface micropore will experience local stress concentrations, and there may be segregation and/or compositional variations around the pores which can affect their growth [78]. Inclusions and other non-metallic particles may create local residual stress fields which can either increase or reduce the impact of applied stress. Therefore, in a given material, short cracks that initiate at particles can grow at very different rates than cracks formed in slip bands [78]. Multiple crack initiation can also reduce the rate of growth due to the formation of irregular crack fronts and erratic crack initiation, as shown in Figures 2.16 (a) and (b), respectively [55]. The size of the test specimen can also affect the results. For example, if the grain size is coarse and the specimen relatively small, a particular specimen may not contain a group of grains in the *weakest* orientation, and this will suppress growth rates [78]. Similarly, when comparing results obtained from a notched specimen to those from a smooth specimen, the volume of material which experiences the maximum stress will be different. High stress levels throughout the thickness of the test specimen can lead to sub-surface initiation [83]. In this case a large crack can suddenly appear at the surface, giving very high apparent growth rates, along with an unpredictable crack shape [83].

For a variety of alloys, the experimental crack growth rate data for short and long cracks converge at a crack depth of approximately 5D to 10D, where D is the grain size [10,77]. Once the crack grows beyond about 5 to 10 grains, growth is constrained by an averaging effect that renders short cracks equivalent to long cracks which encompass many grains [77,84]. As shown in experimental and theoretical studies of long cracks, the monotonic plastic zone at a fatigue crack tip is approximately given by [77]:

$$r_p = 0.15 \left(\frac{K_{\max}}{\sigma_{ys}} \right)^2 \quad (2.8a)$$

where σ_{ys} is the yield strength, r_p is the plastic zone size and $K_{\max}$ is the maximum cyclic stress intensity for a halfpenny-shaped ($a/c = 1$) surface crack of length $2c$, given by:

$$K_{\max} = 1.3\sigma_{\max}\sqrt{a} \quad (2.8b)$$

where $\sigma_{\max}$ is the maximum cyclic applied stress.

At convergence, it has been suggested when $a \approx 5D$ to $10D$ and setting $a = 5D$, the plastic zone size ahead of a short crack can be determined from [77]:

$$r_p = 0.15 \left[1.3 \left(\frac{\sigma_{\max}}{\sigma_y} \right) \sqrt{5D} \right]^2 \quad (2.9)$$

In most short crack studies to date $\sigma_{\max}/\sigma_{ys} \approx 0.8$ [77], and substituting this value into equation 2.9, yields $r_p \approx 0.81D \approx D$. Equations 2.8 and 2.9 suggest that short cracks grow at different rates relative to long cracks when the plastic zone ahead of the crack tip is smaller than the grain size [77,82,84].

Standard linear elastic fracture mechanics used to determine the stress intensity factor for a long crack in a notched specimen assumes that the crack length is the sum of the notch depth and the precrack length. Below this precrack length, or as the crack depth approaches zero, the notch depth becomes much greater than

the crack depth and the stress intensity factor calculated through the standard ASTM method approaches a constant value. In contrast, for short cracks, the plastic strain is localized in the plastic zone immediately ahead of the notch root and it can be assumed that these local stress effects at the notch root can be approximated by an edge cracked specimen as shown in Figure 2.17 [38,79,85]. Therefore, the short crack growth effect makes it necessary to evaluate the stress intensity factor by a different equation than that used in the ASTM method for notched specimens because the crack geometry is different. Two separate equations have been employed in the literature to determine ΔK for short and long cracks, and the concept of a transition point between the short and long crack regimes has also been defined [38,43,79,85]. One equation is used to describe the short cracks below the transition point while the other is used above this point. For example, Wang and Lu [79] developed a technique for determining ΔK for short cracks assuming a through crack in a notched specimen using the following approximation:

$$K = \begin{cases} K_1 & \text{if } K_1 < K_2 \\ K_2 & \text{if } K_2 < K_1 \end{cases} \quad (2.10)$$

The short crack stress intensity factor, K_1 was given by Wang and Lu [79] as:

$$K_1 = 1.12\sigma_y\sqrt{\pi a} \quad (2.10a)$$

where σ_y is the notch root stress, while the long crack stress intensity factor, K_2 was calculated using:

$$K_2 = Y \left(\frac{a + r}{W} \right) \sigma_{nom} \sqrt{\pi(a + r)} \quad (2.10b)$$

where σ_{nom} is the remote applied stress, a is the crack depth, r is the notch radius and Y is the linear elastic fracture mechanics geometric factor.

Swain and Newman [85] assume that a semi-elliptical surface crack is located at the centre of a semi-circular edge notch, as shown in Figure 2.18, which leads to the following equation for the short crack stress intensity factor:

$$K_1 = \sigma_y \sqrt{\pi \frac{c}{Q}} F_{sn} \quad (2.10c)$$

where Q is a shape factor and F_{sn} is the short crack geometric factor which is a function of c , a and t as shown in Figure 2.18.

Both techniques are covered in more detail in Appendix A.

2.4.3 Paris Region

Region II (Figure 2.14) or the *Paris* region, is the midrange of the crack growth rates and can be described by linear elastic fracture mechanics using the power law or Paris relationship:

$$\frac{da}{dN} = C \Delta K^n \quad (2.11)$$

where da/dN is the FCGR, ΔK is the stress intensity factor range ($\Delta K = K_{max} - K_{min}$) and C and n are a function of material properties, environment, frequency, temperature and stress ratio. Significant scatter can occur in FCGR tests [4,49,86], ASTM 647 [58] reports that a scatter of a factor of two in da/dN versus ΔK results is common. When determining the life limits of a component, a number of authors have used conservative C and n values using a three conditional standard deviation upper bound limit of the Paris region of the FCGR curve in life calculations [4,11,34,49]. The three conditional standard deviation upper bound line was calculated using the following [11,34,87]:

$$Upper\ Bound_{99\%} = Mean + 3 \times S_{y/x} \quad (2.12)$$

where,

$$S_{yy} = \Sigma y^2 - \frac{(\Sigma y)^2}{k} \quad (2.12b)$$

$$S_{y/x} = \left[\frac{S_{yy} - \frac{S_{xy}^2}{S_{xx}}}{k - 2} \right]^{1/2} \quad (2.12a)$$

$$S_{xx} = \Sigma x^2 - \frac{(\Sigma x)^2}{k} \quad (2.12c)$$

$$S_{xy} = \Sigma xy - \frac{(\Sigma x \Sigma y)}{k} \quad (2.12d)$$

The fracture surface within the Paris region is often flat showing abundant striations and the plastic zone encompasses many grains with several slip systems active within each grain [1]. A mixed mode fracture, comprising intergranular and transgranular features, occurs in this region in Inconel 718 at elevated temperatures [11,20,34]. Elevated temperature fatigue crack propagation (FCP) properties of Inconel 718 from the literature are summarized in Table 2.7.

The FCGRs of nickel-base alloys in the Paris region are affected by a number of different factors including temperature, frequency, stress ratio and microstructure. In Inconel 718 FCGR's increase with increasing temperature as shown in Figure 2.19 [88], which is attributed to the decreasing tensile strength and modulus of elasticity with increasing temperature. Since the lowest aging temperature used for most of the heat treatments of Inconel 718 is 621°C, a test which is carried out in excess of this temperature will overage the γ' and γ'' causing changes in the material properties. In particular, frequency effects on FCGR's are most pronounced at elevated temperatures with FCGR increasing with decreasing frequency, as shown in Figure 2.20 [83,88]. The increase in FCGR is more pronounced at frequencies less than 0.1 Hz, with only a small change occurring between 0.1 to 5 Hz, as shown in Figure 2.20. In elevated temperature FCGR tests carried out at frequencies below 0.1 Hz, creep effects contribute to crack propagation which causes an increase in the fatigue crack extension per cycle. For a given temperature, the FCGRs increase as

the stress ratio, R , is increased [88]. A number of empirical relationships have been proposed to describe the stress ratio effects on FCGR behaviour and one of the simplest and most widely used relationship was proposed by Walker [88] which is given by:

$$\Delta K_{eff} = \Delta K_{max} (1 - R)^m \quad (2.13)$$

where ΔK_{eff} is the adjusted stress intensity factor and m for Inconel 718 is given by:

$$m = 0.7251 - 3.112 \times 10^{-4} T \quad (2.13a)$$

where T is the temperature in °C. In Inconel 718, less than five percent adjustment of ΔK would be required on the basis of equation 2.13 at a temperature of 650 °C and a load ratio of 0.1. A five percent adjustment is within the scatter of FCGR testing and therefore, an adjustment of ΔK is not considered necessary under these test conditions.

The structure of individual grains becomes less significant as compared to the overall microstructure in the Paris region since it lies beyond the short crack region [46] (greater than 10 grain sizes), see section 2.4.2. Inconel 718 FCGRs are also reduced at elevated temperatures as the grain size increases as shown in Figure 2.21 [88,89]. Figure 2.21 also shows a reduction in FCGR due to the presence of a *necklace* structure in forgings, which is a microstructure containing large grains that are surrounded by smaller ones [88-90]. The larger grains in a necklace structure reduce FCGR as just mentioned, while the smaller grains further reduce FCGRs due to crack tip branching [89]. It is also believed that the zig-zag or *serrated* grain boundaries formed with the help of grain boundary Ni_3Nb needles reduce FCGR at elevated temperatures by suppressing grain boundary sliding [4,36], see section 2.1.2.

2.4.4 Final Failure Region

In the final failure region III (Figure 2.14), the crack growth rate accelerates as the maximum stress intensity factor (K_{max}) approaches the critical stress intensity

factor (K_I) or fracture toughness (K_{IC}) for Mode I crack growth. Ductile fracture dominates in region III and occurs due to the reduction in the net section which significantly increases the stress. Extensive yielding including the formation of large shear lips and a highly *dimpled* fracture surface occur during the last few cycles. Mills et al. [91] report fracture toughness values for Inconel 718 to be in the range of 82 to 117 MPa $\sqrt{m}$ at 538°C, while Wilkinson [29] reports values of the order of 82 to 165 MPa $\sqrt{m}$ for 25 to 650°C.

2.4.5 Fatigue Crack Propagation Testing using The Direct Current Potential Drop Technique for Crack Lengths Monitoring

The DEN specimen is not commonly used for FCP testing, but it has been extensively used for initiation testing where initiation is monitored with the help of a DCPD technique, see section 2.3.1 [60-62,76,85]. It is however possible to carry out FCGR testing on these specimens by continuing the tests beyond the initiation stage. In this way, the crack length could be continuously monitored using the DCPD technique. An optical calibration of the DCPD technique, in which the change in voltage is plotted against the optical crack length, must be carried out in order to extend the calibration curve to include crack lengths beyond the crack initiation stage. The precision of the DCPD technique can be calculated using three standard deviations from the mean, equation 2.5, similar to the initiation tests (section 2.3.1).

Once a crack has initiated at one of the notches there will be a significant increase in stress concentration due to the presence of a sharp crack and a bending load will be present due to the crack opening displacement. Therefore, once a crack has initiated at one notch it has been found that this crack alone will grow to failure without a crack initiating at the other notch [60-62]. Since a crack initiates at only one notch, to calculate ΔK after initiation has occurred it has been assumed that the equations for a single edge notch specimen can be used by adjusting the width of the specimen without including the other notch, Figure 2.22. The equation for calculating ΔK for a single edge notch specimen in tension is [38]:

$$\Delta K = \frac{\Delta P \sqrt{a}}{B W_s} Y_{SEN} (a/W) \quad (2.14)$$

where W_s is the adjusted specimen width, see Figure 2.22, and $Y_{SEN}(a/W)$ is the linear elastic fracture mechanics geometric factor for the single edge notch specimen:

$$Y_{SEN} (\alpha) = 1.99 - 0.41\alpha + 18.7\alpha^2 - 38.48\alpha^3 + 53.85\alpha^4 \quad (2.14a)$$

where $\alpha = a/W$. FCGRs can be determined from the crack length versus number of cycles using the incremental polynomial method, given in ASTM E647-83 Appendix XI [58].

The compact tension specimen is the most widely used specimen for FCGR testing and it can provide data over the entire Paris region. Under certain conditions the compact tension specimen can also provide limited data in the threshold region, just prior to the Paris region, and it can also provide limited data just above the Paris region which can be used to approximate K_{IC} [4]. ASTM standard E647-83 [58] provides details for FCGR testing of compact tension specimens including specimen geometry, apparatus, procedure, parameters and the calculation of results. The drawing of a standard specimen geometry is provided in ASTM E647-83 and the only other requirement is to ensure that *the specimen be predominantly elastic at all values of applied load* [58]. The minimum specimen width (W) requirement is calculated using the following relationship [58]:

$$W - a = \frac{4}{\pi \left(\frac{K_{max}}{\sigma_{ys}} \right)^2} \quad (2.15)$$

where W is the specimen width, a is the crack length, σ_{ys} is the 0.2% offset yield strength and K_{max} is the maximum stress intensity factor calculated using:

$$K_{max} = \frac{P_{max}}{B \sqrt{W}} \frac{(2 + a/W)}{(1 - a/W)^{3/2}} Y_{CT} (a/W) \quad (2.15a)$$

where $P_{\max}$ is the maximum load, B is the specimen thickness, and $Y_{CT}(a/W)$ is the linear elastic fracture mechanics geometric factor for the compact tension specimen, given by:

$$Y_{CT}(\alpha) = 0.886 + 4.64\alpha - 13.32\alpha^2 + 14.72\alpha^3 - 5.6\alpha^4 \quad (2.15b)$$

where $\alpha = a/W$.

Upon substituting ΔK for $K_{\max}$ and ΔP given in equation 2.15c for $P_{\max}$ in equation 2.15a, the stress intensity factor range can be determined.

$$\Delta P = P_{\max} (1 - R) \quad (2.15c)$$

The DCPD technique has been successfully used in a number of FCGR tests to monitor the growth of a crack in compact tension specimens [11,34,49,50]. The voltage probes and current leads placement for the compact tension specimen have been considered by a number of authors [11,34,49,52,56,57,92,93]. A uniform current profile can be obtained when the current leads are positioned on the top and bottom surfaces of the specimen in line with the load line [11,49,52]. The sensitivity of the technique increases when the voltage probes are located close to the crack front and for reproducibility, these are commonly positioned on the notch face of the specimen [11,49,52]. The precision of the DCPD technique is calculated using three standard deviations from the mean (equation 2.5) which is similar to the initiation tests (section 2.3.1). The DCPD technique is covered in more detail in section 2.3.1 and references [11,34,49,52,56,57,92,93].

2.5 Creep Properties

The response of a material under constant tensile loading at an elevated temperature is given by creep strain versus time curve as shown in Figure 2.23 [38]. It can be seen from this figure that the creep curve can be divided into three stages, (i) primary, (ii) secondary or steady-state and (iii) tertiary. In the primary stage, after the initial instantaneous strain (ϵ_0) resulting from the application of the

constant load, materials often undergo a transient period where $d\epsilon/dt$ decreases with time to a steady-state value. The steady-state value may continue for a significant portion of the test and is referred to as the secondary stage. During the tertiary stage, the creep rate increases until final failure occurs at the rupture life (t_R). The creep response of a material varies to reflect a continually changing interaction between the strain hardening and softening (recovery) processes which strongly affect the overall strain rate of the material at a given temperature and stress. The decrease in strain rate in the primary stage may be related to substructure changes that increase resistance to dislocation movement [38]. The secondary stage represents a stable structure where the hardening and softening processes are balanced. The tertiary stage on the other hand, is dominated by microstructural changes and deformation mechanisms such as localized necking, grain boundary sliding, grain elongation and microvoid formation [38]. Elevated temperature creep properties of Inconel 718 obtained from the literature are given in Table 2.8.

There are two major problems associated with creep tests: the significant amount of scatter in the results and the long test durations [37,38,94-97]. Koul et al. [38] stated *that to establish a scatter band approximately 100 data points per stress level are required to ensure that 95% of the population falls above about minus two standard deviation limit with 95% probability.* Due to the significant scatter associated with creep test data, the statistical character of the data must be considered. In order to avoid long test periods, techniques have been developed to extrapolate short term test data to predict the long term characteristics with the Larson-Miller parameter being the most widely used [38]. The Larson-Miller parameter has been shown to give reliable predictions so long as no microstructural changes occur in the alloy during prolonged exposure in a high temperature environment [96]. The Larson-Miller is a temperature compensated rupture life function where the constant, C_{LM} , has a unique value for a given material which depends only on the applied stress [94], where:

$$\log \sigma_a = T \times 10^{-3} (C_{LM} + \log t_R) \quad (2.16)$$

where σ_a is the applied stress, T is the test temperature (°K), t_R is the rupture life (hours). The constant C_{LM} is generally assumed to have a universal value of 20 [38,94], however Barker [18] found a better fit using a value for C_{LM} of 25 for Inconel 718.

Chapter 3

Experimental Material and Methods

3.1 Material

The Inconel 718 material used in this study was obtained from two sources: (i) stock material purchased from Universal-Cyclops Corporation; and (ii) turbine disc material provided by Hawker Siddeley Canada Incorporated, Orenda division.

The stock material was obtained in 9.52 mm (0.375 in) diameter wrought bar, that had been solution treated for one hour at 954°C and water quenched. The chemical composition of the stock material as quoted by Universal-Cyclops is presented in Table 3.1.

Four retired J85-CAN15 second stage turbine discs, conforming to General Electric specified chemical composition given in Table 3.1 and time since new (TSN) values given in Table 3.2, were provided by Orenda. In addition to the retired discs, a total of five tensile and creep specimens, previously machined from a new J85-CAN15 first stage disc, were also made available for testing.

3.1.1 Heat Treatments

The two heat treatment schedules used in this study are given in Table 2.2. An equal number of specimens prepared from stock material were treated with each of the two heat treatments. All J85-CAN15 turbine discs had been given the conventional heat treatment, Table 2.2, during the manufacturing process. The four retired discs were cut in halves and one half of each disc was further heat treated to compare the microstructure and mechanical properties in service exposed and reheat treated conditions. Half of the 2000 TSN disc was reheat treated using the conventional heat treatment while the halves of the remaining three discs (1650, 2090 and 2300 TSN) received the damage tolerant heat treatment.

3.1.2 Hardness Testing

A radial slice from all discs was removed for hardness testing as shown in Figure 3.1(a). Rockwell C hardness tests were performed at 6 mm intervals from the bore to the rim using a Clark US 12 hardness tester and industrial diamond indenter.

3.1.3 Microstructural Characterization

The radial slices cut from the disc halves were further split lengthwise as shown in Figure 3.1(b) and then one of the two pieces from each section was again cut into four smaller pieces, see Figure 3.1(c). These pieces were prepared using standard metallographic techniques to carry out optical microscopy, including linear intercept grain size measurements in accordance with ASTM standard E112-84 [98] with the aid of two optical microscopes (Olympus M061-PGM3 and Leitz Wetzlar MM6). Details of the etchant used for optical metallography are provided in Appendix B.

The remaining sections of each disc were machined into 3 mm diameter rods for thin foil preparation for transmission electron microscopy. The lengthwise rods allowed foils to be prepared at five evenly spaced locations in the disc from the bore to the rim. A Philips EM200 Transmission Electron Microscope equipped with a PW 6554/01 Double Tilt Holder, a EM200 Goniometer and a PW 6540/00 Dark Field Unit operated at 100 kV was used for all thin foil examinations. Details of thin foil preparation are provided in Appendix B.

3.2 Tensile Tests

3.2.1 Tensile Specimen

The use of material from retired turbine discs for testing created special specimen machining constraints due to the complex geometry of these discs. To

make the most efficient use of the disc material, a small diameter specimen was chosen since the shoulders were only fifty percent larger in diameter than the gauge length. The tensile specimen dimensions and their machining location in the disc are shown in Figure 3.2 and Figure 3.3, respectively. Six stock specimens were machined to determine the tensile properties of materials subjected to conventional and damage tolerant heat treatments (three per treatment). The stock material specimens were used initially to calibrate the test system. Four specimens each were machined from the 2000 and 2300 TSN discs, two from the service exposed halves and two from the rejuvenated halves in order to compare the tensile properties of the service exposed discs with the rejuvenated discs.

3.2.2 Tensile Test Apparatus and Parameters

The tensile tests were carried out in accordance with the ASTM E21-79 procedures for elevated temperature tension tests of metallic materials [40]. A computer controlled Material Test System (MTS) servo hydraulic test frame, shown in Figure 3.4(a), was used for all tensile tests.

The elevated temperature tensile tests were carried out at 650°C, using an Applied Test System (ATS) three zone furnace and controller. The test temperature was maintained within the required ASTM specifications of $\pm 1^\circ\text{C}$ by using two thermocouples, one located in air and the other positioned on the specimen surface shielded by a thin nickel foil.

The tensile load was monitored by a computer through an MTS load cell, with a load capacity of 25 metric tonnes. It should be noted that a small manual preload, less than 1 percent of the yield, was required to eliminate slack in the threaded connections. The specimen elongation was measured using an ATS extensometer and linear variable displacement transducer (LVDT). In small specimens, ASTM allows the displacement to be measured at the specimen grips [40]. To accomplish this, an adaptor system was required to attach the ATS extensometer to the specimen grips. A photograph of the extensometer assembly is shown in Figure 3.4(b), and the

drawings of the ring clamp and attachment plates are presented in Figure 3.5(a) and (b), respectively.

The computer controller allowed the loading rate to be decreased during the test to approximate a constant strain rate of $1 \times 10^{-4} \text{ s}^{-1}$. The reduction in the specimen cross section was not monitored, therefore this test was not a true stress-strain test. The computer controller also allowed collection of ten data points per second and automatic calculation of stress, strain, yield stress, ultimate tensile stress and the elastic modulus.

3.3 Fatigue Crack Initiation Testing

The fatigue crack initiation tests were carried out using two test specimens, smooth cylindrical and DEN, requiring different heating systems and definitions of initiation, while all other parameters remained the same. Tests were carried out at 650°C under load control using a triangular waveform and a stress ratio (R) of 0.1 at a frequency of 0.1 Hz unless otherwise noted. The test temperature and waveform were selected to simulate J85-CAN15 engine operating conditions, see sections 2.2 and 2.3 respectively.

3.3.1 Smooth Cylindrical Specimen

The smooth cylindrical specimen dimensions were chosen such that it could be machined from retired discs while satisfying the ASTM standard E606 requirements [47]. The gauge length was the same as that used for the tensile specimens with some modifications made to use existing equipment as shown in Figure 3.6. ASTM standard requires that no circumferential scratches be visible at $\times 25$ [47]. This was achieved in two steps, first by specifying very light cuts during the machining process and then by polishing the specimens in the longitudinal direction with 400 and 600 grit abrasive paper. Twelve smooth specimens were machined from

the bar stock, six conventionally heat treated material and six subjected to the damage tolerant heat treatment.

3.3.2 Smooth Cylindrical Test Apparatus and Parameters

The smooth cylindrical tests were carried out on an MTS servo hydraulic testing system, with a load cell capacity of 10 metric tonnes, using stock material for comparison with the DEN initiation tests that will be discussed later, Figure 3.7(a). Three different maximum initial stress levels were used: 740, 950 and 1100 MPa which corresponded to load levels of 9.72, 12.5 and 14.4 kN for the smooth cylindrical specimen cross section. These stress levels were selected using the damage tolerant heat treatment tensile data for strain values ranging from 0.005 to 0.01 in order to include a full range of stress values from just below σ_{ys} to just below UTS. The stresses were selected from the damage tolerant material data because its UTS was smaller than for the conventionally heat treated material thus avoiding tensile failure. Due to extremely long test periods, the test frequency for the 740 MPa tests was increased to 5 Hertz. Since there was no crack monitoring technique available for the existing test apparatus, the criteria for crack initiation was specimen failure.

The test temperature of 650°C was achieved and maintained by an induction heater coil, shown in Figure 3.7(b), powered by a Lepel high frequency generator and controlled using an IRCON closed loop controller equipped with an infrared radiation pyrometer. The radiation pyrometer was calibrated against thermocouple readings and the thermocouple was held against the specimen surface just above the pyrometer location. The test was paused to avoid damaging the specimen surface as shown in Figure 3.7(c). The emissivity setting was then adjusted to match the controller temperature with the temperature measured by the thermocouple. By repeating the calibration procedure at regular intervals the change in emissivity due to material oxidation was taken into account and the temperature was maintained within ten degrees of the required test temperature throughout the test duration.

3.3.3 Double Edge Notch Initiation Specimen

The standard geometry specified for a DEN initiation specimen could not be used for the disc material due to dimensional limitations. Therefore, it was used as the basis for designing the specimen for this study. The controlling geometric constraint for machining a specimen from the disc was the minimum thickness of the disc. A scaling factor of 0.70 was used to reduce the specimen thickness to within the minimum disc thickness. A drawing of the final specimen configuration is presented in Figure 3.8. The specimens were positioned radially such that the necessary material for the threaded ends would be provided by the bore and bolt hole regions as shown in Figure 3.9. The notches were machined using two different methods due to equipment availability and expense. The twenty specimens machined from the stock material had the notches cut by a lathe, ten from conventionally heat treated material and ten from damage tolerant heat treated material. The notches cut by the lathe were examined with a travelling microscope and found not to be as deep as those specified in Figure 3.8. A fatigue stress concentration factor (K_t) of 1.31 was calculated using equations 2.6 to 2.8 at the measured notch depth of 0.6 mm.

Fourteen DEN specimens were machined from the 2090 hour TSN service exposed disc, out of which 7 specimens were subjected to the damage tolerant reheat treatment. In the disc specimens, the notch was machined using a computer controlled wire electron discharge machine, which created an accurate notch geometry. The disc specimen notches were examined using a travelling microscope and found to be within the specification given in Figure 3.8, and therefore a K_t of 1.93 was obtained from Figure 2.11.

3.3.4 Double Edge Notch Initiation Test Apparatus and Parameters

A computer controlled MTS servo hydraulic system was used for testing DEN specimens, shown in Figure 3.10(a). The crack initiation was monitored using a

computer controlled, three channel, DCPD technique developed at IAR. The stock specimens were tested at maximum initial stress levels of 775, 1000 and 1060 MPa which corresponded to load levels of 13.5, 17.5 and 18.5 kN for the stock specimen notch geometry. The disc specimens were tested at maximum initial stress levels of 740, 950 and 1100 MPa which corresponded to load levels of 8.65, 11.14 and 12.9 kN for the disc specimen notch geometry. The stress levels were selected from tensile data for the damage tolerant heat treatment following methodology similar to that used for the smooth specimens, see section 3.3.2. Due to extremely long test periods, the test frequency was increased to 2 Hz for specimens tested at 740 MPa.

Tests were carried out at 650°C and the specimens were heated with an ATS three zone furnace. A thermocouple could not be mounted on the specimen surface during an actual test because electrical leakage would have affected the DCPD results. Therefore, the furnace was calibrated prior to carrying out any tests by positioning five thermocouples on a specimen surface covering the entire gauge length of the specimen. The furnace settings were adjusted until all thermocouples read within 650°C $\pm$ 5°C, and the thermocouple at the notch was within 650°C $\pm$ 1°C over a period of 24 hours.

The DCPD technique previously developed at IAR [49] had only been used for monitoring crack growth. Therefore, the DCPD set up required modifications prior to using the technique for monitoring fatigue crack initiation. Four modifications were made: 1) increasing the number of potential drop channels, 2) improving the load bearing capacity of the load electrical insulation, 3) minimizing the scatter in voltage readings and 4) improving the detection limit of the DCPD technique.

The DCPD technique, used previously, was developed to monitor crack lengths in compact tension specimens which required only two channels, one for the specimen and one for the reference block [49]. The DEN specimen, on the other hand, has two notches which must be monitored concurrently, as well as a reference block. Therefore an additional monitoring channel was added to the set up. This

additional channel required a third bridge amplifier and some modification of the computer program which operated the test.

The electrical insulation from the load frame, previously used for insulating the MTS frame from the load train was unable to withstand the larger loads required for crack initiation tests. Therefore, the material used for the insulating washers was changed from a thin plastic to higher strength linen phenolic. These washers were then mounted between two flat metallic surfaces that always kept them in compression, as shown in Figure 3.12. The linen phenolic tube, used to insulate the loading rod, was lengthened to insulate the lower stainless steel washer as well as the grip from the loading rod.

The DCPD technique used previously at IAR was found to have a scatter band of ± 3 percent, and voltage spikes in the system had been associated with electromagnetic interference [49]. Modifications were required to eliminate the voltage spikes and to reduce the scatter in the voltage to one percent or less [99,100]. The difficulties were assumed to be related to grounding problems because of increased error observed when other large test equipment was in operation. To reduce ground loops, which may have caused the scatter, a common ground to the test console was used for all related equipment including the computer and interface.

The DCPD technique was previously calibrated using optical tracking of the crack tip which can only be used accurately for cracks longer than approximately 200 μm , when the crack first becomes visible at the surface. In order to calibrate the voltage versus crack length for the DEN crack initiation tests, stock specimens were cycled to a specific percent change in voltage, stopped and then stress ruptured. The specimens were then examined using a scanning electron microscope, where, the distinct differences in fracture surface between the fatigue and stress rupture portions of the crack length were discerned. The crack length was then plotted against the specific change in voltage which provided a single calibration point.

The DEN specimen testing required modifications to the DCPD technique with respect to current input and voltage probe location. Due to the large contact surface at the threaded connections, the current leads were connected to the loading

rods. Heavy gauge copper wires connected the direct current power supply to current rings on the upper and lower loading rods, as shown in Figures 3.10(a) and 3.12. The voltage probes were located on the opposite corners of both notches, as shown in Figure 3.10(b) which allowed a larger area of the notch to be monitored for crack initiation. The reference leads were located on opposite edges of the lowest point of the reduced section of the test specimen as suggested in the AGARD round robin test programme [60], and shown in Figure 3.10(b). The separation of the reference leads had to be at least as large as the specimen leads, including the curvature of the notch, otherwise the signal to noise ratio would have been less than that for the specimen leads which would have caused increased scatter in the results.

3.4 Fatigue Crack Propagation Testing

The FCP tests were carried out following the ASTM E647-83 standard for constant-load-amplitude FCP testing where applicable [58], and the DCPD technique was used to monitor the crack length. Two test specimens were used to obtain FCP data over a range of stress intensity factors, the DEN and the compact tension specimens. The DEN specimen tests covered stress intensity factors in both the threshold and Paris regions. The compact tension specimen tests were carried out at a lower stress level with data covering the entire Paris region including part of the fracture toughness region.

3.4.1 Double Edge Notch Fatigue Crack Propagation Test Specimen and Apparatus

The FCP tests carried out on the DEN specimens were a continuation of the initiation tests described in detail in sections 3.3.3 and 3.3.4. The only modification made to the existing test apparatus was to calibrate the DCPD technique for crack lengths above 200 μm by optical tracking of the crack tip. As in the initiation tests, the FCP tests were carried out under load control using a triangular waveform at a frequency of 0.1 Hz, a load ratio of 0.1 and a test temperature of 650°C. Since the

double edge notch FCP tests were a continuation of the fatigue crack initiation tests, the stress levels used were the same as described in section 3.3.4.

3.4.2 Compact Tension Specimen

In order to maintain consistency with the previous tests carried out at IAR, the same ASTM specimen geometry as that tested by Pishva et al [49] was used, as shown in Figure 3.13. Compact tension specimens were machined from the 2000 and 2300 hour TSN service exposed discs, with the notch tip located near the bolt holes of the disc, Figure 3.3. Three specimens were machined from each disc half, the two service exposed halves and the two reheat treated halves corresponding to a total of twelve specimens.

3.4.3 Compact Tension Test Apparatus and Parameters

The compact tension tests were carried out using a computer controlled MTS servo hydraulic testing system, shown in Figure 3.14. The test equipment was the same as that used by Pishva et al, the details of which are given in reference [11,34,49]. The load controlled tests were carried out at a maximum initial stress of 52 MPa, which corresponds to a load of 5 kN for the compact tension geometry, using a triangular waveform at a frequency of 1 Hz, a load ratio of 0.1 and a test temperature of 650°C. The stress level was selected to provide the largest ΔK range for the Paris region using the ASTM minimum size requirement, equation 2.15. The triangular waveform and test temperature of 650°C were chosen to simulate actual engine operating conditions, see sections 2.3 and 2.2 respectively. A calibration test was initially carried out to verify the DCPD calibration equation used by Pishva et al [49].

3.5 Creep Testing

3.5.1 Creep Specimen

The ASTM standard creep specimen refers to the same specimen geometry as that used for the tensile tests [101]. A detailed description of these specimens was given in section 3.2.1 and shown in Figures 3.2 and 3.3. Six specimens were machined to examine the creep properties of two service exposed discs, 2000 and 2300 TSN, as well as the effects of the two rejuvenation reheat treatments.

3.5.2 Creep Test Apparatus

The ASTM E139-83 standard procedure for creep testing of metallic materials was followed [101] using two constant load creep tests frames, shown in Figure 3.15, and an initial stress of 620 MPa (90 ksi) at 650°C. The initial stress was selected in order to approximate a 500 hour creep-rupture test using data from reference [42], Figure 3.16. A 500 hour test was selected as the longest practical test that could be performed considering the number of specimens to be tested, availability of equipment and time constraint.

An ATS three zone furnace and controller was used to heat the specimens. Temperature was monitored by two thermocouples, one located in air and the other positioned on the specimen surface protected by a thin nickel foil. The temperature was maintained to within the required ASTM variation of $\pm 2^\circ\text{C}$.

The temperature, time and specimen elongation were recorded every hour by a Datalogger printer. The specimen elongation was measured, according to the ASTM standard for small size specimens, with the same system used for the tensile tests which was described in detail in section 3.2.2.

Chapter 4

Results

4.1 Material Response to Heat Treatments

4.1.1 Hardness Profiles

The hardness of the Inconel 718 stock material after conventional and damage tolerant heat treatments, was found to be 42 ± 2 Rc. The hardness of the service exposed discs was found to have an average value of 46 ± 2 Rc, while the reheat treated discs had an average hardness value of 45 ± 2 Rc. A profile of the hardness from the bore to the rim for all discs, both service exposed and reheat treated, is shown Figure 4.1. Hardness values for a first stage new disc from reference [11], are included in Figure 4.1 for comparison.

4.1.2 Optical Microstructures

Optical micrographs showing the microstructure of the stock material subjected to conventional and damage tolerant heat treatments, are presented in Figure 4.2. The average grain size for the stock material was found to be $8.1 \mu\text{m}$ for the conventional heat treatment and $51 \mu\text{m}$ for the damage tolerant heat treatment.

Optical micrographs showing the microstructural profiles of a service exposed disc (2300 hour TSN), a conventionally reheat treated disc half and a damage tolerant reheat treated disc half are presented in Figure 4.3. The average grain size profile from the bore to the rim of the discs is shown in Figure 4.4. The average grain size of the service exposed and conventionally reheat treated discs, range between 10 and $20 \mu\text{m}$, while the average grain sizes of the damage tolerant reheat treated discs range between 45 and $65 \mu\text{m}$. Grain size measurements previously carried out on a new first stage disc [11] are also presented in Figure 4.4.

4.1.3 Transmission Electron Microscopy Structures

Thin foils were prepared from five different radial regions of the disc, (1) near the bore, (2) from the thinnest section of the disc, (3) from the location of the compact tension specimen notch tip, (4) near the bolt hole, and (5) from the rim just below the dovetail slots. Microstructural characteristics and precipitate size measurements were determined for each region.

The microstructural characteristics found using transmission electron microscopy for the service exposed, conventional and damage tolerant reheat treated discs are summarized in Figure 4.5. Damage accumulation was present in all service exposed discs examined. Planar dislocations were found emanating from the grain boundaries and from both carbides and Ni_3Nb precipitates as shown in Figure 4.5(a), 4.5(b) and 4.5(c). Dislocation structures were found in the form of pileups and entanglements, as shown in Figure 4.5(d) and 4.5(e), respectively. Low angle grain boundaries were also found in some regions of the service exposed discs. Transmission electron microscopy of the conventionally reheat treated disc (2000 hour TSN) showed similar damage to the service exposed disc but with a more homogeneous distribution and a reduced density of dislocations. The damage tolerant reheat treated disc (2300 hour TSN) showed changes similar to those found optically, increased grain size and longer and more profuse Ni_3Nb precipitates at the grain boundaries, as shown in Figure 4.5(f). The damage tolerant reheat treatment also produced a more homogeneous dislocation distribution similar to the conventional reheat treatment but with a greater reduction in the overall dislocation density, Figure 4.5(g). Both reheat treatments were found to have reduced the number of pileups and entanglements of planar dislocations that were found in the service exposed discs.

Dark field transmission electron microscopy technique, was used to image the precipitates in order to determine their size, as shown in the dark field micrographs presented in Figure 4.6. This figure provides an example of the [001] diffraction pattern, a dark field image of γ' and γ'' , (010) reflection, and γ'' , $(1\frac{1}{2}0)$ reflection,

and the corresponding bright field image. The distribution of the γ' and γ'' precipitates was homogeneous for both the service exposed and reheat treated discs, as shown in Figure 4.5 and 4.6, respectively. The sizes of the γ' and γ'' strengthening precipitates before and after the conventional and the damage tolerant reheat treatments are presented in Table 4.1. The γ' and γ'' precipitate sizes were essentially similar in the service exposed and reheat treated discs.

4.2 Tensile Properties

4.2.1 Precision of Tensile Test Apparatus

The results of the tensile tests are affected by the precision of the measured load and displacement values. The load cell had a specified linearity of ± 0.15 percent of full scale. This corresponded to an error of ± 0.037 kN for the tensile tests. The linear variable displacement transducer (LVDT) had a specified linearity of ± 0.1 percent or ± 0.006 mm, but the calibration was conducted within ± 0.01 mm. The computer had a maximum error of ± 3 digital units or 0.01 volts which corresponded to an error of less than one percent.

4.2.2 Tensile Test Results

The results of the tensile tests are presented in Table 4.2 which includes: yield strength (0.2% offset), strain at yield, modulus of elasticity (slope of offset), ultimate tensile strength, reduction of area, percent elongation and failure strain. The tensile curves for each condition are shown in Figure 4.7 to 4.10, where in each figure, (a) shows the region prior to yield and (b) represents the entire tensile curve.

The results obtained from the tensile tests of the stock material, for both heat treatments, are given in Table 4.2 and shown in Figure 4.7. These results are within the range of the literature values given in Table 2.4. The scatter in the yield strength and UTS was limited to approximately one percent for most of the tensile tests. The

scatter in the Young's modulus and failure strain values, approximately ten percent, arises from the extensometers and the calibration of the LVDT.

The tensile test data generated from the J85-CAN15 turbine discs are similar to those from the stock material, with the service exposed and conventionally reheat treated disc yield strength, Young's modulus and UTS higher than for the damage tolerant reheat treated specimens, Figures 4.8 to 4.10 and Table 4.2.

The tensile test results obtained from the first stage new disc are shown along with the results from other discs in Figures 4.8 to 4.10. The new disc yield strength, Young's modulus and UTS values are as much as ten percent lower than the conventional reheat treated discs but similar to the damage tolerant disc (Table 4.4).

4.3 Fatigue Crack Initiation Properties

4.3.1 Precision of Smooth Cylindrical Test Apparatus

The only data obtained from the smooth cylindrical tests was the cycles to initiation and therefore the sources of error were limited to the errors caused by the load cell and function generator. The specified linearity of the load cell and function generator was ± 0.1 percent, which corresponds to an error of 0.02 kN and 0.01 Hz, respectively.

4.3.2 Smooth Cylindrical Test Results

The number of cycles to initiation (failure) for the smooth cylindrical tests are presented in Table 4.3 and plotted in Figure 4.11 and 4.12. Scanning electron micrographs of the fracture surfaces from both the conventional and damage tolerant heat treated specimens at 740, 950 and 1100 MPa are presented in Figure 4.13, 4.14 and 4.15, respectively.

The fatigue crack initiation data generated from both conventional and damage tolerant stock material smooth specimens, are shown in Figures 4.11 and 4.12

as dotted lines and are also given in Table 4.3. The results given in Table 4.3 were not directly comparable to the values found in the literature, Table 2.5, since the literature values were obtained from rotating bending and fully reversed axial fatigue tests. It can be seen from Figure 4.11 that the life to initiation increased significantly as the stress level was decreased for all specimens.

At the lowest stress level, 740 MPa, the initiation life of the conventionally heat treated material was an order of magnitude higher than the damage tolerant material, with one of the conventional tests considered to have run out at ten million cycles. Micrographs of the smooth specimen fracture surfaces tested at the low stress level for both the conventional and the damage tolerant heat treatments are shown in Figure 4.13. A single internal crack initiation site, as shown in Figure 4.13(a) and 4.13(b), was found in all specimens tested at this stress level which is not unusual for high cycle fatigue testing [38,83]. The cracks appeared to have propagated through the specimen, Figure 4.13(a) and 4.13(b), until they reached the surface at which time the specimen failed. The surface of the damage tolerant material showed extensive cleavage fracture and striations which are associated with transgranular cracking, Figure 4.13(b) and 4.13(d) respectively, while the surface of the conventionally heat treated material showed secondary cracking associated with intergranular crack growth, Figure 4.13(c).

At the medium stress level, 950 MPa, the difference between the conventional and damage tolerant material initiation lives was less than for the lower stress as shown in Figure 4.11. This difference in the number of cycles to initiation between the damage tolerant and conventional materials ranged from ten percent to a factor of three, Table 4.3. Micrographs of the fracture surfaces for the medium stress level tests are shown in Figure 4.14. Both heat treatments showed signs of multiple initiation sites where the cracks initiated at the surface and propagated into the specimen, see Figures 4.14(a) and 4.14(b). However, the surface of the conventional material showed separate multiple initiation sites and propagation regions, while the damage tolerant material surface had multiple initiation sites which coalesced into a single propagating crack. Although both fracture surfaces showed signs of

intergranular and transgranular cracking, the dominant fracture features were different. The damage tolerant material exhibited transgranular crack growth, showing regions of cleaved grains and striations, Figure 4.14(b) and 4.14(d), while the conventional material showed intergranular fracture with abundant secondary cracking, Figures 4.14(a) and 4.14(c).

At the higher stress level, 1100 MPa, for the conventional and damage tolerant material initiation lives were within the scatter as shown in Figure 4.11. Micrographs of the fracture surfaces for the smooth specimens tested at the high stress level are shown in Figure 4.15. Similar to observations for the medium stress level the cracks initiated at the surface and propagated into the material in both cases.

4.3.3 Precision of Double Edge Notch Initiation Test Apparatus

The results obtained from the DEN tests were affected by the load cell error and the DCPD calibration technique. The load cell had a specified linearity of ± 0.15 percent of full scale, which corresponded to an error of ± 0.037 kN. The computer had a maximum error of ± 3 digital units or 0.01 volts which corresponds to an error of less than one percent on the load waveform and the data acquisition.

Four stock material specimens were used to calibrate the DCPD technique and one disc material specimen was used to verify the validity of the techniques for different notch geometries. The long term, high temperature stability of the DCPD technique was monitored during the first calibration test over a 125 hour test duration, Figure 4.16. This figure shows a sample of voltage readings for both notches of the DEN specimen and includes the maximum and minimum measurements from the 2468 data acquisitions. The notch voltage was within +0.5 percent and -0.6 percent of the baseline voltage over the duration of the test with the majority of data being within approximately ± 0.3 percent. The maximum and minimum values obtained during this calibration test are shown in Figure 4.16 as dotted lines, and these values lie within ± 0.75 percent of the chosen value of 400

digital units. A definite increase in voltage beyond scatter occurred when the one percent initiation criterion, shown by the solid line in Figure 4.16, was crossed. The stock calibration tests were stopped and stress ruptured at four different voltage increases ranging from 1 to 1.75 percent and the fatigue cracks were measured using a scanning electron microscope, Table 4.4. Micrographs of the crack fronts, notch areas and initiation regions of the calibration specimens are shown in Figure 4.17. The initiation region of the crack was then identified from the fracture surface, using scanning electron microscopy, by the change in the mode of failure that occurred between the fatigue and stress rupture regions, Figures 4.17(a) and 4.17(b). The resulting sensitivity of the calibration was an equivalent straight fronted initiation crack of $50 \pm 20 \mu\text{m}$ ($50 \pm 3\sigma$) at a voltage increase of one percent. For thumbnail shaped initiation cracks, the actual crack depth was found to be as high as $180 \mu\text{m}$ with the crack front approximating half the specimen thickness.

4.3.4 Double Edge Notch Crack Initiation Results

The fatigue crack initiation results obtained from the stock DEN specimens, using conventional and damage tolerant materials, are shown along with the smooth specimen results in Figure 4.12. There was significant scatter in the double edged notch results which increased with decreasing stress level, a trend similar to that observed in smooth specimens [9,38,102]. The conventional and damage tolerant heat treated material results were similar, within scatter, for all three stress levels, Figure 4.12. The initiation life of the conventionally heat treated DEN specimens at the low stress level was two orders of magnitude lower than the conventionally heat treated smooth specimens and an order of magnitude lower at the medium stress level. However, the initiation life at the high stress level was within the scatter of the smooth specimen results.

The fatigue crack initiation results for the service exposed (2000 hour TSN) and damage tolerant reheat treated disc DEN specimens are shown with the smooth specimen results in Figure 4.11. This figure shows that the conventionally treated

smooth specimen trends compared well with the DEN service exposed disc material, which was conventionally heat treated prior to service.

Scanning electron micrographs of the initiation regions for each stress level from the service exposed discs and the damage tolerant reheat treated disc are shown in Figure 4.19. The cracks initiated with a generally larger crack front curvature in the disc specimens than in the stock material DEN specimens. Similar to the stock specimens shown in Figure 4.17, crack initiation was observed to be relatively straight across the notch face with only small out of plane transitions. Limited secondary cracking on the notch face was found for the disc specimens relative to the stock material specimens. It was observed that most DEN specimens formed limited secondary cracks, with no apparent secondary cracks found for some specimens.

4.4 Fatigue Crack Propagation Results

4.4.1 Precision of Double Edge Notch Fatigue Crack Propagation Test Apparatus

The results of the double edge notch FCP tests are affected by the load cell error and the calibration of the DCPD technique. The 250 kN load cell has a specified linearity of ± 0.15 percent of full scale, which corresponds to an error of ± 0.037 kN for the DEN tests. The computer has a maximum error of ± 3 digital units or 0.01 V which corresponds to an error of less than one percent on the load waveform and the data acquisition.

Three stock specimens were used to optically calibrate the DCPD technique for the FCP tests, two at room temperature and one at high temperature to confirm the validity of using the room temperature calibration at high temperatures. The crack length was measured by optical tracking of the crack tip with a travelling microscope and was plotted against the nondimensional β parameter in Figure 4.21. The five data points used to calibrate the technique for initiation testing are also included to plot the calibration curve, Figure 4.21. The calibration equation was

formulated by fitting a second order polynomial through the calibration data, including the zero crack length point, and this equation is given by,

$$a = -0.44276 + 0.53831\beta - 0.09555\beta^2 \quad (4.1)$$

The precision of the DCPD technique and therefore the calibration equation 4.1 was calculated to be $20 \mu\text{m}$ ($\pm 3\sigma$) using equation 2.5 and the data shown in Figure 4.16.

4.4.2 Double Edge Notch Fatigue Crack Propagation Results

Fatigue crack propagation curves, da/dN versus ΔK using standard LEFM for ΔK calculations, for the stock specimens, subjected to conventional and damage tolerant heat treatments, are presented in Figures 4.22(a), 4.22(b) and 4.22(c) for the three different stress levels of 775, 1000 and 1060 MPa, respectively. The DEN stock specimens appear to show a threshold for both conventional and damage tolerant heat treatments, as shown in Figure 4.22. The data shows a rapid increase in crack growth rate for the threshold region which levels off when the crack enters an apparent Paris region. Scanning electron micrographs of the notch and fracture surfaces for the stock specimens are presented in Figures 4.23. This figure shows both intergranular and transgranular crack growth on the fracture surfaces. The transgranular crack growth appears to dominate in the initial regions of the fracture surfaces, especially for the damage tolerant heat treated materials.

The FCP results for the 2090 TSN disc material, before and after the damage tolerant reheat treatment, determined using standard linear elastic fracture mechanics relationships for calculating ΔK for a single edge notch specimen, equation 2.14, are presented in Figures 4.24 to 4.25. The disc results for both service exposed and damage tolerant reheat treated DEN specimens are shown in Figure 4.24 for each stress level and the data are combined together in Figure 4.25. The scatter appears to have increased with increasing stress level, with limited scatter at 740 MPa, Figure 4.24(a), and more significant scatter at 1100 MPa, Figure 4.24(c). A

larger data base was created by linking the initiation data to the propagation data which created a single data set that included initiation to failure. This increase in data obtained for the disc specimens, especially below the crack growth rate of 10^{-7} m/cycle, made it possible to estimate ΔK_{th} from the data presented in Figure 4.24 for each of the three different stress levels using standard LEFM for ΔK calculations. Threshold stress intensity factor ranges of 20, 28 and 33 MPa $\sqrt{m}$ were found for the three stress levels of 740, 950 and 1100 MPa, respectively, Figure 4.24. The Paris region slope (n), intercept (C) and correlation coefficients for both the mean and the three standard deviation (+3 σ) upper bound lines determined from these results are given in Table 4.6.

The FCP results for the 2090 hour TSN disc material are presented again in Figures 4.26 and 4.27 where ΔK was determined on the basis of the Swain and Newman model [85]. The model was applied to crack lengths below the transition ΔK which was specified by the model as the point where ΔK is equal to the standard linear elastic fracture mechanics ΔK . The transitional ΔK for both heat treatments, for a through crack, was determined to be 35.2, 45.4 and 52.5 MPa $\sqrt{m}$ for the three stress levels of 740, 950 and 1100 MPa, respectively. These transitional ΔK levels correspond to a crack depth of approximately 280 μm for a straight through crack. The short crack data obtained at all three stress levels were plotted together in Figure 4.27. The slope, intercept and correlation coefficients from the short crack results for both the mean and upper bound lines are given in Table 4.7. Scanning electron micrographs showing the fracture surfaces of the DEN disc specimens are presented in Figure 4.28. Intergranular and transgranular crack growth was observed for both service exposed and damage tolerant reheat treated materials. Secondary cracks appear just following initiation, with intergranular cracking dominating in the service exposed specimens. Transgranular cracking dominates the damage tolerant fracture surface until just prior to failure where severe secondary cracking was found.

4.4.3 Precision of Compact Tension Test Apparatus

The results from the compact tension FCP tests are affected by the load cell error and the calibration of the DCPD technique. The 15 kN load cell had a specified linearity of ± 0.10 percent of full scale, which corresponds to an error of ± 0.007 kN for the compact tension tests. The computer had a maximum error of ± 3 digital units or 0.01 V which corresponds to an error of less than one percent on the load waveform and data acquisition.

A calibration test was carried out using one specimen in order to verify the calibration equation determined in a previous study [11,49]:

$$a/W = -0.47150 + 0.78505\beta - 0.11415\beta^2 \quad (4.2)$$

The data obtained from the verification of the calibration curve developed by Pishva et al. [49] is shown in Figure 4.29 along with a second order polynomial curve fit (Andrews) and Pishva's equation (equation 4.2). This figure shows that the two equations were similar. However, the equations did separate marginally as the nondimensional crack length increased. Pishva's [49] equation was verified since the calibration data followed this equation closer than the polynomial curve fit, Figure 4.29. The precision of the DCPD technique for the compact tension specimen was reported by Pishva to be $69 \mu\text{m}$ ($\pm 3\sigma$) [6,11,49].

4.4.4 Compact Tension Fatigue Crack Propagation Results

The compact tension FCGR data for service exposed disc specimens before and after the conventional and damage tolerant reheat treatments are shown in Figures 4.30(a) and 4.30(b), respectively. Despite the presence of scatter, the data showed clear trends for each material condition. Three standard deviation ($+3\sigma$) upper bound lines, shown in Figure 4.31, were used to represent the trends of the Paris region. The two service exposed discs had equivalent FCGR's which were similar to the values obtained from literature, given in Table 2.7. Figure 4.31(a)

showed an increase in slope and therefore FCGR's for the service exposed discs with respect to the new disc. Previous crack propagation results from a new first stage disc [11] are included in Figure 4.31 for comparison. The slope, intercept and correlation coefficients for the Paris regime for all the compact tension tests are given in Table 4.6. Optical micrographs of the service exposed material before and after the conventional heat treatment are shown in Figure 4.32. This figure shows an increase in the length of the Ni_3Nb precipitates parallel to the grain boundary which results from the conventional reheat treatment.

Scanning electron micrographs of the fracture surfaces for some of the compact tension specimens tested are presented in Figure 4.33. Conventionally reheat treated specimens revealed a predominantly intergranular fracture including extensive secondary cracking whereas the damage tolerant reheat treated specimens showed clear evidence of fatigue striations along with a mixed mode fracture, Figures 4.33(a) and 4.33(b).

4.5 Creep Tests

4.5.1 Precision of Creep Test Apparatus

The results of the creep tests are affected by the precision of the displacement measurement. The LVDT has a specified linearity of ± 0.1 percent or ± 0.006 mm and it was calibrated to within ± 0.01 mm. The load adjustment was limited to one half pound intervals, which corresponded to an error of ± 0.3 percent on the load or initial stress.

4.5.2 Creep Test Results

Creep curves (strain versus time) for the 2000 TSN disc specimen in service exposed and conventionally reheat treated condition are shown in Figure 4.34(a) and Figure 4.34(b) whereas the creep curves for the 2300 TSN disc specimen in service

exposed and damage tolerant reheated condition are shown in Figure 4.34(c) and 4.34(d). The new disc data, for a first stage disc in Figure 4.34(e), are included for comparison. Stress rupture lives of specimens from the retired discs, before and after reheated treatments (conventional and damage tolerant) and from a first stage new disc are compared in Figure 4.35. The creep curves for the 2000 and 2300 hour TSN service exposed discs showed limited scatter and the creep curves depicted similar shapes between specimens and between the two discs. The rupture life of the first stage new disc was significantly lower than the service exposed disc, as shown in Figure 4.34(e) and 4.35. Similarly, the data from the conventionally reheated disc showed a reduction in rupture life with respect to the service exposed discs, Figure 4.34(b) and 4.35. In contrast, the damage tolerant reheated improved the rupture life with respect to the service exposed discs as shown in Figure 4.34(d) and 4.35. The service exposed discs revealed the largest rupture strain of approximately 10 to 12 percent, while the conventional reheated was only marginally lower at approximately 9 percent. The damage tolerant reheated and first stage new disc rupture strains were significantly lower at approximately 6 and 4 percent respectively.

A Larson Miller plot for the tests carried out in this study and for unpublished results from Pishva [11] is presented in Figure 4.36. Larson Miller parameters were determined using the generally assumed universal value of 20 for constant C_{LM} [38,94] and using the constant value of 25 as suggested by Barker [18] and these results are shown in Figure 4.36(a) and 4.36(b), respectively. The creep test results, including minimum creep rate, stress rupture life, primary and secondary life, tertiary life, rupture strain, primary and secondary strain and tertiary strain are presented in Table 4.8.

Chapter 5

Discussion

5.1 Inconel 718 Response to Heat Treatment and Reheat Treatments

The surface hardness of Inconel 718 stock material increased from 24 to 42 Rc after both conventional and damage tolerant heat treatments. This increase in hardness occurs as a result of the double aging sequence (Section 2.1.2). Since the hardness is the same after both heat treatments, it suggests that the partial solution treatment of 926°C for 1 hour in the damage tolerant heat treatment dissolves any overaged γ' and γ'' that form during direct aging at 843°C. These overaged strengthening precipitates would have otherwise decreased the hardness of the damage tolerant stock material [19,28,38].

The hardness of the turbine discs before and after both reheat treatments was found to remain relatively unchanged with only limited scatter between discs, an observation similar to that reported by Pishva [11], Figure 4.1. The engine manufactures specification for the hardness of these turbine discs is 42 Rc, which is lower than the values found for the service exposed discs. Although the hardness values were close, Figure 4.1, the service exposed discs may have strain hardened during service due to cyclic hardening. The average hardness of the first stage new disc, 36 Rc, was less than the engine manufacturer's specification. The exact history of this disc is not known and therefore the cause of the lower hardness values is unclear. The lower average hardness value suggests that the disc may not have been fully aged.

The conventional heat treatment solution temperature of 955°C lies below the solvus temperatures of both NbC and Ni₃Nb precipitates, which restrict grain coarsening [4], Table 2.1. As a result grain growth did not occur during the heat treatment of the stock material, Figure 4.2(a). In contrast, the damage tolerant solution temperature of 1032°C lies above the Ni₃Nb solvus temperature which

causes some grain coarsening, Figure 4.2(b). The direct aging step in the damage tolerant heat treatment produces a large number of plate like Ni_3Nb needles at the grain boundaries, which produce a serrated grain boundary structure, Figure 4.2(b). Laves phase was absent in all reheat treated specimens indicating that the heat treatment procedure was accurately controlled [29].

A *necklace* microstructure was present in both the bore and rim regions of the 2300 hour TSN disc, while the central region showed a uniform grain size distribution $\sim 15\mu\text{m}$. The disc rim region showed a marginally larger grain size with abundant twinning, Figures 4.3(a) and 4.4. Some variation in the grain size from the bore to the rim of this disc is considered to have occurred during the disc forging, Figure 4.4. The conventional reheat treatment did not significantly increase the grain size, Figure 4.3(b), whereas the damage tolerant reheat treatment resulted in a uniform microstructure with a larger average grain size of $42\mu\text{m}$ from the bore to the rim, Figure 4.3(c) and 4.4. This increase in grain size also formed abundant annealing twins in the damage tolerant reheat treated discs. The increase in grain size along with the precipitation of grain boundary Ni_3Nb needles and the formation of serrated grain boundaries was similar to that observed for the stock material.

Transmission electron microscopy revealed the presence of planar dislocations arrays, dislocation pileups, large entanglements of dislocations and low angle grain boundaries in the service exposed discs, Figure 4.5. These results are similar to those reported by Pishva [11]. The presence of planar dislocation pileups and entanglements has been found to degrade material properties [1,11,38]. It is felt that the planar dislocations shown in Figure 4.5(b) formed as a result of service exposure rather than during the initial forging process because the γ' and γ'' precipitate sizes were not different in the vicinity of these arrays. Had these dislocations arrays been present after the forging and solution treatment of the discs, the γ' or γ'' precipitates would be expected to be larger along these arrays relative to the rest of the matrix because of increased dislocation core diffusion during the double aging treatment. The conventional reheat treatment appears to remove the dislocation pileups and entanglements, Figure 4.5(i). These observations further suggest that the

dislocation arrays and pileups formed due to cyclic straining during service. The reheat treated discs reveal a homogeneous dislocation distribution with a much reduced dislocation density, especially in the damage tolerant reheat treated discs, Figure 4.5(i). The γ' and γ'' precipitates remained relatively unchanged by the reheat treatments with only a marginal increase in the size range for the damage tolerant reheat treatment, Table 4.1 [65].

5.2 Tensile Properties

Relative to the damage tolerant heat treatment, the conventional heat treatment resulted in approximately a ten percent increase in yield strength, Young's modulus and UTS values in the stock material Table 4.2 [2]. The loss of strength in the damage tolerant heat treated material is a result of the increased quantity and size of the Ni_3Nb precipitates which formed at the grain boundaries.

Both service exposed discs had similar microstructures which resulted in similar tensile properties, Figure 4.9 and Table 4.2. The conventional and damage tolerant reheat treatments, resulted in a marginal increase in ductility in the service exposed discs. Since this improvement in ductility occurred for both reheat treatments, the annealing out of dislocation substructures at elevated temperatures may have induced this effect.

The lower strength of the new disc relative to the stock material or service exposed discs in conventionally heat treated condition is consistent with its lower hardness values which were attributed to under aging of the disc during the manufacturing process. The tensile results further confirm this hypothesis. The necklace grain structure of the new disc is however believed to be responsible for increasing the failure strain by as much as fifty percent relative to the stock material or service exposed discs [38]. The necklace structure results in a torturous fracture path which enhances disc material ductility.

5.3 Fatigue Crack Initiation

5.3.1 Smooth Specimen Fatigue Crack Initiation

The stress controlled fatigue tests conducted in the present study reveal a similar trend to that found by Au et al. [33] where Inconel 718 lives after the conventional heat treatment were longer than in the damage tolerant condition in strain controlled tests, Figures 4.11 and 4.12. A longer initiation life of the conventionally heat treated specimens can be attributed to their smaller grain size [33,38,103]. This increase in initiation life is however more pronounced at the lower stress levels. Figures 4.11 and 4.12 also show a large amount of scatter, a feature inherent in fatigue crack initiation testing [9,38,102]. Fatigue cracks can initiate from defects, inhomogeneities in the material or microstructure and from stress concentrations including those formed by slip [38]. Although the surface of the smooth specimens were polished until they were scratch free up to a magnification of 25 times, defects could still have been present in some specimens which would have resulted in a reduction in initiation life. Inhomogeneities in the material and/or microstructure could also have resulted in a reduction in initiation life, especially at the lower stress levels where microstructure plays a more significant role [38]. These effects can change the initiation life from specimen to specimen by as much as an order of magnitude [33,38]. Therefore, the scatter of a factor of two or three observed in Figure 4.11 and 4.12 is not unreasonable.

The lowest test stress level of 740 MPa corresponds to less than 80 percent of the yield strength for both materials and given the fact that the initiation lives were in excess of 100,000 cycles, the test condition resulted in high cycle fatigue failure. The cleavage fracture found for the damage tolerant surface may have been due to failure along the annealing twin boundaries which increase in number with increasing grain size. The medium stress level was lower than the yield strength of the conventionally heat treated material but approximately equal to the damage tolerant material yield strength. The crack initiation in the damage tolerant

specimens occurred under high strain low cycle fatigue conditions since the number of cycles to initiation was below 20,000. The conventional specimens on the other hand, were subjected to medium strain low cycle fatigue conditions because the initiation lives were lower than 50,000 cycles. At the highest stress level, the damage tolerant material initiation life was an order of magnitude lower than the conventional material. This can be explained by the fact that the stress level used was equal to the UTS of the damage tolerant material and therefore, the ductility exhaustion occurred sooner during cyclic strain accumulation. In contrast, the high stress level was marginally above the yield strength for the conventionally heat treated material and the initiation lives were longer and the ductility exhaustion occurred after a larger number of fatigue cycles. As can be seen from Figure 4.15, both materials revealed similar fracture features showing multiple crack initiation sites, including evidence of transgranular fracture and large shear lips along the maximum shear stress directions.

5.3.2 Errors in The Direct Current Potential Drop Technique for Fatigue Crack Initiation

The DCPD technique measures changes in area rather than crack length or depth, equation 2.4, and the changes in voltage given in Table 4.4 are therefore representative of this change in area [49,52,56,57,92,93,104]. However, the DCPD technique was calibrated for a change in crack length. Therefore, to relate crack area to crack length, an equivalent straight fronted crack depth was calculated from the measured crack area, Table 4.4. The equivalent crack depth was used to estimate the crack length except when a thumbnail crack was present with a large crack front curvature which resulted in an error of a factor of 3 or more as shown in Table 4.4. The theoretical minimum detectable crack area and length for a one percent increase in voltage, which can be resolved by the DCPD technique, were estimated to be 0.0812 mm^2 and $34 \text{ } \mu\text{m}$, respectively, the details of which are given in Appendix C. The calibration of the DCPD technique resulted in a crack length

of 34.6 μm for a one percent increase in voltage as shown in Table 4.4. The scanning electron microscope measurements revealed the crack length at one percent increase in voltage to be, 50 μm , which is 15 μm longer than the theoretical value, Table 4.4. However, the theoretical value of crack length is the minimum detectable crack length for the equipment used and therefore the 15 μm error in crack length is considered acceptable because it lies within $\pm 3\sigma$ error [100].

Secondary cracks in the notch region can also increase the resistance across the notch face [52,56,92,93,104] and the DCPD based crack length could then be greater than the actual crack length. Secondary cracks were in fact found in the notch region of the calibration specimens, Figure 4.17(d). The secondary cracking occurred as a result of the increased stress concentration due to machining marks on the notch face. These cracks were however small relative to the notch area, a maximum of 100 μm at the surface which corresponds to a crack front depth of less than 20 μm for the geometry of the cracks found in these specimens. This depth is within the given scatter of the DCPD technique and therefore these secondary cracks only had a small effect on the DCPD measurements. Out of plane crack surfaces can also increase the DCPD measured crack length, but the crack fronts shown in Figure 4.17(c) were relatively straight with only limited out of plane initiation sites which coalesced to form the crack front and thus did not affect the accuracy of the technique.

The DCPD estimated crack length for conventionally machined specimens compares well with the equivalent crack depth calculated from the measured crack area, with less than 10 μm difference in all five calibration tests. The DCPD calculated crack depth also correlates well with the actual crack depth, deviating by a maximum of 30 μm for all calibration tests. In the electron discharge machined disc specimen, which initiated as a thumbnail crack, Figures 4.17, the area of the thumbnail crack was measured accurately by the DCPD technique as indicated by the 2 μm difference between the DCPD crack length and the measured equivalent crack depth. Since these crack initiated as a thumbnail crack the equivalent crack depth was not an accurate measurement of the actual crack depth, Table 4.4.

5.3.3 Double Edge Notch Specimen Fatigue Crack Initiation

The reduction in stock material initiation life relative to the smooth specimen results at the low and medium stress levels may have been caused by material notch sensitivity or defective machining of the notch, Figure 4.12 [26]. Inconel 718 has been found to become notch sensitive as a result of the increase in the size, amount and distribution of Ni_3Nb precipitates [26]. Despite significant differences in the Ni_3Nb precipitate distribution between the conventional and damage tolerant microstructures of the DEN stock material specimens, Figure 4.2, their initiation lives were similar. Therefore, the reduction in initiation life relative to the smooth specimens was not caused by the Ni_3Nb precipitates. The lathe machined notches were rounded in the specimen cross section and machining marks were present on the notch face in the stock material. The rounding of the notch was small, only a maximum loss in width of 0.15 mm, and since the specimen was four times larger in width than its thickness, this rounding did not appear to affect the initiation behaviour, Figure 4.17(a). On the other hand, cracks had apparently initiated at the machining marks which span the notch face, Figure 4.17(c). It is suggested that these machining marks acted as stress raisers which significantly increased the local notch root stress and resulted in a decrease in the number of cycles to initiation relative to the smooth specimens for both heat treatments. All failed specimens were examined in a scanning electron microscope revealing similar crack initiation features, Figure 4.17. The crack fronts were straight in all cases with only a small amount of out of plane initiation visible. Secondary cracks in the notch area were found in all specimens, but secondary cracking was not extensive. These observations further confirm that stress concentrations caused by the machining marks influenced crack initiation forming straight crack fronts with marginal curvature through the specimen thickness.

The smooth specimen has been extensively used for initiation testing [33,46,47,103]. The fact that the turbine disc DEN specimen LCF initiation lives were generally similar to the smooth specimen lives for the conventionally heat

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treated discs suggests that the notch geometry is not overshadowing the material behaviour. The corollary to this is that the notch specimen results support the assumption that the smooth specimen can be used to simulate the initiation of a crack at the notch root of a defect for materials which are not notch sensitive. However, the initiation life of the disc material notch specimens was as much as 50 percent higher than the smooth specimens at the high stress level. This increase in life occurs as a result of the stress level approaching the material UTS which perhaps accentuates localized strain effects in the notch area. There is a direct relationship between initiation life and UTS at the high stress level, with the stock material and damage tolerant disc material lives being shorter than the service exposed disc material due to their lower UTS values.

The notches for the disc material specimens were electron discharge machined and then finished with 600 grit paper to remove any debris from the notch surface. However, electron discharge machining leaves a thin cast surface layer which is often brittle. This brittle layer was only partially removed by the 600 grit paper and can reduce the crack initiation life particularly under high cycle fatigue conditions where surface effects dominate. The notched service exposed disc specimen lives were lower than the smooth specimen values at the low stress levels which could have been caused by the cast layer or by the accumulation of damage during service. However, it is unclear from this study which effect was dominant. The notched damage tolerant disc material and stock material initiation lives were much shorter than the damage tolerant smooth specimen lives at low and medium stresses, Figures 4.11 and 4.12. However, unlike the notched damage tolerant stock material, there were no machining marks on the notch face of the disc specimen to justify the shorter initiation lives. Since the damage tolerant microstructure produced large and profuse Ni_3Nb precipitate at the grain boundaries, this may have resulted in a notch sensitive material [26,65]. The results from both the stock (dotted lines) and disc (solid line) material notched specimens are shown in Figure 4.18. This figure shows that the damage tolerant results from the stock and disc material are similar and compare well with the conventional stock notched specimens. The crack front

reveals a curvature in the absence of machining marks in the electron discharge machined disc specimens. The damage tolerant specimens initiated at cleaved grains at the notch root, while the conventional specimens appear to have initiated at the surface. These observations suggest that a reduction in initiation life results from the machining marks for the stock conventional specimens, and from notch sensitivity due to large Ni_3Nb precipitates for both the stock and disc damage tolerant specimens.

The use of the DCPD technique to determine the life to initiation for the DEN specimens also provides for the measurement of the failure lives (initiation plus propagation). These failure lives for both the stock and disc material notch specimens are shown in Figure 4.20 and presented in Tables 4.3 and 4.5 along with the percentage of the life spent in initiation. The failure results followed the same trend as the initiation results with limited increase in magnitude. The percentage of total life spent in initiation ranged from a minimum of 80 percent for the high stress level to 99 percent for the low stress level. By using the notch root radius of 1 mm for the DEN specimens Figure 2.12, the percentage of the life spent in initiation could be estimated to be above 50 percent [38]. The high percentage of life spent in initiation was due to the small net section of the specimen. If the net section is increased, the crack must then propagate further to reach failure [105].

The results of the initiation tests appear to suggest that the damage tolerant heat treatment significantly reduced the initiation life of both stock and disc Inconel 718 materials compared to the service exposed and conventionally reheat treated turbine disc materials. However, when considering the application of the material for turbine discs, the number of low cycle fatigue cycles which a disc is expected to withstand is well below 100,000 cycles. If the data is examined in this region, above the medium stress level (950 MPa), the difference in results is greatly reduced especially for the stock material where it is less than a factor of two, while the damage tolerant reheat treatment still results in a maximum loss in initiation life of an order of magnitude for the disc specimens. The percentage of total life spent in initiation is not representative of the discs in service due to the small net section of the DEN specimen compared to the large section of the disc, Figure 3.3. The DEN

specimen in conjunction with the DCPD technique provides initiation data which is closer to actual service conditions because it considers the effect of a stress concentration arising from bolt holes and rim serrations.

5.4 Fatigue Crack Propagation

5.4.1 Double Edge Notch Fatigue Crack Propagation

To obtain fatigue crack propagation data in the threshold region for the DEN specimens the DCPD technique was calibrated using a travelling microscope and the results are shown in Figure 4.20. The scatter in the calibration data was significant but a clear trend was evident. The calibration is described by a second order polynomial with a precision of $20\text{ }\mu\text{m}$ ($\pm 3\sigma$), equation 4.1. As the crack length decreases, the scatter in the calibration data increases because of the difficulties associated with identifying the tip of small cracks when using the optical microscope. The combination of the calibration data obtained from the initiation tests and the data determined optically from the crack propagation tests provided a calibration equation which was used to determine the Threshold region. Upon using standard linear elastic fracture mechanics formulations for ΔK calculations for a DEN specimen, the crack growth was observed to be present in the Paris region only for a short period prior to specimen failure and the threshold stress intensity factor range, ΔK_{th} , was not clearly discernable from the stock material test data, Figure 4.22. The absence of a ΔK_{th} indicates that the minimum crack growth rates were very high, above those expected for the ΔK_{th} region. The conventional and damage tolerant stock material specimens revealed a similar trend in the Threshold region.

Intergranular cracking accompanied crack growth shortly after the crack initiation region in both conventional and damage tolerant microstructures of the stock material as indicated by the secondary cracking in Figure 4.23. This secondary cracking was more pronounced in the conventionally heat treated specimens and as the crack propagated, intergranular cracking became dominant and the amount of

transgranular cracking decreased. Intergranular crack growth tends to be faster than transgranular crack growth [4,20,38]. Therefore, if the damage tolerant microstructure results in more transgranular crack growth, the material crack growth rate would be expected to be low [4,20,83,102,106].

In contrast to stock material specimens, the service exposed and damage tolerant reheat treated disc specimens revealed a sharply defined ΔK_{th} at all test stresses, Figure 4.24, and ΔK_{th} increased with increasing initial stress, Figure 4.25. The increase in ΔK_{th} with increasing stress is a reflection of the direct relationship between ΔK and stress shown by equation 2.14, although the relevance or irrelevance of the ΔK_{th} will become more evident in the following text.

Upon using Swain and Newman's model [85], for determining ΔK for short cracks, which are defined as cracks below a transitional point where the transitional crack length is expected to be within a range of 5 to 10 grain diameters [10,77], section 2.4.2, the transitional crack length fell within this range (7 grains) in the case of the damage tolerant microstructure but it was above this range for the conventional microstructure (19 grains). This does not imply that Swain and Newman's model is not applicable to conventional microstructure because their model is only based on the geometry of an embedded thumbnail crack and the transitional crack length correlation with the material grain size is not a prerequisite for the application of this model.

The FCGR data, using both linear elastic fracture mechanics methods and Swain and Newman's model for ΔK calculation, are shown in Figure 4.26 with mean lines and three standard deviation upper bound lines included for each stress level. The ΔK values were considerably lower than those determined through conventional linear elastic fracture mechanics formulations for notched specimens with the apparent ΔK_{th} disappearing. Standard LEFM formulations for ΔK are applied assuming a finite precrack length and as the crack length becomes small they can no longer be used. Estimates of ΔK_{th} found by extending the FCGR curve have been found to underestimate the actual FCGR found for short cracks. Short crack growth testing has been found to result in considerable scatter in the FCGR data which can

sometimes overshadow the overall trend [10,77,78,81]. Although the Inconel 718 results did show considerable scatter in the present study, the largest scatter in Figure 4.26 was observed at very low ΔK values, for crack lengths below the 50 μm initiation criterion that were determined by extrapolating back the DCPD calibration curve. The mean and upper bound lines were determined using data above the minimum detectable crack length (50 μm). The ΔK values corresponding to 50 μm crack length were determined to be 6.24, 8.00 and 9.27 $\text{MPa}\sqrt{\text{m}}$ for the three initial stress levels of 740, 950 and 1100 MPa, respectively. Despite the presence of some scatter the mean and the upper bound lines through the Paris region appear to represent the trend in the short crack growth rate data for all DEN results. In addition, the C and n values determined from these lines compared well with the compact tension specimen Paris region results and to the values reported in the literature, Tables 2.7, 4.6 and 4.7. The similarity in these results suggests that it may be possible to extend the Paris region back towards initiation in order to predict the FCGR of Inconel 718 in the short crack regime.

The data presented in Figure 4.26 showed faster FCGR's than those predicted by the edge notch linear elastic fracture mechanics equation and this observation is in agreement with the results reported by other workers in the literature, Figure 2.15 [7,77,78,81]. Limited out of plane initiation and growth described in section 5.3.3 for the DEN specimens could have resulted in an increase in the short crack growth rate, Figure 2.16(a). The growth of short cracks can be affected by both crack closure and specific local changes in microstructure which may also cause a large scatter in FCGR's because small differences in grain structure and/or precipitate morphology can have a significant influence on crack extension [10,77,82,84]. In addition, the cracks in the DEN specimens appear to have initiated at or near the surface as a result of grain cleavage in the damage tolerant material and by either slip or surface inclusions in the conventional specimens. These different initiation sites can also have a significant influence on the short crack growth [78] because they force a certain crack propagation path and therefore may cause an increase in scatter. Multiple initiation sites on the other hand can reduce the short crack growth rate,

as shown in Figure 2.16(b), therefore the absence of secondary cracking on the notch face (multiple initiation), found in Section 5.3.3 for the disc specimens, could also have resulted in an increase in the short crack growth rate. Figure 4.27 shows an envelope of FCGR's in the short crack regime for the various stress levels which all converge on a similar Paris region with reduced scatter. The reduction in scatter can be attributed to the crack tip extension beyond a certain point where the influence of microstructural variations average out with increasing crack length [10,78]. The disc DEN specimens showed similar trends to those found for the stock DEN specimens.

5.4.2 Compact Tension Fatigue Crack Propagation

The higher FCGR's of the service exposed discs relative to new disc specimens may have been a result of the damage accumulation (section 5.1.3) [5,6,11]. The increase in FCGR's for the service exposed discs did not appear to be proportional to service hours or time since new values since the 2000 and 2300 hour TSN disc FCGR's were similar. Damage accumulation in turbine discs has been found to be proportional to the number of low cycle fatigue cycles rather than hours in service [4,6,11] and therefore, it is more likely that these discs had been exposed to similar cyclic fatigue histories.

The upper bound lines for the 2000 hour TSN disc showed that the conventional reheat treatment resulted in a marginal increase in the FCGR's relative to the service exposed disc, Figure 4.31(b). This increase is attributed to the coarsening of the Ni_3Nb precipitate which occurred when the disc was conventionally reheat treated, Figure 4.32. Unlike the damage tolerant reheat, which formed Ni_3Nb precipitates perpendicular to the grain boundaries, the Ni_3Nb precipitates in the conventionally reheat treated discs were parallel to the grain boundaries. These parallel Ni_3Nb precipitates enhance intergranular cracking which increase the FCGR. Scanning electron micrographs, Figure 4.33(a), showed abundant secondary cracking for the conventionally reheat treated material which is indicative of grain boundary

weakness. Although the crack growth was accompanied by a mixed mode fracture, intergranular cracking appeared to dominate for the conventionally reheat treated material.

The upper bounds lines for the 2300 hour TSN disc showed that the damage tolerant heat treatment reduced the FCGR's by 40 percent at a $\Delta K = 24 \text{ MPa}\sqrt{\text{m}}$ and 25 percent at a $\Delta K = 50 \text{ MPa}\sqrt{\text{m}}$, Figure 4.31(c). The formation of needle shaped Ni_3Nb precipitates gave rise to a serrated grain boundary structure in the damage tolerant reheat treated disc, which suppress grain boundary sliding and force the crack to follow a torturous path which reduce the FCGRs [4,6,19,20]. Intergranular cracking can also be inhibited by the presence of a coarse grain size which also suppresses grain boundary sliding at elevated temperatures [83,102,106]. Scanning electron micrographs, Figure 4.33(b), showed fatigue striations, and some secondary cracking for the damage tolerant disc material. Transgranular crack growth appeared to be dominant at lower ΔK values for the damage tolerant reheat treated material, reducing the FCGR. Intergranular cracking did occur as the crack length increased which resulted in an increase in FCGRs. The reduction in FCGR's in the damage tolerant reheat treated disc may also have occurred as a result of service induced damage removal as discussed in Section 5.1.3 [19,35].

5.4.3 Fatigue Crack Propagation in Inconel 718 Turbine Discs

The data from the short crack region was combined with the data from the Paris region for both DEN and the compact tension specimens, Figure 5.1. This figure represents the entire FCGR behaviour of a crack in Inconel 718 disc material where service exposed material behaviour is shown in Figure 5.1(a) and the influence of the damage tolerant reheat treatment is given in Figure 5.1(b). To summarize, once a crack *initiates*, the crack growth enters the short crack region where it is dependent on the initiation process and the specific microstructural features ahead of the crack tip [10,77,78]. Variations between and within specimens resulted in the large amount of scatter in this region. The short crack data then converges into the

Paris region of both DEN and compact tension specimen data, Figure 5.1. The DEN and compact tension specimen data show a straight Paris region with limited scatter as compared to the short crack region. It should be noted that different test frequencies were used to test the two specimen geometries. The DEN specimens were tested at 0.1 Hz while the compact tension specimens were tested at 1.0 Hz. It has been found that an order of magnitude reduction in frequency would result in only a small increase in the FCGR's in the Paris region for Inconel 718 at elevated temperature, Figure 2.20 [88,107]. This reduction in scatter is said to occur as a result of an averaging effect of the microstructure as the crack length increases [10,77,78,84]. The scatter may also be reduced due to increased stress resulting from the increased crack length since at higher loads the effect of small variations in microstructure is not very noticeable. As the fracture toughness of the material was approached both the DEN and compact tension specimens failed. The fracture toughness values estimated from the ΔK values at failure were found to be within the range of 90 to 120 MPa $\sqrt{m}$ which lie within the range of 82 to 165 MPa $\sqrt{m}$ quoted in literature, section 2.4.4.

Regression analysis of the combined compact tension and DEN specimen data was performed to determine a mean line and a three standard deviation upper bound line, Figure 5.1. This analysis was performed separately for the service exposed and the damage tolerant reheat treated specimen data, where the equations of these lines are,

Service Exposed Disc:

$$\begin{aligned} \text{Mean Line: } \frac{da}{dN} &= 0.205 \times 10^{-9} \Delta K^{2.43} \\ \text{Upper Bound Line: } \frac{da}{dN} &= 0.697 \times 10^{-9} \Delta K^{2.43} \end{aligned} \quad (5.1)$$

Damage Tolerant Reheat Treated Disc:

$$\begin{aligned} \text{Mean Line: } \frac{da}{dN} &= 0.195 \times 10^{-9} \Delta K^{2.40} \\ \text{Upper Bound Line: } \frac{da}{dN} &= 0.914 \times 10^{-9} \Delta K^{2.40} \end{aligned} \quad (5.2)$$

The C and n values determined from these lines compare well with the Paris region values for the compact tension specimen results of the present study and the values quoted in the literature, Table 4.6 and 2.9 respectively. Upon comparing equations 5.1 and 5.2, it appears that the damage tolerant reheat treatment does not influence the FCGR resistance of the service exposed disc. However, these equations were fitted through both short and long crack data and considering the magnitude of scatter in the short crack growth rate data the beneficial effects of the damage tolerant reheat treatment may have been masked by the scatter at lower ΔK values.

5.5 Creep

The stress rupture lives of the new disc specimen, Table 4.8, compared well with literature values at similar temperature and stress level, Table 2.8. The new disc specimens revealed the lowest t_R because of the presence of a necklace type grain structure which facilitates grain boundary sliding and leads to early specimen failure. The t_R and ϵ_R values of both service exposed discs were higher than the new disc because their grain size distribution was more uniform than the new disc. The conventional reheat treatment of the 2000 hour TSN service exposed disc led to a decrease in t_R and this reduction occurred due to coarsening of Ni_3Nb precipitates that were parallel to the grain boundaries and these Ni_3Nb precipitates do not offer resistance to grain boundary sliding. The 2300 hour TSN discs subjected to the damage tolerant reheat treatment revealed the lowest rupture life values because of the presence of a large grain size along with a serrated grain boundary structure which suppress grain boundary sliding and force transgranular deformation. The damage tolerant microstructure appears to have reduced the rupture strain, Figure

Damage Tolerant Reheat Treated Disc:

$$\begin{aligned} \text{Mean Line: } \frac{da}{dN} &= 0.195 \times 10^{-9} \Delta K^{2.40} \\ \text{Upper Bound Line: } \frac{da}{dN} &= 0.914 \times 10^{-9} \Delta K^{2.40} \end{aligned} \quad (5.2)$$

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4.34(d) and 4.34(e), but the reduction in ϵ_R primarily occurs towards the end of tertiary creep, suggesting that crack propagation towards the latter part of the creep test is relatively fast.

The data used to plot the Larson Miller parameter as a function of stress, includes the creep results generated in this study and unpublished data obtained by Pishva [11]. The data provided in Figure 4.36 could be used to estimate the rupture life at a lower stress without performing such long term tests. The damage tolerant and conventionally reheat treated specimens appear to have similar slopes, Figure 4.36 suggesting that the improvement in rupture life as a result of damage tolerant reheat treatment would occur at all stress levels. The service exposed data was only available at 620 MPa (90 ksi) and therefore no trend could be predicted.

2

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Chapter 6

Conclusions

6.1 The Effect of Heat Treatment on the Microstructure of Inconel 718

- The conventional heat treatment produced an average grain size of approximately 10 μm along with small grain boundary Ni_3Nb precipitates and carbides, and a homogeneous γ' and γ'' precipitate distribution. Since the service exposed turbine discs had been conventionally heat treated prior to service there were only marginal changes in the microstructure when the discs were reheat treated. These changes included an increase in the size of the Ni_3Nb precipitates which lay parallel to the grain boundaries, and a reduced density of dislocation pileups and entanglements that had formed during service.
- The damage tolerant heat treatment increased the grain size to approximately 50 μm , and produced large Ni_3Nb precipitates perpendicular to the grain boundaries which resulted in a serrated grain boundary morphology. The damage tolerant reheat treatment also produced a homogeneous dislocation structure with a significant reduction in the density of dislocation pileups and entanglements in the service exposed discs. The γ' and γ'' precipitate distributions were essentially similar to those found in the conventionally heat treated material.

6.2 Elevated Temperature Tensile Properties

- The precipitation of a larger volume fraction of Ni_3Nb precipitates in the damage tolerant heat treated stock material resulted in an approximately 10 percent decrease in the yield strength, modulus of elasticity and ultimate tensile strength than the conventionally heat treated material because a lower Nb content was available for γ'' precipitation.

- The conventional reheat treatment did not alter the tensile properties of the service exposed discs while the damage tolerant reheat treatment reduced all tensile properties with the exception of failure strain. Both reheat treatments increased the failure strain of the service exposed turbine disc material by reducing accumulated damage effects caused by service exposure.

6.3 Elevated Temperature Fatigue Crack Initiation

- The DCPD technique was stable and at one percent increase in voltage a crack was always found, the length of which in terms of an equivalent straight fronted crack was estimated to be $50 \pm 20 \mu\text{m}$. Using the one percent increase in voltage definition for crack initiation, the DCPD results were always reproducible.
- The technique used to machine the notches in the DEN specimens was found to significantly affect the crack initiation life. The lathe cut notches of the stock specimens significantly reduced the initiation life because the machining marks on the notch face acted as stress raisers, while the electron discharge machined notches in the disc specimens did not affect the initiation life and these results compared well with the smooth cylindrical specimens results.
- The similarity in the initiation lives of the serviced exposed DEN specimens and the conventionally machined smooth cylindrical specimens suggests that the DEN specimen initiation lives are valid. This similarity also suggests that the assumption that a smooth cylindrical specimen can be used to simulate the notch root of a stress concentration is valid for materials which are not notch sensitive.
- The initiation life of the damage tolerant heat treated material was lower than the conventionally heat treated specimens in the case of both smooth cylindrical and DEN specimens. This loss in initiation life was considerably larger for the DEN specimens because grain boundary Ni_3Nb precipitation during the damage tolerant heat treatment renders Inconel 718 notch sensitive. The notch sensitivity of the damage tolerant microstructure reduces with increasing stress or under medium to high strain low cycle fatigue conditions. Since the life of the turbine

discs is based on low cycle fatigue life, a minor loss in initiation life due to the damage tolerant heat treatment is considered insignificant.

6.4 Elevated Temperature Fatigue Crack Propagation

- The DCPD technique was calibrated for the DEN specimen to monitor crack lengths above 50 μm with a precision of 20 μm .
- The transition from short to long crack growth was determined to be 280 μm , using a model developed by Swain and Newman [85]. This model reduced the ΔK 's below the transition point by assuming an edge cracked plate at the root of the notch, which was found to provide a short crack growth region.
- The DEN specimen short crack growth results merged into a Paris region and the Paris region constant C and n compared well with the values quoted in the literature. These results suggest that the DEN specimen results are valid for crack growth rate monitoring. No threshold was observed when Swain and Newman's model was used for the ΔK calculations.
- The Paris region crack growth rate data obtained for the compact tension specimens machined from the service exposed discs showed that the conventional reheat treatment marginally increased FCGR's while the damage tolerant reheat treatment reduced FCGR's. The increase in FCGR's resulting from the conventional reheat treatment was due to an increase in the Ni_3Nb precipitates which lay parallel to the grain boundaries and these precipitates reduced the resistance to intergranular cracking. The serrated grain boundaries and a larger grain size in the damage tolerant reheat treated discs reduced FCGR's by suppressing grain boundary sliding and forcing the crack to follow a transgranular path.
- The combination of DEN and compact tension specimen results provided the crack growth rate data base for a wide range of ΔK values for the service exposed and damage tolerant reheat treated turbine disc material. The Paris region mean and three standard deviation upper bound lines were found to fit the entire crack

growth rate data. This suggests that for Inconel 718 turbine disc material the entire growth of a crack, once initiated, can be predicted using the Paris region equation. The equations for the three conditional standard deviation upper bound lines of the combined data were found to be:

Service Exposed

$$da/dN = 0.697 \times 10^{-9} \Delta K^{2.43} \quad (6.1)$$

Damage Tolerant

$$da/dN = 0.914 \times 10^{-9} \Delta K^{2.40} \quad (6.2)$$

6.5 Elevated Temperature Creep

- The conventional reheat treatment of service exposed discs resulted in a loss in rupture life due to an increase in the density of Ni_3Nb precipitates which lay parallel to the grain boundaries and these precipitates reduced the resistance to grain boundary sliding.
- The serrated grain boundaries in the damage tolerant heat treated material significantly reduced grain boundary sliding which resulted in an increase in rupture life. This reduction suggests that if a trapezoidal waveform, which simulates a turbine engine mission profile, is used for FCGR testing the damage tolerant reheat treatment may further reduce the FCGR's.
- The Larson Miller plot suggests that the increase in rupture life for the damage tolerant reheat treatment will occur over a wide range of stresses.

Chapter 7

Recommendations for Future Work

- The DEN specimen could be improved by modifying the notch such that it can better simulate the turbine disc bolt hole geometry. A modified DEN specimen shown in Figure 7.1, would allow the notch to be machined without the use of electron discharge machining. Excess material on the specimen blank, shown in Figure 7.1 by the dotted line, would make it possible to machine the notches as holes. The excess material could then be removed by electron discharge machining without affecting the notch root. The geometry of the notch and specimen cross section could also be improved by designing to simulate the actual stresses at the bolt hole, which could in turn be estimated by performing a finite element stress analysis of the disc.
- Only a small number of specimens were tested at each stress level in the present investigation. Further tests should be carried out at the same stress levels and at additional stress levels from 950 to 1100 MPa for improving the statistical significance of the data bases.
- Other short crack models should be examined in order to evaluate their usefulness in describing FCGR data. The possibility of using a mixed mode, K_I and K_{II} , parameter for describing the short growth behaviour should also be investigated.

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Table 2.1: Solvus temperatures of various phases in Inconel 718

<i>Phase</i>	<i>Solvus Temperature °C</i>
γ'	843-871 [22]
γ''	900 [22]
Ni ₃ Nb	982-1037 [31]
NbC	1040-1093 [31]
Laves	1148-1200 [19]

Table 2.2: Conventional and damage tolerant heat treatment schedules

1) Conventional

$$955^{\circ}\text{C}/1\text{h}/\text{AC} + 718^{\circ}\text{C}/8\text{h} \xrightarrow[1^{\circ}\text{C}/\text{min}]{\text{FC}} 621^{\circ}\text{C}/10\text{h}/\text{AC}$$

2) Damage Tolerant

$$1032^{\circ}\text{C}/30\text{min} \xrightarrow[1\text{h } 3\text{min}]{\text{FC}} 843^{\circ}\text{C}/4\text{h}/\text{AC}$$

$$+ 926^{\circ}\text{C}/1\text{h} \xrightarrow[1\text{h } 9\text{min}]{\text{FC}} 718^{\circ}\text{C}/8\text{h} \xrightarrow[1\text{h}]{\text{FC}} 621^{\circ}\text{C}/8\text{h}/\text{AC}$$

Table 2.3: Precipitate sizes resulting from the standard double aging sequence

<i>Reference</i>	<i>Conditions</i>	γ' (Å)	γ'' (Å)
Pishva [11] (J85-CAN15)	Service Exposed New Disc	200 950*	— —
Radavich [30]	Average	< 200	**
Andrews [34]	Damage Tolerant	~ 150	—
Paulonis [22]	Fully Heat-Treated	250	600 (ϕ)
Brooks [39]	Conventional	< 200	~ 200 (Width)

* Heat Treatment Not Available

** Diameter of γ'' approximately 5 to 6 times its width

Table 2.4: Elevated temperature tensile properties of Inconel 718

<i>Reference</i>	<i>Heat Treatment</i>	<i>Temp</i> (°C)	σ_{ys} (MPa)	ϵ_{ys} (%)	<i>UTS</i> (MPa)	ϵ_{uts} (%)	<i>Elong</i> (%) (2")	<i>RA</i> (%)
Alloy Digest [41]	annealed, aged 718°C/16h	649	382	—	1130	—	22.2	—
Mil Std [42]	996°C solution + con dbl age	649	860	0.6	1050	8	—	—
Yven & Roy [72]	conventional 1093°C solution + con dbl age	538 538	920 848	— —	1164 1080	— —	33.3 21.5	42.7 42.5
Saslananda & Shahinian [35]	conventional Wilson - modified	650 650	1004 874	— —	1158 1090	9.5 11.0	18.7 14.3	15.0 5.6
Mills & Jones [86]	conventional/ 1093 solution + con dbl age	649 649	954 827	— —	1011 926	— —	10.2 5.6	— 14.0
Hall [31]	conventional	650	950	—	—	—	19	—
Mills & Blackburn [91]	conventional - high conventional - low	538 538	1064 898	— —	1189 1118	— —	10 17	38 29
Brooks & Bridges [39]	conventional	650	965	—	1156	—	25	60
Wilkinson [29]	conventional - high conventional - low	649 649	931 1103	— —	1103 1276	— —	18 32	25 65
Barker [18]	conventional plate	650 649	1103 898	— —	— 1010	— —	— 9.2	— —
Kruegar [108]	conventional	649	965	—	1117	—	—	—
Coles et al [111]	conventional	649	862	—	999.7	—	12	18

con dbl age = conventional double age

Table 2.5: Elevated temperature fatigue crack initiation properties of Inconel 718

<i>Reference</i>	<i>Heat Treatment</i> (Grain Size)	<i>Temperature</i> (°C)	<i>Applied Stress</i> (MPa)	<i>Cycles to Initiation</i> (N _i)
Cook et al [46] [*]	conventional	650	480	10 ⁵
		650	550	10 ⁴
		650	600	10 ³
Au et al [33] ^{**}	conventional (15 μm)	650	812 ^{***}	3800
		650	900	1100
		650	967	475
		650	1010	150
	damage tolerant (45 μm)	650	750	1800
Mossei [103]	(32 μm)	650	800	650
		650	850	300
		650	890	70
		550	1200	2 x 10 ⁴
	(16 → 90 μm)	550	1100	~ 10 ⁵
		550	1100	2 x 10 ⁴
		550	1100	5 x 10 ⁴

* S - N curves (high cycle fatigue)

** fully reversible axial loading in strain control

*** maximum stress of first load cycle

Table 2.6: Sensitivity and precision of the direct current potential drop technique

<i>Reference</i>	<i>Sensitivity</i> (μm)	<i>Precision</i> (μm)
Jones et al [64]	50	—
Riddell et al [63]	70	25
Knott [56]	10-50	—
Brudin & Policella [92]	50	10
Gangloff [93]	100	0.8
Pishva [11]	—	25

Sensitivity - shortest crack measurable

Precision - minimum interval measurable

Table 2.7: Elevated temperature fatigue crack growth rate properties of Inconel 718

Reference	Heat Treatment	Temp (°C)	Freq (Hz)	Load Ratio R	Slope 'n'	Intercept 'C' (m/cycle)		Corr. Coeff	FCGR @ ΔK	
						mean	upper		(mm/cycle)	(MPa√m)
Andrews [34]	conventional IAR-modified	650	1	0.1	2.53	1.39×10^{-10}	1.85×10^{-10}	0.9889	2.76	50
		650	1	0.1	2.45	1.38×10^{-10}	1.84×10^{-10}	0.9943	2.01	50
Pishva [11]	conventional con-1520 TSN con-2045 TSN	650	1	0.1	1.81	1.44×10^{-9}	1.63×10^{-9}	0.9860	1.94	50
		650	1	0.1	2.88	5.70×10^{-11}	6.49×10^{-11}	0.9858	5.07	50
		650	1	0.1	1.86	1.17×10^{-9}	1.26×10^{-9}	0.992	1.82	50
Koul et al [4]	conventional IAR-modified damage toler	650	0.1	0.1	—	—	—	—	3.0	45
		650	0.1	0.1	—	—	—	—	1.7	45
		650	0.1	0.1	—	—	—	—	1.8	45
James & Mills [75]	conventional INEL-modified	649	0.667	0.05	2.52	$\sim 1.2 \times 10^{-7}$	—	0.9239	~ 2.6	50
		649	0.667	0.05	2.95	$\sim 2.3 \times 10^{-8}$	—	0.9588	~ 2.8	50
Smith & Michel [109]	conventional	649	0.17	0.05	~ 2.2	$\sim 1.9 \times 10^{-8}$	—	—	~ 9.0	50
Sadananda & Shahinian [35]	conventional wilson	650	0.17	0.05	~ 7.4	$\sim 4.5 \times 10^{-6}$	—	—	~ 30	50
		650	0.17	0.05	~ 3.1	$\sim 1.3 \times 10^{-5}$	—	—	~ 13	50
Nicholas et al [110]	conventional	649	0.01	0.1	—	$\sim 1.0 \times 10^{-9}$	—	—	~ 0.2	50
Mid Std [42]	conventional	638	0.667	0.05	~ 6.4	$\sim 1.0 \times 10^{-10}$	—	—	~ 2.0	50

INEL = Idaho National Engineering Laboratory

Corr Coeff = Correlation Coefficient

FCGR = Fatigue Crack Growth Rate

Table 2.8: Elevated temperature creep properties of Inconel 718

<i>Reference</i>	<i>Grain Size</i>	<i>Temperature (°C)</i>	σ (MPa)	t_R (HRS)	<i>Elong (%)</i>
Stroup & Heacox [23]	45 μ m	650	758	114	8.5
	12 μ m	650	758	73	15.7
Mil std [42]	982 sol + con dbl age	649	~690	100	—
		649	~590	1000	—
		649	~450	10000	—
Muller [26]	conventional	649	690	53.8	5.3
	1038/1h sol + 954/3h + con dbl age	649	690	38.7	4.1
	1038/1h sol + 917/10h + con dbl age	649	690	50.6	7.8
Notched:	conventional	649	690	153	—
	1038/1h sol + 954/3h + con dbl age	649	690	0.5	—
	1038/1h sol + 917/10h + con dbl age	649	690	0.9	—
Brooks [39]	conventional	730	415	197	38.0
Notched:	conventional	650	730	55	30.8
Wilkinson [2]	conventional	649	758	25-240	15-35
		649	724	60-300	10-35
		649	690	100-300	10-35
Kruegar [108]	conventional	649	620	100	—
Feldstein [65]	conventional	649	620	299	—

Table 3.1: Chemical compositions of Inconel 718

<i>Weight %</i>	<i>Stock (Wrought Bar)</i>	<i>General Electric Forgings</i>
Carbon (C)	.04	0.02 - 0.08
Manganese (Mn)	.09	0.35 max
Silicon (Si)	.12	0.35 max
Sulfur (S)	.002	0.015 max
Phosphorus (P)	.008	0.015 max
Chromium (Cr)	18.26	17.00 - 21.00
Iron (Fe)	18.87	15.00 - 21.00 (Balance)
Cobalt (Co)	.27	1.00 max
Molybdenum (Mo)	2.92	2.80 - 3.30
Niobium (Nb) + Tantalum (Ta)	5.14	4.75 - 5.50
Titanium (Ti)	1.00	.75 - 1.15
Aluminum (Al)	.58	0.30 - 0.70
Boron (B)	.0034	0.006 max
Copper (Cu)	.05	0.30 max
Nickel (Ni)	52.57	50.00 - 55.00
Calcium (Ca)	< .005	————
Magnesium (Mg)	.11 PPM	————

Table 3.2: Retired J85-CAN15 turbine disc time since new values

<i>Serial Number</i>	<i>Time Since New (Hours)</i>
First Stage WYOR-0123	Part # 12A00288P01 0.0
Second Stage 11J781-61 11J781-51 11J781-57 11J781-41	Part # 12A00289P01 1647.2 (1650) 1999.5 (2000) 2089.0 (2090) 2293.3 (2300)

Table 4.1: Strengthening precipitate γ' and γ'' sizes

Precipitate Diameter ($\text{\AA}$)					
<i>Precipitate</i>	Disc Region				
	<i>Bore Region</i>	<i>Thin Section of Disc</i>	<i>Compact Tension Notch Tip</i>	<i>Bolt Hole Region</i>	<i>Rim Region</i>
Service Exposed 2000 hour TSN disc					
γ'	100-200	150-200	--	150-200	200
γ''	200-300	250-300	150-250	200-300	350-400
Conventional Reheat Treated (2000 hr TSN)					
γ'	100-150	150	150-200	200-300	200
γ''	200-250	300	200-250	350-400	300
Service Exposed 2300 hour TSN disc					
γ'	150	200	200	150-200	200-250
γ''	300	250-300	300-350	300-350	300-400
Damage Tolerant Reheat Treated (2300 hr TSN)					
γ'	150-200	150-200	250-300	200-300	200
γ''	250-300	250-300	300-400	350-400	350-400

Table 4.2: Tensile results at 650°C

<i>Material</i>	$\sigma_{y.2\%}$ (MPa)	ϵ_y (%)	<i>E</i> (MPa)	<i>U.T.S.</i> (MPa)	ϵ_f (%)	<i>R.A</i> %	<i>Elongation</i> (%)
Stock Material							
DT	939.5	.617	202500	1101.8	9.1	38.6	8.89
	937.3	.734	176900	1092.8	11.0	41.7	——
CON	1066.2	.583	257900	1170.8	9.7	34.9	9.12
	1059.7	.636	237900	1172.0	8.2	35.4	——
2300 hour TSN disc							
SE	1103.4	.917	174900	1189.3	7.9	41.9	7.28
	1090.2	.828	204300	1090.3	7.6	45.4	6.91
DT	952.6	.818	152600	1101.2	8.5	35.4	7.69
	943.9	.766	177600	1102.1	9.4	38.6	8.70
2000 hour TSN disc							
SE	1064.0	.761	195300	1163.2	9.0	54.4	8.07
	1061.8	.777	182300	1161.4	8.3	52.6	7.00
CON	1057.5	.783	186800	1162.5	10.7	49.1	9.94
	1068.4	.765	204300	1172.7	11.0	40.4	——
First Stage New Disc							
NEW	935.1	.763	158600	1091.2	14.7	40.9	12.2
	954.8	.922	129900	1086.3	14.4	36.7	9.3

Table 4.3: Smooth cylindrical and double edge notched disc material fatigue crack initiation results at 650°C

<i>Stress Level</i>		<i>Smooth Cylindrical Specimens</i>		<i>Double Edge Notched Disc Specimens</i>					
Maximum Stress (MPa)	Stress Range (MPa)	Cycles to Failure (Initiation)		Cycles to Initiation		Cycles to Failure		Percentage Life for Initiation	
		CON	DT	SE	DT	SE	DT	SE	DT
740	666	4605745 1x10 ⁷ +	257120 369530	1698139 —	44212 25404 11764	1698139 —	— 27049 13130	99.9 —	— 93.9 83.6
950	855	53954 14361 22445 33.27	15450 13357 10196 —	3635 [†] 31263 33675 —	4424 [*] 3649 — —	4362 31890 34217 —	5033 4562 — —	83.3 98.0 98.4 —	87.8 80.0 — —
1100	990	2012 1054	261 234	2548 2957	2724 2185	3019 3338	3177 2747	84.4 88.6	85.7 80.0

[†]Specimens considered to have run out

^{*}100,000 cycles at 740 MPa prior to increasing load

^{**}100,000 cycles at 830 MPa prior to increasing load

Table 4.4: Direct current potential drop elevated temperature initiation calibration results at 650°C

Specimen	Voltage Increase (%)	DCPD Crack Length (μm)	Measured Crack Length (μm)		Difference (μm)	
			Equivalent Crack Depth	Actual Depth	DCPD vs Equivalent	DCPD vs Actual
Stock Conventional	1.75	60.4	58	60	2.4	0.4
Stock Conventional	1.5	51.9	55	55	3.1	3.1
Stock Damage Tolerant	1.3	44.9	49	77	4.1	32
Stock Damage Tolerant	1.0	34.6	46	50	11.4	15.4
Disc Damage Tolerant	1.67	67	65	180*	2	113

* Thumbnail crack

Table 4.5: Double edge notched stock material fatigue crack initiation results at 650°C

<i>Double Edge Notched Stock Specimens</i>							
Maximum Stress (MPa)	Stress Range (MPa)	Cycles to Initiation		Cycles to Failure		Percent Initiation	
		CN	DT	CN	CT	CN	DT
775	698	53221	23218	54377	24034	97.7	96.6
		19111 *	28201	19279	29568	99.1	95.4
		16157	—	—	—	—	—
1000	900	2210	1886	2520	2230	87.7	84.6
		4269	2707	4490	3013	95.1	89.8
		4955	—	5562	—	89.1	—
1060	950	1848	2053	2199	2419	84.0	84.9
		1866 **	1975	2032	2272	91.8	86.9
		1641	—	1992	—	82.4	—

* failed at threads

** failed between reference leads

Table 4.6: Fatigue crack propagation results at 650°C

Compact Tension Specimens					
Stress Level	Number of Specimens	n	Cupper (m/cycle)	C _{mean} (m/cycle)	Corr. Coeff.
52MPa	ΔK _{max} = 56, ΔK _{min} = 30 (MPa√m)				
2000 hour TSN Disc					
SE	3	2.37	0.277x10 ⁻⁹	0.185x10 ⁻⁹	0.9713
CON	3	2.36	0.325x10 ⁻⁹	0.241x10 ⁻⁹	0.9925
2300 hour TSN Disc					
SE	3	2.33	0.289x10 ⁻⁹	0.215x10 ⁻⁹	0.9869
DT	3	2.49	0.125x10 ⁻⁹	0.955x10 ⁻⁹	0.9972
First Stage New Disc [a]					
NEW	4	1.84	0.165x10 ⁻⁸	0.131x10 ⁻⁸	0.9863
Double Edge Notched Specimens					
2090 hour TSN Disc					
740 MPa	ΔK _{max} = 100, ΔK _{min} = 50 (MPa√m)				
SE	2	1.32	0.225x10 ⁻⁷	0.228x10 ⁻⁷	0.9932
DT	2	0.60	0.893x10 ⁻⁶	0.587x10 ⁻⁶	0.9662
950MPa	ΔK _{max} = 100, ΔK _{min} = 55 (MPa√m)				
SE	3	1.94	0.364x10 ⁻⁸	0.231x10 ⁻⁸	0.8953
DT	2	1.28	0.460x10 ⁻⁷	0.359x10 ⁻⁷	0.9429
1100MPa	ΔK _{max} = 100, ΔK _{min} = 60 (MPa√m)				
SE	2	2.71	0.146x10 ⁻⁹	0.832x10 ⁻¹⁰	0.8918
DT	2	2.12	0.182x10 ⁻⁸	0.913x10 ⁻⁹	0.8159

SE - Service Exposed
 CON - Conventionally reheat treated
 DT - Damage tolerant reheat treated

Table 4.7: Short fatigue crack propagation results at 650°C

<i>Double Edge Notched Specimens</i>					
<i>Stress Level</i>	<i>Number of Specimens</i>	<i>n</i>	<i>C_{upper}</i> (m/cycle)	<i>C_{mean}</i> (m/cycle)	<i>Correlation Coefficient</i>
740 MPa		$\Delta K_{\max} = 100, \Delta K_{\min} = 6.24 \text{ (MPa}\sqrt{\text{m}})$			
SE	1	2.49	0.600×10^{-9}	0.176×10^{-9}	0.9677
DT	2	2.40	0.129×10^{-8}	0.266×10^{-9}	0.9405
950 MPa		$\Delta K_{\max} = 100, \Delta K_{\min} = 8.00 \text{ (MPa}\sqrt{\text{m}})$			
SE	3	2.35	0.113×10^{-8}	0.356×10^{-9}	0.9604
DT	2	2.22	0.200×10^{-8}	0.575×10^{-9}	0.9560
1100 MPa		$\Delta K_{\max} = 100, \Delta K_{\min} = 9.27 \text{ (MPa}\sqrt{\text{m}})$			
SE	2	2.12	0.248×10^{-8}	0.914×10^{-9}	0.9596
DT	2	1.99	0.384×10^{-8}	0.139×10^{-8}	0.9514

Table 4.8: Creep results at 650°C

Specimen	$t_p + t_s$ (hr)	t_t (hr)	t_R (hr)	$\epsilon_p + \epsilon_s$ (%)	ϵ_t (%)	ϵ_R (%)	ϵ_{min} ($1/s \times 10^{-9}$)
SE	2300 hour TSN disc						
1	100	182	282	1.0	8.073	9.073	6.389
2	125	124	249	0.8	6.861	7.661	8.056
3	140	139	279	0.75	10.95	11.696	6.111
DT	2300 hour TSN disc						
1	175	125	300	1.45	4.688	6.138	9.722
2	—	—	348	—	—	—	—
3	225	181	406	1.47	4.686	6.156	7.778
SE	2000 hour TSN disc						
1	110	160	270	1	7.06	8.065	6.667
2	100	161	261	0.83	8.46	9.287	6.389
3	113	157	270	1.162	9.25	10.411	6.389
CON	2000 hour TSN disc						
1	84	86	170	1.14	7.036	8.176	11.111
2	86.2	101.8	188	0.80	7.856	8.656	11.667
3	100	95	195	1.07	8.003	9.073	13.056
NEW	First stage new disc						
1	57.9	28.1	86	1.98	1.815	3.795	4.778
2	49.6	42.4	92	1.55	1.921	3.471	3.389
3	35	56.0	91	1.33	2.442	3.772	3.833

SE - service exposed

CON - conventional reheat treatment

DT - damage tolerant reheat treatment

Table 5.1: Swain and Newman short to long crack growth transition points [85]

<i>Transition Point (μm)</i>		
Crack Geometry	a (μm)	ΔK ($\text{MPa}\sqrt{\text{m}}$)
Constant 'c' varying a, t = 1.19mm		
c = t	280	45
c = $\frac{2}{3}t$	240	44
c = $\frac{1}{2}t$	210	43
c = $\frac{1}{3}t$	180	42
Proportional a/c		
a = c	135	42
a = $\frac{2}{3}c$	150	42
a = $\frac{1}{2}c$	180	43
a = $\frac{1}{3}c$	240	44

a - Crack depth
2c - Surface Crack length
t - Specimen thickness

Figure 1.1: Schematic representation of the damage tolerance based life cycle management methodology [4]

Figure 2.1: The microstructure of Inconel 718 [17]

Figure 2.2: Time-Temperature-Precipitation diagram of Inconel 718

Figure 2.3: Grain coarsening behaviour of Inconel 718 [4]

Figure 2.4: Inconel 718 grain boundary Ni_3Nb time-temperature-precipitation curves for various solution temperatures [4]

Figure 2.5: Schematic illustrating the effects of microstructure on Inconel 718 tensile properties [21]

Figure 2.6: Loading waveform and definitions

Figure 2.7: Constant load amplitude fatigue data for 7075-T6 aluminum alloy notch specimens [38]

SMOOTH CYLINDRICAL SPECIMEN

Figure 2.8: Notch root discontinuity simulation using a smooth specimen

Figure 2.9: Various crack geometries at initiation

Figure 2.10: Stress concentration factors for grooves of various depths in an axially loaded flat bar [66]

Figure 2.11: The effect of tensile strength and notch radius on relative notch sensitivity [38]

Figure 2.12: Ratio of cycles to initiation to total cycles to failure as a function of notch root radius in 0.23 C, 0.88 Mn, 0.04 Si steel [38]

Figure 2.13: Crack propagation data as a function of number of cycles (N) versus crack length (a) [38]

Figure 2.14: Schematic da/dN versus ΔK curve for superalloys [1]

Figure 2.15: Summary of ΔK -correlated micro crack versus large crack growth in 7075 Al alloy [77]

(a) Slower crack growth due to an irregular crack front

(b) Low initial crack growth rates due to erratic crack initiation

Figure 2.16: The effects of multiple initiation on crack growth [55]

Figure 2.17: Edge crack assumption for short crack modelling of notch specimen

Figure 2.18: Nomenclature of Swain and Newman short crack growth model [85]

Figure 2.19: The effect of temperature on the fatigue crack propagation behaviour of Inconel 718 in Air [88]

Figure 2.20: The effect of cyclic frequency on the fatigue crack propagation behaviour of Inconel 718 in Air at 538°C [88]

Figure 2.21: (a) The effect of environment on the fatigue crack growth rate of Inconel 718 at 650°C (•○ small grain size; ▼▽ large grain; ◆◇ necklace grain structure) [89]

Figure 2.21: (b) The effect of grain size on the fatigue crack growth rate of Inconel 718 at 538°C (ASTM grain sizes: Heat A=11.5, Heat E=5, Heat F=10.5, Heat G=10, Heat G Forging=5 with a necklace) [88]

Figure 2.22: Single edge notch assumption for double edge notch specimen after crack initiation

Figure 2.23: Typical creep curve showing the three regions in strain-time-space [38]

Figure 2.24: Larson Miller plot of 5-590 iron based alloy [38]

(a) Cross section from disc

(b) Split sections for optical and transmission electron microscopy

(c) Final sections for optical microscopy

Figure 3.1: Sections from Inconel 718 second stage J85-CAN15 turbine disc

NOTE 1: ALL DIMENSIONS IN INCHES

NOTE 2: THE REDUCED SECTION MAY HAVE A GRADUAL TAPER FROM THE ENDS TOWARD THE CENTER, WITH THE ENDS NOT MORE THAN 0.003 IN LARGER IN DIAMETER THAN THE CENTER

DRAWN BY ROBERT ANDREWS

DATE JULY 25, 1988

NOT TO SCALE

Figure 3.2: Tensile specimen geometry

(a) Test frame, controller and direct current potential drop technique

(b) Specimen and voltage probe connections

Figure 3.10: Double edge notch test apparatus used for fatigue crack initiation and propagation studies

(a) Test frame and controller

(b) Load line, specimen and extensometer

Figure 3.4: Tensile testing apparatus used

Extensometer Grip Mount
 Drawn By Robert Andrews
 Date May 3, 1969

Note 1 All Dimensions in Inches
 Note 2 Material - Hastelloy-X
 Note 3 Not To Scale

(a) Ring clamp for specimen grip

Extensometer Rod Mounts
 Drawn By Robert Andrews
 Date May 2, 1969

Note 1 All Dimensions In Inches
 Note 2 Material - Hastelloy-X
 Note 3 Not To Scale

(b) Plates which attach to extensometer rods

Figure 3.5: Extensometer detail for tensile and creep testing

SMOOTH CYLINDRICAL SPECIMEN

NOTE 1: ALL DIMENSIONS IN INCHES

NOTE 2: MATERIAL - IN 718, PROVIDED

NOTE 3: NO UNDERCUT AT FILLETS

NOTE 4: USE VERY LIGHT PASSES ON LAST SEVERAL PASSES TO MINIMIZE CIRCUMFERENTIAL MACHINE MARKS IN GAUGE LENGTH AND AT RADIUS

NOTE 5: GAUGE LENGTH TO BE POLISHED TO 600 FINISH

DRAWN BY: ROBERT ANDREWS

DATE: APRIL 2, 1990

NOT TO SCALE

Figure 3.6: Smooth cylindrical specimen geometry

(a) Test frame, controller and induction heater

(b) Smooth cylindrical specimen, grips and heater coil

(c) Thermocouple for temperature calibration

Figure 3.7: Smooth cylindrical test apparatus used for fatigue testing

DOUBLE EDGE NOTCHED SPECIMEN

DRAWN BY: ROBERT ANDREWS

DATE: MAY 10, 1989

NOTE 1 ALL DIMENSIONS IN INCHES

NOTE 2 MATERIAL - IN 718, PROVIDED

NOTE 3. NO UNDERCUT AT FILLETS

NOTE 4 REDUCED SECTION TO BE SMOOTH AND FREE OF SCRATCHES

NOTE 5 RADIUS MARKED '*' CAN BE INCREASED IF NECESSARY TO STANDARD TOOL SIZE

Figure 3.8: Double edge notch specimen geometry

J85 CAN-15 SECOND STAGE TURBINE DISC
 DOUBLE EDGE NOTCHED SPECIMENS
 DRAWN BY ROBERT ANDREWS
 DATE: MAY 10, 1989

- NOTE 1: ALL DIMENSIONS IN INCHES
 NOTE 2: EDGE NOTCHED SPECIMENS ARE TO BE MACHINED AS SHOWN IN DRAWING 2
 NOTE 3: SPECIMENS LABELLED 'A' ARE OFFSET 25 DEGREES CLOCKWISE FROM RADIAL
 NOTE 4: SPECIMENS LABELLED 'B' ARE OFFSET 25 DEGREES COUNTER CLOCKWISE FROM RADIAL

SECTION VIEW

Figure 3.9: Drawing of double edge notch specimen locations in second stage J85-CAN15 turbine disc

(a) Test frame, controller and direct current potential drop technique

(b) Specimen and voltage probe connections

Figure 3.10: Double edge notch test apparatus used for fatigue crack initiation and propagation studies

Figure 3.11: Schematic diagram of direct current potential drop technique

Isolation of Direct Current Potential Drop Technique

Drawn By Robert Andrews
Date June 1, 1990

Note Drawing Not To Scale - Schematic Only

Figure 3.12: Electrical isolation of direct current potential drop assembly from test frame

Figure 3.13: Compact tension specimen geometry

(a) Test frame, controller and direct current potential drop technique

(b) Specimen and reference locations as well as current lead and voltage probe placement

Figure 3.14: Compact tension test apparatus used for fatigue crack growth rate testing

Figure 3.15: Creep test apparatus used

Figure 3.16: Creep testing stress selection for Inconel 718

Figure 4.1: Hardness profile of turbine discs

(a) Conventional heat treatment ($8\mu\text{m}$)

(b) Damage tolerant heat treatment ($50\mu\text{m}$)

Figure 4.2: Optical micrographs of stock Inconel 718 material

(a) Service exposed disc (2300 hour TSN)

Figure 4.3: Optical micrographs showing turbine disc microstructural profile

(b) Conventionally reheat treated disc

Figure 4.3: Optical micrographs showing turbine disc microstructural profile

(c) Damage tolerant reheat treated disc

Figure 4.3: Optical micrographs showing turbine disc microstructural profile

Figure 4.4: Grain size variation of turbine discs from bore towards the rim

(a) Grain boundary structure

(b) Dislocations in the vicinity of grain boundary precipitates

(c) Dislocations emanating from Ni_3Nb precipitate

Figure 4.5: Transmission electron micrographs showing typical dislocation characteristics found in turbine discs

(d) Dislocation pileup

(e) Dislocation entanglement

(f) Serrated grain boundary

(g) Homogenous damage tolerant microstructure

Figure 4.5: Transmission electron micrographs showing typical dislocation characteristics found in turbine discs

(a) [001] diffraction pattern

(b) Bright field [001] direction

(c) Dark field $(\bar{1}00)$ reflection of (b)

(d) Dark field $(\frac{1}{2}10)$ reflection of (b)

Figure 4.6: Typical dark field transmission electron micrographs

(a) Yield region

(b) Tensile curve

Figure 4.7: Stock material tensile results at 650°C

(a) Yield region

(b) Tensile curve

Figure 4.8: Service exposed and new disc tensile results at 650°C

(a) Yield region

(b) Tensile curve

Figure 4.9: Tensile results of a service exposed disc before and after the conventional reheat treatment at 650°C

(a) Yield region

(b) Tensile curve

Figure 4.10: Tensile results of a service exposed disc before and after the damage tolerant reheat treatment at 650°C

Figure 4.11: Smooth cylindrical and double edge notch disc material fatigue crack initiation results at 650°C

Figure 4.12: Smooth cylindrical and double edge notch stock material fatigue crack initiation results at 650°C

(a) Conventional specimen initiation region

(c) Conventional specimen initiation site and secondary cracking

(b) Damage tolerant specimen initiation region showing some existence of cleavage fracture

(d) Damage tolerant specimen striated surface

Figure 4.13: Scanning electron micrographs of smooth cylindrical specimen fracture surfaces tested at 650°C and 740 MPa:

(a) Conventional specimen initiation site

(b) Damage tolerant specimen initiation site

(c) Conventional specimen secondary cracking

(d) Damage tolerant specimen striated surface

Figure 4.14: Scanning electron micrographs of smooth cylindrical specimen fracture surfaces tested at 650°C and 950 MPa:

(a) Conventional specimen initiation region

(b) Damage tolerant specimen initiation site

(c) Conventional specimen striated surface

(d) Damage tolerant specimen striated surface

Figure 4.15: Scanning electron micrographs of smooth cylindrical specimen fracture surfaces tested at 650°C and 1100 MPa:

Figure 4.16: Long term stability of the direct current potential drop technique for fatigue crack initiation testing at 650°C

(a) Initiation site at low magnification

(b) Initiation site at high magnification

(c) Notch face

(d) Notch face secondary crack

Figure 4.17: Scanning electron micrographs of double edge notch calibration specimens tested at 650°C and 775 MPa.

(e) Disc specimen initiation region

(f) Disc specimen initiation region

Figure 4.17: Scanning electron micrographs of double edge notch calibration specimens tested at 650°C and 775 MPa.

Figure 4.18: Elevated temperature double edge notch specimen fatigue crack initiation results at 650°C

(a) Service exposed at 740 MPa

(b) Service exposed at 950 MPa

(c) Service exposed at 1100 MPa

Figure 4.19: Scanning electron micrographs of double edge notch disc specimen initiation regions at 650°C

(d) Damage tolerant at 740 MPa

(e) Damage tolerant at 950 MPa

(f) Damage tolerant at 1100 MPa

Figure 4.19: Scanning electron micrographs of double edge notch disc specimen initiation regions at 650°C

Figure 4.20: Elevated temperature double edge notch specimen cycles to failure results at 650°C

Figure 4.21: Direct current potential drop technique calibration curve for double edge notch specimen

Figure 4.22: Double edge notch stock material fatigue crack propagation results using standard linear elastic fracture mechanics for ΔK calculations at 650°C

(c) 1060 MPa

Figure 4.22: Double edge notch stock material fatigue crack propagation results using standard linear elastic fracture mechanics for ΔK calculations at 650°C

(a) Secondary cracking on fracture surface
in conventionally heat treated material

(b) Cleavage, striations and secondary cracking
on fracture surface in damage tolerant heat
treated material

Figure 4.23: Scanning electron micrographs of stock material
double edge notch specimens

(a) 740 MPa

(b) 950 MPa

Figure 4.24: Double edge notch turbine disc fatigue crack propagation determined using standard linear elastic fracture mechanics for ΔK calculations at 650°C

(c) 1100 MPa

Figure 4.24: Double edge notch turbine disc fatigue crack propagation determined using standard linear elastic fracture mechanics for ΔK calculations at 650°C

Figure 4.25: Double edge notch turbine disc fatigue crack propagation results combined for all of the three stress levels at 650°C

(a) Mean line at 740 MPa

(b) Upper bound line at 740 MPa

Figure 4.26: Double edge notch turbine disc short crack propagation determined using the Swain and Newman model [85] for ΔK calculations at 650°C

(c) Mean line at 950 MPa

(d) Upper bound line at 950 MPa

Figure 4.26: Double edge notch turbine disc short crack propagation determined using the Swain and Newman model [85] for ΔK calculations at 650°C

of/de

3

(e) Mean line at 1100 MPa

(f) Upper bound line at 1100 MPa

Figure 4.26: Double edge notch turbine disc short crack propagation determined using the Swain and Newman model [85] for ΔK calculations at 650°C

Figure 4.27: Double edge notch turbine disc short crack propagation results combined for all three stress levels at 650°C

(a) Secondary cracking on fracture surface in conventionally reheat treated specimen

(c) Striations on fracture surface in conventionally reheat treated specimen

(b) Secondary cracking on fracture surface in damage tolerant reheat treated specimens

(d) Striations on fracture surface in damage tolerant reheat treated specimens

Figure 4.28: Scanning electron micrographs of disc material double edge notch fracture surfaces

Figure 4.29: Direct current potential drop calibration curve for the compact tension specimen

(a) 2000 hour TSN disc

(b) 2300 hour TSN disc

Figure 4.30: Compact tension specimen fatigue crack propagation data for Inconel 718 at 650°C

(a) 2000 and 2300 hour TSN service exposed and new disc

(b) 2000 hour TSN service exposed, conventional reheat and new disc

Figure 4.31: Compact tension specimen fatigue crack propagation upper bound lines for Inconel 718 at 650°C

(c) 2300 hour TSN service exposed, damage tolerant reheat and new disc

Figure 4.31: Compact tension specimen fatigue crack propagation upper bound lines for Inconel 718 at 650°C

(a) 2000 hour TSN service exposed

(b) 2000 hour TSN conventional reheat treatment

Figure 4.32: Optical micrographs showing an increase in Ni_3Nb precipitate size at the grain boundaries due to the conventional reheat treatment of service exposed discs

(a) Secondary cracks and striations on conventional fracture surface

(b) Secondary cracks and striations on damage tolerant fracture surface

Figure 4.33: Scanning electron micrographs of compact tension specimen fracture surfaces

(a) 2000 hour TSN service exposed disc

(b) 2000 hour TSN conventional reheat treatment disc

Figure 4.34: Creep rupture curves at 650°C

(c) 2300 hour TSN service exposed disc

(d) 2300 hour TSN damage tolerant reheat treated disc

Figure 4.34: Creep rupture curves at 650°C

(e) First stage new disc

Figure 4.34: Creep rupture curves at 650°C

Figure 4.35: Histogram of stress rupture lives at 650°C

(a) Constant (C_{LM}) = 20 [94]

(b) Constant (C_{LM}) = 25 [18]

Figure 4.36: Larson-Miller parameter plots for Inconel 718 in different heat treatment and service exposed conditions at 650°C

(b) Damage tolerant disc material

(a) Service exposed disc material

Figure 5.1: Fatigue crack propagation curves for combined double edge notch and compact tension specimen results at 650°C

PROPOSED DOUBLE EDGE NOTCHED SPECIMEN

DRAWN BY ROBERT ANDREWS

DATE MAY 30, 1990

NOT TO SCALE

NOTE 1 ALL DIMENSIONS IN INCHES

NOTE 2 SPECIMENS ARE TO BE MACHINED ACCORDING TO THE HIDDEN LINE. FIRST WITH HOLES PREPARED ACCORDING TO DISC SPECIFICATIONS (DRILLED, REAMED AND SHOT PEENED) FOLLOWING THESE OPERATIONS THE MATERIAL REPRESENTED BY THE HIDDEN LINES SHOULD BE REMOVED

NOTE 3 THE REDUCED SECTION SHOULD BE SMOOTH AND POLISHED TO A 600 FINISH

Figure 7.1: Recommended modified double edge notch specimen geometry

Appendix A: Short Crack Growth Models

(1) Wang and Lu Model [79]:

The stress intensity factor, K , is determined for both short and long crack growth in a notched specimen using the following approximation:

$$K = \begin{cases} K_1 & \text{if } K_1 < K_2 \\ K_2 & \text{if } K_2 < K_1 \end{cases} \quad (1)$$

where K_1 is the short crack stress intensity factor, equation 2 and K_2 is the long crack stress intensity factor, equation 3.

$$K_1 = 1.12\sigma_y\sqrt{\pi a} \quad (2)$$

$$K_2 = Y \left(\frac{a + r}{W} \right) \sigma_{nom} \sqrt{\pi(a + r)} \quad (3)$$

where σ_y is the notch root stress, σ_{nom} is the remote applied stress, a is the crack depth, r is the notch radius and Y is the linear elastic fracture mechanics geometric factor.

This technique assumes that for short crack growth in notched specimens, the local effects at the notch root are dominate.

Results from the this model using data from the double edge notch fatigue crack propagation tests are given in Table A1. The stress intensity factor range, ΔK , are calculated using the stress range, $\Delta\sigma$. The transition point from short to long crack growth is defined by equation (1) as the point where the short crack growth ΔK_1 is greater than or equal to the long crack growth ΔK_2 . Table A1 shows that for the specimens and test conditions used, no transition occurred for this model since ΔK_1 was less than ΔK_2 for the entire crack growth region.

(2) Swain and Newman Model [85]:

The model proposed by Swain and Newman uses a different equation for the short crack stress intensity factor, assuming a semi-elliptical surface crack located at the centre of a semi-circular edge notch, as shown in Figure 2.18, and given by:

$$K_1 = \sigma_y \sqrt{\pi \frac{c}{Q}} F_{sn} \quad (4)$$

where Q is a shape factor, equation 5 and F_{sn} is the short crack geometric factor given in equation 6 in which c is a function of a , c and t as shown in Figure 2.18 and ϕ the parametric angle of the notch:

$$Q = 1 + 1.464 \left(\frac{c}{a} \right)^{1.65} \quad \text{for } \frac{c}{a} < 1 \quad (5a)$$

$$Q = 1 + 1.464 \left(\frac{a}{c} \right)^{1.65} \quad \text{for } \frac{c}{a} > 1 \quad (5b)$$

$$F_{sn} = \left[M_1 + M_2 \left(\frac{c}{t} \right)^2 + M_3 \left(\frac{c}{t} \right)^4 \right] g_1 g_2 g_3 g_4 f_\phi f_w \quad (6)$$

for $c/a < 1$:

$$M_1 = 1 \quad (6a)$$

$$M_2 = \frac{0.05}{0.11 + \left(\frac{c}{a} \right)^{\frac{3}{2}}} \quad (6b)$$

$$M_3 = \frac{0.29}{0.23 + \left(\frac{c}{a} \right)^{\frac{3}{2}}} \quad (6c)$$

$$g_1 = 1 - \frac{\left(\frac{c}{r} \right)^4 \left[2.6 - 2 \left(\frac{c}{r} \right) \right]^{\frac{1}{2}}}{1 + 4 \left(\frac{c}{a} \right)} \cos \phi \quad (6d)$$

$$g_2 = \frac{1 + 0.358\lambda + 1.425\lambda^2 - 1.578\lambda^3 + 2.156\lambda^4}{1 - 0.08\lambda^2} \quad (6e)$$

where,

$$\lambda = \frac{1}{1 + \frac{a}{r} \cos (0.9 \phi)} \quad (6f)$$

$$g_3 = 1 + 0.1 (1 - \cos \phi)^2 \left(1 - \frac{c}{r} \right)^{10} \quad (6g)$$

$$g_4 = 1.14 - \frac{0.1}{\sqrt{1 + \frac{a}{r}}} \quad (6h)$$

$$f_w = 1 - 0.2\gamma + 9.4\gamma^2 - 19.4\gamma^3 + 27.1\gamma^4 \quad (6i)$$

where,

$$\gamma = \frac{(a + r)}{w} \quad (6j)$$

$$f_\phi = \left[\left(\frac{c}{a} \right)^2 \cos^2 \phi + \sin^2 \phi \right]^{\frac{1}{4}} \quad (6k)$$

and for $c/a > 1$ all functions remain the same except for the following:

$$M_1 = \sqrt{\frac{a}{c}} \quad (6l)$$

$$f_\phi = \left[\left(\frac{a}{c} \right)^2 \sin^2 \phi + \cos^2 \phi \right]^{\frac{1}{4}} \quad (6m)$$

Due to the complexity of the model a mathematical software package, *TK solver*, was used to perform the calculations. The *Rule Sheet* and *Variable Sheet* entered into TK solver are given in Table A2, where ΔK was calculated for both the short and long crack equations. The calculations were repeated for a variety of different crack geometries for crack depths, a , ranging from 50 to 350 μm , see table A1. Initially it was assumed that half the crack length at the surface, c , was a constant percentage of half the specimen thickness, t ($c=0.33t$, $c=0.5t$, $c=0.66t$, $c=t$). It was then assumed that half the surface crack length, a , varied proportionally to the crack depth, a ($a=c$, $a=0.66c$, $a=0.5c$, $a=0.33c$). A summary of the resulting transition points determined for each of the series of calculations are given in Table A3.

Table A1: Wang and Lu short crack model [79]

a (mm)	a + r (mm)	ΔK_1 (MPa $\sqrt{m}$)	ΔK_2 (MPa $\sqrt{m}$)
Alternating stress (σ_a) = 855 MPa			
0.135	1.325	19.32	41.85
0.511	1.701	38.36	51.05
1.42	2.61	63.95	78.23
1.898	3.088	72.95	97.16
3.81	5.0	104.77	170.3
$\sigma_a = 666$ MPa			
0.280	1.47	22	35
3.022	4.212	72.68	129
$\sigma_a = 990$ MPa			
0.246	1.43	30.82	51.5
1.529	2.719	76.0	95.0

Table A2: TK Solver rule and variable sheets

Rule Sheet:				
C	c=	.33*t		
	a=	2.0*c		
	K=	S*.qrt(pi *c/Q)*f		
	Q=	1+1.464*(a/c)^1.65		
	f=	(m1+m2*(c/t)^2+m3*(c/t)^4)*g1*g2*g3*g4*fp*fw		
	m1=	sqrt(a/c)		
	m2=	0.05/(0.11+c/a)^(3/2))		
	m3=	0.29/(0.23+c/a)^(3/2))		
	g1=	1-(c/t)^4*(2.6-2)*c/t)^(1/2)*cos(phi)/(1+4*(c/a))		
	g2=	(1+0.358*L+1.425*L^2-1.578*L^3+2.156*L^4)/(1+0.08*L^2)		
	L=	1/(1+a/4*cos(.9*phi))		
	g3=	1+0.1*(1-cos(phi))^2*(1-c/t)^10		
	g4=	1.14-0.1/sqrt(1+a/r)		
	fw=	1-.2*gam+9.4*gam^2-19.4*gam^3+27.1*gam^4		
	gam=	(a+r)/w		
	fp=	((a/c)^2*(sin(phi))^2+(cos(phi))^2)^(1/4)		
	cd=	a*1000*1000		
	Ksen=	RL/B/ws*sqrt(as)*(1.99-0.41*as/w+18.7*(as/w)^2-38.48*(as/w)^3+53.85(as/w		
	as=	a+r		
Variable Sheet:				
L		K	31.33193	Short Crack delta K (MPa(m)^1/2)
	855	S		Stress at Notch Root (MPa)
	3.14159	pie		Pie
		c		Length of Crack face = 2c (m)
		Q	1.0739339	
		f	.62083928	
L	.000195	a		Crack Depth (m)
		m2	.00328852	
	.001191	t		Specimen Thickness (m)
		m3	.01892407	
		g1	.99999707	
		g2	2.9640029	
		g3	1	
		g4	1.0473066	
		fp	.40463315	
		fw	1.1579654	
	1.5707	phi		Notch Ellipse Axis - Pie/2 (radians)
		L	.07499313	
	.00119	r		Radius of Notch (m)
		gam	.16616677	
	.008335	w		Specimen Width (m)
L		cd	195	Crack Depth (microns)
L		Ksen	73.270666	Single Edge Notch deltaK (MPa (m) ^1/2)
		as	.001385	Crack Depth plus Notch Radius (m)
	10026	RL		Nominal Load Range (N)
	2 3835	B		Specimen Thickness (mm)
	8.335	ws		Specimen Width (mm)

Table A3: Swain and Newman short crack model [85]

<i>Transition Crack Depth (a) and Stress Intensity Factor Range(ΔK)</i>		
Crack Geometry	a_t (μm)	ΔK MPa $\sqrt{\text{m}}$
$c = t$	280	45
$c = \frac{2}{3}t$	240	44
$c = \frac{1}{2}t$	210	43
$c = \frac{1}{3}t$	180	42
$a = c$	135	42
$a = \frac{2}{3}c$	150	42
$a = \frac{1}{2}c$	180	43
$a = \frac{1}{3}c$	240	44

Appendix B: Metallographic Preparation

Optical Etchant

After preparing specimens using standard metallographic techniques they were etched using the following etchant:

Etchant	(1)	40 ml distilled water
	(2)	20 ml Hydrogen Peroxide (H_2O_2)
	(3)	60 ml Hydrochloric Acid (HCl)

After mixing let stand for approximately five minutes until boiling rapidly, submerge specimen in etchant for approximately thirty seconds. This etchant was only useable for approximately five hours before more is required.

Transmission Electron Microscope Thin Foils

Three millimetre diameter rods were sectioned into 0.4 mm thick foils which were then thinned further to between 0.13 to 0.18 mm using 600 grit abrasive paper. These foils were then thinned using a Struers Tenupol 2 jet polisher using one of the following electrolytes and parameters.

Room Temperature:

Electrolyte

94% Acetic Acid (CH_3COOH)

6% Perchloric Acid ($HClO_4$)

Jet Polishing Parameters:

Current = 0.1 to 0.13 amperes

Voltage = 40 to 60 volts

Flow Rate = 2.4

Sensitivity = as high as possible and still operate

Low Temperature:

Electrolyte

95% Methanol

5% Perchloric Acid (HClO_4)

Jet Polishing Parameters:

Current = 45 milliamperes

Voltage = 8 volts

Temperature = -40 to -50°C

Flow Rate = 3.5

Sensitivity = as high as possible and still operate

The low temperature apparatus was developed in order to provide better consistency and larger thin sections than for the room temperature polishing. A 20 litre liquid Nitrogen container was fitted with a top which was held in place by springs. The pressure of the evaporating liquid Nitrogen created pressure in the canister which forced the liquid Nitrogen up through a pipe with a flow valve and then through the electrolyte cooling coil. The valve was used to control the flow of liquid Nitrogen through the cooling coil which in turn controlled the temperature. The canister top was also fitted with a pressure gauge so that the pressure could be monitored and two relief valves for safety reasons.

The low temperature apparatus was successful in producing superior foils with more consistency than could be achieved using the room temperature apparatus.

Appendix C: Direct Current Potential Drop Technique
Theoretical Minimum Detectable Crack Area
For A One Percent Increase In Voltage

Digital Processor Resolution:

The PDP11/04 digital processor is an eleven bit machine, therefore it had a resolution of 2048 digital units (du) over a 10 volt range. With an initial value of 400 du a one percent increase over a 10 volt range represented a change in voltage, ΔV , given by:

$$\frac{\Delta V}{10} = \frac{4}{2048}$$

$$\therefore \Delta V = 19.53 \text{ mV}$$

When a gain of 850 was used this change in voltage was reduced to:

$$\Delta V = 22.98 \text{ } \mu\text{V}$$

MTS quotes an inherent noise in the acquisition, interface and processing systems of ± 3 du which was less than the one percent change if the initial value selected was greater than 300 du. The initial value was obtained by adjusting the gain to produce a signal of 400 du in order to reduce the percent error of the system. The signal was not amplified further since when the gain was increased, the error of the amplifier signal also increased.

Theoretical Minimum DCPD Crack Area and Length:

Theoretical Double Edge Notched Specimen Resistance (ΔR):

The change in resistance of a material is given by:

$$\Delta R = \frac{L}{A} d\rho + \frac{\rho}{A} dl - \frac{\rho L}{A^2} dA \quad (1)$$

where L is the shortest path between voltage probes,

A is the minimum cross sectional area of the current path,

ρ is the material resistivity,

$d\rho$ is the change in material resistivity,

dl is the change in length of the specimen between the voltage probes,

and dA is the change in minimum cross sectional area.

The temperature of the test specimen was maintained to within $\pm 1^\circ\text{C}$ in the notch region where the DCPD voltage probes were located, therefore the change in material resistivity, $d\rho$, was negligible. Since the voltage probes were located at the edges of the notch and the strain varied over the radius of the notch, it was assumed that the effect of strain on the resistance of the specimen, dl , was neglected. Therefore by substituting $d\rho \rightarrow 0$ and $dl \rightarrow 0$, into equation (1) and rearranging:

$$dA = - \frac{\Delta R A^2}{\rho L} \quad (2)$$

- i) Resistivity (ρ) is given in the Military Handbook, for Inconel 718 at 650°C as [42]:

$$\begin{aligned} \rho &= 52 \mu\Omega \text{ inch} \\ \rho &= 52 \mu\Omega \text{ inch} \times 25.4 \frac{\text{mm}}{\text{inch}} \\ \therefore \rho &= 1320 \mu\Omega \text{ mm} \end{aligned}$$

- ii) The area (A) of the smallest cross section for the double edge notched specimen is given by:

$$\begin{aligned} A &= [\text{Width} - (2 \times \text{NotchRadius})] \times \text{Thickness} \\ A &= [9.525 - (2 \times 1.191)] \times 2.3825 \\ \therefore A &= 17.018 \text{ mm}^2 \end{aligned}$$

- iii) The change in resistance (ΔR) at the one percent increase in voltage of 22.98 μV using a gain of 850 and 14 amperes is given by:

$$\begin{aligned}\Delta R &= \frac{\Delta V}{I} \\ \Delta R &= \frac{22.98 \mu V}{14 \text{ Amps}} \\ \therefore \Delta R &= 1.641 \mu \Omega\end{aligned}$$

- iv) The length of the minimum path (L) between voltage probes was the hypotenuse of the notch, where one side was the thickness and the other was the circumference of the notch (C_n) and is given by:

$$\begin{aligned}L &= \sqrt{C_n^2 + \text{Thickness}^2} \\ L &= \sqrt{(\pi r)^2 + 2.3825^2} \\ L &= \sqrt{(\pi 1.191)^2 + 2.3825^2} \\ \therefore L &= 4.436 \text{ mm}\end{aligned}$$

Substituting all the values determined above into equation (2), the theoretical minimum crack area was determined to be:

$$\begin{aligned}\Delta A &= \frac{(1.641)(17.018)^2}{(1320)(4.436)} \\ \therefore \Delta A &= 0.0812 \text{ mm}^2\end{aligned}$$

and therefore the theoretical minimum crack length was found to be:

$$\begin{aligned}a &= \frac{\text{Crack Area}}{\text{Thickness}} \\ a &= \frac{0.0812}{2.3825} \\ a &= 0.0341 \text{ mm} \\ \therefore a &= 34 \mu m\end{aligned}$$

END

28-05-92

FIN