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Damage Tolerance of Inconel 718 Turbine Disc Material

by
Min Chang
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
September, 1991
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A handwritten signature in black ink, appearing to read 'M. de la...', with a long horizontal flourish extending to the right.

Thesis Supervisor

A handwritten signature in black ink, appearing to read 'J. Smith', with a horizontal line underneath.

Chairman

Department of Mechanical
and Aerospace Engineering

Abstract

In this project, the influence of heat treatment on the microstructure and mechanical properties of Inconel 718 turbine disc material has been studied in detail. The microstructures produced by a conventional (standard) heat treatment (SHT) and a modified heat treatment (MHT) are described and their effects on low cycle fatigue (LCF), creep rupture, and fatigue crack growth rate (FCGR) properties are discussed. The modified heat treatment significantly improves the creep rupture life at 650 °C but results in a loss in low cycle fatigue life at room temperature and 525 °C. The fatigue crack propagation test results show that fatigue crack growth rates of MHT and SHT materials are very close to each other at room temperature and 525 °C.

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Nomenclature

AC: Air Cooling

ASM: American Society for Metals

ASTM: American Society for Testing and Materials

BCT: Body Centred Tetragonal

CCGR: Creep Crack Growth Rate

CRSS: Critical Resolved Shear Stress

CT: Compact Tension

FC: Furnace Cooling

FCC: Face Centred Cubic

FCGR: Fatigue Crack Growth Rate

G.B.: Grain Boundary

G.S.: Grain Size

HCP: Hexagonal Close Packed

hr: hours

k : Boltzmann's constant, 13.8×10^{-24} J/atom·K or 8.31 J/mole·K

LCF: Low Cycle Fatigue

MHT: Modified Heat Treatment

NDI: Non Destructive Inspection

N_f : number of cycles to final fracture

ppm: parts per million

P_{min} : minimum load (kN)

P_{max} : maximum load (kN)

R : stress ratio (P_{min}/P_{max})

R : gas constant, 1.987 cal/mole·K or 8.31 J/mole·K

RA: Reduction of Area

RT: Room Temperature

SII: Safe Inspection Interval

SHT: Conventional (Standard) Heat Treatment

TCP: Topologically Close Packed

T_m : melting point (Kelvin)

t_R : rupture time in creep rupture test

TTT: Time Temperature Transformation

V_f : volume fraction of precipitate

$\Delta\epsilon_T$: total strain range (%)

$\Delta\epsilon_p$: plastic strain range (%)

$\Delta\epsilon_e$: elastic strain range (%)

E : Young's modulus (MPa)

σ_{ys} : 0.2% offset yield stress (MPa)

σ_{UTS} : ultimate tensile stress (MPa)

ϵ_o : instantaneous strain in creep test (%)

ϵ_R : rupture strain in creep test (%)

σ_o : initial stress in creep test (MPa)

$\dot{\epsilon}_s$: minimum (steady state) strain rate in creep test (%/hour)

τ_c : critical resolved shear stress

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Chapter 1

Introduction

1.1 Background

Superalloys are capable of withstanding high stresses at elevated service temperature in corrosive environments. Besides these properties, reliability is another critical concern for aircraft gas turbine operation. The materials must possess high strength, excellent corrosion resistance, good creep and fatigue strength, and good crack propagation resistance.

Superalloys have been used in jet engines for critical components such as turbine discs, blades, vanes and burner cans. The service temperatures for turbine discs vary from around 150 °C at the hub to about 760 °C at the rim and during certain periods the discs may be loaded in excess of 80% of the material's yield strength. Gas inlet temperatures may exceed 1480 °C [1]¹. Coupled with high temperature gradients, load variations arising from rotational speed variations during service give rise to thermomechanical fatigue problems. Because an engine operates for considerable periods of time during which the rotational speed remains constant, the effect of creep loading is also of prime importance.

Turbine discs have always been the reliability limiting components in gas turbine engines. Progress in disc materials development has been made by designing higher strength superalloy materials and controlling the metallurgical processing techniques.

¹ Numbers in [] designate reference numbers.

Turbine discs operate at elevated temperatures, under cyclic stresses up to 1100 MPa, in an oxidizing or hot corrosion atmosphere. Inconel 718, Inco 901, A-286 and Waspaloy, are the most common disc materials in use in gas turbine engines today. Amongst these materials, Alloy 718 (also known as Inconel 718, Udimet 718, UNS N0718, Allvac 718, Pyromet Alloy 718, Haynes Alloy 718, Lescalloy 718, etc.) stands out as the most dominant alloy in superalloy production, indeed a workhorse alloy. For example, Alloy 718 comprises 51% of the total weight of the Space Shuttle Main Engine for more than 1500 individual parts operating in hydrogen, oxygen, and hydrogen-rich steam at temperatures ranging from -253 to 760 °C [2]. Currently, this grade accounts for 45% of the total production of nickel-base alloys in wrought form and 25% in cast form.

1.2 Damage Tolerant Design

The key criteria for component design are durability, safety, and economy is a major consideration. Safety is dictated by the factoring down of the critical life, i.e. by margins of safety. However, a damage tolerance approach is inherently safer since the assumption is made that there may be defects in the material and the parts are suspected to prevent the defects to grow to critical sizes. Three tests can be used to evaluate durability and safety of a disc material: Creep Crack Growth Rate (CCGR) or creep rupture life and low cycle fatigue (LCF) life for durability, Fatigue Crack Growth Rate (FCGR) for safety. Further improvement of these properties can prolong the service life of discs.

The life of compressor and turbine discs is limited by LCF because of high cyclic stresses. LCF behaviour is a complex phenomenon which is not only affected by factors

such as service conditions, but also by the statistical variation in properties of the manufactured components. This causes scatter in the fatigue performance of materials exposed to a given service environment. Premature crack initiation followed by crack propagation can cause disc failure which will have serious consequences on the operational safety of the aircraft. This provides strong motivation to quantitatively evaluate and control the risk of premature disc failure. Hence, the traditional methods used to predict the life of gas turbine engine rotor components have resulted in an overly conservative estimates of useful life [3]. In a safe life design approach, disc life limits are established on the basis of the number of LCF cycles required to form a detectable crack (0.8 mm long) typically in 1 in 1000 components [10]. At the highest stress levels the cyclic margin between the first "engineering" crack and component failure severely restricts component life. These considerations require an alternative design philosophy and a new approach to disc material technology, that is, the damage tolerant design [4,5].

In damage tolerance design, the critical fracture locations of discs are assumed to contain defects of a size corresponding to the detection limit of the non destructive inspection (NDI) technique [6,7,8] used to inspect the components. These defects are then assumed to propagate and their rates of propagation during service are established on the basis of fracture mechanics principles, using experimental crack growth rate data. The time or number of cycles to grow these inherent cracks to a predetermined dysfunction² size are then used to establish a safe inspection interval (SII) on the basis of which discs are repeatedly returned to service until a crack is eventually detected. Crack propagation is more important than crack initiation in damage tolerant design since the FCGR may

²Note: dysfunction crack size is not a critical crack size. It is a fraction of the actual critical size for a given component. A critical crack size is such a crack at which the component will fail.

be too high to obtain a practical or economical SII [9]. The most important concept here is that there are crack like defects in the components and these cracks will propagate during service. A "perfect" disc (no defects in the component) does not exist.

Presently forged Inconel 718 gas turbine discs (used for example, in J85-CAN-15 engines to power CF-5 and in G.E. F404 engines to power the CF-18) are subjected to conventional or standard heat treatment to produce a uniform fine grained microstructure together with a uniform intragranular strengthening precipitate distribution which leads to good resistance to LCF crack initiation. The standard heat treatment has proved adequate from a traditional safe life point of view. It has been argued that the conventional microstructure may not prove adequate in damage tolerant design concepts which are implemented in new engine designs.

In damage tolerance design, the crack propagation life and the SII are governed by the initial crack length, stress level, stress distribution/concentration or the geometry of the component, and above all, the microstructure of the disc material. The stress intensity factor (K) relates the crack and component geometry parameters through a relationship of the form $K = \sigma \sqrt{\pi a} f(\frac{a}{W})$, where a is the crack length, W is the width of the sample, and σ is the applied stress. The crack growth rate varying as a function of K or ΔK is strongly dependent on the material microstructure.

In the turbine disc rim area, the service stress imposed at elevated temperatures also produce time dependent straining of the component and results in creep. By definition, creep is time-dependent strain or deformation occurring under stress at elevated temperature. Over a period of time, creep terminates in fracture and this is called stress rupture. In general, creep occurs at a temperature slightly above the recrystallization

temperature of the material because they have similar mechanisms which involve atom diffusion at grain boundaries. Therefore, in addition to LCF and FCGR properties, the creep properties (creep crack growth rate CCGR and stress rupture life) must also be considered while designing disc microstructures.

1.3 Objectives of this Project

In turbine discs, damage tolerance based life prediction methods use the time for LCF cracks to grow to a dysfunction size to calculate a safe inspection interval. The fatigue crack growth rate is a critical parameter for SII prediction. A modified heat treatment (MHT) has been developed at the Structures and Materials Laboratory of the Institute for Aerospace Research (IAR) that produces a microstructure which confers higher FCGR resistance to Inconel 718 than the conventional/standard heat treatment (SHT) does for this alloy.

Relative to the conventional heat treated microstructure, this novel microstructure is successful in decreasing the fatigue crack growth rates and creep crack growth rates of the material without substantial loss in low cycle fatigue life at 650 °C. However, as noted earlier, a turbine disc is subjected to a wide range of stress and temperature histories during service. Temperature in the disc bore is very low (marginally above the room temperature) while the stresses are very high. In contrast, in the bolt hole regions the temperature may be as high as 500 - 600 °C and the stresses are considerably lower than those encountered in the bore region. In view of these observations it is important to assess the damage tolerance of Inconel 718 material subjected to the MHT, at room temperature and 500 - 550 °C.

The purpose of this research project on Inconel 718 disc material was to conduct an experimental study:

1. To examine and compare the microstructure of Inconel 718 disc material subjected to two different heat treatments, i.e. the conventional (standard) heat treatment and the modified heat treatment.
2. To compare the tensile strengths of the disc material subjected to two heat treatments (MHT and SHT) at room temperature (RT) and 525 °C.
3. To compare the LCF life of MHT and SHT material at RT and 525 °C.
4. To compare the FCGR data of the two heat treatments at RT and 525 °C.
5. To compare the effect of the two heat treatments on creep behaviours at 650 °C.
6. To discuss property differences, in 2 to 5 above, in terms of microstructures produced by SHT and MHT.

Chapter 2

Literature Review

Ni-base superalloys are compositionally complex, containing as many as a dozen elements, but microstructurally simple when compared with steels or titanium alloys. The microstructure consists of an austenitic γ matrix containing γ' and/or γ'' intermetallic precipitates, various types of carbides, borides and other secondary phases. These phases may be distributed throughout the matrix and/or located at the grain boundaries, depending on the alloy type and its processing history. Appropriate compositional and processing control result in alloys that are extremely corrosion resistant and possess excellent high temperature strength and very good creep resistance and ductility.

Nickel is the principal element (>50%) in Ni-base superalloys and some of them contain about 17 - 34% iron. In addition to chromium for corrosion/oxidation resistance, they contain aluminium and titanium to form γ' [Ni₃Al,Ti], niobium to form γ'' [Ni₃Nb] as precipitation strengtheners, which have ordered face centred cubic (FCC) and body centered tetragonal (BCT) structures respectively and are coherent with the parent FCC γ austenitic matrix. Carbon is added to form (Nb,Ti)C carbides as grain boundary strengthening phases. Substantial amounts of molybdenum, tungsten, and tantalum are introduced to promote solid solution strengthening in both matrix and second phases at both room and high temperatures. In addition to σ , μ , and Laves, geometrically close packed η [Ni₃Ti], and δ [Ni₃Nb] may form which can reduce alloy strength and cause embrittlement. Other elements such as B and Zr are added to improve properties of the

grain boundaries [1].

2.1 Inconel 718 and Its Microstructure

Inconel 718 consists of γ matrix, γ' and γ'' precipitate phases which produce coherency strains in the γ matrix, and small amounts of NbC and TiC carbides, TiN nitrides, δ , Laves, α -Cr, σ , M_6C and $M_{23}C_6$ phases. Because of controlled silicon contents and lower temperatures used for alloy homogenization, σ and M_6C are not commonly found [11]. However, depending on the material processing history, the δ phase can be present. In rare cases, complex carbides, Laves, and other phases have also been observed, but these are generally the result of poor ingot homogenization practice, abnormal heat treatment, or long time exposure at high temperature.

2.1.1 γ' and γ'' phase

Aluminium and titanium are added in amounts required to precipitate ordered FCC γ' [Ni_3Al,Ti] phase, which precipitates coherently with the FCC austenitic γ matrix. Other elements, notably niobium, tantalum, and chromium, also enter γ' . This phase is required for high-temperature strength and creep resistance.

The γ' phase is quite stable with respect to temperature and has a very unique property — positive temperature-strength relationship up to temperatures as high as 900 °C [12]. Its melting point is 1385 °C. Since the degree of order in [Ni_3Al,Ti] increases with temperature, alloys with a high volume fraction (V_f) of γ' show a remarkable increase in strength with increasing temperature up to about 800 °C [13]. With the amount of work hardening increasing with temperature a peak yield strength of 1000 - 1100 MPa,

Fig.2.1, occurs in the temperature range 700 - 900 °C depending on the alloy composition.

Higher Ti content causes FCC Ni₃Ti based γ transformation to η (HCP Ni₃Ti), such as in A-286 [12]. Since Ni has an FCC structure, Al is FCC too but Ti is HCP and as a result, Ni₃Al type of FCC γ phase is more stable than Ni₃Ti type of FCC γ . Al is the major γ forming element in Ni-base superalloys while Ti is the major γ forming element in Ni-Fe-base superalloys. The reason is related to the atomic size of Al and Ti and the lattice parameters of Ni and Fe. Atomic diameter of Al is 2.862 Å and Ti is 2.92 Å and the lattice parameter of Fe is smaller than Ni. Since Ti in Ni base γ matrix causes much larger misfit than Al (16.28% for Ti in γ matrix and only 2.48% for Al in γ matrix), substitution of Fe for Ni in γ results in the Ti based γ structure being more stable. The shape and structural stability of γ depend on the mismatch parameters. The γ contributes 10-20% to the strengthening of Inconel 718 [14].

The γ' , the major strengthening phase in Inconel 718 forms disc like precipitates with a diameter of 600 Å and thickness of 50 - 90 Å at 650 - 760 °C [15]. The γ' forms as coherent platelet in the {100} planes of γ matrix. When Nb depletion occurs due to heavy δ or γ'' precipitation, γ can form preferably in the matrix. The Body-Centred-Tetragonal (BCT) phase γ'' [Ni₃Nb] is coherent with the γ matrix and induces large mismatch strains between γ and γ'' . Unlike γ , which causes strengthening through the necessity to disorder the particles as they are sheared (high APB energy), γ'' strengthens alloys by virtue of high coherency strains (2.86%) in the lattice [16]. This phase provides high strength at low to intermediate temperatures, but is unstable at temperatures above 650 °C due to the presence of large mismatch strains. The practical use of γ' precipitation is restricted to Ni-Fe-base superalloys used below 650 °C such as Inconel

718.

The γ' and γ'' phases are metastable in Inconel 718. The γ'' is the least stable phase. The stable terminal phase in Inconel 718 is the δ phase [11]:

Both conversions are slow at low temperature but faster at high temperature. In about 1 hour $\gamma \rightarrow \eta$ transformation takes place at 788 - 815 °C and $\gamma'' \rightarrow \gamma$ transformation at 650 - 850 °C. The $\gamma'' \rightarrow \delta$ occurs at 750 - 1000 °C after long time exposure and this transformation is accelerated by deformation. It is the δ transformation, accompanied by the overaging and partial dissolution of the precipitation strengthening phases that sets the temperature-time limit for the jet engine application of Inconel 718.

The lack of notch ductility in Inconel 718 has been associated with the presence of a γ'' precipitate free zone. The γ'' precipitate free zone can be eliminated, and ductility can be restored by appropriate heat treatments. The conversion of γ'' to γ to δ causes a sharp reduction in strength [17].

In wrought Inconel 718, the γ''/γ precipitation during aging heat treatment is controlled by the temperature of hot working, the amount of Nb + Ta present in the matrix, the amount of δ phase present and the remanent forging strain after the solution treatment, but the percentage of niobium and aging temperature are the most important variables [11].

2.1.2 δ phase [18,19]

In Inconel 718 the equilibrium precipitate is incoherent orthorhombic delta (δ) $\text{Ni}_3(\text{Nb,Ti})$ phase which forms upon aging in the 760 - 982 °C temperature range and usually is most pronounced between 927 - 954 °C. The δ is brittle and needle-like in

morphology. The preferred precipitates are γ' (major) and γ (minor), which develop at temperatures below the δ formation region but which are not thermodynamically stable phases. Solution treatments above 1010 °C and aging at 871 - 982 °C causes reversion to δ [11]. Aging temperatures in most heat treatments are below the δ forming temperature. In cast Inconel 718, needle-like δ growing from the Laves phase can be dissolved after 1038 °C/10hr or 1093 °C/2hr. No δ phase remains up on solution treating at 1121 °C or higher temperatures.

This phase forms as intragranular laths along {111}, or grows into plates lying on the matrix {111} planes at grain boundaries. The δ solvus temperature is the key to grain size control. This solvus temperature is controlled by the amount of Nb present and decreases as the Nb content is decreased. So, if Nb distribution is not uniform, local grain growth will occur [11]. The commercial heat treatment for Inconel 718 consist of a solution temperature just below the δ solvus to dissolve the γ' and γ phases in order to maintain the fine grain size developed during forging.

Too much δ phase out of solution will rob the matrix of Nb, resulting in inadequate volume fraction of γ' to affect full hardening during subsequent heat treatment, particularly when using a direct aging process (wrought processing followed by aging, no solution). Creep strength is usually affected by too much of δ phase and insufficient γ' phase. If insufficient δ is present during forging, the grain boundary pinning may be inadequate resulting in coarser grains than desired. Continuous δ phase along grain boundaries and in long needle across the grain width are more deleterious in coarser grained structures due to the longer δ laths. This morphology has been shown to have an adverse effect on stress rupture ductility and fracture toughness. Smaller particles tending toward a more spheroidal form are generally preferred [11].

2.1.3 Laves

A metastable hexagonal Laves phase, $\text{Fe}_2(\text{Nb,Ti})$ may form interdendritically in Inconel 718 at 1148 - 120 °C during solidification or at 787 - 843 °C/10hr during heat treatment. The composition of Laves is not constant and can be influenced by minor elements such as silicon [15]. It can be removed by hot working above 1093 °C. Powder metallurgy processes can avoid Laves. What is important about Laves is that its formation is dependent up on large concentration of slow diffusing niobium, thus reducing the volume fraction of γ' . This makes Inconel 718 unresponsive to homogenization treatment within a reasonable time [11]. The highest strength versions of the alloy at both room temperature and 650 °C require the availability of maximum niobium content for γ' precipitation without creating the Laves phase. Laves is more resistant to solutioning compared with other phases in Inconel 718 such as δ . Too much Laves phase is not good for the properties of a material [20]. Laves islands are crack initiation sites and LCF life is impaired when Laves phase is present [18].

2.1.4 Carbides and Nitrides

Because high Nb and Ti can stabilize MC carbides and nitrides, the primary carbide and nitride in Inconel 718 is NbC and TiN respectively. NbC carbides are deleterious for mechanical properties, especially for LCF life [11]. Since Niobium forms a very stable carbide with a weight ratio of Nb:C, 93:12, small changes in the carbon activity of the alloy can greatly influence the distribution of Nb [11]. Since carbide takes niobium away, the presence of carbon decreases the strength of the material. So, a small amount of C (<0.05 wt.%) provides a limited beneficial effect as a grain boundary pinning agent. NbC occurs in this alloy in large quantity and TiN in small quantity. NbC

dissolves above 1038 °C and precipitates at 704 - 982 °C as a grain boundary film [14].

2.2 Heat Treatment of Inconel 718

Strengthening by precipitates (γ' , γ'') is related to (1) particle size r ; (2) anti-phase boundary (APB) energy; (3) stacking fault energy of γ' , γ'' ; (4) coherency strain (size misfit) between matrix and precipitates γ' , γ'' ; (5) volume fraction V_f of γ' , γ'' ; (6) γ - γ' , - γ'' modulus mismatch. These factors are influenced by heat treatment and some are controlled by the heat treatment procedure.

To some extent, a γ' (γ'') - strengthened superalloy should, after similar heat treatment, have reproducible microstructural features and mechanical properties. The heat treatment is used:

- 1, to control the grain size;
- 2, to control the size and distribution of γ' , γ'' ;
- 3, to control the carbides distribution;
- 4, to control the grain boundary morphology; and
- 5, to control other phases such as δ , Laves, etc.

Inconel 718 is normally used in the precipitation hardened condition which requires solutioning at high temperature followed by aging at intermediate temperatures. Numerous combinations of temperatures, times, and cooling rates have been used. From this point of view, the only differences between any modified heat treatment and conventional heat treatment are differences associated with solution temperature, hold time, and cooling rate.

2.2.1 Solution Treatments

Inconel 718 is usually solution treated at temperatures ranging between 927 - 1066 °C. Solutioning above 1149 °C is to be avoided because the δ phase, MC carbides, γ' and γ'' will dissolve into solution and lead to rapid grain growth and film like NbC precipitates at the grain boundaries during subsequent aging. Solutioning at 1066 °C results in lower notch strength at 650 °C. Stress rupture life is also affected. For instance, creep life at 704 °C/517 MPa is tripled (from 45 to 136 hours) by raising the solution temperature from 954 to 1066 °C. However, maximum notched stress rupture life is obtained upon solution treating at 954 °C rather than 1066 °C [21].

Solutioning at temperatures in the heterogeneous $\gamma/\gamma'/\gamma''$ phases region is done with the aim of controlling the γ' and γ'' precipitate volume fraction, size, morphology and distribution. These features are dependent on temperature, holding time, and cooling rates employed during different heat treatment stages [1]. The fine grain structure is usually produced by a partial solution heat treatment below the γ' solvus in order to retain a fraction of γ' which pins the grain boundaries. Usually the undissolved γ' , γ'' fraction is kept low since it reduces the amount of γ' , γ'' available later for precipitation strengthening. Even for the same alloy, the solution temperature changes with the percentage of certain elements [22].

The higher solution treating temperature is used for optimizing the creep and stress-rupture properties but will result in some grain growth and more extensive solution of carbides in wrought alloys. After aging, the resulting microstructure of these wrought alloys consists of large grains that contain the principal strengthening phases and a heavy concentration of carbides at the grain boundaries. The lower solution treating temperature is required for optimum tensile properties at elevated temperatures and dissolves the

principal strengthening phases without extensive grain growth or causing carbides to go into solution. An intermediate solution treatment temperature is selected to produce a compromise of properties [17,21].

Solutioning at 1066 °C and aging at 816 - 982 °C results in δ phase precipitation at the grain boundaries in a plate morphology. Solutioning at 982 °C and aging at 816 - 899 °C leads to δ platelet spherodization which suppresses grain growth. Fig.2.2 shows the effect of solution treatment temperature on strength and ductility in Inconel 718. Fig.2.3 shows the variation in strength of Inconel 718 from RT to 650 °C as a function of different heat treatment procedures [21].

Above 1030 °C grain growth will occur and will be limited only by carbide precipitation such as NbC (solvus 1040 - 1093 °C); only MC and TiN are stable at this temperature in Inconel 718. Below 1030 °C, grain growth may be inhibited by δ precipitates (their effective pinning range is 915 - 995 °C [1,12]). This is the only useful but very important purpose of δ phase especially during forging [11]; A background of overaged γ' is formed. Heating above the γ' solvus but below the δ solvus (996 °C) causes recrystallization due to the loss of γ' (dissolved), but the grain boundaries are still pinned by δ which formed during solution treatment at about 954 °C. Heating above 1030 °C permits grain growth since the pinning effect of δ is lost [12]. Care should be taken if solutioning above this temperature is used. Table 2.1 shows the effects of different solution temperatures on the tensile and stress rupture properties of Inconel 718.

2.2.2 Aging (Precipitation) Treatments

Strengthening is a consequence of the precipitates hindering the movement of dislocations within the metal. Most precipitation strengthened superalloys need an aging

cycle to strengthen the alloy but there are some alloys which are only solution treated and their aging occurs in service at elevated temperatures such as cast Hastelloy B and C.

Precipitation is a diffusional process. Higher precipitation temperatures usually are associated with a lower nucleation rate and a coarser precipitate distribution. As the aging temperature approaches the precipitate solvus temperature, the amount of precipitates present decreases rapidly. Usually the maximum hardness does not correspond to the formation of an equilibrium phase but to the metastable, transition phases, which form in a considerably finer distribution than does the equilibrium phase [21]. The higher the precipitation temperature, the lower the maximum hardness because less precipitate forms as the solvus temperature is approached. However, the higher the temperature, the higher the rate of precipitation, and hence the maximum hardness is attained in less time. Higher final aging temperature is used to obtain better creep-rupture properties due to coarse particles size. For alloys used at low to intermediate temperatures, lower final aging temperatures are employed to retain their tensile strengths [17].

Aging at temperatures above 816 °C can form a substantial amount of δ which ties up much of the available Nb in the alloy thus limiting the amount of γ' that can form. As the amount of δ increases, the strengthening of the alloy decreases due to decreasing amounts of γ' . Usually, at lower temperatures, age hardening requires a very long period (months) and a better hardening result is obtained. At higher temperature, the peak hardness is reached in a short time period (days or hours) but the peak hardness is lower than that at lower temperature. That is why some high strength alloys need double age hardening [23]. Age hardening at 718 °C for 8 hours, furnace cooling to 621 °C at about 55 °C/hr and holding for 8 hours prior to air cooling was developed by Eiselstein [11,24] and has been chosen over thirty years by many researchers as the best aging cycle for

Inconel 718. Even some other superalloys (Inconel 706, for example) are aged with this cycle:

$$718\text{ }^{\circ}\text{C}/8\text{hr}/\text{FC} \rightarrow 55.6\text{ }^{\circ}\text{C}/\text{hr} \rightarrow 621\text{ }^{\circ}\text{C}/8\text{hr}/\text{AC}.$$

This aging cycle will be referred to as "Standard age" from now on. The γ has never been observed to form before γ' precipitation because the precipitation temperature of γ' is higher than γ . Fine γ' precipitates during the first aging step and γ precipitates during the second aging step. It has been shown [21] that the best combination of properties can be obtained upon solution treating at 927 $^{\circ}\text{C}$ followed by Standard Age in sheet alloy. Heacox [25] compared the effect of different aging cycles on the properties of Inconel 718 and concluded that double aging at 788 $^{\circ}\text{C}$ + 621 $^{\circ}\text{C}$ overaged the material but produced the highest ductility and the best balance of rupture strength and rupture elongation. However, 690 $^{\circ}\text{C}$ + 621 $^{\circ}\text{C}$ yielded the highest tensile strength while 718 $^{\circ}\text{C}$ + 621 $^{\circ}\text{C}$ yielded the best combination and balance of mechanical properties. Maximum strength with a single aging step can be obtained at 704 - 760 $^{\circ}\text{C}/10 - 20\text{hr}$. Double aging increases the strength by 70 - 140 MPa and ductility is not appreciably decreased. A very low cooling rate (10 $^{\circ}\text{C}/\text{hr}$) from first aging step will produce the same effect as a double aging [14].

Actually almost every double aging procedure for superalloys consists of a high temperature aging treatment followed by a low temperature aging treatment. The reason is the balance and combination of different properties at different temperatures. Another reason for two or three step aging sequence, in addition to controlling the γ , γ' precipitates, is the need to precipitate or control grain boundary carbide morphology and

distribution [17]. The intermediate temperature aging treatment is used to modify the grain boundaries for improved creep-rupture properties. The cooling rate between aging steps is usually furnace cooling.

2.2.3 Standard (Conventional) Heat Treatment (SHT)

The following procedures are now widely standardized by most alloy producers and by ASM and ASTM specifications [26].

1. The heat treatment procedure recommended for optimum stress-rupture properties, room temperature tensile and yield strengths, and fatigue strength is:

927 - 1010 °C/(0.5-1hr)/AC or faster + Standard Age.

Test results indicate that 954 °C solution results in higher tensile strength and rupture elongation but lower tensile ductility and rupture strength than 1004 °C solution treatment [25].

2. The following heat treatment is used for optimum ductility in heavy sections, impact properties, and low-temperature toughness; however, this treatment has a tendency to produce notch brittleness in stress-rupture [26]:

solution: ***1038 - 1066 °C/(0.5-1hr)/AC or faster;***

aging: ***760 °C/10hr/FC → 649 °C/20hr/AC.***

These heat treatments result in good LCF life in Inconel 718 which is strongly influenced by the grain size of the material. Small grain size is a major factor in achieving high fatigue strength in Inconel 718. Table 2.2 lists the effect of grain size on HCF properties. The lower-temperature heat treatment schedule (1 in above) tends to result in fine grain size and optimum fatigue properties [26]. The conventional heat treatment produces a uniform grain structure with a uniform intragranular strengthening precipitate distribution

which leads to good resistance to low cycle fatigue crack initiation. Standard (conventional) heat treated Inconel 718 has proved adequate from a traditional safe life point of view. SHT alloy has a 30 - 40 μm grain size. Direct aged material can have a grain size as fine as 3 - 5 μm .

The SHT used for Inconel 718 turbine discs material comprises [27, 42]:

$$955\text{ }^{\circ}\text{C}/1\text{hr}/\text{AC} + 718\text{ }^{\circ}\text{C}/8\text{hr}/\text{FC} \rightarrow 2\text{ }^{\circ}\text{C}/\text{min} \rightarrow 621\text{ }^{\circ}\text{C}/8\text{hr}/\text{AC}.$$

This heat treatment produces a uniform grain structure together with a uniformly distributed intragranular strengthening precipitates which leads to good resistance to LCF crack initiation. It has proven adequate for optimum fatigue strength, stress-rupture properties, room temperature tensile strength and yield strength. Fig.2.4 shows the microstructure of SHT Inconel 718.

2.2.4 Modified Heat Treatment (MHT)

A series of heat treatments have recently been studied at IAR of NRC for improving creep crack growth rate (CCGR) and fatigue crack growth rate (FCGR) properties. After extensive studies, the following modified heat treatment (MHT) was found to improve CCGR and FCGR at 650 $^{\circ}\text{C}$ [10, 42]:

$$1032\text{ }^{\circ}\text{C}/1\text{hr}/\text{FC} \rightarrow 3\text{ }^{\circ}\text{C}/\text{min} \rightarrow 843\text{ }^{\circ}\text{C}/4\text{hr}/\text{AC} +$$

$$926\text{ }^{\circ}\text{C}/1\text{hr}/\text{FC} \rightarrow 3\text{ }^{\circ}\text{C}/\text{min} \rightarrow 718\text{ }^{\circ}\text{C}/8\text{hr}/\text{FC} \rightarrow 2\text{ }^{\circ}\text{C}/\text{min} \rightarrow 621\text{ }^{\circ}\text{C}/8\text{hr}/\text{AC}$$

The reason for designing this heat treatment for Inconel 718 disc material is to improve the damage tolerance capability of the discs. This microstructural design philosophy is

based on:

1. selecting a solution treatment to dissolve Ni_3Nb (δ and γ') and γ precipitates into the γ matrix without inducing excessive grain growth which lowers the LCF life;
2. developing a post solution treatment sequence to precipitate δ needles along the grain boundaries to produce zig-zag grain boundaries for grain boundary sliding resistance;
3. choosing an aging treatment to obtain the optimum γ' and γ'' precipitate distributions to strengthen the grain interiors without changing the serrated grain boundary structures.

Solutioning at 1032 °C (below the grain coarsening temperature) was considered optimum in achieving full solution of the Ni_3Nb and γ precipitates. Temperatures higher than 1032 °C cause rapid grain growth because NbC begins to dissolve. A solution treatment time of 1 hour was considered adequate for homogenizing the microstructure without causing extensive grain growth. Fig.2.5 shows the grain growth as a function of temperature and time for Inconel 718.

Furnace cooling to 843 °C from the solution treatment temperature and holding for 4 hours yields profusely serrated grain boundaries by precipitating δ phase along the boundaries with needle like morphology. The nose of the Ni_3Nb TTT (Time Temperature Transformation) curves lies in the vicinity of 843 °C in all cases. Direct aging at 818 °C formed short Ni_3Nb needles with limited serration at the grain boundaries whereas direct aging at 917 °C led to Laves phase precipitation and a longer aging time led to intragranular Ni_3Nb precipitation.

Direct aging that produces serrated grain boundaries also causes coarsening of the matrix γ'' and γ (overaging) which lowers strength and is harmful to LCF life. Therefore,

the material is re-solutioned at 926 °C to partially dissolve the coarse γ' and γ without dissolving the grain boundary Ni_3Nb . Re-solutioning at 917 °C does not completely dissolve the coarse γ' whereas 955 °C partial solutioning leads to excessive growth of Ni_3Nb needles which reduces the γ' volume fraction. Finally, the material is subjected to the Standard Age to give optimum γ' , γ precipitation. Fig.2.6 shows the microstructure of serrated grain boundaries of MHT material.

The MHT was considered promising for the combination of properties needed in gas turbine discs. This heat treatment leads to a substantial improvement in elevated temperature (650 °C) crack propagation resistance with limited loss in resistance to LCF crack initiation as compared to SHT [10].

2.3 Microstructure and Creep Resistance

Creep in superalloys is dependent on a number of microstructural parameters. Of primary importance are the γ , γ' precipitates size, lattice mismatch, size misfit, volume fraction, morphology, and grain size. Creep life commonly increases roughly linearly with the γ , γ' volume fraction, Fig.2.7. The outstanding creep resistance of conventional Ni-base superalloy is primarily a function of the precipitation of γ and/or γ' . However, when the γ/γ' solution temperature (solvus) is approached during service, the particles may grow or even go into solid solution. When this occurs, strength is rapidly lost. This limits the operational capacities of γ , γ' precipitation strengthened superalloys.

Lattice mismatch between precipitates and matrix adversely influences the creep strength. A small mismatch, can prolong creep-rupture life and no mismatch can greatly prolong it [12,23] but Inconel 718 is primarily strengthened by large coherency strains

associated with γ' precipitate which hinders dislocation motion. The γ and γ' precipitate sizes and distributions is essential parameters in the SHT and MHT Inconel 718 material. This implies that the improved creep resistance in modified heat treated Inconel 718 can be attributed to the presence of serrated grain boundaries and large grain size. Koul et al [10] concluded that grain boundary serrations improve grain boundary strength and relieve boundary stress concentrations by suppressing grain boundary sliding. Serrated grain boundaries also provide a tortuous path for a growing creep crack under intergranular cracking conditions. Lasalmonie et al [28] have also pointed out that the serrated grain boundaries are beneficial to creep crack propagation resistance.

Grain size also has an important effect on creep properties. Stroup et al [29] pointed out that a fine grain size increases rupture ductility but decreases rupture strength. Grain boundary sliding in fine grained alloys is the major cause of reduced creep strength. It is, however, difficult to separate the influence of grain size from that of other microstructural features such as texture, size and volume fraction of precipitates on creep properties. Increasing grain size accompanied by increasing γ size decreases rupture ductility; increasing γ' size or decreasing its volume fraction increases the tendency toward notch brittleness [29]. For Inconel 718 it has been suggested that the optimum particle size for creep resistance is diameter $d\gamma' = 23$ nm, $d\gamma = 14$ nm, but for ambient tensile strength, $d\gamma' = 27$ nm, $d\gamma = 15$ nm [11]. Other factors that govern creep properties include grain boundary carbide precipitation and control of deleterious phases such as δ , σ , η etc..

When creep deformation occurs by slip and climb of dislocations, the minimum creep rate is theoretically independent of grain size [28], and can be expressed as $\dot{\epsilon}_s = A \sigma^n D$ [12,16], where σ is the applied stress, D is the diffusivity, A and n are

constants. But grain size does influence the creep rate under grain boundary sliding and diffusional creep conditions [28]. At intermediate to high stress levels and above $0.5T_m$, creep deformation is believed to be controlled by the power law mechanism [30].

$D = D_0 e^{-Q/RT}$ is the diffusivity, R is the gas constant, T is the absolute temperature in Kelvin, and the activation energy $Q = 264$ kJ/mol for grain size in the range of 10 - 40 μm and $Q = 285$ kJ/mol for $d = 100$ μm [11]. These values are close to the lattice self diffusion activation energy for Ni, 280 kJ/mol (67 k cal/mol) [31]. Therefore, it is reasonable to assume that the value of the coefficient of diffusion in Inconel 718 is similar to that in nickel [12]. D_0 is a proportionality constant independent of temperature but depends on grain size and surface area, since diffusion rates at grain boundaries and surfaces are much higher because these are zones of crystal imperfection and lower packing density [31,32]. D_0 is in the range of 0.48 - 1.90 cm^2/sec [23,31,32]. The value used in [11] is

$$D = 1.98 e^{-\frac{280000}{RT}} \quad \text{cm}^2/\text{sec}.$$

The measured steady state creep rate in [11] is 0.49%/100hr under test condition of 650 °C/648 MPa, and grain size 40 μm . No numerical comparison between the measured and calculated $\dot{\epsilon}$, was made.

For most metals, when $T > 0.5T_m$, the creep activation energy is close to the diffusion activation energy. This fact is very important in understanding the creep mechanism for most alloys [20,30,31]. For example, creep rate of BCC α -Fe is found to be approximately 200 times faster than that experienced by FCC γ -Fe, since the

diffusivity of γ -Fe is only 1/350 of α -Fe. That is why most high temperature materials are FCC-base alloys. Elements like W, Mo and Cr have higher melting points, but they are not the best high temperature basic materials due to their BCC structure with a lower packing density.

It has been suggested that diffusional and power law creep are the dominant deformation mechanism in Inconel 718 in creep crack growth [11]:

$$\dot{\epsilon}_s = A \frac{D G b}{k T} \left(\frac{b}{d} \right)^m \left(\frac{\sigma_a - \sigma_o}{G} \right)^n \quad (2.3.1)$$

where b is Burgers vector, d is grain size, G is shear modulus, k is the Boltzmann constant, σ_a is the applied stress, σ_o is the back stress opposing the dislocation motion during steady state creep, $\sigma_o = 0.6 - 0.9 \sigma_a = 620 - 865$ MPa in 580 - 670 °C region [33], and A , m , n are constants which can be experimentally determined. n is about 1 for diffusional creep and 3 - 6 for dislocation power law creep. From above creep rate equation, we can see that (1) lower diffusivity is beneficial to creep resistance, so particles or large solutes can reduce diffusivity; (2) large grain size can reduce creep rate; (3) large modulus (shear modulus G or elasticity modulus E) can increase creep resistance.

In a constant load creep-rupture test the applied stress increases with time. This stress increase promotes damage in the material and more damage results in increasing the stress further. This vicious circle accelerates the creep rate during stage III. Little is discussed in detail about the mechanisms of the transformation from steady state creep (stage II) to accelerating creep (stage III).

The fractographic characteristics of creep depend on the applied load and test temperature. At low temperatures and high stresses (short creep life), transgranular fracture almost always occurs, as in the normal tensile test. At high temperatures ($>0.5T_m$) and low stresses (long creep life), intergranular failure occurs. The microstructure of specimens or components fractured under these test conditions is characterized by the presence of numerous pores and incipient crack nucleation that lead to microscopic crack growth and final rupture.

During creep, voids develop on the grain boundaries parallel to the stress direction, while solute atoms move to the grain boundary which is perpendicular to the stress direction, resulting in necking [34]. Two types of cracks are observed in creep tests, which are schematically shown in Fig.2.8:

1. Wedge-shaped cracks at grain boundary triple points, which can be attributed to stress concentrations arising from grain boundary sliding. They are favoured at high stress in relatively short term creep test at low temperatures;
2. Round creep cavities form on grain boundaries perpendicular to the stress axis which eventually coalesce. They usually favour at low stress in long term tests at high temperatures [34].

Development of creep cavities is a heterogenous process, thus, cavities coalesce locally to form micro-cracks long before final fracture. The grain boundary creep fracture process involves the nucleation, growth, and coalescence of grain boundary cavities. Usually cavities initiate at inhomogeneities at the grain boundary such as jogs and precipitates, or at grain boundary triple-points. Grain boundary sliding, grain boundary diffusion, surface diffusion, and power-law creep ($\dot{\epsilon}_c \propto \sigma^5$ [30]) are all thought to play a role in cavity nucleation and the subsequent growth of cavities [35].

2.4 Factors Influencing the LCF and FCGR Properties

2.4.1 The influence of Strain Range on LCF

Fatigue life may be divided into crack initiation and propagation phases since the kinetic properties of these two phases are known to be different from LCF and FCGR tests. Pure fatigue is produced by time-independent plastic deformation. However, superalloys are used at high temperatures in aggressive environments, and damage is often due to the combined effects of fatigue, creep, and environment. If other factors are equal, fatigue is governed by the plastic true strain range $\Delta\epsilon_p$. A number of methods describing the relationship between strain range and LCF life can be used to predict the LCF life [36]. The Manson - Coffin (M - C) equation is often employed and defined as [12]:

$$\frac{\Delta\epsilon_p}{2} = \epsilon_f (2N_f)^c \quad (2.4.1)$$

Where $\Delta\epsilon_p/2$ is the true plastic strain amplitude, ϵ_f is the fatigue ductility coefficient proportional to the fracture ductility in monotonic tension defined by the strain intercept at $2N_f = 1$, $2N_f$ is the number of load reversals to failure, and c is the fatigue ductility exponent, a material property varying between -0.5 to - 0.7 for many metals and alloys; for superalloys, more negative values are often reported. On a log-log scale, this equation gives a straight line when $\Delta\epsilon_p \gg \Delta\epsilon_e$ or N_f is small, say $N_f < 1000$, where plastic strain is predominant.

The application of M-C equation is difficult because plastic strain alone cannot be obtained from a component but total strain can be obtained by stress analysis. The empirical relationship between true elastic strain (or true stress amplitude) and the number

of cycles to failure N_f is given by [30]

$$\frac{\Delta \epsilon_e}{2} = \frac{\Delta \sigma}{2E} = \frac{\sigma_f (2N_f)^b}{E} \quad (2.4.2)$$

Where $\Delta \epsilon_e/2$ is the true elastic strain amplitude; $\Delta \sigma$ is the true stress range; σ_f is the fatigue strength coefficient related to the true fracture stress in monotonic tensile test, defined by the stress intercept at $2N_f = 1$; b is the fatigue strength exponent; On log-log scale, it also yields a straight line which holds only when $\Delta \epsilon_e \gg \Delta \epsilon_p$ or plastic strain is small or N_f large, for instance, $N_f > 10^4$.

The fatigue resistance of a material subjected to a given total strain range could be estimated by superposition of the elastic and plastic strain components, then the total strain amplitude is

$$\frac{\Delta \epsilon_T}{2} = \frac{\Delta \epsilon_e}{2} + \frac{\Delta \epsilon_p}{2} = \frac{\sigma_f (2N_f)^b}{E} + \epsilon_f (2N_f)^c \quad (2.4.3)$$

A simplified equation is Manson's "universal slopes" equation [37,38]:

$$\Delta \epsilon_T = D^{0.6} N_f^{-0.6} + 3.5 \frac{\sigma_{UTS}}{E} N_f^{-0.12} \quad (2.4.4)$$

where D is the ductility in a tensile test (true fracture strain). This equation relates the fatigue life to the ambient temperature tensile properties and can be used for a lot of materials. It is convenient for quick primary estimation of fatigue life of a material.

Usually equation 2.4.3 and 2.4.4 are used for fatigue life prediction at room temperature and symmetric (tension strain equal to compression strain) continuous (no

hold time) cyclic fatigue. As for high temperature, these equations result in over prediction of LCF life by using high temperature tensile properties [38]. The actual life is lower than expected because at high temperatures, oxidation and cyclic creep attack the grain boundaries. Short time tensile properties are unaffected by these factors and hence are no longer directly appropriate for use in estimating cyclic life time. At very high temperature, however, oxidation may increase fatigue life by crack tip blunting [12]. A number of methods have been developed for high temperature LCF life prediction such as the SRP (strain range partition) method [38] and the four point method [37].

In the regime where $\Delta\epsilon_T$ is mostly elastic, fatigue life increases roughly with increasing tensile strength (σ_f and σ_{UTS} in above equations); in the regime of large $\Delta\epsilon_p$, fatigue life increases roughly with increasing tensile ductility (ϵ_f and D). These are rough correlations only because fatigue behaviour does not simply reflect tensile behaviour. Tensile behaviour is much less sensitive to material defects and does not reflect the cyclic and time-dependent damage mechanism present during high temperature fatigue [12].

The cyclic stress-strain curve of polycrystalline materials from the onset of yielding to the maximum load is approximated by a straight line on log – log scale, which is similar to the Holloman relation in a tensile test [12,39]:

$$\Delta\sigma = A(\Delta\epsilon_p)^n \quad (2.4.5)$$

A and n (0.05 - 0.5 for most materials) are constants, for superalloys $n \approx 0.1$ at low and intermediate temperatures. The low value of n for precipitation-hardened, planar slip materials reflects the easily reversed deformation in softened slip bands. More homogeneous deformation at high temperatures is reflected in increased values of n for superalloys, to as high as about 0.3. The relation between the constants of the above

equations are [39]:

$$b = n \cdot c, \quad A = \frac{2\sigma_f}{(2\varepsilon_f)^n} \quad (2.4.6)$$

2.4.2 The Influence of Frequency on LCF Life

Although the loading rate (strain rate $\dot{\varepsilon}$) is a meaningful parameter when considering the physical behaviour of a material, from an engineering viewpoint the frequency appears to be more meaningful since the most important concern is generally how the structure will behave under cyclic operation where the strain rate is complex; in other words, frequency includes various strain rates. It is believed that the frequency influence is not negligible at elevated temperature [11,40].

Manson-Coffin equation is usually suitable for low temperature. A frequency modified cyclic stress-strain relationship has been developed by Coffin [40] where

$$\Delta\sigma = A(\Delta\varepsilon_p)^{n'} v^i \quad (2.4.7)$$

to modify the M-C equation for high temperature application:

$$\Delta\varepsilon_p = C_2(N_f v^{k-1})^{-\beta} \quad (2.4.8)$$

where v is the frequency, A , n' , i , C_2 , β , k are coefficients determined from test data by regression analysis, and n' is the cyclic strain hardening exponent. A , n' , and β increase with increasing temperature but i decreases. The $\Delta\sigma v^i$ term is called the frequency modified stress range. k is a frequency related constant. When $k = 1$ this equation reduces to the M-C equation 2.4.1 at ambient temperature. When $k = 0$, then v^{-1} is the period, N_f

$v^1 = t_f$, the fracture time, and the equation turns into a creep equation. So when both fatigue and creep mechanisms are present, the range of k is $0 < k < 1$. But one should note $\Delta \epsilon_p \propto t_f$ in creep and $\Delta \epsilon_p \propto 1/N_f$ in fatigue.

From the stress-strain hysteresis loop we can get the elastic strain range:

$$\Delta \epsilon_e = \frac{\Delta \sigma}{E} = B N_f^{-\beta'} v^j \quad (2.4.9)$$

where $B = AC_2^{n'}/E$, $\beta' = \beta \cdot n'$, $j = -\beta n'(k - 1) + i$. Then the total strain range now can be found as:

$$\Delta \epsilon_T = \Delta \epsilon_p + \Delta \epsilon_e = C_2 (N_f v^{k-1})^{-\beta} + B N_f^{-\beta'} v^j \quad (2.4.10)$$

letting $k = 1$, $i = 0$, $\beta = 0.6$, $n' = 0.2$, $C_2 = D^{0.6}$, and $A = 3.5 \sigma_{U19}/D^{0.12}$, this equation transforms to the Manson's "universal slopes" equation 2.4.4. From this comparison we can get some feel for the coefficients in the above equation.

2.4.3 The Influence of Temperature and Microstructure on LCF

The effect of increasing temperature on fatigue life for a given $\Delta \epsilon_p$ is not always to decrease it. Usually at elevated temperatures, a material has a minimum fatigue life. In vacuum this is not evident, but in an air environment the fatigue life decreases with increasing temperature from room temperature to some minimum (in air) and after the minimum point the fatigue life increases with increasing temperature under the same test condition of plastic strain range and cyclic frequency [12,39]. Around the minimum point either the decrease or increase is very rapid. This trend is not obvious in vacuum. The curve is similar to the curve of elongation versus temperature for most superalloys, Fig.2.9.

This phenomena was also mentioned in a paper of LCF behaviour of Inconel 718

by Fournier and Pineau [41], in which it was observed that the decrease in fatigue life with increasing temperature occurs rapidly around 550 °C and seems to slow down above 600 °C. By comparing their data with the results in reference [42] at 650 °C, it is shown that from RT to 500 °C the loss of life is not dramatic, but from 500 to 600 °C it drops rapidly, after 600 °C the rate of reduction slows down. It seems that the minimum in fatigue life of Inconel 718 occurs above 650 °C. It should be noted that E and σ_y will decrease significantly as the temperature is increased above 650 °C. It has, therefore, been concluded that 600 °C represents the upper temperature limit for application of Inconel 718 [43].

Fatigue strength decreases with increasing grain size [26]. Direct aged (Standard Age after hot rolling) material has the highest fatigue strength compared to conventional heat treatment. Fine grains can delay crack nucleation and increase ductility by reducing the slip length and therefore the stress concentration and promoting homogeneous deformation [28]. On the contrary, fatigue crack propagation is faster in fine grained materials. Au et al [42] compared the effect of grain size and grain boundaries on LCF of Inconel 718 and concluded that the serrated grain boundary morphology has little influence on LCF resistance. After two heat treatments which produced planar grain boundaries with grain size of ASTM #5 and serrated grain boundaries with grain size of ASTM #4, the LCF life of serrated grain boundary material was only marginally better than its planar grain boundary counterpart.

It is pointed out that when r_p/d (r_p is the radius of plastic deformation zone, d is grain diameter) is small (< 1), cracks propagate along grain boundaries; when r_p/d is large (> 1), transgranular cracking is the predominant mode [44]. Hertzberg also predicted an intergranular crack path at lower K level and transgranular fracture at intermediate to high

K level [30]. For fine grain size ($r_y > d$), dislocations are trapped by grain boundaries and the pile-ups cause an accumulation of damage during cyclic loading [28]. This will cause intergranular cracking when the plastic zone becomes large enough, i.e. crack length is relatively long. For large grain size and short crack length, the stress concentration at crack tip is high and slip occurs locally which will not spread to grain boundaries and as a result, crack initiation and propagation is straight and transgranular.

2.5 The Effects of Environment and Microstructure on FCGR

Fatigue crack growth rate da/dN is expressed as a function of the crack tip stress intensity factor range ΔK and characterizes a material's resistance to stable crack propagation under cyclic loading. An expression often used is Paris equation:

$$\frac{da}{dN} = C \Delta K^n \quad (2.5.1)$$

The theoretical analysis of FCGR data is based on linear elastic fracture mechanics (LEFM). The FCGR under constant-load-amplitude is primarily a function of ΔK (or K_{max}), load ratio R , temperature, environment (especially at elevated temperature). Other loading variables, such as cyclic frequency, waveform (hold time), heat treatment, product form, microstructure, thickness, heat to heat variation, even different environments have effects on FCGR.

Because of the scatter of the data, an upper bound line of the Paris equation is sometimes used for comparison between test results. The linear regression of the coefficients (constant C and slope n) in Paris equation is

$$\log\left(\frac{da}{dN}\right) = \log C + n \log(\Delta K) \quad (2.5.2)$$

The upper and lower bound lines then are

$$\log\left(\frac{da}{dN}\right) = \log C + n \log(\Delta K) \pm 3s \quad (2.5.3)$$

where s is the standard deviation and $3s$ is usually selected as a confidence of 99.7% (when enough data points are sampled). It means that 99.7% of the data are within the band between upper and lower bound lines.

Expressing da/dN as a function of ΔK provides results that are independent of planar geometry (but not thickness), thus the comparison of data obtained from a variety of specimen configurations and loading conditions can be made. Moreover, this feature enables da/dN versus ΔK data to be utilized in the design and evaluation of engineering structures.

Temperature is the most important test parameter. The effects of other factors such as microstructure, stress ratio R and waveform etc. are strongly dependent on temperature. FCGR increases with increasing temperature in air environments but the slope n in Paris equation decreases slightly and the coefficient C increases greatly with increasing temperature. This means that the FCGR at low ΔK level (near threshold) is accelerated more than at high ΔK level at elevated temperature. This is consistent with the conclusion reached by Clavel and Pineau [45] that increasing temperature accelerates FCGR, particularly at low ΔK level. At high ΔK level and elevated temperature, FCGR is increased by decreasing frequency. Test results obtained by James and Mills [46] on Inconel 718 also show that the slope n decreases as temperature increases from 5.2 at 24

°C to 1.9 at 650 °C and the constant C increases from about 10^{-12} to 10^{-7} . Other data [11] also shows that the n value reduced from about 4 at RT to 3 at 649 °C even other material has the same behaviour trend [40]. The explanation for this trend is that C and n are functions of temperature and n decreases with increasing temperature [39].

Thamburaj et al [47] compared the FCGR at 650 °C in vacuum and in air. The FCGR in vacuum is much slower (3 - 6 times) than in air. This indicated that the elevated temperature FCGR resistance is influenced primarily by oxidation. The CCGR (creep crack growth rate) in vacuum is also much slower than in air (10 - 100 times). Shahinian and Sadananda [48] also compared the FCGR of Inconel 718 in vacuum and air, and concluded that FCGR at 538 °C in vacuum is essentially cycle-dependent, i.e. unaffected by hold time, but in air is both cycle-dependent and time dependent, as a result of an environmental effect. At 650 °C the FCGR is predominantly time-dependent, in vacuum due to creep and in air due to both creep and environmental effects. As temperature and hold time increase (or frequency decreases), the creep effect becomes more pronounced. It was also pointed out that at elevated temperature creep mainly contributes at high ΔK values to FCGR while environmental effect mainly contributes at lower ΔK levels. An increase in temperature from 25 to 650 °C produces a reduction in LCF life and a significant increase in FCGR, especially in the threshold regime.

Clavel and Pineau [45] pointed out, that the temperature mainly has an effect in the slow crack growth regime since the "threshold" is decreased from about $14 \text{ MPa}\sqrt{\text{m}}$ at 25 °C to approximately $8 \text{ MPa}\sqrt{\text{m}}$ at 550 °C. In the higher crack growth regime ($da/dN > 10^{-5} \text{ mm/cycle}$) the temperature has only a slight effect and the value of n in Paris expression is close to 3.5.

Frequency effects are most pronounced at elevated temperatures and in air

environments. FCGR increases with decreasing frequency at elevated temperature. Actually, at high temperature, the effect of frequency is time dependent and cycle dependent crack propagation. At very low frequency (<0.01 Hz), intergranular cracking is dominant; at very high frequency (>100 Hz), transgranular (striation formation) crack propagation is dominant. The range of $0.01 < f < 100$, is the mixed region, where the transgranular and intergranular crack propagation occurs. Pedron and Pineau [49] concluded that first, at high frequency (say 20 Hz) there is no pronounced effect of microstructure on FCGR; secondly, a decrease in test frequency leads to an increase in FCGR; and thirdly, when a hold time is included in the fatigue cycling, the FCGR is greatly increased.

FCGR increases with increasing hold-time. This effect is closely related to temperature and is related to superposition effects of creep-fatigue mechanisms. Andrieu, Cozar, and Pineau [11] examined the effect of frequency, waveform (or hold time) and concluded that these factors have little influence on FCGR at 550 °C, but at 650 °C high frequency results in the lowest FCGR, low frequency produces intermediate FCGR, and low frequency with hold time causes the fastest FCGR. James [11] pointed out that grain size has only a moderate effect on FCGR behaviour during continuous cycling, but it has a profound effect under hold-time conditions at elevated temperatures. Grain size strongly influences the creep properties and LCF resistance. But Sims et al [12] pointed out that in the intermediate growth rate regime, $da/dN \propto d^k$, where d is grain size and k is a constant from -1 to -0.5 . Pedron and Pineau [49] found that fine grained (ASTM #6-8) Inconel 718 with β (Ni_3Nb) on grain boundaries reduces FCGR at 650 °C. Other results

[46] have also shown very little effect of grain size (ASTM #4 in CHT³ and ASTM #11.5 in INEL⁴) on FCGR from 24 to 649 °C.

The stress ratio R ($R = P_{min}/P_{max} = K_{min}/K_{max}$) has some effects on FCGR through its influence on ΔK :

$$\Delta K_{eff} = K_{max}(1 - R)^m \quad (2.5.4)$$

where $m = 0.6 - 0.7$ and decreases slightly with increasing temperature [11]. Substituting this effective stress intensity factor range in Paris expression

$$\frac{da}{dN} = C(\Delta K_{eff})^n \quad (2.5.5)$$

This modified Paris equation can also be written as

$$\frac{da}{dN} = C \left[\frac{\Delta K}{(1 - R)^{1-m}} \right]^n \quad (2.5.6)$$

All the parameters m , n and C are determined empirically from test data. In the high ΔK region, Forman proposed [36]

$$\frac{da}{dN} = \frac{A(\Delta K)^n}{(1-R)K_c - \Delta K} \quad (2.5.7)$$

where K_c is the critical stress intensity factor at which fracture occurs. James [11] summarizes many test results and concluded that generally, on the basis of ΔK , FCGR increases with increasing R . Increasing R means increasing average stress amplitude, in

³ CHT (conventional heat treatment) = SHT [37,111]:
955C/1hr/AC + 718C/8hr/FC → 55C/hr → 621C/8hr/AC.

⁴ INEL (Idaho National Engineering Laboratory) heat treatment [3,80,81,111]:
1093C/1hr/FC → 55C/hr → 718C/4hr/FC → 55C/hr → 621C/16hr/AC

other words, increasing the hold time and hold stress level. Mathematically, $-1 \leq R \leq +1$, but when $R = 1$, the test turns to be a pure tensile or creep test because of $P_{\max} = P_{\min}$. The above conclusion will not hold under these extreme situations because the fracture mechanism changes. Koul et al's results [10] show that at the same stress intensity factor level, CCGR is lower than FCGR due to the absence of cycling under creep.

As mentioned earlier, large grain size, grain boundary carbides, eliminating of Laves and δ phases, can all help to improve FCGR but very low FCGR has also been associated with a "necklace" microstructure where large grains are surrounded by necklaces of very small grains [11]. Coarse grained material results in much rougher fracture surface which needs more energy to propagate compared with a fine grained material. Pedron and Pineau [49] compared the FCGRs of the fine grained, coarse grained, and "necklace" structure of Inconel 718 material and found that the "necklace" structure exhibits the most attractive FCGR properties at elevated temperature whereas the coarse grained material (ASTM #3-4) is slightly superior to the fine grained structure (ASTM #6-8). It was suggested that in the finer grained material the grain boundary area exposed to oxygen penetration is larger than in coarse grained structure and the air environment attack is the essential factor in controlling FCGRs at elevated temperature. Koul et al [10] showed the FCGR comparison of SHT and MHT material at 649 °C. The MHT doubled the grain size and formed serrated grain boundaries. FCGR was reduced by a factor of 2 in MHT. All these results seem to indicate that a coarse grain size and serrated grain boundaries can increase the FCGR resistance at elevated temperature.

Chapter 3

Experiment and Material

3.1 Chemical Composition of Inconel 718

The nominal composition of Inconel 718 turbine disc material is given below (wt.%) [26]:

Cr	Fe	Nb+Ta	Mo	Ti	Al	C	B	Ni
18.1	17.5	5.35	3.0	0.95	0.5	0.3	0.03max	balance.

The forms and compositions of the materials used in the present study are listed in Table 3.1. The composition of turbine discs in the table is just for comparison.

Three types of tests were conducted for comparing the effects of the SHT and MHT on material properties: creep rupture tests; low cycle fatigue (LCF) tests; and fatigue crack growth rate (FCGR) tests.

3.2 Creep Testing

Creep test specimens were cut from stock plate material with the axis of the specimen oriented in the rolling direction. Fig.3.1 shows the dimensions of the specimen

conforming to ASTM E-8 specifications. The tests were conducted on a 9080 kg Applied Test System Inc. (ATS) load frame fitted with a linear variable displacement transducer (LVDT) monitor. The creep furnace shuts off automatically when the specimen breaks. A thermocouple was attached to the specimen gauge length and another thermocouple was attached to the furnace wall for control of the temperature. Error between the two thermocouples was about ± 1 °C at steady state.

Creep-rupture tests on SHT and MHT materials were conducted at 650 °C in laboratory air. The axial creep strain was recorded with the help of an extensometer attached to the specimen shoulders (notches). The extensometer is connected to the LVDT and the data was recorded digitally as a function of time. The sampling interval was set at one hour. The gauge length (initially 28.575 mm) includes the fillets. The gauge length changes during the test and equals to the initial length plus the elongation of the specimen. The LVDT was calibrated at

$$\Delta l = 2 \text{ mm} = 7.00 \text{ V.}$$

When the specimen length change is <10%, there is little difference between the engineering strain and true strain [50]. Then, the engineering strain at the calibrated point is

$$2 \text{ mm}/28.575 \text{ mm} = 0.06999 \approx 0.07 = 7\% = 7 \text{ V.}$$

If the strain-voltage relationship is considered linear, then 1.0 V $\approx$ 1.0% engineering strain. Fig.3.2 shows the linearity of the transducer. Its maximum deviation in 0 – 2.5 mm is 0.006 mm (0.19 V). The test is under constant load of 763 kg and the initial stress is

593 MPa (86 ksi). This was chosen to yield approximately a 500 hour rupture life on SHT Inconel 718 material. The reduction of cross section area was calculated according to ASTM E-139 (9.8.2 and 10.4): $RA = (A_b - A_a)/A_b$, where A_a is the cross section area after testing and A_b is the cross section area before testing.

3.3 LCF Testing

The LCF specimen used this test was the standard ASTM E-606 axial fatigue specimen, Fig.3.3, cut from hot rolled bar (diameter = 22 mm). The test section of each specimen was polished manually in the axial direction with successively finer grit emery papers (grade 320, 400, 600) to remove circumferential machine marks. The tests were conducted in tension - compression ($R = -1$) under total axial strain control, using a fully reversed, constant amplitude, triangular waveform and a constant strain rate of 0.2%/sec. in laboratory air. The total strain range varied between 0.65 - 2%. The test equipment was a MTS closed loop electro-hydraulic testing system. The specimen gage length was heated by a high frequency induction coil and the test temperature was monitored and controlled by an infrared radiation pyrometer. The error in temperature control is less than 2% if the requirements of the temperature control system are satisfied (specimen surface oxidation getting relatively stable). Test were conducted at room temperature and 525 °C.

During the constant amplitude strain controlled LCF tests, the cyclic stress response was obtained from a series of load-strain hysteresis loops recorded by an X-Y recorder. The crack initiation in each case was defined by the number of cycles for a 5% drop in the steady state tensile stress value.

All LCF tests were started in tension. The maximum tensile stress at $N = 1/4$ (the first quarter of the first cycle) in the test (initial tensile loading) is equivalent to a simple tension test. Therefore, the monotonic stress-strain curve of a material can be constructed from its strain controlled LCF test results by plotting the maximum tensile stress and applied tensile strain values recorded at $N = 1/4$ for all strain ranges considered.

Similarly, the cyclic stress-strain curve was obtained. Since the tests were conducted at a constant total strain range and the cyclic stress continually varies due to hardening and/or softening; the cyclic plastic strain also continually varies; the values of $\Delta\epsilon_p/2$ and $\Delta\sigma/2$ or σ were taken at half life $N/2$, in conformity with the procedure followed in the literature and for comparison with other worker's results. The half life is a good first approximation of the steady state value since sometimes there is no steady stress during the test. Details of LCF tests are given in [51].

3.4 FCGR Testing

Twenty two compact tension (CT) specimens, Fig.3.4, were cut from the stock plate ($12.7 \times 76.2 \times 2000$ mm) for FCGR testing. The notch was oriented parallel to the rolling direction (T-L position), that is, the cracking plane was kept parallel to the rolling direction. The test parameters were selected with regard to previous tests undertaken at IAR, available equipment in the laboratory and ASTM E-647 specifications. The specimens used conformed to a standard size with $W = 50.8$ mm (2.0 in.) in width and $B = 12.7$ mm (0.5 in.) in thickness as required in ASTM E-647 specifications.

Test Conditions

All CT specimens were precracked at room temperature at $f = 2 - 5$ Hz and the precrack length a_0 was > 3.17 mm. All FCGR tests were conducted under the following conditions:

Test Frequency	$f = 1.0$ Hz;
Maximum Load	$P_{max} = 13$ kN;
Stress Ratio	$R = P_{min}/P_{max} = K_{min}/K_{max} = 0.1$;
Environment	Lab. Air;
Wave Form	Sawtooth;
Test Temperature	RT and 525 °C;

The maximum load was chosen to yield ΔK values in the range of 20 - 100 MPa $\sqrt{\text{m}}$. At these ΔK values, the size of the plastic zone at the crack tip was presumed small compared to the crack length, therefore, LEFM (Linear Elastic Fracture Mechanics) was applied conforming to ASTM E-647 specifications.

Crack Monitoring

A direct current potential drop (DC-PD) technique was used to monitor crack length. Two optical tele-microscopes were also used to measure the crack length on both sides of the specimen when tests were conducted at RT. At 525 °C, the DC-PD technique was used to measure the crack growth. The calibration of the relationship between the potential drop (voltage across the crack tip) and the crack length is critical for the measurements, particularly at high temperatures. It has been shown elsewhere [52] that the calibration curve established at room temperature (RT) can be used at elevated temperature.

A reference CT specimen of the same material and configuration was connected in series with the test specimen in order to eliminate the effects of variations in test temperature, current and material etc. A voltage ratio parameter, β (or VR), which is thus independent of these variables, was used to calibrate the increase in the voltage cross the crack tip with increasing non-dimensional crack length a/W , where a is the crack length and W is the specimen width. This parameter is defined as:

$$\beta = VR = \frac{(V_{on} - V_{off})_{test\ specimen}}{(V_{on} - V_{off})_{reference\ specimen}} \quad (3.4.1)$$

where V_{on} and V_{off} are the potential drop when the pulsed current is on and off respectively (sampling period is 0.5 second). A pulsed current was used to minimize the heating of the current leads ($I = 9 - 10$ A). Although the D.C. current is large, the current density is low because of the relatively large specimen size. Low current density produces a very small potential drop on the crack tip: 4 - 5 mV at large crack length (before fracture) and 1 - 2 mV during crack initiation and early propagation stage. These are small input signals for the amplifiers and a high gain was set to amplify the signals which then are sent to the computer to calculate the crack length. The zero point drifting (V_{off}) and the gain drifting (V_{on}) of the two amplifiers (one for test specimen and one for the reference specimen) may not be equivalent. These variations result in scatter in the voltage ratio β which results in scatter in the crack lengths because of the relatively small input signals. Under normal conditions, the DC-PD measured crack length is within $\pm 2\%$. This scatter is allowable by ASTM E-647. Fig.3.5 shows the calibration curve between the optically measured crack length a/W and the voltage ratio VR (β). Fig.3.6 shows the difference

between the optical and DC-PD crack length. Other notable problems are system grounding and signal cable shielding. The calibration curve is mainly used at 525 °C.

The expression for the calibration curve was obtained by fitting a second order polynomial through the data of optically measured crack length (precision = 0.01 mm) a/W and DC-PD measured voltage ratio β (VR):

$$a/W = x_1 + x_2 \beta + x_3 \beta^2 \quad (3.4.2)$$

where x_1 , x_2 , x_3 are constants determined by polynomial regression. The calibrated curve for this test was found to be (Fig.3.5a):

$$a/W = -0.37026 + 0.65941 \beta - 0.08212 \beta^2 \quad (3.4.3)$$

3.5 Optical and Electron Metallography

The microstructure of the material and fracture surfaces analysis were carried out on both optical (Olympus PGM3 and Leitz Wetzlar) and scanning electron (SEM, Novascan 30) microscopes.

Samples for optical metallography were prepared using standard metallographic techniques to reveal grain boundaries, precipitates, and carbides. They were cut from as received stock material (12.7 mm plate), SHT, and MHT materials for the microstructures comparison. After mounting in a mould with bakelite under 4200 psi and heated to 150

°C for 4 - 5 minutes then cooled to at least below 90 °C, the samples were ground through grade 300, 400, and 600 SiC sand papers and then polished on a polish cloth wheel through 9, 3, and 1 μm diamond paste. Thorough ultrasonic cleaning was carried out between each steps. After checking if there is any scratch still on the surface under low magnification optical microscope, the samples were then etched in a solution of 40 ml distilled water, 20 ml hydrogen peroxide (H_2O_2), and 60 ml hydrochloric acid (HCl). This solution is quite effective at revealing microstructure in Inconel 718. The solution was mixed in the above order and boiled for several minutes before use. With this solution etching time is of the order of seconds. Under-etched surface can be etched again but over-etched surface must be repolished at least from grade 600 SiC sand paper. Optical metallography was used for determination of grain size and general comparison of grain boundaries among as received, SHT, and MHT materials. The grain size was determined in accordance with the ASTM E112-84 standard practices. Polished and etched samples were carbon coated for SEM metallography used for more in-depth analysis of the microstructure and to establish composition of the precipitates.

Samples for fracture surface analysis under SEM were cut from tested specimens (CT specimens, LCF specimens, and creep rupture specimens) and were ultrasonically cleaned in acetone. They were then mounted on aluminum holders ensuring that the specimens were electrically conductive. For specimens tested at room temperatures, the fracture surfaces can be directly examined under SEM. Gold coating may be needed if the sample is tested at elevated temperature because the fracture surface is oxidized. Oxides have the tendency to charge up in the microscope which interferes with SEM imaging.

Chapter 4

Results and Discussion

The modified heat treatment (MHT) for Inconel 718 was developed to reduce CCGR and FCGR without substantial loss in LCF life and other mechanical properties relative to standard heat treatment (SHT), especially at elevated temperatures. This chapter will concentrate on the creep, FCGR and LCF results obtained on the two microstructures.

4.1 Basic Mechanical Properties

Hardness: the average hardness of Inconel 718 subjected to the two heat treatments at RT in terms of Rockwell hardness number was measured to be:

SHT 46.5 Rc

MHT 44.5 Rc

The MHT has no significant effect on hardness of the material.

Modulus of elasticity E : E was recorded in the first cycle during the LCF test. The average E is:

at RT: SHT 198×10^3 MPa

MHT 199×10^3 MPa

at 525 C: SHT 173×10^3 MPa

MHT 172×10^3 MPa

Clearly, the MHT does not change the modulus of elasticity of Inconel

718. The E values are the same as those given in the literature.

Yield stress σ_y : the yield stress is measured as the offset strength at 0.2% plastic deformation in the first cycle of the LCF test:

at RT: SHT 1204 MPa

MHT 1036 MPa

at 525 °C: SHT 1066 MPa

MHT 876 MPa

The yield strength is reduced by 14% at RT and 18% at 525 °C for the MHT material. This decrease results from the grain size increase (about ASTM #7 or $\sim 30 \mu\text{m}$ for SHT and ASTM #4 or $\sim 60 \mu\text{m}$ for MHT) since Hall-Petch equation indicates that metals with finer grain size have higher yield strength at temperatures below $0.5T_m$. The yield strength of SHT material at RT and 525 °C are the same as those given in reference [26].

4.2 Creep Behaviour

The creep results are tabulated in Table 4.1. The modified heat treatment almost doubled the stress rupture life and the rupture strain of MHT material is smaller than that of the SHT material. Fig.4.1 compares the creep curves of the SHT and MHT materials at 650 °C. These curves clearly indicate that the modified heat treatment extends the creep life of Inconel 718 material.

At the beginning, elastic and perhaps some plastic deformation occur and this is followed by a period of transient creep during which the creep rate decreases due to flow hardening. The strain rate decreases with time to a steady-state value ($\dot{\epsilon}_s$) that usually

persists for a substantial portion of the material's life. The decrease in strain rate is related to substructure changes that increase the overall resistance to dislocation motion. This period was only of the order of about 1 hour in both SHT and MHT materials, Fig.4.1. The constant (or minimum or steady state) strain rate creep persisted for about 150 hours for SHT and 300 - 400 hours for MHT materials, Fig.4.1. The minimum (steady) creep rate for every specimen (either for SHT or for MHT) is almost the same.

The final stage of creep was characterised by an increase in strain rate which is believed to result from an increase in the extent of creep damage that ultimately leads to failure. This stage constitutes about half of the total creep life for both SHT and MHT materials, including the transition period from steady creep to the beginning of tertiary creep, approximately 300 - 400 hours for MHT and 200 - 250 hours for SHT. At high stress and/or high temperatures, the accelerating creep rate in the tertiary stage is dominated by a number of weakening metallurgical instabilities. Barbosa et al [15] concluded that there are three main damage factors operative in tertiary creep: intrinsic softening due to increasing dislocation density; loss of internal section caused by microvoid nucleation on grain boundaries; loss of external section when localized necking occurs. Other features like internal oxidation and/or corrosion, intercrystalline fracture, precipitation of brittle second phase particles, transformation of second phases that originally contributed to strengthening the alloy into less effective strengthening phases, are contribute to damage accumulation and ultimately failure of the material.

The creep fracture strain is lower than the tensile strain (for both SHT and MHT materials). Fig.4.1 and Table 4.1 show that the creep fracture strain is about 5 - 8% but in a tensile test the final fracture strain was of the order of 8 - 11% at 650 °C. This phenomenon is generally observed in many superalloys under creep rupture testing

conditions. The fracture strain of $\approx 5.4\%$ for MHT material is marginally smaller than that of the SHT material ($\approx 6.4\%$). It is interesting to note that necking was evident in the SHT specimens but not in MHT specimens, Fig.4.2. These results indicate that the serrated grain boundaries impart tensile stability to the material.

The MHT material revealed a predominantly transgranular fracture, Fig.4.3 whereas in SHT material the fracture was primarily intergranular, Fig.4.4. These observations indicate that a large grain size and the serrated grain boundaries suppress grain boundary sliding in the MHT material and encourage the creep deformation within the grain interior. This is verified by the presence of slip lines within the grain interior of the MHT material, Fig.4.3b and Fig.4.5.

4.3 Low Cycle Fatigue Behaviour

The results of the LCF tests for Inconel 718 subjected to SHT and MHT at RT and 525 °C are given in Table 4.2. The cyclic stress response of the materials are shown in Fig.4.6 to Fig.4.11. Fig.4.6 and Fig.4.7 show the room temperature LCF results whereas Fig.4.8 and Fig.4.9 are the LCF results at 525 °C. Fig.4.10 and Fig.4.11 show the comparison of the LCF lives of the SHT and MHT materials.

From these figures, it can be seen that SHT material always softens cyclically but the MHT material initially hardens and then softens. Most cyclic hardening or softening occurs quickly within the first 10 cycles and the strength generally stabilizes to a steady state value by approximately half life. It is believed that whether a material cyclically strain softens or hardens depends on the ratio of monotonic ultimate tensile strength σ_{UTS}

to 0.2% offset yield strength σ_y . When $\sigma_{UTS}/\sigma_y > 1.4$, a material will harden, but when $\sigma_{UTS}/\sigma_y < 1.2$, softening will occur [30]. For a ratio between 1.2 and 1.4, prediction becomes difficult. Usually, hardening or softening is more pronounced at high stress strain conditions. This can be seen in Fig.4.6 to Fig.4.11. As stress and strain increase, the softening or hardening becomes more pronounced and the hardening peak appears earlier. Large σ_{UTS}/σ_y ratio means also that the material has a relatively large work hardening range because there is a bigger increase in flow stress from yield to maximum stress. For Inconel 718 the ratio $\sigma_{UTS}/\sigma_y \approx 169.8/154 = 1.1$ for SHT and $\sigma_{UTS}/\sigma_y \approx 159.1/128 = 1.25$ for MHT at 525 °C. Another characteristic of softening or hardening is that initially hard and strong material generally will cyclically strain soften, and initially soft material will harden. The present test results also agree with these trends. On a linear scale, the $\sigma - N$ curve can be divided into three stages: the first stage exhibits a high rate of hardening or softening but is of short duration followed by a much longer period of steady state stress response; the final stage consists of rapid decrease in the stress response due to macrocrack propagation. Rupture is imminent at the end of this phase.

From the point of view of microstructure, it has been suggested that work hardening involves an increase in dislocation density whereas cyclic softening is associated with the shearing of precipitates [53]. This conclusion was based on the comparison of the interaction between dislocation and γ' precipitates because the Orowan bypass stress is much higher than particle shear stress, Table 4.3. Generally, in LCF tests, when the stress level is relatively low, particle shearing mechanism predominates. After shearing, the particles are easier to be cut again and their size gets smaller.

Some differences can be observed when the test temperature is changed. At RT the cyclic stress of the MHT material is lower than that of the SHT material even though

the MHT material hardens at the beginning, Fig.4.10. At 525 °C the cyclic stress response of the MHT material is higher than that of the SHT material, especially at high applied strain values, Fig.4.11. This is a positive feature of the MHT material from a design point of view since its strength is increased under cyclic loading and it also indicates that the MHT material has more stable strength than SHT material under cyclic loading when exposed to elevated temperature.

Fig.4.12 and Fig.4.13 show the monotonic and cyclic stress-strain behaviour of the two microstructures. Cyclic stability of the MHT material is quite obvious at both temperatures. The complex mode of monotonic-cyclic stress-strain curve for SHT Inconel 718 at 25 °C has also been noted by Pineau [54]. Clearly, the MHT reduces the large difference between monotonic and cyclic tensile stress-strain curves observed in the SHT material.

The LCF results show that the crack initiation occurs more easily in the MHT material than in the SHT material due to the large grain size of the MHT material, Fig.4.10 and Fig.4.11 and Table 4.2. At comparable strain ranges, the SHT material has a longer life and higher initial strength than the MHT material. However, the half-life strength of the MHT material is higher. These results show that when the applied load is static the SHT material is superior but when the load is alternating the SHT material softens considerably. Thus, the MHT material is superior to the SHT material from a cyclic loading point of view.

A comparison of the LCF behaviour of SHT and MHT materials can be made in terms of the ratio of the number of cycles to rupture at comparable strain range,

$$\text{at RT: } N_{\text{MHT}} / N_{\text{SHT}} \approx 25\%$$

$$\text{at 525 °C: } N_{\text{MHT}} / N_{\text{SHT}} \approx 60\%$$

This means that at elevated temperature the loss in LCF life of the MHT material is less than the loss at room temperature.

$$\text{for SHT: } N_{(525\text{ }^{\circ}\text{C})} / N_{(RT)} = 31\%$$

$$\text{for MHT: } N_{(525\text{ }^{\circ}\text{C})} / N_{(RT)} = 60\%$$

This means the effect of temperature is more severe on the SHT material but the LCF life of the MHT material still is only about 60% of the SHT material.

The MHT for Inconel 718 was developed to reduce CCGR and FCGR without substantial loss in LCF life relative to the SHT. Table 4.4 lists the ratio of LCF life of the MHT and SHT materials. At RT, the ratio $N_{(MHT)}/N_{(SHT)}$ increases when total strain decreases. In contrast, at 525 °C the ratio $N_{(MHT)}/N_{(SHT)}$ decreases when total strain range $\Delta\epsilon_T$ decreases. At high temperature and low strain level the fatigue life decreases considerably by about 80%. But at high strain levels the LCF life loss is 30 - 40%. This trend shows MHT does reduce the LCF life of Inconel 718. However, other mechanical properties such as hardness, modulus of elasticity, and yield strength are not affected to any great extent.

Low cycle fatigue cracks in smooth specimens can have three types of origins: 1) surface slip, 2) internal stress concentration, 3) localized surface attack. Ni-base superalloys display all three types of crack initiation sites but at elevated temperatures, in an oxidizing environment, the main mechanism responsible for crack initiation are internal stress concentration and localized environment attack. At RT, crack initiation is formed by intrusion-extrusion of slip bands, but at very high temperatures, crack usually starts from grain boundaries [44].

The fracture surface of the SHT Inconel 718 material shows typical appearance of fatigue fracture regimes (Fig.4.14) at both RT and at elevated temperature. Three stages

of fatigue fracture are very clear: I, crack initiation, II, crack propagation, III, final rupture of the specimen. The fracture surface of the MHT Inconel 718 material is generally rough, Fig.4.15. This appearance is related to the coarse grain size of the MHT material.

Under SEM the fracture mechanism of the SHT and MHT materials are very similar. Fig.4.16 shows the fatigue crack initiation site and Fig.4.17 shows the fatigue striations during early crack propagation in both materials. The crack initiation morphology is transgranular cleavage and most of the cracks originated at the surface or just below the surface. The mode of subsequent fatigue crack propagation is mixed. Note the dimples on the grain boundaries. The large dimples are caused by precipitates (Fig.4.18) that have a cubic structure and the small dimples are caused by microvoid coalescence. Some other studies also showed that LCF cracks originate from specimen surface in Ni-base alloys [54,55,56]. Mills [11] reports that the coalescence of the primary microvoids (dimples) were nucleated at broken MC type carbides during the early stage of plastic straining, regardless of heat treatment (MHT or SHT) and this is confirmed by the present results.

Two main variables that influence the LCF life are the temperature and plastic strain range. Fig.4.19 shows the influence of plastic strain range on N_f at RT and 525 °C for the two heat treatments. For Inconel 718, a bilinear life has been found and this means there are two slopes in M-C equation. Bilinear relation between $\Delta\epsilon_p$ and LCF life at 25 °C has been previously observed in Inconel 718 and Waspaloy by Clavel and Pineau [54,12]. For low $\Delta\epsilon_p$, life appears to be reduced (c is more negative). More negative the c , the less is the effect of plastic strain on LCF life. Because of the bilinear feature, the regression analysis using M-C equation 2.4.1 is not satisfactory. But the total strain range

versus fatigue life, Fig.4.20, shows a better linearity. Other researchers [10,55,56] have also found that the power law relationship (M-C equation) was better obeyed when the total strain range, instead of the plastic strain range, was used to correlate data. Fig 4.21 and Fig.4.22 show the regression results for the fatigue properties of Inconel 718 and Table 4.5 lists all the parameters determined by equation 2.4.3. The room temperature results in Fig.4.21 indicate a higher fatigue life for the SHT than the MHT material, but the asymptotes in Fig.4.22 are not consistent with the test result at 525 °C. These two figures seem to indicate that equation 2.4.3 is perhaps more suitable for room temperature life prediction.

4.4 Fatigue Crack Growth Rate (FCGR) Behaviour

Fig.4.23 shows that at room temperature there is no obvious difference between the FCGRs in the SHT and MHT materials. The MHT material, however, has slower FCGRs at 525 °C, Fig.4.24.

On average, the total number of cycles to failure in the MHT material is marginally higher than the SHT material. This small improvement can also be seen in Fig.4.25 where the crack lengths are plotted as a function of fatigue cycles. There is little difference between the results of SHT and MHT materials at room temperature, but the cracks in the SHT material (dotted line) propagate faster than in the MHT material at 525 °C. The final crack length at fracture is about the same for both materials indicating that the fatigue strength and critical fracture toughness (K_{IC}) are about the same for SHT and MHT materials.

Upon comparing the FCGR results at low ΔK values ($<25 \text{ MPa}\sqrt{\text{m}}$), in Fig.4.23

and Fig.4.24, the so-called "threshold" value is reduced as the test temperature increases. In other words, the crack growth enters the second stage (steady crack growth stage) earlier or at lower ΔK level. This trend has also been observed by Clavel and Pineau in Inconel 718 [45].

In the steady crack growth stage, the Paris equation (Equation 2.5.1) is usually used to describe crack growth behaviour. Tables 4.7 and 4.8 list the results of the regression analysis carried out on FCGR data generated at RT and 525 °C and the values of Paris constants are compared with the values reported in the literature. The ΔK range also influences the regression results and this is why the data point at very low ΔK and very high ΔK values are excluded. In the present case, the range of ΔK used was 28 - 60 MPa $\sqrt{\text{m}}$. The slope n changes from 4.8 at RT to 3.6 at 525 °C and C increased by about 100 times at 525 °C. Upon comparing these results with the results given in references and considering the differences in test conditions and specimen orientation with respect to the rolling direction, the values of Paris constants C and n are within the range specified in the literature. On an average, the C value is lower by a factor of 2 in the MHT material relative to the SHT material. The upper bound FCGR lines (99.7% confidence) are compared in Fig.4.26 where it is clearly shown that the MHT reduced the FCGR at elevated temperatures.

Fig.4.27 to Fig.4.29 show the fracture surfaces at crack initiation and short crack propagation stage in the CT specimens. Clearly, the crack growth in early stage is characterized by a transgranular type of cracking. Fig.4.29b shows the fracture features within the Paris region of relatively higher ΔK level (about 40 MPa $\sqrt{\text{m}}$) at 525 °C and Fig.4.30 shows the fracture features during final stage of cracking in the CT specimens at 525 °C. The final rupture was characterised by ductile dimples and broken particles.

These features were also observed during the later stage of Paris region (Fig.4.29b) to final fracture.

It is well known that different modes of cracking are observed over a wide range of ΔK values in FCGR testing. At low ΔK ($< 20 \text{ MPa}\sqrt{\text{m}}$), the fracture surface displays a highly faceted or cleavage-like appearance (Fig.4.27 to Fig.4.29) indicative of crystallographic fracture along intense heterogeneous deformation bands [11]. Clavel and Pineau [45] have analyzed the fracture surfaces of Inconel 718 and suggested that the crystallographic aspect of the fracture surfaces observed at room temperature in threshold regime is characteristic of the presence of a high density of smooth and planar areas or facets. These facets are transgranular and have a strong crystallographic aspect — decohesion of the matrix along crystallographic planes like $\{111\}$ and along the twin deformation bands.

At intermediate ΔK levels (30 - 60 $\text{MPa}\sqrt{\text{m}}$), well-defined striations are observed, as shown in Fig.4.29b. In this stage, the plastic zone becomes bigger, striations and secondary cracks form together. These features indicate that cracks propagate with a mixed mode mechanism — cleavage and intergranular fracture. Striation direction is not always along the global crack propagation direction. Locally, cracks may grow in different directions. At higher temperature and lower frequency (or longer hold-time) the transition from transgranular to intergranular crack propagation decreases, i.e. the transition value of ΔK decreases with increasing temperature and hold time. The transition from transgranular cracking to intergranular cracking is also dependent on grain size. Fig.4.29a shows that the transition in SHT material occurred earlier than that in MHT material. When $r_y < d$, where r_y is the size of the plastic zone at the crack tip and d is the grain size which is usually stable during testing, the accumulation of damage on grain

boundaries is small. However, when $r_y > d$, dislocation and slip lines are trapped by grain boundaries and causes damage accumulation. According to Lasalmonie's observation [28], this transition is the starting point of a steady state crack propagation stage and there is a change from a faceted crystallographic fracture mode to a striated fracture surface. It is well known that the incidence of grain boundary fracture increases with increasing test temperature. As temperature increases the plastic zone increases, more grain boundaries are involved in slip and plasticity [57].

At high ΔK values ($>60 \text{ MPa}\sqrt{\text{m}}$), fracture surfaces generally exhibit rather poorly defined microvoids coupled with large striations, Fig.4.30. In the final stages of crack propagation, precipitate particles break and fracture occurs along the interphase boundaries between precipitates and the matrix, Fig.4.31. Striations were never observed near these particles. This implies that these particles are hard and microvoids coalesce around them to cause fracture.

The size of the plastic zone has a significant influence on FCGR. Fig.4.32 shows the effect of a single overload on FCGR. The sharp drop in FCGR resulted from a single shock (overload, $P_{max} > 26 \text{ kN}$, nominal $P_{max} = 13 \text{ kN}$). The overload enlarges the plastic zone from its normal size. When the test continues under normal load, stress at crack tip is much lower than normal value, so, it takes many more cycles for the cracks to grow out of the large plastic zone.

Conclusions and Recommendation

Conclusions

In this project, the following conclusion can be drawn upon comparing the mechanical properties of the SHT and MHT materials:

- The creep rupture test results have shown that the MHT increases the creep rupture life of Inconel 718. Compared with the SHT material, the creep rupture life of the MHT material increases by a factor of about 2 at 650 °C and 593 MPa. However, the creep ductility is reduced by about 14%.
- The LCF test results have shown that relative to the SHT material, a significant loss in LCF life occurs in the MHT Inconel 718 material at both room temperature and 525 °C. The loss at 525 °C is less pronounced than that at RT. The LCF life of the MHT material is shorter by a factor of 3 relative to the SHT material at RT and by a factor of 1.5 to 2 at 525 °C.
- The fatigue crack propagation tests revealed similar FCGRs at room temperature in both SHT and MHT materials. The FCGRs were however decreased by a factor of 2 by the MHT at 525 °C.
- No obvious decrease in hardness and elasticity modulus was observed as a result of MHT.
- Relative to the SHT material, the yield strength were lower by about 18% at 525 °C and 14% at RT in the MHT material.
- Fracture analysis has shown that cracking during creep is predominantly

intergranular in SHT material and mixed trans-intergranular in the MHT material. LCF cracks initiate transgranularly from the specimen surfaces or subsurface inclusions and along crystallographic planes in both SHT and MHT materials. The FCGR test results at elevated temperatures indicate that a transition from transgranular at early propagation stage to final intergranular fracture is influenced by the grain size.

- It is clear that fine grain size improves resistance to crack initiation and results in higher tensile strength. Large grain size and a serrated grain boundary structure decrease FCGRs at 525 °C and increase creep rupture life at 650 °C.

Recommendation

- It is recommended to study the LCF and FCGR properties of the SHT and MHT materials at RT and elevated temperatures using specimens that duplicate the surface conditions of rim serrations, bolt holes and hub area of a turbine disc. Most of these locations are broached or reamed and shot peened prior to service. Specimens should be designed to study the influence of the MHT material subjected to reaming and shot peening conditions for LCF crack initiation and crack propagation resistance relative to the SHT material.

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Table 2.1 Properties of 718 as a Function of Heat Treatment

Solution Treatment ^a	Tensile at 21 °C (ksi)				Stress-Rupture at 650 °C/100 ksi		
	σ_{ys}	σ_{UTS}	ϵ (%)	RA (%)	Life (hr)	ϵ (%)	RA (%)
None (Direct Aged)	193	221	19	34	95	24	31
940 °C/1hr/AC	180	212	18	34	194	11	16
955 °C/1hr/AC	177	206	20	38	122	14	19
970 °C/1hr/AC	166	204	23	41	218	13	15
980 °C/1hr/AC	170	204	24	43	200	6	10
1010 °C/1hr/AC	172	202	22	46	270	6	12
1040 °C/1hr/AC	169	198	25	48	225 ^b	2	8

a: All standard aged 718 °C/8hr/FC, 55.6 °C/hr, 621 °C/8hr/AC.

b: Notch sensitive at $K_t = 3.8$; ϵ = elongation, RA = reduction of area.

Data from [12].

Table 2.2 Effect of Grain Size on HCF (454 °C)

Grain Size	fatigue strength ksi at 10^7 cycles	FS/ σ_{UTS}
ASTM 2	55	0.33
ASTM 5	80	0.45
ASTM 12	115	0.59

FS: fatigue strength. Source: [12].

Table 3.1 Chemical Compositions of Inconel 718 (wt.%)

Stock Type	Ni	Cr	Nb+Ta	Mo	Ti	Al	Si	C	Fe
Bar	53.2	18.3	5.17	3.0	1.0	0.42	1.12	0.03	Bal.
Plate	53.1	17.9	5.11	3.06	0.96	0.47	0.21	0.05	Bal.
Turbine disc	50-55	17-21	4.75-5.5	2.8-3.3	0.75-1.15	0.3-0.7	0.35*	0.02-0.08	Bal.

*. maximum. Source: [11,15].

Table 4.1 **Creep-Rupture Test Results for Inconel 718 at 650 °C**

Initial Stress: 593 MPa (86 ksi); Constant Load; Initial Gauge Length: 28.575 mm (1.125 in.).						
	Modified Heat Treatment			Conventional Heat Treat		
Specimen Number	2#	4#	6#	1#	3#	5#
Rupture Time (hours)	792	612	638	441	494	426
Rupture Strain (%)	5.85	5.04	5.57	6.67	7.59	4.85
Min Creep Rate (%/100h)	0.1519	0.2046	0.2033	0.1597	0.1558	0.1708
Reduction of Area (%)	11.64	9.75	10.58	26.0	24.31	12.69
Intercept of Tangent ¹	0.1058	0.3968	0.4602	0.3551	0.2457	0.0972
Instantaneous Strain (%)	0.1	0.3	0.42	0.3	0.15	0.05

1. Strain at intercept of the tangent at steady creep rate with strain axis at time zero.

Table 4.2 Inconel 718 LCF Test Results

Specimen	$\Delta \epsilon_T$	$N=1/4$		$N=N_f/2$			$N_{5\%}$	N_f
		E	σ_{ys}	$\Delta \epsilon_o$	$\Delta \epsilon_p$	S_{max}		
525 °C								
SHT9	0.86	178	>965	0.80	0.06	738		>34456
SHT8	1.02	174	≈ 1069	0.86	0.16	793	16038	16070
SHT7	1.56	172	1047	0.98	0.58	834		1218
SHT6	1.25	174	1088	0.93	0.32	821	2900	2956
MHT9	1.04	178	>855	0.90	0.14	793	2628	2660
MHT8	1.03	175	>855	0.89	0.14	793	>2500	2646
MHT7	1.56	172	882	1.05	0.51	896	780	820
MHT6	1.10	172	882	0.91	0.19	807	2120	2155
MHT5	1.25	172	876	1.00	0.25	841	1633	1741
Room Temperature								
SHT5	1.08	198	1229	0.95	0.13	910		>13800
SHT4	0.87	194	1196	0.84	0.03	910		>22200
SHT3	1.25	200	1269	0.97	0.28	979		11300
SHT2	1.97	197	1218	1.24	0.73	1199	1170	1271
SHT1	1.53	201	1111	1.07	0.46	1007	3900	3936
MHT4	0.88	199	1025	0.84	0.04	841	8500	9055
MHT3	1.18	199	1047	0.97	0.21	958	4000	4101
MHT2	1.96	199	1047	1.15	0.81	1124		181
MHT1	1.53	199	1047	1.07	0.46	1089		1223

NOTE: Unit of E is 10^3 MPa and σ_{ys} is MPa at 0.2% offset at $N=1/4$. S_{max} (MPa) is the maximum tensile stress at half life ($N=N_f/2$). $N_{5\%}$ is the number of cycles that corresponds to a decrease of 5% in the maximum tensile stress at half life. Unit of the strain range is in %.

Table 4.3 Estimated Values of Orowan and Shear Threshold Stresses for Different γ' Particle Sizes at Room Temperature

γ' (diameter)	Orowan Stress	Shear Stress
10 nm	1692 MPa	461 MPa
21 nm	1053 MPa	574 MPa
30 nm	799 MPa	660 MPa

Data from [53].

Table 4.4 LCF Life Comparison of Inconel 718

	Room Temperature				525 °C			
$\Delta \epsilon_T$ (%)								
SHT	1.97	1.53	1.25	1.08	1.56	1.25	1.02	0.86
MHT	1.96	1.53	1.18	0.88	1.56	1.25	1.03	1.10
Cycles N_f								
SHT	1271	3936	11300	15800	1218	2956	16070	34456
MHT	181	1223	4104	9055	820	1741	2646	2155
$N_{\text{MHT}}/N_{\text{ISHT}}$ (%)	14.2	31.1	36.3		67.3	58.9	16.5	

Table 4.5 Parameters of Manson-Coffin Equation for Inconel 718

		σ_f/E	b	ϵ_f	c
RT					
	MHT	0.0092	-0.075	0.3381	-0.696
	SHT	0.0141	-0.106	0.4865	-0.611
525 °C					
	MHT	0.0151	-0.141	3.561	-0.989
	SHT	0.0077	-0.058	0.377	-0.622
**		0.0083	-0.059	0.389	-0.631
**		0.00899	-0.0729	0.731	-0.741

NOTE: $E = 200 \times 10^3$ MPa at RT and $E = 172 \times 10^3$ MPa at 525 °C.

** Comparison data, conventional heat treatment, tested at 500 and 550 °C. After [72,77].

Table 4.6 FCGR test results of Inconel 718 on CT specimens

specimen No. Test at RT	cycles/crack length (cm) in pre-crack	cycles/crack length (cm) final break	total cycles
M1	227,275/1.263	174,818/3.444	402,000
M2	193,911/1.267	188,664/3.348	382,500
M3	125,176/1.192	184,316/3.322	309,500
M4	195,241/1.281	137,715/3.267*	333,000
M5	233,845/1.442	117,628/3.467	351,500
S1	201,962/1.411	120,835/3.517	323,000
S2	184,709/1.347	132,104/3.333*	317,000
S3	178,568/1.359	121,913/3.562	288,000
S4	185,736/1.349	130,733/3.419	316,500
S5	213,770/1.352	135,066/3.447	349,000
Specimen No. Test at 525 °C	precrack at RT		
M6	273,653/1.307	77,394/3.4521	351,000
M7	280,244/1.305	70,595/3.4796	351,000
M8	204,670/1.196	106,248/3.4277†	311,000
M9	271,749/1.366	90,455/3.3895	362,000
M10	198,804/1.306	81,343/3.5776	280,000
S6	214,668/1.405	51,290/3.2594	266,000
S7	249,218/1.3515	57,942/3.3365	307,000
S8	259,000/	112,200/3.3723†	371,000
S9	240,790/1.3715	54,100/3.4732	295,000
S10			

* Actual crack length is a little bit longer than the number.

† Overloaded to 26 kN (nominal 13 kN) 3 times during test; system halted and specimen exposed to 525 °C for 2 - 4 hours with small load (self weight of the connections and the piston of the test machine).

Table 4.7 Comparison of Coefficients in Paris Law at RT

	MHT	SHT	Ref.[36]	Ref.[46]	Ref.[54]
$C (x10^{11})$	1.28	3.0	8.0	39.6	12.8
n	4.75	4.48	4.0	3.66	4.0
da/dN	0.91	0.76	0.328	0.449	0.525

* $x10^{-3}$ mm/cycle at $\Delta K = 45 \text{ MPa}/\sqrt{\text{m}}$.

Ref.[46], CHT, 12.7 mm plate, at 24 °C, $R = 0.05$, $f = 0.667 \text{ Hz}$ (40 cpm).

Ref.[54], CHT, at 24 °C.

CHT = SHT: 955 °C/1hr/AC + 718 °C/8hr/FC - 55 °C/hr - 621 °C/8hr/AC.

Table 4.8 Comparison of Coefficients in Paris Law at 525 °C

	MHT	SHT	Ref.[54]	Ref.[46]	Ref.[46]
$C (x10^{11})$	98.9	130	32.9	880	3025
n	3.68	3.64	3.95	3.07	2.5
da/dN	1.20	1.36	1.115	1.038	0.41

* $x10^{-3}$ mm/cycle at $\Delta K = 45 \text{ MPa}/\sqrt{\text{m}}$.

Ref.[54], CHT, 550 °C.

Ref.[46], CHT, 538 °C, $R = 0.05$, $f = 0.667 \text{ Hz}$.

Ref.[46], INEL heat treatment: 1093 °C/1hr/FC - 55 °C/hr -
- 718 °C/4hr/FC - 55 °C/hr - 621 °C/16hr/AC.

Fig.2.1 Temperature - flow stress curves of γ' and the influence of some solute elements. Source: [82].

Fig.2.2 Effect of varying solution temperature on yield strength and ductility of Inconel 718; solution as indicated + "Standard Age". Source: [21].

Fig.2.3 Tensile and yield strengths of Inconel 718 versus temperature. Source: [21].

Fig.2.4 SHT microstructure of Inconel 718. x200.

Fig.2.5 Grain size versus solutioning temperature for Inconel 718. Source: [15].

Fig.2.6 MHT microstructure of Inconel 718. x500.

Fig.2.7 Stress-rupture strength of Ni-base superalloys in relation to volume fraction (V_f) of γ' . Source: [32].

2

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Fig.2.8 Schematic representation of creep cavities.

Fig.2.9 Variation of tensile elongation of selected superalloys with temperature. Data from [12].

Fig.3.1 Configuration of specimen for creep rupture tests.

Fig.3.2 Linearity of calibrated LVDT to measure creep elongation in creep - rupture test.

Fig.3.3 Specimen configuration for low cycle fatigue (LCF) tests.

Fig.3.4b Orientation and dimensions of the block cut from 12.7 mm Inconel 718 plate for the CT specimen.

Fig.3.5a Calibration curve of β (VR) parameter versus optical crack length. Second order regression based on three specimens tested (S1 for SHT and M4, M5 for MHT).

Fig.3.5b Measured optical crack length (average of the values on both sides of the specimen) versus potential drop at the crack tip (S1 - S4 for SHT and M4 for MHT).

Fig.3.6 Comparison of the optically and DC-PD measured crack lengths (S1 - S4 for SHT and M2, M5 for MHT).

Fig.4.1 Strain - time curves of creep rupture test of Inconel 718 at 650°C.

Fig.4.2

Inconel 718 after creep rupture test at 650 °C, MHT 6, $t_R = 638$ hours, $\epsilon_R = 5.57\%$; SHT 3, $t_R = 494$ hours, $\epsilon_R = 7.59\%$. Note many secondary cracks near the fracture.

Fig.4.3a Creep rupture morphology of Inconel 718 at 650 °C, MHT 4, $t_R = 612$ hours. external (above) and fracture surface (below).

Fig.4.3b Slip lines inside grains of MHT Inconel 718, creep rupture tested at 650 °C.

Fig.4.4 Intergranular creep cracking of SHT Inconel 718.

Fig.4.5 Intergranular crack in the SHT material (below) and transgranular crack in the MHT material (above).

Fig.4.6 LCF results of conventional heat treated (SHT) Inconel 718 at room temperature (RT).

Fig.4.7 LCF results for modified heat treated (MHT) Inconel 718 at room temperature (RT).

Fig.4.8 LCF results for SHT Inconel 718 at 525 °C.

Fig.4.9 LCF results for MHT Inconel 718 at 525 °C.

Fig.4.10 Comparison of the LCF results for SHT and MHT Inconel 718 at RT.

Fig.4.11 Comparison of the LCF results for SHT and MHT at 525 °C.

Fig.4.12 Comparison of the SHT and MHT Inconel 718 monotonic and cyclic stress - strain curves at RT.

Fig.4.13 Comparison of the SHT and MHT Inconel 718 monotonic and cyclic stress - strain curves at 525 °C.

Fig.4.14

Fracture surface of SHT Inconel 718.

Above: SHT5, at RT, $N_f > 13800$ cycles, $\Delta \epsilon_T = 1.07\%$;

Below: SHT8, at 525 °C, $N_f > 16070$, $\Delta \epsilon_T = 1.02\%$.

Fig.4.15

Fracture surface of MHT Inconel 718.

Above: MHT4, at RT, $N_f > 9055$ cycles, $\Delta \epsilon_T = 0.89\%$.

Below: MHT6, at 525 °C, $N_f > 2155$, $\Delta \epsilon_T = 1.11\%$.

Fig.4.16 Crack initiation at the surface of the specimen in LCF tests at 525 °C. above: SHT8; below: MHT9.

Fig.4.17 **Fatigue striations in MHT Inconel 718 in LCF tests.**

Fig.4.18 Characteristics of Inconel 718 fracture surfaces. Microvoids coalescence around particles appeared in both SHT and MHT materials.

Fig.4.19 The influence of plastic strain range on LCF life. Bilinear relation between them was noted in Inconel 718.

Fig.4.20 Total strain range versus LCF fracture life.

Fig.4.21 Fatigue properties of Inconel 718 after MHT and SHT, tested at RT. Straight lines are regression of M-C equation and small r is the correlation coefficients.

Fig.4.22 Fatigue properties of Inconel 718 after MHT and SHT, at 525 °C. Straight lines are regression of M-C equation and r is the correlation coefficients.

Fig.4.23 Fatigue crack growth rate test results at RT.

Fig.4.24 Fatigue crack growth rate test results at 525 °C.

Fig.4.25a Optical crack length increases with cyclic loading at RT.

Fig.4.25b DC-PD measured crack length versus cycles at 525 °C. The irregular shape of the curves resulted from the overload.

Fig.4.26a Comparison of the upper bound lines of Paris equation in SHT and MHT materials tested at RT.

Fig.4.26b Comparison of the upper bound lines of Paris equation in MHT and SHT materials tested at 525 °C.

Fig.4.27

Crack initiation of CT specimens in FCGR test. The arrows indicate the laminal band. Above: SHT; below: MHT.

Fig.4.28 Optical microscopy showing the laminar structure in 12.7 mm Inconel 718 plate after MHT, x500.

Fig.4.29a Early stage of fatigue crack propagation at RT.
Above: SHT, striations in large grains but not in banded area;
below: MHT, note the crack along the twins.

Fig.4.29b Early stage fatigue crack propagation at 525 °C.
Above: SHT; below MHT.

Fig.4.30 Final stage fatigue crack propagation at 525 °C.
Above: SHT; below: MHT.

Fig.4.31 Final fracture of Inconel 718 in FCGR test.
Above: SHT at RT; below: MHT at 525 °C.

Fig.4.32 Detail record of FCGR versus ΔK . Overload resulted in sharp drop in crack growth rate. It took many more cycles for the crack to grow out of the enlarged plastic zone.

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