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Development and Characterization of a Damage Tolerant Microstructure for a Ni-base Turbine Disc Alloy

by

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B. Eng.**

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

Master of Engineering

**Department of Mechanical and Aerospace
Engineering, Ottawa-Carleton Institute for
Mechanical and Aerospace Engineering**

**Carleton University
Ottawa, Ontario
January 2000**

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**Development and Characterization of a Damage Tolerant
Microstructure for a Ni-base Turbine Disc Alloy**

submitted by
Richard M. Kearsey

in partial fulfillment of the requirements for
the Degree of Master of Engineering.

Thesis Supervisor

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Abstract

A modified heat treatment is developed for the nickel-base superalloy, PWA 1113. The main objective for the development of this new heat treatment is the design of a damage tolerant microstructure (DTM). The DTM is designed with the aim of imparting improved fatigue crack growth resistance without forfeiture of other vital properties such as tensile strength, stress rupture life, and low cycle fatigue lifetimes. This is achieved by optimizing the material's grain size, grain boundary morphology, and the intragranular precipitate size and distribution. Fatigue test results demonstrate that, when compared to the conventional microstructure (CM), the short crack growth rate for the DTM is slower by a factor of 3 at room temperature and 2.2 times slower at 482 °C. Creep test results show that at 690 MPa (100 ksi) and 705 °C, the creep-rupture life is extended by a factor of almost 4 for the new DTM. Tensile test results indicate minimal strength losses for the DTM with respective YS and UTS values of 80% and 90% of the CM baseline values at both test temperatures. A critical aspect of the damage tolerant design philosophy for advanced disc materials is the early detection of small fatigue cracks and lifetime prediction based on established fatigue crack growth rate data. For this reason, an alternating current potential difference (ACPD) technique is also developed for fatigue testing, and is shown to be extremely reliable in short crack growth measurement applications.

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Nomenclature

LCF:	low cycle fatigue
DTD:	damage tolerant design
NDI:	non-destructive inspection
FCGR:	fatigue crack growth rate
SII:	safe inspection interval
PWC:	Pratt and Whitney Canada
NRC:	National Research Council of Canada
CM:	conventional microstructure
DTM:	damage tolerant microstructure
FCC:	face-centered cubic
V_f:	volume fraction
K_T:	stress concentration factor at a notch
ΔK:	stress intensity range
ΔK_{th}:	stress intensity threshold
ΔK_{eff}:	effective stress intensity level
LEFM:	linear elastic fracture mechanics
σ_{op}:	minimum crack opening stress
CTOD:	crack tip opening displacement
ω:	plastic zone size
SEM:	scanning electron microscope
TEM:	transmission electron microscope

ACPD:	alternating current potential drop
DCPD:	direct current potential drop
I:	electrical current
R:	electrical resistance
V:	potential difference
δ:	alternating current skin depth
ENSIP:	Engine Structural Integrity Program
SEN:	single edged notch specimen
h:	serration height
λ:	serration wavelength
V_{crack}:	crack probe voltage
$V_{\text{reference}}$:	reference probe voltage
δV:	change in crack probe voltage
V_o:	initial crack probe voltage (crack free)

1. Introduction

1.1. Background

As the technology of today's gas turbine industry progresses, the advancement in material design concepts and processing must also coincide. Ever-increasing service temperatures and stresses are resulting in the need to augment presently used engine materials as well as existing safety guidelines. For instance, emerging safety standards for rotating engine components are now starting to demand properties in engineering materials that may not meet the required specifications.^{1, 2} In particular, the damage tolerance specifications imposed upon turbine disc materials have prompted researchers to investigate ways to improve fatigue and creep properties of presently used alloys without compromising the traditional safe life limits. As a result, a microstructural design philosophy has been successfully developed that is aimed at improving the damage tolerance of conventional disc materials.³

A damage tolerant disc material has the ability to exhibit greater resistance to the growth of inherent or service induced flaws under creep and/or fatigue loading conditions, while still maintaining adequate low cycle fatigue properties, creep rupture strength, and tensile strength.⁴ Consequently, improving the damage tolerance of conventional Ni-base disc alloys has become an area of substantial interest amongst disc material manufacturers and engine designers alike.

Ni-base superalloys were initially incorporated into the gas turbine engine because of their exceptional strength and oxidation resistance at elevated service temperatures. Material design concepts were aimed at improving these two properties to simultaneously improve engine operating efficiencies. Over the last 50 years, improvements in mechanical properties have largely been achieved by the introduction of numerous alloying additions in conjunction with modified processing techniques. With time, manufacturers have been able to produce lighter alloys with superior strength and crack initiation properties. However, one unattractive characteristic of these newer disc materials has been decreased resistance to crack growth.⁵ This problem of depreciating crack growth resistance is further magnified as modern turbine components are subjected to progressively higher service temperatures and stresses. For this reason, changes in material properties and disc-life design philosophies have become imperative if any additional advancement in engine optimization is to be achieved.

1.2. Disc-Life Design Philosophies

Due to the critical impact of component failure in gas turbine engines, strict attention has been given to the lifing methodology of these parts. Namely, how long can the part safely remain in service before it must be retired? From the onset of major superalloy implementation in the 50's and 60's, all turbine components abided by the safe-life design philosophy. This safe-life design approach states that a component must be designed for a finite service life during which no significant fatigue damage can occur.⁶ Specifically, disc life limits are set by the statistical number of low cycle fatigue (LCF) cycles required to form a detectable 800 μm surface crack, in 1 out of 1000 components

tested.⁷ One major drawback of this lifing methodology is that a large number of discs are removed from service prematurely. In some cases, it has been proven that over 80% of these prematurely retired discs have at least 10 lifetimes remaining.⁸ The inefficiency and unnecessary cost of this lifing methodology is evident in this regard. Another disadvantage of the safe-life design approach is the assumption that components entering service are free from defects. This negligent approach could prove to be catastrophic if a rogue flaw is present in a highly stressed region of the disc.⁹ Crack growth from this defect may occur and ultimately result in a disc burst prior to the safe-life expectation. Furthermore, the likelihood of this type of scenario increases, as turbine optimization requires that these same disc materials be used at even more demanding service conditions. Because of the shortcomings with the safe-life design philosophy, a more appropriate lifing methodology has evolved.

The damage tolerant design (DTD) philosophy includes the assumption that a flaw or a rogue defect is present during the initial design of the disc.¹⁰ The initial size of this flaw, a_i , is assumed to be the minimum detectable crack size of the non-destructive inspection (NDI) instrument used during engine maintenance. Unlike the safe-life design methodology that is dependant on pure statistical analysis of LCF tests, the DTD philosophy predicts disc lifetimes based on fatigue crack growth data predicted by fracture mechanics principles. The fatigue crack growth rate (FCGR) of the material can be determined by performing a sufficient number of fatigue tests that reflect the in-service variables of stress amplitude, stress concentration, loading frequency, and temperature. Once the FCGR of the material is known and a critical crack size, a_c , has

been established, a safe inspection interval (SII) for the component can be determined. This SII is derived from the time it takes for a_i to propagate to a predetermined crack size, a_m , based on the FCGR of the disc material. After each SII, the disc is removed from service for inspection. If no cracks are found the disc is returned to service, otherwise, it is retired. The DTD methodology is simple in nature but effective in application. Not only does this methodology prevent catastrophic failure by assuming the presence of inherent flaws from initial service, but it also prevents unnecessary discarding of discs that are retired prior to their actual full-life capability. A schematic of this inspection process is provided in Figure 1.¹¹

It should be mentioned that this life management philosophy requires the use of very reliable NDI techniques during engine inspection intervals. This is to ensure that cracks will not be missed during inspection resulting in unexpected catastrophic failures. In addition, the need to remove the discs from the aircraft for inspection does pose a problem with respect to applying the DTD philosophy to civil aircraft engines. Civil aircraft are in flight much more than military aircraft, which usually results in shorter maintenance periods that do not allow for major engine inspection processes. For this reason, most civil aircraft still abide by the safe-life design philosophy. However, by incorporating damage tolerant microstructures into civil aircraft turbines, improvements can still be made with respect to extending the engine's safe life limits.

It is apparent that the critical aspect of the DTD philosophy is the length of the SII. If the SII were too short, the maintenance downtime would make the process impractical.

Therefore, materials that have consistently superior crack growth rate properties are essential to the application of this lifing methodology. Hence, researchers must not restrict themselves solely to the area of damage tolerant microstructural design. They must also pay attention to the methods used to characterize the crack growth rate properties, which ultimately influence the SII predictions. This includes the crack monitoring techniques used to measure crack initiation and the initial stages of early crack growth, as well as the fracture mechanics principles used to define the results obtained.

1.3. Thesis Objectives

As previously mentioned, damage tolerant design has become an area of great interest for gas turbine developers everywhere. Recently, Pratt and Whitney Canada (PWC) has requisitioned the development of a damage tolerant microstructure for their Ni-base disc alloy designated as PWA 1113. In collaboration with the National Research Council of Canada (NRC) and Carleton University, a project was initiated with the aim of developing a damage tolerant microstructure for PWA 1113 by thermal processing. Specifically, the goal was to improve the creep-rupture life and short crack growth rate of the material without causing detrimental loss to other mechanical properties such as tensile strength and low cycle fatigue life. To clearly define the main objectives of this thesis, they have been broken down in point form as follows:

- 1) **Develop a heat treatment sequence that produces a damage tolerant microstructure for the Ni-base turbine disc alloy, PWA 1113.**
- 2) **Develop an alternating current potential difference (ACPD) technique that is capable of measuring the short crack growth rates of the CM and DTM during fatigue tests at room and elevated service temperatures.**
- 3) **Determine the effects of the new microstructure on short crack growth rate by performing fatigue tests on both the conventional microstructure (CM) and the damage tolerant microstructure (DTM).**
- 4) **Perform tensile and creep-rupture tests on the CM and DTM to verify that the improved fatigue and creep resistance of the DTM was not gained with significant loss of either proof strength and ultimate tensile strength.**

2. Literature Review – Properties of Gas Turbine Discs

2.1. Turbine Disc Requirements

Since their earliest applications, turbine discs have always been one of the most critical components of the gas turbine engine, strongly influencing the overall engine life and reliability characteristics. By virtue of the mechanical rotation of these discs and their location in the hot-gas section of the engine, they are subjected to high stresses and hot corrosive environments. Thermal and stress gradients observed are quite diverse, and are largely dependent on the size of the disc and its rotational speed. Maximum temperatures are located at the outer edge or rim of the disc, where the overall nominal stresses are relatively low. Conversely, the inside bore of the disc is where the temperature is lowest and is also the area of highest direct stress.¹² Figure 2 illustrates the cross-sectional geometry of the turbine disc used in this study, as well as the temperature and stress allocations mentioned. Initially, disc materials were chosen based on tensile strength properties required for the bore, and creep strength requirements for the rim. However, during the 1960's, it was determined that LCF strength was also an important factor due to cyclic stresses experienced during normal flight operations.¹³ Furthermore, modern turbine engines require turbine disc materials that are capable of exhibiting crack resistance over a long period of time while subject to variable loads, temperatures, and environments.¹⁴ Since early development, many material derivatives have been produced to meet the constantly intensifying list of disc property requirements.

2.2. Turbine Disc Materials

The first group of turbine disc materials used in the gas turbine engine were the austenitic stainless steels. One commonly referenced event is the 1939 test flight of Whittle's W-1

gas turbine engine, which possessed a gas turbine section made primarily of stainless Stayblade steel.¹⁵ At the time, these alloys proved to be tolerable with respect to corrosion resistance and tensile strength. With time, however, the demand for higher service temperatures and stresses resulted in the need for materials with superior high temperature properties. Higher operating temperature demands ultimately led to the replacement of turbine steels with more creep resistant superalloys. The first superalloys were initially conceived for turbosupercharger applications in reciprocating engines, and due to their superior properties they were quickly incorporated into the hot gas section of the gas turbine engine.¹⁶ From the moment they were first implemented into the aircraft engine, material developers promptly went to work designing new and improved versions of the simple Ni-Cr parent alloy. Over the years, numerous derivatives were produced largely because of the demand for faster, more efficient, turbine-driven aircraft by military and civil markets alike. Some commonly used superalloys, along with their respective creep and proof strength properties, are presented in Table 1.¹⁷ As can be seen from this table, the general trend for disc material selection is the use of Ni-base superalloy derivatives. These superalloy derivatives completely dominate the turbine disc market and warrant a brief overview.

2.2.1. Ni-base Superalloys

As the name suggests, Ni-base superalloys are those alloys that have nickel as the major constituent (30 to 75%), along with appreciable amounts of chromium (up to 30%).¹⁸ Over the years, numerous alloying additions have been incorporated into the base alloy. Initially the development of different alloy compositions was generated on a trial-and-error basis, with the anticipation of producing an alloy with improved operational

properties. However, as modern metallurgy methods developed, material producers were eventually able to characterize the effects of different alloying additions.

2.2.1.1.   *Chemical Composition*

The remarkable improvements in high temperature strength, corrosion resistance, and creep-rupture properties, have been largely due to alloying with various combinations of elements. The addition of aluminum and titanium was first implemented in the production of the earliest precipitation hardened Ni-base superalloy, Nimonic 80. This alloy consisted of a Ni-20% Cr solid solution matrix (commonly referred to as the austenitic or γ phase), with 2.25% Al and 1% Ti.¹⁹ The addition of aluminum and titanium to the Ni-base matrix was perhaps the most critical development stage in the evolution of Ni-base superalloys, as it was responsible for the creation of the precipitation hardening phase, γ' . Other performance enhancements have been obtained by adding such elements as molybdenum, cobalt, niobium, tantalum, zirconium, boron, iron, and tungsten. The effects of these more commonly used alloying additions added to Ni-base alloys have been provided in Table 2 for review.

2.2.1.2.   *Microstructure*

As previously mentioned, mechanical property enhancement of Ni-base alloys was largely due to the composition of alloying additions. More precisely, these compositional adjustments were responsible for the microstructural changes produced; which is the underlying reason for improved properties. Figure 3 illustrates the history of microstructural evolution of Ni-base alloys, showing the major beneficial and detrimental

phases that have emerged over the years as well as the gradual change in γ' morphology.²⁰

The major constituents present in most Ni-base superalloys are:

- 1) γ phase, the face-centered cubic (FCC) austenitic matrix,
- 2) γ' phase, the ordered FCC intermetallic precipitate, and
- 3) Carbides, commonly MC, M_6C , M_7C_3 , and above all $M_{23}C_6$.

The γ phase, as already mentioned, is the continuous austenitic matrix generally consisting of nickel and chromium, as well as smaller amounts of other solid-solution strengthening elements. Although a significant amount of strength is attained by the solid-solution strength of the matrix, the majority of Ni-base alloys gain strength from the existence of the γ' precipitate phase.

The γ' phase is an ordered intermetallic FCC structure, empirically represented as an A_3B -type compound. Generally, the precipitate has the nominal composition of $Ni_3(Al, Ti)$, with nickel occasionally being substituted by cobalt and titanium with tantalum. Due to their similar lattice parameter features, the matrix-precipitate mismatch is usually less than 1.3%.¹⁹ Precipitate morphology is governed by the degree of this γ/γ' mismatch. Small mismatches of $\sim 0.05\%$ result in spherical γ' . As the degree of mismatch increases, γ' occurs in cuboidal form, followed by a semicoherent plate morphology when

mismatches are greater than 1.25%.¹⁹ Fundamentally, the precipitate shape is determined by its ability to lower the overall free energy of the system. The free energy of the system is governed by the competing variables of surface energy and strain energy. In brief, the surface energy value varies with the square of the characteristic precipitate dimension, whereas the strain energy term varies with the cube of this dimension. Hence, at small sizes where the surface energy dominates, the precipitates assume a spherical shape to lower this value. Likewise, at larger sizes where the strain energy dominates, the precipitate assumes a more complex shape that reduces the overall strain.²¹

There are several types of carbide phases common to Ni-base superalloys. Primary MC type carbides, which are the first to form during solidification processes, can form at grain boundaries and within the matrix. Secondary MC carbides can also form during heat treatment processes. At lower temperatures, the MC carbides can transform readily into the more stable $M_{23}C_6$ carbide.²² This carbide phase can influence the mechanical properties of the alloy as will be discussed in the next section. M_6C carbides are similar to $M_{23}C_6$ carbides, and have a general tendency to form when molybdenum and tungsten levels are high in the alloy.

2.2.1.3.   *Strengthening Mechanisms*

Ni-base superalloys are capable of exhibiting very high strength at temperatures close to their physical melting points. In general, to attain the highest level of strength in these alloys it is necessary to apply the proper heat treatment sequence to optimize microstructural features. Standard heat treatment practices usually involve a solutioning

step above the γ' solvus that allows for subsequent controlled cooling required for precipitation of the γ' strengthening phase. The solutioning step is followed by one or more aging sequences required to optimize both inter and intragranular microconstituents. If the proper heat treatment sequence has been implemented then the alloy's strength may be attributed to the mechanisms such as solid solution strengthening, precipitation hardening, and grain boundary strengthening.

Although a significant amount of strength can be attributed to solid solution strengthening, these alloys attain most of their high temperature strength by precipitation hardening. The volume fraction (V_f) and size of the γ' particles are the two main variables that influence the degree of precipitation hardening. The precipitate V_f of these alloys is governed by the relative amounts of Al and Ti added to the parent material. By producing alloys with high V_f of fine precipitate the creep strength of aerospace components can be significantly improved for high temperature applications, as seen in Figure 4.²³ Although the V_f of the γ' precipitate does play an important role in the strength of the alloy, the actual particle morphology is also vital. The interaction of dislocations and precipitates within the stressed component is greatly influenced by the particle size. Deformation due to an applied stress requires the motion of dislocations within the grains. Once a dislocation encounters an impeding precipitate, it must either cut through it, or find some way around it. With very fine precipitate sizes it requires less energy for dislocations to climb to another plane to bypass the barrier rather than to cut through the precipitate itself. In the other extreme, when the precipitate size is too large for cutting, it may be easier for the dislocations to bow between them and loop around the

particles.²⁴ Figure 5 illustrates the processes of dislocation climb, precipitate cutting and looping.²⁵ To attain the highest degree of hardness required for optimum strength levels, suitable aging heat treatments must be implemented to produce the proper precipitate size. An example of this optimum precipitate size / peak hardness level has been provided graphically in Figure 6.²⁶ At smaller precipitate sizes the main strengthening mechanism is by precipitate cutting. As the precipitate size increases, precipitate bypassing dominates and the strength begins to decrease. Peak hardness for most materials is usually observed at the transition point between the two strengthening mechanisms.

The intergranular features of Ni-base superalloys are also important to the alloy's overall strength. By controlling the precipitation of second phase particles (either carbide or γ' precipitates) at the grain boundaries their inherent strength can be optimized drastically. Discontinuous $M_{23}C_6$ grain boundary carbides are beneficial as they retard grain boundary sliding, resulting in improved grain boundary strength. On the other hand, if these carbides form as continuous films along the grain boundaries, brittle properties are imparted into the microstructure.²²

3. Literature Review – Fatigue of Gas Turbine Discs

3.1. Characteristics of Fatigue

In addition to the general prerequisites of high temperature strength and corrosion resistance, turbine disc materials must also exhibit substantial resistance to fatigue at elevated temperatures. Fatigue is the general name used to describe failure that occurs in components subjected to repeated loading. This repeated loading may be in the form of a fluctuating stress or strain, a mechanical vibration, a fluctuating temperature, or a combination of one or more of these variables. However, for the most part, service failures in gas turbine components are usually a result of imposed cyclic stresses and resonant vibrations. Fatigue failure can be quite insidious, as it is generally a progressive process usually governed by the growth characteristics of unseen cracks that may fracture suddenly without warning. Hence, the necessity to fully characterize the fatigue properties of critical engine components is of vital importance. The general stages in the fatigue process are crack nucleation, short crack growth, long crack growth, and fracture.

Overall, local surface imperfections such as scratches and machine marks are the most obvious sites for crack nucleation. Inclusions, hard precipitates, grain boundaries and even twin boundaries are also microscopic features susceptible to crack nucleation. On an even finer scale, the orientation of cross-slip planes, dislocation density, and crystallographic defects all ultimately contribute to the mechanism of crack nucleation. It is the presence or absence of one or more of these characteristics that will influence the mode of crack nucleation.

Depending on the material's microstructural features and the imposed loading conditions, crack nucleation may occur either within a grain interior (transgranularly) or along a grain boundary (intergranularly). Transgranular crack nucleation occurs as a result of the build-up of localized regions of plastic deformation (slip) that develop during cyclic loading.²⁷ These crack nucleation sites usually occur along the slip plane of maximum shear, which is oriented at 45° with respect to the axial loading direction. It has been shown that these regions of slip physically manifest as collections of intrusions and extrusions at free surfaces due to the displacement of metal along the slip planes.²⁷ As the number of these surface irregularities increases, a discernable crack will eventually appear. For clarity, a schematic of this process has been provided in Figure 7.²⁸ Studies have indicated that these banded structures cyclically soften with respect to the matrix, resulting in a concentration of plastic strain.²⁹ These bands are commonly referred to as “persistent slip bands” or PSB's. Typically, dislocation motion along these planes ends at a grain boundary or some other impeding structure, at which point a dislocation pile-up results and inhibits crack advancement.³⁰ Zhang *et al.*, through intensive diffraction studies, showed that it was this mechanism of dislocation pile-up that inhibited crack growth, as opposed to the interaction of the actual crack tip with the grain boundary.

Intergranular crack nucleation is usually only observed under elevated temperature conditions for turbine disc materials.³¹ Crack nucleation usually occurs at a surface grain boundary triple point or at a brittle particle located on the grain boundary. In a similar manner to transgranular crack nucleation, intergranular cracks typically form on a plane inclined at 45° to the loading axis. Shearing of bonds along the grain boundaries occurs as

a result of the preferential migration of grain boundary dislocations in this plane, allowing a crack to initiate. At room temperature, grain boundary strength is usually greater than that of the grain interior. However, some instances of this intergranular crack nucleation mode have been reported to occur in superalloys at room temperature. Wang et al have observed intergranular nucleation in the turbine disc material Waspaloy; however, the mode of crack growth usually changed to transgranular after very short distances of less than one grain length.³¹

Although the nucleation stage plays a major part in the fatigue life of a turbine component, a more significant characteristic for material designers is that of crack growth. In particular, the study of short crack growth is certainly important for the improved understanding of the overall fatigue process and is important to the development of materials with improved fatigue resistance. This is largely due to the fact that cracks in engineering components spend the majority of their lives as physically small cracks on the order of 5 to 1000 μm in depth.³²

3.2. Short Crack Growth

Until recently, most fatigue research completed in the gas turbine field concentrated largely on the characterization of relatively long crack growth behavior of disc materials. Although this information is essential in itself, it only supplies the designer with a partial description of a material's overall fatigue behavior. With only the long crack growth behavior known, designers are left in the dark with respect to the initiation and short crack growth properties. As previously mentioned, these two regimes generally account

for a large portion of the total fatigue life of a component, and as a prerequisite for damage tolerance applications this information is quintessential.

There are two main defining properties of short cracks that distinguish them from typical long crack growth features.

1. Short cracks have generally been observed to exhibit faster and more irregular propagation rates compared to long crack data.³³
2. Short crack propagation occurs at stress intensity factors (ΔK) well below the traditional threshold value (ΔK_{th}) for long crack growth.³⁴

These two distinguishing properties of short cracks have been reported by numerous researchers all of whom have reported similar anomalous short crack behavior.^{31- 35} Figure 8 illustrates the anomalous behavior of short cracks as compared to long cracks for stress intensity levels below the threshold limit.³⁶

3.2.1. Classification of Short Cracks

What is a short crack, or more appropriately, what is the short crack regime? There is no one universal definition applicable to all cases, however, a general classification has arisen over the last two decades that can be schematically represented as in Figure 9.³⁷ This figure depicts the short crack regime consisting of two main regions. The microstructurally short crack region refers to those cracks that are small in length relative to such features as grain size. The physically short crack region includes those cracks that are generally less than one millimeter in length.

Although both short crack regions possess many common characteristics, the differences between the two are apparent if one examines the separate stages of growth more closely.

3.2.2. Stages of Short Crack Growth

3.2.2.1. Stage I Growth

Microstructurally short crack growth, often referred to as Stage I growth, deals with the propagation of cracks that do not abide by Linear Elastic Fracture Mechanics (LEFM) principles. It was discovered that when trying to apply LEFM principles to short cracks, linear elastic models largely underestimated the growth rates typically observed under experimental conditions. Since the discovery of anomalous short crack behavior, extensive work has been performed in order to explain the possible reasons for the phenomena.^{38,39} A summary of two of the more important reasons for LEFM discrepancy is as follows:

1. LEFM relies heavily on continuum mechanics principles for crack propagation. However, when dealing with Stage I short crack growth, the crack is so small that features such as second phase particles impede crack propagation. This ultimately results in the breakdown of the material continuum and causes discrete behavior at such small crack sizes.
2. LEFM principles incorporate small scale yielding into its propagation model. In other words, LEFM assumes that the crack tip plastic zone size is orders of magnitude smaller than the actual crack length. In the case of microstructurally

small cracks, the plastic zone size is comparable to that of the crack length and hence the LEFM requirement of confined crack tip plasticity is not supported.

Due to the impediment of Stage I crack propagation by such features as grain boundaries and intragranular precipitates, the resulting growth rate is often irregular in nature and is governed by such features as grain size, particle size and particle distribution.^{30, 40} When an obstacle is encountered by the crack tip plastic zone, the growth rate tends to decrease followed by rapid acceleration upon traversing the barrier. It is this irregular growth rate that gives the Stage-I crack a faceted fracture surface commonly observed during early crack growth.

3.2.2.2.   *Stage II Growth*

The main difference between Stage I and Stage II short growth is the mode of plastic deformation. The onset of Stage II growth occurs when the crack tip plastic zone size surpasses that of the actual grain size. As the crack length increases, microstructural features become less effective, crack tip deformation increases, and propagation starts to tend toward continuum mechanics principles.⁴¹ When the plastic zone size reaches approximately one grain size, plastic deformation becomes more uniform and results in the activation of more than one active slip plane. It has been observed by Radhakrishnan *et al.* that, after approximately 10 grain diameters, the plastic zone size was sufficiently large to allow Stage II growth to commence.⁴⁰ This is illustrated in Figure 10 which depicts a plastic zone size equivalent to that of one grain diameter.⁴⁰ Important to note is that once Stage II commences the mode of shear deformation changes to that of mode I deformation, which occurs normal to the applied stress.

3.2.3. Factors Affecting Short Crack Growth

3.2.3.1. *Microstructure*

The most obvious feature that is capable of altering the short crack growth behavior of a material is its microstructure. Several researchers have shown that by reducing the grain size of a component a decrease in short crack growth rate will result.^{40,42} This is due to the higher yield strength imparted into the material as a result of the finer grain size, which in turn means that the shear strength is also increased during the early stages of short crack growth. In effect, by decreasing the grain size the propagating crack will encounter a higher fraction of grain boundaries that will ultimately slow down the Stage I crack growth rate. Furthermore, a smaller grain size also dictates that the transition from Stage I to Stage II will occur much sooner at smaller crack lengths. This is due to the smaller crack tip plastic zone size required in finer grain sized materials essential for homogeneous deformation to begin.⁴⁰ However, one downfall of materials with finer grain sizes is that the long crack growth rate, as well as the creep crack growth rate, increases proportionally.¹⁰ The increase in creep rate is obvious as finer grain sizes dictate higher fractions of grain boundary area that are susceptible to grain boundary sliding and diffusional creep.

Another important pair of microstructural features that can drastically affect the short crack growth rate properties of a material are the precipitate size and morphology. There is an optimum precipitate size where γ' precipitates effectively impede crack propagation by forcing dislocations to loop around the particles resulting in fatigue hardening.¹³ The main advantage of hardening is the promotion of slip along other possible planes. This

results in more uniform deformation and effectively slows down the early stages of crack growth. However, one study has shown that in the case of decreasing the precipitate size for Waspaloy (0.1 μm to 0.01 μm), significant reduction in crack propagation rates have been yielded for short crack growth.⁴³ The results suggest that the shearing of finer precipitates produces well-defined planar slip whereas the looping of larger precipitates resulted in wavy glide. The general trend observed was that slower crack propagation rates were observed with increasing slip planarity. However, the exact mechanism for improved crack growth resistance due to improved planar slip was not presented. It should also be mentioned that the fatigue tests performed were conducted using compact tension specimens, which are not ideal for short crack growth monitoring.

Studies have also shown that γ' precipitates located at the grain boundaries have been effective in decreasing short crack growth rate properties. These intergranular precipitates can effectively modify the grain boundaries from a planar morphology to a more wavy structure providing for a more torturous crack path.¹⁰ Ideally, a combination of a fine grain size with increased grain boundary irregularity can lead to substantial improvements in short crack growth resistance.⁴³

3.2.3.2. Crack Closure Effects

In addition to microstructural effects, the crack closure behavior of fatigued components has also been known to influence short crack growth properties. Numerous researchers have reported that the absence of closure effects during early fatigue are partially responsible for the faster short crack growth rates observed.^{44,45,46} In short, closure

effects occur as a result of the crack faces contacting before the crack opening load drops back to zero. This implies that upon subsequent loading there is a minimum stress (σ_{op}) required to separate the crack faces, i.e., the crack does not open instantaneously and is not exposed to the atmosphere at stress levels below σ_{op} . A schematic of this minimum stress level required to open the crack is provided in Figure 11.⁴⁷ In terms of crack growth, this closure effect also influences the minimum stress intensity level required for crack growth to occur. The effective stress intensity factor (ΔK_{eff}) can be defined by the following equation,

$$\Delta K_{eff} = K_{max} - K_{op} \dots\dots\dots(1)$$

where K_{max} is the maximum stress intensity factor observed at peak loading and K_{op} is the stress intensity factor corresponding to the minimum load required to open the crack faces.

There are three possible mechanisms that may be accountable for the crack closure effect. These are 1) plasticity induced closure, 2) oxide induced closure, and 3) roughness induced closure. Plasticity induced closure is caused by the plastic wake generated at the crack tip.⁴⁷ This plastic deformation causes a permanent expansion of the material within the crack tip plastic zone. Closure occurs behind the crack tip as it propagates through the expanded deformation zone.

Oxide induced closure is caused by the build-up of corrosion products in the wake of the crack.³⁹ This corrosive debris acts as a wedge between the two crack faces and allows closure to occur sooner than for non corroded crack surfaces. Roughness induced closure

is caused by the impingement of faceted fracture surfaces on one another.³⁹ The two mating surfaces of the crack do not join perfectly and result in premature closure. A more detailed description of the three possible closure mechanisms is provided in Figure 12.⁴⁸

One of the main reasons suggested for the faster crack growth rates of short cracks, as opposed to long cracks, is the less pronounced effect of plasticity induced crack closure. Short cracks have smaller plastic wakes and hence the mechanism of plasticity induced closure is less prominent.³⁹ This results in a higher ΔK_{eff} for short cracks which in turn implies a faster crack growth rate.

Research has shown that the extent of crack closure is dependant on the actual crack size.⁴⁹ James and Morris monitored crack tip opening displacement (CTOD) at zero load for numerous crack sizes ranging from 50 to 500 μm . They concluded that for cracks less than 160 μm in length the extent of closure decreased with decreasing crack size.

Furthermore, Newman observed that crack closure varied with the extent of plastic deformation at the crack tip.³⁴ His findings showed that as the closure effect increased, ΔK_{eff} decreased, resulting in a smaller plastic zone size (w). The fully reversed plastic zone size corrected with respect to the crack closure effect is given by the following equation,

$$w = \left(1 - \frac{\sigma_{op}}{\sigma_{max}} \right)^2 \cdot \frac{\rho}{4} \dots\dots\dots(2)$$

where ρ is the monotonic plastic zone size, σ_{op} is the minimum stress required to open the crack, and σ_{max} is the stress at maximum load. It is apparent from this equation that as the closure effect increases, σ_{op} increases and results in a reduction of the crack tip plastic zone size.

3.2.3.3. *Notch Effects*

The majority of fatigue failures in gas turbine discs occur due to local stress concentration factors located at notches within the component geometry. These notches are usually present in turbine discs as either bolt holes or rim serrations. Plasticity located at the notch root has been found to increase the short crack growth rate, in particular, the Stage II crack growth rate.^{50,51} The effect of notch root plasticity ceases as the crack passes through the plastic region and exhibits a slower growth rate. Ewalds explains this decrease in growth rate to be directly related to the large plastic strain range located within the notch root plastic field.⁴⁷ This larger plastic strain range can be tied to the plasticity effects of crack closure as explained in the previous section. In some cases, it has been shown that the effect of notch root plasticity can be so pronounced that when the crack passes through the notch root plastic field, propagation actually terminates.⁵² Upon traversing the notch root plastic zone, crack propagation will only resume if the crack tip plasticity effects are high enough to drive the crack.

3.2.3.4. *Environmental Effects*

The severity of environmental effects on short crack growth behavior can be directly related to the exposure of the crack tip to the testing environment. As previously mentioned, corrosion of the crack faces can result in crack closure phenomena that alter

propagation rates. When crack closure is present, the exposure time of the crack tip is directly related to the stress required to open the two crack faces. Higher σ_o levels imply shorter exposure time of the crack tip to the environment. Hence, crack propagation rates for the short crack regime should be higher than that of long cracks due to the absence of oxidation induced crack closure.³⁹

In addition to corrosive environmental effects, high temperature effects can also influence the early stages of short crack growth. At elevated temperatures, the potential for cross slip deformation to occur increases, which in turn promotes more homogenous slip.⁵³ Statistically speaking, by increasing the number of possible slip planes the likelihood of crack propagation bypassing impeding barriers increases, and thus yields higher crack propagation rates.

3.2.4. Short Crack Growth Models

As previously mentioned, conventional LEFM models commonly used to describe long crack growth behavior are unsuitable for application within the short crack regime. Although there is some discrepancy as to the actual mechanisms responsible for the short crack growth anomaly, the general consensus is that the growth behavior is by no means linear and that more appropriate models are required.

To date, there have been numerous models proposed that incorporate nonlinear, or elastic-plastic fracture mechanics concepts. Among these, some of the more commonly used are the J-integral⁵⁴, crack closure⁵⁵, and crack tip opening displacement (CTOD)⁵⁶ methods. Perhaps one of the most suitable models proposed for short crack

characterization is that proposed by Newman that describes semi-elliptical cracks growing from a semi-circular notch.³⁴ Newman's model characterizes short crack growth in a notched specimen with the aid of a non-linear stress intensity factor (ΔK) that was formulated from numerous sources. These sources include equations generated by means of finite element analysis, weight function methods, and from previously developed equations that were initially formulated for similar crack configurations in an open hole. A more detailed description of Newman's short crack growth model has been provided in Appendix A. In an attempt to validate Newman's work, Beres and Koul compared their results obtained using J-integral and displacement approximation methods to those determined using Newman's ΔK model.⁵⁷ After comparisons were made for several crack sizes, it was concluded that the model was indeed accurate for determining the stress intensity of short cracks in notched members.

Inasmuch as it is important to have an accurate model for describing the crack tip stress state during testing, it is just as critical to be able to accurately measure the actual crack size. For this reason, considerable effort has been given to the development of reliable crack measuring techniques during fatigue testing.

3.3. Short Crack Measurement Methods

Accurate knowledge of the crack size and propagation rate in a turbine disc as a function of the applied stress field are critical qualifying factors for the damage tolerant approach to engineering fatigue assessment. Initially crack measurements were made utilizing optical methods and simple fracture surface analysis. Since then, the three most popular

methods for monitoring crack growth have been in-situ microscopy, replication, and potential difference techniques.

3.3.1. In-situ Microscopy

Scanning electron microscopy (SEM) and transmission electron microscopy (TEM) have been used extensively for monitoring crack initiation and early crack growth behavior. It is a useful crack measurement technique as it not only supplies the observer with changes in crack growth, but it can also provide other important characterizing features such as the mode of crack initiation. Stephens *et al.* used in-situ SEM microscopy to characterize the short crack growth behavior of a Ni-base superalloy at both room and elevated temperatures.⁵³ The advantages of using this technique include such features as extremely high resolution ($\sim 0.1 \mu\text{m}$) and complete microstructural characterization; however there are several disadvantages that arise. Fatigue specimen size, as well as load train attachments, are severely restricted in terms of size as they must be able to fit within the confines of the SEM environment. In addition to size restrictions, the inert nature of the SEM chamber also eliminates the possibility of fatigue testing under normal corrosive environments. One final disadvantage to using in-situ microscopy is that the technique is only capable of measuring surface crack features and completely disregards subsurface crack depth measurements. This restricts the crack growth to a 2D plane and does not represent true crack front properties.

3.3.2. Replication

Among the different replication techniques used for crack measurement, the plastic replication method is probably the most widely used. Cellulose acetate films that have

been softened with acetone are used to replicate the surface of the specimen for the purpose of early crack detection. Swain was able to successfully characterize the short crack behavior of several different aluminum alloys utilizing this technique.³² The procedure involves test interruption followed by acetate film application that is capable of conforming to the crack opening topography. The replica is then removed for microscopic examination. A schematic of the replica procedure used to monitor crack growth in notched members is depicted in Figure 13.³² The resolution of this procedure is extremely high and is comparable to that of in-situ microscopy techniques (0.1 μm). The most obvious drawback to using the replica technique is the necessity to interrupt the fatigue test for acetate film application. The specimens must be held under significant loads for various dwell times that may result in deviations from true crack growth behavior.

3.3.3. Potential Difference Methods

The existence of a crack in an electric field can introduce a perturbation which, if measured, can be interpreted in terms of crack size and shape.⁵⁸ The source of the electric field may be supplied by means of direct potential or alternating potential. Essentially, a constant current is passed through the fatigue specimen and potential drop measurements are made via a voltage probe that is strategically placed in the vicinity of probable crack initiation location. As the crack begins to grow, the uncracked cross-sectional area of the specimen decreases and as a result the electrical resistance of the specimen increases. From another viewpoint, one may look at the presence of a crack resulting in an increase in the path length of current flow within the specimen. As the current path length increases with increasing crack size the conductivity decreases proportionally. Unlike the

crack monitoring techniques of in-situ microscopy and replication, potential drop methods are suitable for fatigue testing conditions that do not permit test interruption for acquisition and are very suited for testing in high temperature corrosive environments.

3.3.3.1. Direct Current Potential Difference (DCPD)

The main requirement for the DCPD technique is the input of a stable, constant direct current throughout the body of the test specimen. The stability of the current source is critical in order to assure that the electric field generated within in the specimen does not fluctuate. As the crack advances, the cross sectional area through which the current can flow decreases. This results in an increase in the specimen resistance that can be described by the following equation ⁵⁹,

$$R = \frac{\rho L}{A} \dots\dots\dots (3)$$

where ρ is the material resistivity, L is the current path length, and A is the cross sectional area available for current flow. If the current source (I) is maintained constant and the resistance (R) continues to rise with increasing crack size, the potential drop (V) measured across the crack is determined by Ohm's law ⁵⁹,

$$V = I \cdot R \dots\dots\dots (4)$$

The correlation between the actual crack size and the potential drop measured can be made after performing several calibration tests.

The main advantage of the DCPD technique is its simple, and relatively low cost instrumentation that is very suitable for long term, high temperature testing. ⁵⁸ It has also

been proven to be a reliable means of measuring propagation rates for short cracks ranging in size from 50-5000 μm .⁶⁰ However, since most metals are excellent conductors of electricity, high direct currents are required to generate a reasonable signal-to-noise ratio. The requirement for high d.c. current levels is further magnified as specimen size increases and the relative voltage differences between cracked and uncracked specimens are low. In some cases, direct currents up to 100 amperes have been used for high temperature testing.⁶¹ Using high direct currents are disadvantageous as they result in extreme heating of current leads and specimen contact points. Not only does this directly affect the resistance of the lead material but it may also contribute to thermal emf problems.⁶² In order to compensate for such problems, as well as instrumentation drift and current variation, a reference voltage is generally used. This reference voltage is measured from a location that does not incur crack growth but is exposed to the exact same environmental parameters.⁵⁹

3.3.3.2.   *Alternating Current Potential Difference (ACPD)*

Fundamentally the ACPD crack monitoring procedure is comparable to that of the DCPD technique. Similarly, the procedure involves passing a current through the test specimen with potential drop measurements taken across the site of crack initiation. However, the use of an alternating current source produces more remarkable electrical phenomena than those generated by DCPD instrumentation.

When a direct current is passed through a conductor the current density is uniformly distributed throughout the whole of the specimen's cross-section. However, when an

alternating current is passed through a conductor the current density is only distributed throughout the specimen surface. For clarity, the difference between AC and DC distribution is depicted in Figure 14.⁶⁶ The current density is strongest at the specimen surface and decays to zero at some characteristic distance into its thickness. The confinement of the current density to a thin surface layer is known as the “skin effect”.⁶³ This skin effect is a direct result of the greater fraction of induction occurring in the mid-section of the conductor. As the alternating current changes direction, the magnetic field collapses from the surface exterior towards the center of the specimen. As a result, more magnetic lines of flux collapse through the center of the specimen, producing an area of higher impedance.⁶⁴ This area of high impedance forces the current to flow through the path of least resistance, which is the outer surface layer of the specimen. Figure 15 depicts the skin effect and the balloon-shaped current distribution that is generated between the two current leads.⁶⁵

The extent of the skin effect is usually quantified in terms of skin depth. The characteristic skin depth (δ) is defined by the following equation as,

$$\delta = \sqrt{\frac{\rho}{\pi \mu_r \mu_o f}} \dots\dots\dots (6)$$

where ρ is the resistivity of the current carrying conductor (Ωm), μ_r is the relative permeability, μ_o is the permeability of vacuum ($4\pi \times 10^{-7} \text{ H/m}$), and f is the frequency of the alternating current (Hz). This skin depth measurement corresponds to the distance into the specimen thickness where the current density is only 37% of the surface value, as shown in Figure 16.⁶³ From this equation it is obvious that the skin depth is inversely

proportional to the square root of the applied frequency. In other words, higher alternating frequencies result in smaller skin depths. This feature is highly advantageous for short crack measurement as it confines the majority of the current to the specimen surface, making it highly sensitive to surface breaking cracks. However, one drawback to using extremely high frequencies is the strong magnetic field that is induced. In high magnetic fields, the voltage probes used to measure crack size are more susceptible to picking up induced voltage readings, that when superimposed on the actual crack voltage results in an overestimation of the true crack size.⁶⁶

Another advantage of the ACPD technique, with regards to the skin effect, is the ability to use much lower currents compared to those used in DCPD applications. Tiku found this to be an extremely useful feature for highly conductive metals such as aluminum, due to the relatively large current inputs that are normally required to generate detectable voltage measurements.⁶⁷ Also, the ability to use lower input currents negates excessive overheating in the lead material and at the contact junction points of the specimen.

As already mentioned, when using high frequency current sources there is a problem with magnetic field induction. Charlesworth *et al.* reported pick-up voltages directly proportional to the frequency used during testing.⁶⁸ They reported that the main cause for voltage pick-up was due to looping in the voltage probe wires. Significant reduction in voltage pick-up was attained by minimizing the loop area between the voltage leads and the specimen surface. This was effectively done by twisting the lead wires together as close as possible to the specimen surface, leaving enough lead material for attachment

across the crack area. Extensive experimental work performed by Tikunov showed that the voltage probe design, as well as probe attachment to the specimen surface, were critical factors in obtaining repeatable, low-noise, ACPD measurements.⁶⁷

In addition to optimizing voltage probe design, considerable effort has been expended in enhancing the current lead design used in ACPD applications. As already shown in Figure 15, due to the skin effect the AC current distribution is confined to a thin surface layer in the specimen. From this figure it can be reasoned that the specimen thickness will not diversely affect the current distribution; however, the specimen surface area will have a direct effect. Clearly, as the specimen surface area is increased, the current density decreases if the current input is maintained constant. Lower current densities will result in smaller potential drop measurements, which in turn results in lower signal-to-noise ratios. This eventuates lower ACPD resolution and impairs short crack measurement capabilities. Unvala has patented a novel technique known as current focusing to overcome this problem.⁶⁹

The aim of the current focusing technique is to prevent current distribution over the total specimen surface area and instead, confine it to a narrow region around the crack initiation site. The technique is based on the principle that when a conductor carrying an AC current is placed in close proximity to the return conductor, e.g. specimen, then the distribution of the AC current in the return conductor is confined and tends to be a mirror image of the current flowing in the upper conductor.⁷⁰ The current focusing technique is shown in Figure 17, which depicts the much narrower region of current distribution

residing below the current lead wire. The much more confined current distribution along the specimen surface is a direct result of the electric field being concentrated between the current carrying wire and the specimen. Effectively, the magnetic field generated forces the current in the specimen to flow directly below the current wire as illustrated in the cross-sectional view provided in Figure 18.⁷⁰

Although the ACPD technique appears to have more advantageous features for short crack measurement applications, it should be noted that the initial setup is far more laborious than DCPD systems. Voltage probe design, current lead design, and lead placement and attachment must be scrupulously carried out and capable of repetition for subsequent fatigue tests. Another aspect that requires meticulous attention is the crack calibration procedure. Prior to actual testing, the correlation between ACPD voltage and crack length must be determined. Numerous researchers have determined that the correlation between probe voltage and crack length follows a linear relationship.^{50,68,71} However, Tikunov, Hwang and Ballinger have reported that their calibration studies exhibited a 2nd order polynomial relationship.^{67,72} Furthermore, Lu *et al.* have reported that, after calibration testing, the correlation between crack size and voltage followed a power law relationship.⁷³ It is apparent that considerable discrepancy exists between the aforementioned studies which further magnifies the need to pay careful attention during calibration testing.

4. Literature Review – Damage Tolerance of Gas Turbine Discs

The concept of damage tolerance was first introduced into the gas turbine field as a direct result of the Engine Structural Integrity Program (ENSIP), which was a program started by the U.S. Air Force in the early 1970's.⁷⁴ The main focus of the ENSIP program was to ensure the ability of critical engine components to resist failure due to the presence of flaws, cracks, or other damage for a specified period of unrepaired usage. The main driving force behind the initiation of the program was the growing number of engine structural integrity problems that were arising due to material property deficiencies. It was becoming more evident that, due to increasing operational requirements, traditional safe-life design concepts were unsuitable for engine design and component life management processes.

4.1. Damage Tolerant Design Requirements

Engine structural integrity is accomplished by ensuring safety through the use of damage tolerant design. As mentioned in section 1.2, the critical aspect of the DTD philosophy is the length of the SII used during maintenance. Short inspection intervals dictate longer maintenance downtime and make the process unfeasible. Hence, materials that have improved creep resistance and slower crack growth rate properties are not only compulsory for new safety standards, but are essential to the successful application of the DTD lifing methodology. To fulfill these requirements, researchers have attempted to improve the fatigue/creep properties of commonly used disc materials through microstructural modification techniques.

4.2. Microstructural Modification Techniques

Conventional thermal processing methods for turbine disc forgings generally involve solutioning above the γ' solvus temperature followed by single or multiple-step aging heat treatments. Disc manufacturers have concentrated on maximizing the proof strength and LCF properties of wrought discs by producing microstructures of very fine grain size with ultra-fine secondary precipitate size. With time, manufacturers have been able to produce lighter alloys with superior strength and crack initiation properties. However, one unattractive characteristic of these stronger disc materials has been decreased resistance to crack growth.⁷⁵ Hence, disc producers must now design components with a compromise between improving the fatigue crack growth resistance for improved damage tolerance while maintaining acceptable levels of other mechanical properties, such as tensile strength and low cycle fatigue crack initiation lifetimes. Over the past fifteen years, researchers have given considerable thought to microstructural modification techniques required to attain well-balanced mechanical properties. Damage tolerant microstructures have been developed by numerous methods including, thermal-mechanical processing, compositional modifications, as well as modifications to conventional heat treatment practices.^{3, 76, 77, 78} Provided in Table 3 is a short list of some studies done by various researchers in the damage tolerant field.

For the most part, successful damage tolerant designs have been achieved by modifying conventional heat treatment procedures to optimize such microstructural features as grain size, grain boundary morphology, as well as precipitate size and morphology. The following case studies have been selected to show the various thermal processing

techniques implemented to attain suitably tailored microstructures for damage tolerance applications.

4.2.1. Case Study 1: Damage Tolerant Design of IN 718

Inconel 718, in its conventional form, has been used extensively as a disc alloy for numerous gas turbine engines. In an attempt to improve the damage tolerance of this commonly used alloy, Koul *et al.* have designed a modified heat treatment that promotes considerable improvements in crack growth rate properties at elevated temperatures.³ The improved damage tolerance of this alloy was achieved by thermal processing methods that produced optimum combinations of grain size, grain boundary morphology, and precipitate morphology. The conventional heat treatment for Inconel 718 involves solutioning for one hour at 955 °C, then direct aging at 718 °C for 8 hours with a final direct aging step at 621 °C for 8 hours. Although this treatment proved to be satisfactory for safe-life design applications, the resulting microstructure exhibited high crack growth rate properties inadequate for damage tolerant design criteria.

Koul *et al.* showed through laborious experimental work, that a serrated grain boundary morphology could be attained by solutioning at a slightly higher temperature (1032°C/1hr) with controlled furnace cooling to 843 °C and holding for 4 hours.³ The modified solutioning heat treatment and subsequent controlled cooling promoted the precipitation of δ -Ni₃Nb needles along the grain boundaries that resulted in a zig-zag boundary morphology, as opposed to the relatively straight morphology of the conventional microstructure. Figure 19 is the typical zig-zag grain boundary serration that

results from intergranular precipitation. It should also be noted that the modified solutioning heat treatment was chosen based on experimental results which showed that at 1032 °C the second phase intermetallic precipitates could be fully solutioned without any excessive grain growth effects.³ The prevention of excessive grain growth was vital to the preservation of LCF initiation life.

In addition to optimizing the grain boundary morphology, Koul *et al.* subsequently implemented an additional heat treatment sequence for optimizing γ' and γ'' precipitates to further strengthen the intragranular region without degenerating the serrated grain boundary morphology. The sequence involved partially solutioning the new serrated microstructure to dissolve the coarse intragranular γ'' precipitates (as seen in Figure 19) that formed during the serration heat treatment step.³ Removing these coarse precipitates was deemed necessary as they may act as initiation sites for premature LCF failure. This was subsequently followed by the conventional 2-step aging sequence required to optimize γ' and γ'' size for optimum strength.

After metallographic examination, the overall microstructural changes included a new serrated grain boundary morphology, slightly larger grain size, and an optimized intergranular precipitate size and shape. Mechanical testing showed that the new DTM improved the FCGR by a factor of 2 at 650 °C and the CCGR was decreased by a factor of 5 at the same temperature. These improvements were made with minimal penalties to LCF properties and marginal decrease in the material's proof strength (21% drop).³ It can

be deduced that the presence of grain boundary serrations helped to inhibit grain boundary sliding and provide a more tortuous crack path in the case of intergranular fatigue. In addition, the larger grain size may have also aided in the material's improved creep resistance but adversely affected the yield strength. The author also suggested that the heavy grain boundary precipitation was responsible for decreasing the proof strength of the new DTM, as it depleted Nb from the matrix, which lowers the level of solid solution strengthening.³

4.2.2. Case Study 2: Damage Tolerant Design of MERL 76

In another study performed by Chang *et al.*, a successful DTM was produced for the P/M Ni-base disc alloy, MERL 76. The DTM was also developed by modifying the conventional heat treatment procedure. The conventional heat treatment for MERL 76 involves solutioning at 1143 °C/2hr/OQ followed by a stabilization step of 982°C/1hr/fan cooled, with a final aging sequence at 732 °C/8hr/AC.⁷⁶ Again, it was discovered that through supersolvus solutioning at 1190 °C and controlled furnace cooling to an intermediate temperature, the straight grain boundaries were transformed into a serrated morphology.⁷⁶ An SEM image of this new serrated morphology has been provided in Figure 20. In contrast to the serrations formed in the previous study, the grain boundaries for this DTM have a more wavy-like nature due to the precipitation and growth of elliptical γ' precipitates in the grain boundary region.

Furthermore, optimum γ' size and shape were attained by omitting the stabilization step and implementing the conventional single-stage aging sequence. Although the same

aging sequence was applied to the DTM, the resulting γ' morphology is quite different from the CM. The new DTM primary γ' have a homogeneously dispersed rosette structure along with a much finer secondary γ' precipitate size as seen in Figure 21.⁷⁶

In addition to the new serrated grain boundary morphology and optimized γ' , there was also a substantial increase in grain size (10 μm to 30 μm). Tensile tests performed showed that this larger grain size resulted in a 7% decrease in proof strength. However, LCF tests conducted at 700 °C revealed that this larger grain size did not diversely affect the crack initiation lifetimes of the DTM with respect to LCF properties of the CM.⁷⁶ Fatigue tests showed that, at 700 °C, the FCGR was approximately 10 times slower and creep tests performed indicated that at 760 °C the CCGR was also lowered by a factor of 10.⁷⁶

Slower CGR's were directly linked to the new mixed mode of crack growth for the new DTM. For the CM, the mode of crack propagation was always intergranular, whereas for the DTM crack growth occurred both intergranularly and to a greater degree, transgranularly. It was also concluded that the slower transgranular fracture mode was due to the suppression of grain boundary sliding by the serrated grain boundary morphology.⁷⁶

It is obvious from both case studies presented that the addition of a serrated grain boundary morphology was vital to improving the damage tolerance of both disc

materials. For this reason, the possible mechanisms of serration formation will be examined more closely.

4.2.3. Mechanisms for Serrated Grain Boundary Formation

Controlled furnace cooling heat treatments have been implemented by numerous researchers to impart grain boundary serrations in a wide range of Ni-base superalloys^{79,80,81}. In the various studies performed, the resulting serrated boundaries have proven to be beneficial for improving such features as impact strength, fatigue resistance, and creep ductility. There have been numerous proposed theories that deal with how grain boundary serrations form. Several models that propose to explain the formation of wavy grain boundary serrations (as seen in Case 2 in the previous section) will be examined in more detail for review.

4.2.3.1. Grain Boundary Migration Model

The mechanism of the formation of serrated grain boundaries in Ni-based superalloys is associated with the heterogeneous nucleation of γ' precipitates at the grain boundaries. One theoretical model proposed by Larson and Floreen, relates the formation of serrations to the migration of grain boundary sections in between the primary γ' particles at the grain boundaries.⁸² This grain boundary migration theory is depicted in Figure 22. The basis of this model is that the grain boundary segments continue to migrate until γ' homogeneously nucleates throughout the intragranular region at some lower temperature during the controlled cooling sequence. The main fault with the proposed mechanism is that no scientific explanation can be given as to why the grain boundary segments

maintain mobility when cooling from elevated to lower temperatures. Opponents to this model have argued that once a material has reached a stable grain size at a specific solutioning temperature, cooling from above the γ' solvus is not likely to cause significant boundary migration between the primary γ' precipitates due to migration time dependency.⁸³

4.2.3.2. *Precipitate Migration Model*

Koul *et al.* have proposed another serration formation model that is also related to the γ' precipitation, however, their approach does not concentrate on boundary migration phenomena.⁸⁴

From their research, they have determined that the majority of primary γ' particles responsible for serrations are elliptical in shape and are always located within the concave side of the serration.⁸³ The core of this proposed model is that the primary γ' that have precipitated adjacent to the grain boundary actually partake in migration as opposed to the grain boundaries themselves. This is graphically depicted in Figure 23, which shows the proposed sequence of precipitate movement into the grain boundary. The model suggests that the γ' forms on either side of the grain boundary once the solvus temperature range has been encountered during furnace cooling. In brief, the suggested mechanism of precipitate movement is due to the accommodation of coherency strains between the grain boundary/precipitate interface. The net strain energy differential that results, provides enough driving force to promote γ' precipitate movement in the

direction of the grain boundary. Particle movement is stopped once the line tension of the boundary overcomes the strain differential driving force.⁸⁴ However, it has been determined by Koul *et al.* that only high volume fraction alloys with a sufficiently large lattice mismatch are capable of forming serrations during furnace cooling.⁸³ Furthermore, the proposed strain energy differential model by Koul and Gessinger, states that there is a minimum precipitate size required for serration formation based on the following equation⁸⁴,

$$r_c = \frac{3\gamma_{st}}{4\mu\delta^2} \dots\dots\dots (7)$$

where γ_{st} is the interfacial energy due to the structural distortion caused by the misfit dislocations, δ is the particle-matrix misfit parameter, and μ is the shear modulus of the matrix. It should be noted that the model accounts for the varying effects of cooling rates on serration morphology if one considers the correlation between precipitate size and cooling time.

Utilizing this serration formation model as a starting point, Koul *et al.* have developed a mechanistic microstructural design concept that aims at producing the optimum size and morphology of grain boundary serrations required to produce damage tolerant materials. Once it has been established that the alloy is capable of forming grain boundary serrations the next logical step in DTM design is to determine the size of serrations

required to optimize crack growth resistance. Wu and Koul have developed a constitutive model that predicts the optimum grain boundary serration size that effectively inhibits grain boundary sliding.^{85,86} The model accurately describes the transient creep behavior of complex Ni-base superalloys with respect to the serration amplitude (h) and the serration wavelength (λ). The optimum serration geometry to prevent grain boundary sliding can be approximated by the simplified steady-state creep rate equation,

$$\dot{\epsilon} = \phi A_0 \frac{D\mu b}{kT} \left(\frac{b}{d}\right)^q \left(\frac{l+r}{b}\right)^{q-1} \frac{(\sigma_a - \sigma_{ig})(\sigma_a - \sigma_{ic})}{\mu^2} \dots\dots\dots(8)$$

where A_0 is a non-dimensional constant, l is the grain boundary precipitate spacing, r is the radius of the grain boundary particle, d is the grain size, q is the grain size index, μ is the elastic shear modulus, σ_a is the applied tensile stress, σ_{ig} is the tensile stress opposing dislocation glide, σ_{ic} is the tensile stress opposing dislocation climb, and ϕ is the *grain boundary serration factor* defined by,

$$\phi = \frac{2}{\sqrt{1 + \left(\frac{\pi h}{\lambda}\right)^2}} - 1 \dots\dots\dots(9)$$

There is an optimum range of serration amplitude and wavelength values required to reduce creep rate. From these equations, it is then possible to predict the γ' precipitate

size required to form the appropriate serration h and λ , necessary for suppressing grain boundary sliding phenomena. Equations (7) through (9) define the basic microstructural design criteria.

4.2.3.3. *Precipitate Growth Model*

Danflou *et al.* have proposed another serration formation mechanism that suggests the formation of grain boundary serrations due to the growth of γ' precipitates into the moving grain boundary.⁷⁸ From their work, they propose that some of the γ' precipitates dissolve either partially or completely in contact with the “moving” grain boundaries and such a dissolution contributes to the growth of coarser, more stable γ' precipitates in the opposite grain.⁷⁸ They suggest that the complete dissolution of γ' at the moving grain boundary, followed by reprecipitation behind it, can occur at high-angle grain boundaries during recrystallization. The mechanism proposes that the grain boundary acts as a short-circuit path permitting the solute to reach an equilibrium state more quickly. This proposed serration mechanism has some drawbacks, as it does not explain the formation of grain boundary serrations in the grains opposing grain growth (i.e. serrations that protrude into the growing grain).

More recently, the same authors suggest that the γ' growth along the grain boundary is directly related to dendritic growth. This assumption was made based on their observations with serration formation in Astroloy and Henry's alloy.⁸⁷ From their work, they observed arms, which originate from dendritic particles, pushing the grain boundaries on both sides and consequently forming serrations. It is proposed that these

dendritic fan structures grow due to solute diffusion across the grain boundary and the growth rate is determined by the remaining solute concentration in the neighboring grain. Figure 24 shows an example of this dendritic fan structure observed and its associated grain boundary serration.⁸⁷ It should be noted that this more recent work by the authors does not elaborate on the role of grain boundary migration with respect to serration formation.

5. Materials and Experimental Procedures

5.1. Material

The conventional material used in this study was taken from an approximately 0.6 meter (2-ft.) diameter turbine disc that was provided by Pratt and Whitney Canada (PWC) in a heat treated condition. The nominal composition (wt%) for PWA 1113 is provided in Table 4. The alloy is commonly referred to as “modified Waspaloy”; however, it is more comparable in composition to that of Udimet 720 than to Waspaloy. Because this is a proprietary alloy of PWC, no detailed information was supplied with respect to processing parameters for conventional heat treatment practices. However, it was disclosed that the disc was produced by a forging method known as the Gatorizing process. This is a procedure that involves alloy powder consolidation through extrusion of a canned powder compact followed by isothermal forging. As a result, the CM contains a very fine microduplex γ - γ' grain size (ASTM 10) with a relatively planar grain boundary morphology and a poorly distributed, coarse γ' precipitate structure, as seen in Figure 25. Grain size measurements taken along the cross-section of the as-received disc (see Figure 26), revealed no major deviations in mean intercept distance.

A PHACOMP analysis was also performed on the alloy based on the compositional results from Table 4. The computational results predicted a primary γ' volume fraction of 33%. This value is comparable to that obtained using the Clemex image analysis software, which gave an average γ' volume fraction of 36% over the entire disc cross-section.

5.2. Microstructural Development

The DTM design procedure was based on successful work by Koul *et al.* who developed DTM's for Inconel 718 and MERL 76 with marked improvements in fatigue and creep resistance ^{3,10} Based on their past success, a similar methodology was adopted that involved three sets of experimental heat treatment procedures for optimizing the grain size, grain boundary morphology, and the intragranular γ' morphology.

5.2.1. Grain Size Optimization

Grain size optimization was accomplished by determining the solution treatment conditions required to dissolve all second phase γ' precipitates while preventing excessive grain growth. This procedure involved cutting numerous same-size samples from the as-received disc, solutioning at various temperatures, and then quenching in water to maintain the solutioned microstructure for metallographic examination. Care was taken to ensure that each specimen was positioned in the same location of the furnace with a control thermocouple positioned at the specimen surface for each solutioning heat treatment. Solutioning temperatures ranged from 1050 °C up to 1250 °C, with various solutioning times of 1, 2, and 4 hours. Each sample was cut using a high speed abrasive saw, polished to a 1 μm finish, and electrolytically etched to reveal the grain boundary structure. The most suitable procedure for revealing the grain boundaries in the solutioned material was electrolytically etching at 6.0 volts for 1.5 minutes in a solution of 50% methanol and 50% hydrochloric acid. With each specimen polished and etched, grain size measurements were made using an optical microscope following ASTM E112 specifications.

5.2.2. Grain Boundary Morphology Optimization

Grain boundary morphology was optimized by the determination of a post solutioning heat treatment sequence that defines the controlled-cooling rate range required to promote the formation of a serrated grain boundary structure. The mechanism of the formation of serrated grain boundaries in Ni-based superalloys is associated with the heterogeneous nucleation of γ' precipitates at the grain boundaries. This heterogeneous nucleation is governed by the controlled cooling rate that is used to cool the material from the γ' solvus temperature to some lower intermediate temperature.

Numerous heat treatments were performed to determine the best cooling rate and furnace removal temperature that promotes optimum grain boundary serrations. The first step in each heat treatment was to solutionize the material above the γ' solvus temperature (determined from the grain coarsening study) followed by controlled furnace cooling to an intermediate temperature at which point the specimen is removed from the furnace and air cooled. Three different cooling rates (1, 2 and 3 °C /min) were used to investigate the effect of cooling time on grain boundary serration formation and also on the matrix γ' morphology. Table 6 lists the schedule of cooling rates investigated as well as the end temperatures at which the specimens were removed from the furnace. Each heat treated specimen was sectioned, polished to a 1 μm finish, and etched to reveal the grain boundary structure. The electrolytic etchant used in the grain coarsening study did not clearly reveal the intergranular structure of the furnace cooled material. A more suitable method involved electrolytic etching at 2 volts for 45 seconds in a solution of 10g CuCl_2 ,

40 ml HCl, and 60 ml of ethanol. A metallographic examination of each etched specimen was done with the aid of an SEM.

5.2.3. Intragranular γ' Optimization

The final step in DTM development was the determination of an appropriate aging sequence that produces the ideal amount, size, and distribution of γ' precipitates required to strengthen the γ matrix. The proposed aging sequence is a modification of that used for the similar alloy Udimet 720.

The typical aging sequence for Udimet 720 is a two-step heat treatment of 760 °C/12hr/ air-cooled + 650 °C/24hr/ air-cooled. It is optimistic that only the first step of the sequence need be used for aging the serrated PWA 1113 material. It is postulated that the serration heat treatment implemented prior to aging will reduce the required aging time and the necessity of a two-step treatment. For this reason aging heat treatments at 760 °C were applied to the modified disc material for periods ranging from 1 hour up to 60 hours. Each aged specimen was sectioned for microstructural examination and for hardness indentation tests. Rockwell C hardness measurements were taken from each aged specimen to determine the aging time required to impart peak hardness in the material. Two pieces of each heat treated specimen were mounted for microstructural evaluation and were polished to a 1 μ m finish. Grain boundary microstructures were again prepared by electrolytic etching at 2 volts for 45 seconds in a solution of 10g CuCl₂, 40 ml HCl, and 60 ml of ethanol. The γ' precipitate morphology was examined using another electrolytic etching procedure that involved etching at 6.5 volts for 2

seconds in a solution of 12 ml H_3PO_4 , 40 ml HNO_3 and 48 ml of H_2SO_4 . All etched specimens were examined using an SEM at suitable magnifications.

5.3. Fatigue Testing

Short crack growth fatigue tests were performed using single edge notched (SEN) specimens in accordance with the ASTM-E647 specifications. Testing parameters included a loading frequency of 5 Hz, a zero stress ratio with a maximum concentrated stress of 1520 MPa (220 ksi), and a triangular waveform loading pattern. As seen in Figure 27, the SEN notch radius was 1.2 mm (0.049 in.) to simulate the K_T factor of 2.68 that is commonly found in the critical regions of the disc. Three specimens from each microstructural condition were tested at room temperature and 482 °C. A new high temperature (max. 1400 °C) single zone furnace was procured to conduct the 482 °C tests.

5.3.1. Furnace Calibration

Before testing could be started, it was required to determine the temperature profile of the new furnace being utilized. Specifically, the furnace set point that provides a specimen surface temperature of 482 °C in the gauge section had to be determined. This process was done using an SEN specimen with five thermocouples welded to the surface. Each thermocouple is incrementally spaced across the notch side of the specimen covering the entire gauge length. This specimen was then placed in the furnace and monitored using a multi-channel scanner until the appropriate set point was found, as shown in Figure 28. It was immediately obvious that a large heat sink was present in the load train and modifications would be required. Initially it was thought that the water-cooled grips were the source of heat dissipation; however, upon closer examination this possible source was

eliminated. It was found that the specimen extension pieces were the main cause of heat dissipation from the furnace. These small extension pieces conducted the heat from the specimen to the outside atmosphere very effectively. This effect was reduced by slowing down the rate of heat loss by insulating the area of the extension pieces located outside of the furnace. To ensure that this process was repeatable, the insulation was made from solid alumina brick, which was machined to fit around the extension pieces tightly, as depicted in Figure 29. The insulation of the extension pieces effectively reduced the required temperature setting of the furnace by almost 200 °C. Figure 30 shows the temperature profile of the furnace along the length of the specimen at furnace set-point of 685 °C. It was found that at this set-point the area of the specimen notch achieved a temperature of ~ 482 °C as required. The specimen profile has also been included in the graph to show the actual position with respect to temperature.

One of the main reasons for ensuring a repeatable high temperature environment was to reduce any temperature fluctuations during elevated temperature testing. Large deviations in temperature can result in fluctuating thermal emf's produced by the furnace. These thermal emf's must be kept constant to ensure stability in the crack monitoring technique employed. The following section describes the ACPD equipment used for crack monitoring and the technique developed to monitor the short crack growth behavior.

5.3.2. ACPD Crack Monitoring Technique

5.3.2.1. ACPD Equipment

The ACPD unit used during fatigue testing was a CGM-7 crack growth monitor produced by Matelect Ltd. The CGM-7 components consist basically of an oscillator, current power source, pre-amplifier, phase sensitive detector, and filters. A simplified block diagram for the unit has been provided in Figure 31. The following is a brief description of the major components and their function in crack growth monitoring.

The oscillator is capable of generating a very stable sinusoidal waveform of variable frequency ranging from 500 Hz to 240 kHz. The current source is an advanced power amplifier that delivers a constant current of up to 5 amperes independent of the load impedance (i.e. resistance of specimen and lead wires).

The pre-amplifier selected for use with the CGM-7 system was a high-input impedance preamplifier. This particular preamplifier was chosen because it has the ability to eliminate junction effects (i.e. change in contact resistivity between the voltage leads and the specimen), and because it also accounts for any changes of resistivity in voltage lead material (e.g. due to temperature changes etc).

The incorporation of a phase sensitive detector in the CGM-7 unit is required to demodulate the AC voltage readings taken from the specimen during testing. The probe voltage readings taken will be of the same frequency as the input current; however, the

phase of the signal will be offset. This phase difference is due to inductance in the specimen, lead wires and the characteristic skin effect produced by the alternating current. Also, there is an additional amount of voltage pick-up by the probes due to the electromagnetic radiation generated by the current flowing in the specimen. Figure 32 shows a schematic of the ACPD phase components with respect to current and inductance. In this figure, OC represents the in-phase current component and OD is the in-phase inductive component of the ACPD signal. The ideal ACPD signal, OA, (where $OA = (OC^2 + OD^2)^{1/2}$) is at 45 ° to the current phase. In the case of inductive pick-up, the actual ACPD signal is offset by an amount proportional to the voltage pick-up, AB, present in the system. Thus, relating the actual ACPD signal, OB, to the crack size will result in overestimation due to voltage pick-up. This pick-up tends to be more severe with increasing current frequency. It is apparent that in both cases, the 0° current phase component, OC, does not get affected by the inductive pick-up present. Hence, by utilizing the phase sensitive detector, one can measure the 0° current component and relate its changing value to crack growth. In the fatigue tests performed, the Channel 1-0° ACPD component was used to measure the changing crack size, and the Channel 2-0° ACPD component was used as a reference value where no crack growth is observed.

After the signals have been demodulated, they are converted to a DC value. Incorporated in the CGM-7 is a set of filters that are capable of blocking any AC noise generated and allowing the DC signals to pass through. The filters can be set to cut-off frequencies ranging from 1 Hz up to 20 kHz. In the tests performed, all crack monitoring was recorded using the maximum filter setting of 1 Hz.

5.3.2.2.   *ACPD Probe Design*

As previously mentioned, high frequency alternating current has the disadvantage of producing an emf signal that has a large influence on potential measurements. However, to optimize the skin effect and thus increase the sensitivity of short crack detection, it is necessary to use high frequency currents. In theory, the phase sensitive detector should be capable of eliminating all the inductive pick-up present in the system. However, ACPD studies conducted by Tiku, using a similar CGM unit, have shown that signal demodulation is not always 100% effective.⁶⁷ In fact, it was determined that the design and positioning of ACPD probes were vital factors in obtaining accurate crack measurement readings. For this reason, it is essential to design current and voltage probes such that the emf pick-up by the voltage probe wires is minimized.

5.3.2.2.1.   *Current Probe Design*

The most obvious criteria for current probe design was the selection of a lead material that has a good current carrying capacity, as well as high temperature capability. The lead material must also be rigid enough to prevent current probe movement during fatigue testing and yet have enough ductility for wire twisting during probe assembly.

Experiments were performed with various gauge sizes of silver and Chromel (Ni-Cr) lead wire to determine which was most suitable for room and elevated temperature applications. The silver wire had extremely good conductivity; however, it proved to be quite difficult to spot weld to the specimen surface. The silver wire was also very easily deformed which proved to be quite problematic with respect to repetitiveness of wire

twisting and probe placement. To alleviate this problem, the less ductile, more weldable Chromel wire was examined.

The Chromel wire exhibited good weldability characteristics and moderate conductivity ($122.62 \mu\Omega\cdot\text{cm}$). Previous testing at NRC demonstrated that Chromel wire was insufficient as a current lead material. Past tests using this material have shown that its higher resistivity (*c.f.* silver wire) limited the magnitude of current capable of being carried to the test specimen. However, experimentation did not exhibit these same results. The wire was capable of carrying currents of up to 5 amperes without reaching the unit's maximum voltage output. This is probably due to the difference in electrical properties of the older ACPD unit used, (*i.e.* lower maximum current and potential outputs of the CGM-5).

The choice of diameter size was restricted due to the size of the notch of the test specimen. For current-focussing purposes, the wire must travel along the contour of the notch, which has a very small radius of only 1.2 mm (0.049 in.). Also, to minimize the loop area between the current input and output contact points, the current probe had to be placed as close as possible to the specimen surface. Hence, a wire of 0.813 mm (0.032 in.) diameter was chosen as it was capable of carrying enough current (up to 5 amperes) to the specimen and could easily be shaped to fit along the contour of the notch.

As previously mentioned, it is required to bend the current carrying wire along the contour of the notch for current focussing applications. This current focussing wire must be kept insulated from the specimen, hence the need for a suitable insulating material. In addition, it was also observed that if the current focussing wire is not held stationary, potential readings vary due to changes in electric field properties. Therefore, it would be ideal to use an insulative material with good stiffness properties at high temperatures. Initially, a high temperature ceramic coating was used to insulate the wire, which provided good stiffness; however, it proved to be very brittle once placed under cyclic loading. An alternate solution was found by using very small 26-gauge fiberglass sleeving for electrical insulation, and additional stiffness was gained by reinforcing the weld points with high temperature ceramic adhesive.

To ensure that the current probe lead wires were consistently made identical for each fatigue test, a mechanical jig was designed. The jig was used to create a notch in the Chromel lead wire that fit precisely into the contours of the SEN specimen, as seen in Figure 33. The jig was also used to bend the wires in the exact same place each time, for marking the lead wires for spot welding, and for repeatable twisting of the input and output current wires. Figure 35 and Figure 36 show the top and bottom pieces of the current probe jig respectively. The final current probe design is depicted in Figure 34 illustrating the spot weld positions, the insulative sleeving used, as well as the wire twisting used to reduce current loop area.

5.3.2.2.2. Voltage Probe Design

As previously mentioned, high-temperature Chromel wire was implemented for use as the current probe lead material. Due the success of its application in current probe design, Chromel wires ranging in diameter from 0.127 mm (0.005 in.) up to 1.143 mm (0.045 in.) were tested for suitability as voltage probe material. It was found that the smaller Chromel wire sizes of diameters less than 0.508 mm (0.020 in.) were far too ductile for wire twisting. In addition, these smaller wire sizes tended to break during spot welding procedures when exposed to a minor load. It was determined that Chromel wire of 0.508 mm (0.020 in.) diameter was the minimum diameter required to prevent wire breaking during spot welding and wire twisting.

To minimize induction in the voltage probes, the lead wires were twisted as close as possible to the specimen surface, which minimized the loop area. The spacing between the lead wires (5.89 mm) was determined by the diagonal distance from opposite corners across the specimen notch. A potential probe jig was designed to ensure that the lead wires were separated the exact same distance for each probe constructed. The jig was also used for twisting the lead wires together producing the exact same loop size for each voltage probe. The loop size was determined by the minimum wire length required to spot weld the probes to the specimen surface without making contact to the current probe previously attached. Figure 37 and Figure 38 show the top and bottom pieces of the voltage probe jig respectively. It was necessary to construct two voltage probes for each SEN specimen; one for crack growth measurement and one for reference voltage measurement. Reference voltage measurements were made during each fatigue test to

monitor any changes in AC signal that were not associated to crack growth. Crack probes and reference probes were made exactly the same; however, spot welding of the reference probe leads was made 0.5 cm behind the specimen notch in an area where no crack growth occurs. The final ACPD probe assembly with the voltage probes welded in place can be seen in Figure 39. It should be mentioned that all probe wires were meticulously spot welded in place by hand, with the aid of a low magnification macroscope.

5.3.2.3.   *ACPD Electronics Calibration*

5.3.2.3.1.   *Equipment Tests*

Prior to fatigue testing, it was necessary to examine the ACPD electronics with respect to the new probe design. The systems electronics were examined for stability and noise characteristics with respect to such variables as current level, current frequency, amplifier gain, and filter settings.

The effect of current on the probe voltage readings was determined by varying the current output level from 0 to 5 amperes while maintaining a constant frequency (130 kHz) and signal gain (1000). It is evident from Figure 40 that there is a linear relationship between the ACPD probe voltage and the current. A second test was performed to examine the effect of variable gain on probe voltage readings. In this test the frequency was kept constant at 130 kHz, the current was fixed at 2.0 amperes, and the gain was varied from 1000 to 30000. Figure 41 again shows a linear relationship between signal gain and probe voltage. The results from both of these tests were as predicted and verified that the ACPD equipment, as well as probe assembly, were working properly.

A third equipment test was performed to examine the relationship between current frequency and probe voltage output. The gain was held constant at 1000, current level was fixed at 2.0 amperes, and the frequency was varied from 1 kHz to 240 kHz. According to ACPD theory, the skin depth is inversely proportional to the square root of the current frequency as discussed in section 3.3.3.2. Furthermore, the ACPD signal is inversely proportional to the skin depth. Hence, the probe voltage, in theory, should be directly proportional to the square root of the current frequency. Figure 42 illustrates the relationship between the square root of the current frequency and probe voltage. The results do not confirm a complete linear relationship; in fact, there seems to be an inflection point in the curve at around 20 kHz. This is most evident in the voltage crack probe and is probably due to the larger loop area associated with it (i.e. more susceptible to emf pick-up). However, it should be noted that there seems to be a good linear relationship from 20 kHz up to 240 kHz. Also, a somewhat linear trend (smaller slope) is evident from 1 kHz to 20 kHz. This inflection point should not pose a problem during fatigue testing as tests will be conducted well above the 20 kHz level to optimize the ACPD skin effect.

5.3.2.3.2.   *ACPD Signal Drift Tests*

ACPD drift measurement tests were performed at different combinations of current level and current frequency. A simple test program was written that measures the probe outputs with respect to time. It was observed that at current settings greater than 2.5 amperes the current lead wires began to overheat dramatically. It was feared that at these high temperatures the lead material might break-down with respect to long duration times and result in fluctuating voltage readings. However, after running numerous long-term drift

tests, utilizing the maximum 5 amperes of current, it was found that, although the wire temperatures reached as high as 120 °C, the current stability was still maintained. For this reason, an input current setting of 5 amperes was selected for all future fatigue tests to take advantage of this ACPD unit's high current output capability. This in turn results in greater voltage probe measurements without using large signal gain settings, which in turn provides for a better signal-to-noise ratio.

The current frequency selection was another system parameter that was crucial to crack growth monitoring. The choice of current frequency determines the depth of current penetration (skin depth) below the specimen surface. The higher the frequency, the smaller the skin depth. Having a smaller skin depth effectively increases the current density in the surface of the specimen in the area where short crack initiation would occur. This effect simply allows one to detect crack initiation and growth with much more sensitivity. However, at higher current frequencies the emf pick-up by the probes tends to increase dramatically. For this reason, numerous drift tests starting at 30 kHz up to 240 kHz were performed. Each test ran for over 50 hours to examine the stability of voltage probe output and current fluctuation. At frequencies of 150 kHz and higher, the current stability was very poor which in turn resulted in unacceptable drift and noise in probe measurements. From 30 kHz up to 145 kHz the current was much more stable; however, just above 130 kHz some emf pick-up was evident due to several voltage spike readings that were observed. The end result of all drift test measurements was that a current frequency of 130 kHz at 5 amperes provided the best combination of stability and skin depth. Figure 43 shows the drift test results for both probes monitored over 50 hours.

It is obvious that the CH1-0° readings (crack measurements) are slightly less stable than the CH2-0° readings (reference) as expected. The signal noise for each probe was approximately 0.5 mV with corresponding drift measurements of ~1mV and ~0.7 mV for the crack and reference probes respectively. This is largely due to the fact that the crack voltage probe (wired across the notch) has a larger loop length compared to the reference probe (no notch). Therefore, more noise pickup and a larger voltage signal is expected from the crack probe. To verify this assumption, a second test using the same ACPD settings was performed, however, the channels were reversed to ensure that the difference was not due to internal electronics (e.g. lock-in amplifier or filtering etc.). The results are shown in Figure 44. The data verifies that the difference in both channels is not due to ACPD electronics but is due to the voltage crack probe's greater susceptibility to pick-up stray emf signals.

5.3.2.4.   *ACPD Crack Calibration*

Twelve calibration tests were performed over a wide crack size range (50 μm - 1200 μm) to determine the crack size-potential drop relationship. Each specimen was fatigued at room temperature for a predetermined change in crack voltage, subsequently heat tinted to provide crack area contrast, and then pulled apart to expose the crack surface. The fracture surfaces were inspected using an SEM and digital images were recorded for image analysis purposes. Several SEM images showing the typical thumbnail crack geometry that forms during fatigue testing are shown in Figure 45 through Figure 48. For the most part, each specimen possessed one dominant crack as seen in Figure 45. However, there were a few specimens where multiple cracks developed as seen in Figure 47. In the case where multiple cracks were present, the dominant crack dimensions were

used for relating the change in probe voltage to crack depth. The typical voltage probe response for one such test (thumbnail crack shown in Figure 45) has been provided in Figure 49 to illustrate the effect of crack growth on crack and reference probe voltages.

The crack depth, crack area, and surface crack length measurements were determined for each fractured calibration specimen using the Image Pro™ analysis software package. Provided in Table 5 are the calibration test results relating the change in crack probe potential to crack geometry.

One promising aspect of the calibration results was the observed high resolution of the ACPD technique employed. The smallest detectable crack size was 80.3 μm (Figure 48) at a voltage change of 2.21 mV. Although a noticeable change in voltage can be detected below the 80.3 μm depth, a minimum of 2 mV was deemed the smallest reliable voltage change with respect to resolution. This is a conservative restriction if one considers the extreme stability of the voltage readings (on the order of ± 5 μV).

The main purpose for calibration testing was to determine the relationship between crack depth and change in crack voltage. Figure 50 is the calibration curve generated for the twelve calibration specimens tested. From the results, it was determined that an exponential trend most favorably fit the experimental data as follows,

$$\delta V = 1.9665 \cdot e^{0.0042552 \cdot (a)}, R^2 = 0.995 \quad \dots\dots\dots (10)$$

where δV is the change in crack probe potential (mV) and a is the crack depth (μm). Linear and higher order polynomial fits were also examined but they did not conform as well as the exponential fit. These findings differ from the results of numerous other researchers who have determined that the correlation between probe voltage and crack length follows either a linear relationship^{50,68,71} or a 2nd order polynomial relationship.^{67,88} The most probable reason for such a wide discrepancy is most likely due to the difference in the number of calibration tests performed in these past studies. In the studies where a linear correlation is reported, it was generally observed that the calibration curves were generated based on very few calibration tests; especially in the very short crack size range (i.e. $<300 \mu\text{m}$). In the cases where 2nd order polynomial fits were noted, the reason for discrepancy is less apparent based on the well defined voltage/crack size curves reported. In terms of crack depth, equation (10) can be rewritten in the form,

$$a = 235.007 \cdot \ln(\delta V) - 158.924 \dots\dots\dots (11)$$

Although this equation is a good fit for the crack depth- δV data, it is obvious that there is minimum δV value ($\delta V = 1.9665 \text{ mV}$) below which the equation is invalid. However, this does not pose a problem as it has already been determined that the acceptable ACPD resolution is 2 mV for detectable crack initiation. Hence, crack probe δV measurements of less than 2 mV are not included in fatigue crack growth rate data, (i.e. crack monitoring range is $2 \text{ mV} < \delta V < 300 \text{ mV}$). The upper limit of the crack size range was determined based on calibration test results that showed that when δV is $\sim 300 \text{ mV}$, the crack depth is $\sim 1200 \mu\text{m}$ and the crack shape changes from a semi-elliptical thumbnail geometry to a full through-thickness crack. This restriction was applied because the short

crack growth model chosen for fracture mechanics analysis loses validity in the through-thickness crack regime.

In addition to relating the crack depth to δV , the correlation between crack area and potential change was also investigated. In a similar manner, the crack area measurements recorded in Table 5 were plotted against the respective δV values as seen in Figure 51. From the results, it was determined that a 2nd order polynomial best fit the experimental data as follows,

$$\delta V = 9.3085 (A)^2 + 40.3932 (A), \quad R^2 = 0.996 \quad \dots\dots\dots (12)$$

Although the crack area (A) can be converted easily to crack depth values, it was decided that the crack depth- δV correlation (equation 10) was more appropriate for crack calibration purposes. This is mainly due to the fact that more error is introduced when converting the crack area to crack depth.

The final step of the crack calibration procedure was determining the crack shape characteristics with respect to increasing crack size. Specifically, it was required to determine the relationship between crack depth (a) and surface crack length ($2c$) over the entire crack size range (50 μm – 1200 μm). Figure 52 shows the general linear trend observed over the entire crack size range investigated. The relationship between crack depth and surface crack length can be represented by the following equation,

$$c = 1.4387(a), \quad R^2 = 0.995 \quad \dots\dots\dots (13)$$

The crack aspect ratio was required for ΔK calculations used in Newman's short crack growth model described in Appendix A.

5.3.2.5.   *ACPD Data Acquisition*

As previously mentioned, the fatigue loading conditions incorporate a triangular loading waveform at a frequency of 5 Hz and zero load ratio (max load of 32 kN) for both ambient and 482 °C test temperatures. The data acquisition cycle consisted of fatigue loading for 10 cycles with probe voltage measurements taken at each peak and valley load level as illustrated in Figure 53. The initial time period between data acquisition cycles is every 60 seconds until crack initiation is detected, then the time interval is decreased to every 10 seconds. These peak and valley measurements are then averaged over the data acquisition interval (10 cycles) to generate an average voltage reading for both peak (fully loaded) and valley (zero load) conditions. By over-sampling the voltage readings a more stable measurement can be obtained.

The purpose of recording ACPD voltages at both peak and valley loading conditions was to monitor the loading effects on both the reference and crack probes. Figure 54 shows that there is a 0.4 mV change in crack probe voltage for every 1 kN increment in applied load. This effect was observed for the crack probe only and no offset in voltage was observed for the reference probe channel. Due to this loading effect, it was decided that peak load measurements for the crack probe would be used for crack monitoring (i.e. the zero load measurements for the crack probe were ignored). Due to the fact that there was

no difference between peak and valley load readings for the reference probe measurements, the recorded reference probe value was taken as an average of all 20 readings (peak + valley) over each data acquisition interval. A summary of the data recorded for each data acquisition interval is as follows,

N :	last cycle number in data acquisition interval
V_{crack} :	average of peak voltage measurements (Ch.1-0° component) over total interval
$V_{\text{reference}}$:	average of peak and valley voltage measurements (Ch.2-0° component) over total interval
P_{max} :	maximum load reading at last cycle in interval
δV_{crack} :	difference in V_{crack} from initial voltage, V_o , recorded prior to crack initiation

Fatigue test data acquisition was performed until δV reached ~300 mV, which was previously determined during crack calibration to be the δV value where the semi-elliptical thumbnail crack changes to a through-thickness geometry. Upon test completion, short crack growth rate properties as well as stress intensity factor calculations were determined from the raw data files generated.

5.3.3. Fatigue Data Analysis Method

With fatigue testing completed, the raw data files generated during ACPD crack monitoring must be analyzed to determine the material's short crack growth rate properties. Specifically, the material's crack growth rate (da/dN) as a function of the stress intensity range (ΔK) is required to characterize its resistance to crack extension under cyclic loading. Expressing da/dN as a function of ΔK provides results that are independent of planar geometry.

Stress intensity calculations were determined by applying the short crack growth model developed by Newman *et al.* (described in Appendix A). This model was specifically developed for the determination of ΔK values for semi-elliptical cracks emanating from notched specimens. The relative da/dN values were determined using the suggested seven point polynomial fit method specified in the ASTM E647 standard. Basically the procedure involves incrementally fitting a 2nd order polynomial to a set of seven successive crack depth data points over the entire crack length history. This effectively removes any scatter recorded during crack monitoring and provides a smoother curve fit for da/dN interpolation.

It is obvious from the short crack model described in Appendix A that extensive mathematical manipulation is required for ΔK determination. For this reason, a computer program was written that incorporates Newman's crack model, as well as the seven point polynomial fit method. The data analysis code was written in Visual Basic 6.0 and has proven to be very useful with respect to generating a program that can be modified to conform to user specifications (e.g. if a change in load or specimen geometry is required). The purpose of this software is to convert raw ACPD data files that were generated during short crack growth tests into engineering parameters of interest. Specifically, the crack probe voltage measurements are processed to give crack depth, crack area, surface crack length, crack growth rate, crack area growth rate, and stress intensity factor values. The computer code, as well as the Visual Basic forms, have been provided in Appendix B for review. The program input procedure is described as follows.

First, the user is asked to provide the necessary correlation parameters for δV -crack depth and for δV -crack area relationships that were predetermined by crack calibration testing. The initial voltage reading (V_o) prior to crack initiation as well as the crack aspect ratio parameters must also be provided by the user for stress intensity calculations. The program then allows the user to fill in the specific specimen dimensions, properties, and loading conditions that were employed in the test. The current default settings are set for the PWA 1113 material properties with respect to the SEN geometry used in the present study. If other materials or fatigue specimens are used the user can simply input the changes where prompted.

5.4. Tensile Testing

Tensile tests were performed using a closed-loop servo-hydraulic MTS load frame in accordance with the ASTM-E8 standard, at both room temperature and 650 °C. Tensile test specimens were machined as per Figure 55. All specimen blanks were cut in a radial direction from the disc to ensure uniform mechanical properties. Strain data was gathered using an MTS high temperature quartz-rod extensometer while monitoring simultaneous load readings from the MTS load cell. Multiple tests were conducted for the conventional and damage tolerant microstructures at each test temperature.

5.5. Creep Testing

Creep-rupture tests were performed in accordance with the ASTM-E139 standard and were conducted at 705 °C under a constant load of 690 MPa (100 ksi). A SATEC Model-D constant load creep frame was used with a lever arm ratio of 20:1. Creep specimen blanks were again cut in a radial direction from the disc and were machined to the same 0.635 cm (¼ inch) diameter configuration as used for the tensile specimens depicted in

Figure 55. Both CM and DTM conditions were tested for comparison. All creep extension measurements were made using a high resolution displacement gauge.

6. Microstructural Response to Heat Treatments

6.1. Results and Discussion

6.1.1. Grain Size Optimization

To determine the solutioning temperature required to dissolve all γ' precipitates without the occurrence of excessive grain growth, a detailed grain coarsening study was performed. Table 7 shows the results of the grain size measurements obtained for the 1,2 and 4 hour solution heat treatments made over the 1050 °C to 1250 °C temperature range. These results are illustrated in Figure 56, which shows the grain coarsening behavior of the material over the wide range of solutioning temperatures and hold times. It is apparent from this figure that a major shift in grain size occurs between 1130 °C and 1135 °C for all solutioning hold times. For example, it can be seen from Table 7 that for a 2 hour solutioning time, the grain size changes dramatically from 35 μm to 100 μm . This rapid change in grain size is even more apparent from the two optical micrographs shown Figure 57 and Figure 58. It can therefore be deduced that at 1135 °C, the γ' precipitates are fully dissolved and this effectively removes all grain boundary pinning agents and permits grain growth to occur. Hence, the temperature chosen for solutioning the γ' precipitate phase was 1135 °C. A two hour solutioning time was also deemed an appropriate hold time as it assured uniform solutioning through the whole thickness of the $\frac{3}{4}$ inch specimen blanks used for mechanical testing. The end result of the grain coarsening study is the solutioning sequence of 1135 °C for 2 hours that fully dissolves all the primary γ' precipitates and results in a final DTM grain size of $\sim 100 \mu\text{m}$ (ASTM 4).

6.1.2. Grain Boundary Morphology Optimization

Numerous heat treatment sequences were used to determine the appropriate cooling rate range required to heterogeneously nucleate γ' precipitates along the grain boundaries. The results from the controlled cooling experiments performed (see Table 6) indicate that a cooling rate of 3 °C/min from 1135 °C to 1060 °C provides for the best grain boundary serration morphology. Cooling to 1060 °C at a rate of 3 °C/min produced the optimum precipitate size (~1-3 μm) that generated grain boundary serrations throughout the microstructure's intergranular region, as seen in Figure 59. The average serration amplitude (h) produced was on the order of 2 μm with a serration wavelength (λ) of approximately 5 μm . Similar to the findings by Koul *et al.*, the majority of grain boundary γ' precipitates responsible for grain boundary serrations are elliptical in shape and are always located within the concave side of the serration⁸⁴.

The slower cooling rates of 1 °C/min and 2 °C/min proved to be ineffective in producing useful grain boundary serrations for all furnace removal temperatures investigated. For example, cooling from 1135 °C to 1060 °C at 1 °C/min produced excessively large grain boundary serrations. These large serrations, as illustrated in Figure 60, are not of the typical sinusoidal morphology as described earlier in Figure 59. In fact, they appear to be directly associated with the formation of dendritic γ' fans that push on either side of the grain boundary. Figure 61 shows the same serrated structure; however, the specimen has been etched to reveal the γ' precipitate morphology associated with these excessively large serrations. These large dendritic γ' fans were also observed by Danflou *et al.* in their work with Henry's alloy as shown in Figure 24.⁸⁷ They produced the same dendritic fan

structure by applying an isothermal heat treatment just below the alloy's γ' solvus temperature directly after solution heat treatment. This low undercooling results in a low density of particles hence favoring the growth of arms before supersaturation is lost and coarsening begins.⁸⁷ The same reasoning can also be used for the growth of dendritic fans in PWA 1113 upon slow cooling (1 °C/min and 2 °C/min) from the γ' solvus temperature. The slower cooling rate allows for a longer period of dendritic fan growth to occur, as opposed to faster cooling rates that result in relatively rapid transitions from the solid solution state to complete homogeneous nucleation. Faster cooling rates result in rapid loss of supersaturation and a higher density of precipitate particles are produced, preventing the mechanism of dendritic arm growth.

6.1.3. Intragranular γ' Optimization

Age hardening heat treatments were conducted on the newly serrated material to determine the peak hardness level required for optimum strength. The Rockwell C hardness test results from the aging study are provided in Table 8, along with the associated aging curve attached as Figure 62. The results indicate that an aging heat treatment of 10 hours at 760 °C provides for the peak hardness level in the modified material.

It was observed that the peak hardness level does not exist at a single point, and in fact, the peak hardness is present for some range of aging times. This is typical of most superalloys due to their heat resistant nature. Although the peak hardness is initially observed at an aging time of 10 hours, it was decided to use an aging step of 16 hours at

760 °C. This was done to slightly overage the primary γ' precipitate structure to produce a suitable particle size. By producing a precipitate size that is just larger than optimum size required for peak hardness, the likelihood for dislocation looping to be the predominant dislocation/precipitate interaction mechanism is increased. This should theoretically impede transgranular short crack propagation by forcing dislocations to loop around the particles (as opposed to cutting) resulting in fatigue hardening.¹³ The main advantage of hardening is the promotion of slip in other possible planes. This results in more uniform deformation and effectively slows down the early stages of crack growth.

Both the CM and the newly aged DTM microstructures were examined under an SEM. It is quite obvious from Figure 63, that the primary γ' size and distribution for the CM is quite large and poorly distributed throughout the matrix. Primary γ' precipitates for the CM condition are up to 10 μm in diameter and are usually located along the grain boundaries, likely contributing little to the overall matrix strength. Figure 64 shows the new γ' precipitate morphology and distribution of the DTM. The new features include a smaller primary γ' precipitate size ($\sim 1\mu\text{m}$) that is more homogeneously distributed throughout the intragranular region, as well as a much finer secondary precipitate size ($<0.1\mu\text{m}$).

7. Mechanical Properties of Conventional and Damage Tolerant Microstructures

7.1. Results and Discussion

7.1.1. Short Crack Fatigue Properties

Fatigue test results revealed that the DTM possessed better short crack fatigue resistance than the CM at both room temperature and at 482 °C. As illustrated in Figure 65 and Figure 66 respectively, at a stress intensity (ΔK) of 10 MPa $\sqrt{\text{m}}$, the crack growth rate for the DTM is approximately 3 times slower at room temperature and 2.2 times slower at 482 °C compared to the CM.

It is apparent from both the room and elevated temperature fatigue curves presented, that short crack growth behavior does not conform to traditional long crack growth characteristics. This is most likely due to the two main defining properties of short cracks that distinguish them from typical long crack growth features.

1. Short cracks have generally been observed to exhibit faster and more irregular propagation rates compared to long crack data.³³
2. Short crack propagation occurs at stress intensity factors (ΔK) well below the traditional threshold value (ΔK_{th}) for long crack growth.⁸⁹

These two distinguishing properties of short cracks have been reported by numerous researchers, all of whom have reported similar anomalous short crack behavior.⁹⁰ This is primarily due to the fact that microstructurally short crack growth often deals with the propagation of cracks that do not abide by LEFM principles.

LEFM relies heavily on continuum mechanics principles for crack propagation. However, when dealing with short crack growth, the crack in the beginning is smaller than the actual grain size and microstructural features tend to influence crack propagation. This ultimately results in the breakdown of the material continuum and propagation exhibits a more discrete behavior. Due to the interference of short crack propagation by such features as grain boundaries and intragranular precipitates, the resulting growth rate is often irregular in nature.⁴⁰ When such an obstacle is encountered by the crack tip plastic zone, the growth rate tends to decrease followed by rapid acceleration upon traversing the barrier. This fluctuating crack growth behavior is visible in Figure 65 and Figure 66, which show periods of deceleration and acceleration at relatively low ΔK levels. These inflection points on the curves occur at ΔK levels that are indicative of crack depths equivalent in length to the average microstructural grain size. Evidence of this possible grain boundary interaction can be more clearly seen from Figure 67 which shows the crack propagation rate as a function of crack length for both CM and DTM microstructures. It can be clearly seen that the initial crack propagation rates for both material conditions tend to decrease when the crack sizes approach the respective average grain sizes. This suggests that the initial deceleration is most likely due to dislocation

pile-up at the impeding grain boundary, followed by rapid acceleration once sufficient stress intensity has been attained for slip band extension into the neighboring grain.

The improvements in short crack growth rate properties can be explained on the basis of differences in the fracture modes observed in Figure 68 and Figure 69. In Figure 68, it is obvious that the fracture mode of the CM is transgranular, with evidence of striations or faceted crack growth across the grains. There is also evidence of the ineffectiveness of the large overaged γ' precipitates in retarding crack growth. Figure 68 shows the lack of fatigue striations across these large precipitates suggesting that dislocation interaction occurs by means of climb as opposed to cutting or looping. The DTM fracture surface also exhibits a transgranular fracture; however, due to the finer, more dispersed γ' precipitate morphology, crack growth is decreased within the intragranular regions. This is most likely caused by the combination of crack deflection around the γ' precipitates which prevents slip-band extension within the grain itself, as well as the larger DTM grain size.

7.1.2. Creep Rupture Properties

In addition to the improved short crack growth rate properties of the DTM, the creep-rupture test results demonstrate a marked improvement over the creep properties of the CM material. It was predicted that the DTM would have better creep resistance due to its larger grain size and serrated grain boundary morphology. In fact, the test results shown in Figure 70 indicate that at an initial stress level of 690 MPa (100 ksi) and at a temperature of 705 °C, the creep-rupture life was almost 3 ½ times longer for the DTM.

Furthermore, it is apparent from the creep fractograph in Figure 71 that the fracture mode of the DTM is intergranular, as indicated by the irregular topography and grain boundary relief. Compared to the fracture surface of the CM, Figure 72, the DTM has a more highly featured surface profile. Although the fracture mode of the CM is also intergranular, it does not possess the same degree of rough topography due to its smaller grain size. This small grain size ($\sim 13\text{ }\mu\text{m}$), results in a relatively smooth fracture surface almost indistinguishable from the transgranular overload section of the specimen, as shown in Figure 72. The intergranular fracture in both materials suggests that the test temperature of $705\text{ }^{\circ}\text{C}$ was indeed above the equicohesive limit. In other words, at this temperature the grain interiors retained superior creep resistance compared to the grain boundaries.

Two other important results can be derived from Figure 70,

1. There is a delayed onset of tertiary creep ($\sim 50\text{ hrs}$) for the DTM as compared to the CM ($\sim 5\text{ hrs}$)
2. The tertiary creep time interval is much longer for the DTM.

As for the superior creep resistance of the DTM, it may be attributed to a combination of a coarser grain size, the serrated grain boundary morphology, and the smaller γ' precipitate size that is more uniformly distributed. The coarser grain size ($70\text{-}100\text{ }\mu\text{m}$) of the DTM means that there is a lower grain boundary fraction present that could serve as creep fracture paths in the material. More specifically, having fewer grain boundaries present impedes the ability of vacancy diffusion along the grain boundaries, which is the

rate controlling factor for grain boundary sliding and diffusional creep. The serrated morphology of the DTM may also slow down surface creep crack propagation by presenting a more tortuous crack path, as compared to the relatively planar boundaries of the CM. These microstructural features are known to suppress grain boundary sliding. The finer ($1\mu\text{m}$) γ' precipitate size, as well as their decoration along the grain boundaries, may also retard crack propagation. The tortuous grain boundary paths created by the more dispersed nature of these precipitates, results in higher intergranular strength. In contrast, the CM precipitate size is extremely large ($10\mu\text{m}$) and poorly distributed playing no role in grain boundary strengthening.

7.1.3. Tensile Properties

Tensile tests were conducted at room temperature and at 650°C . Figure 73 and Figure 74 show the tensile test results for room temperature and 650°C respectively. Tensile test results indicated minimal strength losses for the DTM in terms of YS and UTS values with 80% and 90% of the CM baseline values for both room temperature and 650°C . From these results it can be seen that the UTS and YS values of the new DTM are comparable to that of the conventional microstructure. Yield and ultimate strengths are slightly lower in the new DTM as was to be expected due to the larger grain size present.

Room temperature and high temperature tests also showed an increase in specimen elongation for the new DTM. Average elongation values for room temperature tests were 12% and 18% for the CM and DTM respectively. Likewise, at 650°C , the average elongation values were 12% for the CM and 15% for the DTM. This trend is consistent

with the improvement in creep ductility of the DTM as shown in Figure 70. The increased ductility of the new DTM may also be accountable for the improved high temperature fatigue resistance observed. Bain *et al.* have developed an improved DTM for the Ni-base superalloy Udimet 720 by simply increasing the overall grain size.⁹¹ Their findings have shown that at the expense of marginal losses in tensile strength, the increased ductility of the coarser grain sized material produces significant improvements in high temperature fatigue resistance as well as stress rupture life.

8. Summary and Future Work

8.1. Summary

A successful DTM for the Ni-base superalloy, PWA 1113, has been designed. The DTM was produced by implementing a modified heat treatment sequence that optimized the grain size, grain boundary morphology, and γ' precipitate size and distribution. The modified heat treatment procedure developed is as follows:

1. Solutioning for 2 hours at 1135°C,
2. Controlled cooling from 1135 ° to 1060°C at 3°C/min followed by air cooling,
3. And aging at 760°C for 16 hours.

The new heat treatment developed imparted a serrated grain boundary morphology, coarser grain size of ~ 100 μm , and a finer 1 μm primary γ' size that was more uniformly distributed.

Mechanical testing confirmed that the newly developed microstructure had improved damage tolerance compared to the CM. Short crack propagation rates were decreased by a factor of 3 at room temperature and a factor of 2.2 at 482 °C. Grain size had a profound effect on the shape of ΔK versus crack growth rate correlation. At 705 °C and 690 MPa, the creep rupture life of the DTM was extended by a factor of almost 4. Marginal losses in proof strength and ultimate strength were also observed during room temperature and high temperature tensile tests. Improved damage tolerance properties can be directly related to the improved grain boundary strength of the DTM that was invoked by

controlled-cooling heat treatments. The finer γ' precipitate size and distribution also contributed to slower intragranular crack propagation rates by impeding dislocation motion within the grain interiors. These findings for the new DTM are in good agreement to earlier damage tolerance studies for comparable Ni-base superalloys^{3,10}.

The current-focussing ACDP technique developed was very successful in characterizing the short crack growth properties of PWA 1113 for both the CM and DTM conditions. The technique developed showed extremely high resolution and was capable of reliably detecting crack sizes as small as 80 μm in depth. The end result is a crack monitoring technique that can be used to characterize the short crack fatigue properties of critical regions in turbine discs without the necessity of test interruption.

8.2. Recommendations for Future Work

- The completed work concentrated on the characterization of short crack growth properties of the PWA 1113 disc material. Crack propagation rates in the long crack regime ($> 1200 \mu\text{m}$) were not possible due to the SEN specimen geometry used and the crack growth model applied. It would be beneficial to perform additional fatigue testing using compact-tension specimens to generate long crack growth rate properties for a comparative study. The results could be used to confirm the anomalous short crack behavior in the crack range already characterized and determine the transition point where slower long crack growth propagation begins.

- Further metallographic examination using transmission electron microscopy is required to investigate the dislocation deformation mechanisms that occurred during fatigue. TEM foil analysis may confirm that the dominant dislocation-precipitate interaction is one of dislocation looping during short crack growth. This would further reinforce the necessity to optimize the intragranular γ' size to improve short crack growth resistance.
- Another aspect of short crack fatigue testing that was not investigated was the distinctive dependence of short crack resistance on microstructural properties. Specifically, the improved damage tolerance observed in the new DTM could not be completely related to just the larger grain size, precipitate size or grain boundary morphology alone. To determine the effects of each microstructural feature on short crack resistance, tests should be performed on specimens where all microstructural variables are kept constant except the parameter that is being investigated.

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Tables

Alloy	Al	Ti	Nb	Mo	W	Cr (Weight %)	Co	Others	0.2% PS 200°C N/mm ²			Stress (N/mm ²) for 0.1% creep strain in 100 hrs at 650°C		
									700	900	1100	400	600	800
Nimonic 901	3.0			5.7	12.5	-		42 Ni Bal Fe						
Waspaloy	4.4			4.5	19.5	13.5		-						
Inconel 718	6.75			3.0	19.0	-		53 Ni Bal Fe						
APK 1 (Low C Astroloy)	7.5			5.0	15.0	17.0		-						
IN 100	9.3			3.2	12.5	18.5	0.8	V						
Rene 95	9.5			7.0	14.0	8.0		-						

Table 1: Commonly used superalloy disc materials ¹⁷

Effect	Cr	Al	Co	Fe	Mo	W	Ti	Ta	Nb	Hf	C	B	Zr	Mg
<i>solid solution strengtheners</i>	+		+	+	+	+								
<i>precipitate formers</i>		+					+	+	+					
<i>carbide formers</i>	+				+	+	+	+	+		+			
<i>corrosion resistance</i>	+	+												
<i>grain boundary strengtheners</i>										+	+	+	+	+

Table 2: Alloying addition effects on Ni-base alloys

Case #	Researchers	Material	Technique	Microstructural Modifications	Mechanical Results
1	A.K. Koul P. Au N. Bellinger R. Thamburaj W. Wallace J.-P. Immarigeon	IN 718	modification of conventional heat treatment	- new serrated grain boundaries - slightly coarser grain size (35 μm c.f. 20 μm) - typical precipitate morphology with the exception of needle-like precipitates at the grain boundaries	- FCGR 2X slower @ 650 °C - CCGR 5X slower @ 650 °C 21% lower proof strength - comparable LCF @ 650 °C
2	M. Chang A.K. Koul C. Cooper	MERL 76	modification of conventional heat treatment	- new serrated grain boundaries - coarser grain size (30 μm c.f. 10 μm) - new rosette γ' morphology with finer secondary γ' - elliptical γ' at grain boundaries	- FCGR 10X slower @ 700 °C - CCGR 10X slower @ 760 °C 7% lower proof strength - comparable LCF @ 700 °C
3	S. Bashir M.C. Thomas	Alloy 720	modification of conventional composition (reduction of B and C levels)	- no change in grain boundary morphology - no change in grain size - no change in γ' morphology - much smaller boride and carbide stringer concentrations	- FCGR 3X slower at 650 °C - comparable proof strength at all temperatures - much improved LCF life
4	H.L. Danilou M. Marty A. Walder	Astroloy	modification of conventional heat treatment	- new serrated grain boundaries - comparable grain size (65 μm to 90 μm) - new butterfly wing γ' morphology - coarse γ' (~ 6 μm) at grain boundaries	- FCGR 3X slower @ 750 °C - comparable proof strength

Table 3: Examples of Damage Tolerant research from literature

Ni	Cr	Co	Mo	Ti	Al	Zr	B	Fe	Mn	Si	Cu	C
Bal.	16.52	13.96	4.08	4.5	2.02	0.07	0.007	0.06	<0.1	<0.1	<0.1	0.032

Table 4: Nominal Composition of PWA 1113 (wt%).

δV (mV)	crack depth (μm)	surface crack length (μm)	c (μm)	c/a	crack area (mm^2)	δV (mV)
2.21	80.3	300.1	150.05	1.8686	0.0209	2.21
2.89	110.6	402.1	201.05	1.8178	0.0797	2.86
3.86	138.2	522.2	261.1	1.8893	0.1896	3.86
5.67	248.7	718	359	1.4435	0.179	5.67
8.96	279.3	802.1	401.05	1.4359	0.2014	8.96
16	489.1	1433.5	716.75	1.4654	0.4616	16
26	621.4	1949.3	974.65	1.5685	0.7788	26
52	786	2112.7	1056.35	1.3440	1.1121	52
109.36	931.2	2748.4	1374.2	1.4757	1.7587	109.36
282.2	1156.7	3306.2	1653.1	1.4292	3.657	282.2
282.38	1169	3341.4	1670.7	1.4292	3.71	282.38
294	1193.3	3399.7	1699.85	1.4245	3.9734	294

Table 5: Crack calibration results.

Specimen batch	A	B	C
Cooling rate from γ' solvus	1 °C/min	2 °C/min	3 °C/min
Removal temperature. (°C)	1115	1115	1115
followed by A/C	1105	1105	1105
	1095	1095	1095
	1080	1080	1080
	1075	1075	1075
	1060	1060	1060
	1050	1050	1050
	1040	1040	1040
	1030	1030	1030

Table 6: Schedule of cooling rates and furnace removal temperatures investigated.

solution temperature (°C)	mean intercept distance (μm)		
	1 hour duration	2 hour duration	4 hour duration
1050	13.48	14.86	15.10
1075	15.03	15.45	15.90
1100	16.57	17.41	21.65
1115	24.55	25.23	25.70
1125	27.23	30.22	34.40
1130	28.65	34.38	37.60
1135	71.80	99.10	111.96
1140	148.15	168.60	203.70
1150	279.40	391.00	473.10
1165	350.10	501.50	680.50
1175	431.40	624.00	862.70
1200	637.70	780.60	1080.10
1225	758.60	957.00	1115.60
1250	822.43	1010.60	1120.50

Table 7: Results from grain coarsening study.

760 °C aging time (hrs)	Rockwell C Hardness										
	1	2	3	4	5	6	7	8	9	10	average
0	37	35	35	35	36	38.5	36	38.5	37	36.5	36.5
1	36	36.5	36	37	37	37	36	36.5	36.5	37	36.6
2	36	36	36.5	37.5	37.5	37	37	36	37	37.5	37.1
4	38.5	38	38.5	38	38.5	38	38.5	40	38.5	38.5	38.2
5	38.5	38	38	38	38.5	38	41	38	38.5	40	38.4
6	38.5	38	40	38	38.5	40	38.5	38.5	38	40	38.4
10	38.5	41.5	38.5	38.5	38	38	38	41	41	38	38.7
12	38	38	38	40	38	38	38.5	38.5	38.5	38	38.7
14	38.5	40.5	38.5	38	38.5	38	38	38.5	38.5	38	38.6
16	36	38.5	41	38	40	38.5	38.5	41	38.5	38.5	38.7
18	40	38.5	40.5	40	38	40	40	38.5	40	40.5	38.6
24	38.5	40	40	38	38.5	40	38.5	38.5	40	38	38.6
36	37.5	40	40	38	38	40	41	38	40	38.5	38.5
60	38	40	38	38	40	40	40	38	40	38.5	38.5

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Appendix A: Newman's Short Crack Growth Model

Stress intensity factor equation for a semi-elliptical crack growing from the center of a semi-circular notch,

$$K = S \sqrt{\frac{\pi c}{Q}} F_{jn} \left(\frac{c}{a}, \frac{c}{t}, \frac{a}{r}, \frac{a}{w}, \frac{r}{t}, \frac{r}{w}, \phi \right) \quad (1)$$

where F_{jn} is the boundary-correction factor, Q is the shape factor and S is the applied stress. The relevant specimen and crack dimensions from the above equation are provided in the following figure.

Figure 75: Definition of specimen and surface crack dimensions.

The boundary correction factor is given by,

$$F_{jn} = \left[M_1 + M_2 \left(\frac{c}{t} \right)^2 + M_3 \left(\frac{c}{t} \right)^4 \right] g_1 g_2 g_3 g_4 g_5 f_\phi f_w \quad (2)$$

For $c/a \leq 1$

$$M_1$$

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

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

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

$$g_2 = \frac{[1 + 0.358\lambda + 1.425\lambda^2 - 1.578\lambda^3 + 2.156\lambda^4]}{1 + 0.08\lambda}$$

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

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

$$g_4 = K_r \left[0.36 - \frac{0.032}{\left(1 + \frac{a}{r} \right)^{\frac{1}{2}}} \right]$$

Where K_T is the elastic stress concentration factor ($K_T = 2.682$ for uniform displacement) at the semi-circular notch, and

$$g_s = 1 + \sqrt{\frac{c}{a}} \left[0.003 \left(\frac{r}{t} \right)^2 + 0.035 \left(\frac{r}{t} \right) (1 - \cos \phi)^3 \right] - 0.35 \left(\frac{c}{t} \right)^2 \left(1 - 0.5 \frac{c}{a} \right)^3 \cos \phi$$

$$f_w = 1 + 2.17\gamma^2 - 3.4\gamma^4 + 3.7\gamma^6$$

$$\gamma = \left(\frac{c}{t} \right)^{\frac{1}{2}} \left(\frac{a+r}{w} \right)$$

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

$$Q = 1 + 1.464 \left(\frac{c}{a} \right)^{1.65}$$

For $c/a > 1$, the following variables change accordingly

$$Q = 1 + 1.464 \left(\frac{a}{c} \right)^{1.65}$$

$$M_1 = \left(\frac{a}{c} \right)^{\frac{1}{2}} \left(1.04 - 0.4 \frac{a}{c} \right)$$

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

Appendix B: Visual Basic Data Analysis Code and Forms

The following code was written using Visual Basic 6.0 to determine the ΔK and da/dN values from raw data files generated during ACPD crack monitoring. The code incorporates the Newman – Swain short crack growth model for determining ΔK , as well as the seven point polynomial fit method suggested by ASTM for determining da/dN . Because of the programming nature of Visual Basic, the object activated forms have been provided following the basic computer code.

Form frmCheck

```
Dim A, B, Vo As Single
Dim D, E, F As String
Dim G As Single
Dim R, W, T, S, Kt As Single
Dim crackdepth, newcrackdepth, surfacecrack As Double
Dim crackarea, newcrackarea, aspect, deltaV As Double
Dim oldfile, newfile, fcgfile As String
Dim freq, IdRatio As Integer
Dim Time, PCH1Z, VCH1Z, MaxCH2Z, PkLoad, VyLoad As Single
Dim Cycle, SpecTemp As Integer
Dim i, j, k, l, m, n, p, q, z As Integer
Dim QL, QR As Integer
Dim junk As String
Dim A1(2000), AR(2000), delV(2000) As Single, N1(2000) As Long, reading(2000) As Integer
Dim A2(7), AR2(7), dARdN(2000), dAdN(2000) As Single, N2(7) As Long
Dim C1, C2 As Long
Dim X, b0, b1, b2, DEN, DEN2, DENV2, rate As Double
Dim SX1, SX2, SX3, SX4 As Single
Dim Y, SY1, SY2, SY3 As Single
Dim u0, u1, u2, rate2 As Double
Dim V, SV1, SV2, SV3 As Single
Dim Qfactor, Fsn, Fwidth, Fphi As Single
Dim M1, M2, M3, g1, g2, g3, g4, g5
Dim lambda, gamma, cdivt, rdivt, adivr As Single
Dim phi As Integer, PI As Single
Dim delK As Double
Private Sub sysDetectOS_DisplayChanged()
```

```

If Screen.ActiveForm.Width <> sysDetectOS.WorkAreaWidth Then
    Screen.ActiveForm.Left = sysDetectOS.WorkAreaLeft
    Screen.ActiveForm.Width = sysDetectOS.WorkAreaWidth
End If
If Screen.ActiveForm.Height <> sysDetectOS.WorkAreaHeight Then
    Screen.ActiveForm.Top = sysDetectOS.WorkAreaTop
    Screen.ActiveForm.Height = sysDetectOS.WorkAreaHeight
End If
End Sub

Private Sub cmdProcess_Click()
    Unload frmFCGModel
    Unload frmInputParameters
    Unload frmInputParam2
    Unload frmSelectFile
    Open oldfile For Input As #1
    Open newfile For Output As #2
    Open fcgfile For Output As #3

    'Preliminary reading of test info such as filename, test freq. etc. into new file

    For i = 1 To 9
        Line Input #1, junk
    Next i
    Write #2, "Filename and path: ", newfile
    Input #1, junk, SpecTemp
    Write #2, "Test temperature (°C) ", SpecTemp
    Input #1, junk, PkLoad
    Write #2, "Maximum load (kN) ", PkLoad
    Input #1, junk, freq
    Write #2, "Test frequency (Hz) ", freq
    Input #1, junk, IdRatio
    Write #2, "Test Load Ratio ", IdRatio
    Line Input #1, junk
    Line Input #1, junk
    Write #2, "Measured Test Parameters are as follows: "
    Write #3, "Calculated Crack Model Parameters are as follows: "
    Print #2, "event #, Cycles, delta V (mV), a(microns), Area(mm^2), a(microns)reg., Area(mm^2)reg., dA/dN (reg.), delta K, da/dN(mm/cycle)reg. "
    Print #3, "Cycles, a(microns),Qfactor, m1, m2, m3, g1, lambda, g2, g3, g4, g5, Fw, Fphi, Fsn, delta K "
    i = 0

    'Read in raw data file info and perform preliminary processing

```

Do Until EOF(1)

Input #1, Time, Cycle, PCH1Z, VCH1Z, MaxCH2Z, PkLoad, VyLoad, SpecTemp

deltaV = (PCH1Z - Vo) * 1000 'this gives the change in voltage in mV

If deltaV <= A Then

crackdepth = 0

crackarea = 0

Else: crackdepth = (Log(deltaV / A)) / B

F = 0 'for this case F = 0

crackarea = (((E ^ 2) - 4 * D * (F - deltaV)) ^ 0.5) - E) / (2 * D)

End If

i = i + 1

delV(i) = deltaV

A1(i) = crackdepth

AR(i) = crackarea

N1(i) = Cycle

reading(i) = i

If i <= 3 Then

Write #2, reading(i), N1(i), delV(i), A1(i), AR(i)

Write #3, N1(i), A1(i)

End If

Loop

' using secant method and non-regressed data to find dadN

'For z = 1 To i - 1

'dadN(z) = (A1(z + 1) - A1(z)) / (N1(z + 1) - N1(z))

'Next z

' Calculate growth rates using 7 point polynomial method

n = i - 6

QL = 0

For j = 1 To n

l = 0

QL = QL + 1

QR = QL + 6

For k = QL To QR

l = l + 1

A2(l) = A1(k)

AR2(l) = AR(k)

```

N2(l) = N1(k)
Next k
C1 = 0.5 * (N2(1) + N2(7))
C2 = 0.5 * (N2(7) - N2(1))
SX1 = 0: SX2 = 0: SX3 = 0: SX4 = 0: SY1 = 0: SY2 = 0: SY3 = 0
SV1 = 0: SV2 = 0: SV3 = 0:
For k = 1 To 7
    X = (N2(k) - C1) / C2
    Y = A2(k)
    V = AR2(k)

    SX1 = SX1 + X
    SX2 = SX2 + X ^ 2
    SX3 = SX3 + X ^ 3
    SX4 = SX4 + X ^ 4
    SY1 = SY1 + Y
    SY2 = SY2 + Y * X
    SY3 = SY3 + Y * X ^ 2
    SV1 = SV1 + V
    SV2 = SV2 + V * X
    SV3 = SV3 + V * X ^ 2

Next k
DEN = 7 * (SX2 * SX4 - SX3 ^ 2) - SX1 * (SX1 * SX4 - SX2 * SX3) + SX2 * (SX1 * SX3 - SX2 ^ 2)
b0 = (SY1 * (SX2 * SX4 - SX3 ^ 2) - SY2 * (SX1 * SX4 - SX2 * SX3) + SY3 * (SX1 * SX3 - SX2 ^ 2)) / DEN
u0 = (SV1 * (SX2 * SX4 - SX3 ^ 2) - SV2 * (SX1 * SX4 - SX2 * SX3) + SV3 * (SX1 * SX3 - SX2 ^ 2)) / DEN
b1 = (7 * (SY2 * SX4 - SY3 * SX3) - SX1 * (SY1 * SX4 - SY3 * SX2) + SX2 * (SY1 * SX3 - SY2 * SX2)) / DEN
u1 = (7 * (SV2 * SX4 - SV3 * SX3) - SX1 * (SV1 * SX4 - SV3 * SX2) + SX2 * (SV1 * SX3 - SV2 * SX2)) / DEN
b2 = (7 * (SX2 * SY3 - SX3 * SY2) - SX1 * (SX1 * SY3 - SX3 * SY1) + SX2 * (SX1 * SY2 - SX2 * SY1)) / DEN
u2 = (7 * (SX2 * SV3 - SX3 * SV2) - SX1 * (SX1 * SV3 - SX3 * SV1) + SX2 * (SX1 * SV2 - SX2 * SV1)) / DEN
DEN2 = (b1 / C2 + 2 * b2 * (N2(4) - C1) / C2 ^ 2)
DENV2 = (u1 / C2 + 2 * b2 * (N2(4) - C1) / C2 ^ 2)
X = (N2(4) - C1) / C2
newcrackdepth = b0 + b1 * X + b2 * X ^ 2
newcrackarea = u0 + u1 * X + u2 * X ^ 2
If newcrackdepth < 0 Then newcrackdepth = 0
If newcrackarea < 0 Then newcrackarea = 0

Call calcDeltaK

m = j + 3
rate = ((b1 / C2) + (2 * b2 * ((N1(m) - C1) / (C2 ^ 2)))) / 1000
rate2 = (u1 / C2) + (2 * u2 * ((N1(m) - C1) / (C2 ^ 2)))

```

```

If rate < 0 Then rate = 0
If rate2 < 0 Then rate2 = 0
Write #2, reading(m), N1(m), delV(m), A1(m), AR(m), newcrackdepth, newcrackarea, rate2, delK, rate
Write #3, N1(m); A1(m), Qfactor, M1, M2, M3, g1, lambda, g2, g3, g4, g5, Fwidth, Fphi, Fsn, delK
If j = n Then
    For p = 1 To 3
        Write #2, reading(j + p + 3), N1(j + p + 3), delV(j + p + 3), A1(j + p + 3), AR(j + p + 3)
        Write #3, N1(j + p + 3), A1(j + p + 3)
    Next p
End If
Next j

Close #1
Close #2
Close #3

Dim Msg As String
Msg = "  Your new data file has been created  "
MsgBox Msg, , " Just thought you would like to know ..."
'check screen output
'Print crackdepth
'Print M1, M2, M3
'Print g1, g2
'Print g3, g4, g5
'Print surfacecrack
'Print "newcrackdepth", newcrackdepth
'Print aspect
'Print "Fwidth", Fwidth, "Fphi", Fphi
'Print "Fsn", Fsn, "delK", delK
End Sub

Private Sub cmdQuit_Click()
    Unload frmCheck
    Unload frmFCGModel
    Unload frmInputParameters
    Unload frmInputParam2
    Unload frmSelectFile
    Msg = " If you need assistance or require code modifications, contact rick.kearsey@nrc.ca"
    MsgBox Msg, , " Thanks for using this software ....."
End Sub

Private Sub Form_Load()

```

```

List1.AddItem frmSelectFile.txtFilename
List1.AddItem frmInputParameters.txtA
List1.AddItem frmInputParameters.txtB
List1.AddItem Label5.Caption
List1.AddItem frmInputParameters.txtD
List1.AddItem frmInputParameters.txtE
List1.AddItem Label5.Caption
List1.AddItem frmInputParameters.txtVo
List1.AddItem frmInputParam2.txtG
List1.AddItem Label5.Caption
List1.AddItem frmFCGModel.txtR
List1.AddItem frmFCGModel.txtW
List1.AddItem frmFCGModel.txtT
List1.AddItem frmFCGModel.txtS
List1.AddItem frmFCGModel.txtKt
List1.AddItem frmSelectFile.txtNewFile
List1.AddItem frmSelectFile.txtFCGfile

```

```
A = List1.List(1)
```

```
B = List1.List(2)
```

```
D = List1.List(4)
```

```
E = List1.List(5)
```

```
Vo = List1.List(7)
```

```
G = List1.List(8)
```

```
R = List1.List(10)
```

```
W = List1.List(11)
```

```
T = List1.List(12)
```

```
S = List1.List(13)
```

```
Kt = List1.List(14)
```

```
oldfile = List1.List(0)
```

```
newfile = List1.List(15)
```

```
fcgfile = List1.List(16)
```

```
End Sub
```

```
Private Sub List1_Click()
```

```
    ndx = List1.ListIndex
```

```
    Select Case ndx
```

```
        Case 0, 15, 16
```

```
            frmSelectFile.Show
```



```

Case 1, 2, 4, 5, 7
    frmInputParameters.Show
Case 8
    frmInputParam2.Show
Case 10, 11, 12, 13, 14
    frmFCGModel.Show
Case Else
    Exit Sub
End Select
Unload frmCheck

End Sub

Private Sub calcDeltaK()
    PI = 3.14159265359
    surfacecrack = G * newcrackdepth

    If surfacecrack = 0 Then
        aspect = 0
        cdivt = 0
        adivr = 0
    Else: aspect = surfacecrack / newcrackdepth
        cdivt = surfacecrack / (T * 1000)
        adivr = newcrackdepth / (R * 1000)
    End If
    rdivt = R / T

    phi = 0
    lambda = 1 / (1 + adivr * Cos(0.9 * phi))
    gamma = Sqr(cdivt) * (((newcrackdepth / 1000) + R) / W)

    'shape factor determination
    If aspect <= 1 Then
        Qfactor = 1 + 1.464 * aspect ^ 1.65
    Else
        Qfactor = 1 + 1.464 * (1 / aspect) ^ 1.65
    End If

    'boundary correction factor determination

    M2 = 0.05 / (0.11 + aspect ^ (3 / 2))
    M3 = 0.29 / (0.23 + aspect ^ (3 / 2))
    g1 = 1 - ((cdivt ^ 4) * Sqr(2.6 - 2 * cdivt) / (1 + 4 * aspect)) * Cos(phi)

```

```

g2 = (1 + 0.358 * lambda + 1.425 * (lambda ^ 2) - 1.578 * (lambda ^ 3) + 2.156 * (lambda ^ 4)) / (1 + 0.08 * (lambda ^ 2))
g3 = 1 + 0.1 * ((1 - Cos(phi)) ^ 2) * (1 - cdivt) ^ 10
g4 = Kt * (0.36 - 0.32 / Sqr(1 + adivr))
g5 = 1 + Sqr(aspect) * (0.003 * (rdivt ^ 2) + 0.035 * rdivt * (1 - Cos(phi) ^ 3)) - 0.35 * (cdivt ^ 2) * ((1 - 0.5 * aspect) ^ 3) *
Cos(phi)
Fwidth = 1 + 2.17 * (gamma ^ 2) - 3.4 * (gamma ^ 4) + 3.7 * (gamma ^ 6)

```

```

If aspect <= 1 Then

```

```

    M1 = 1

```

```

    Fphi = ((aspect ^ 2) * ((Cos(phi)) ^ 2) + ((Sin(phi)) ^ 2)) ^ (1 / 4)

```

```

End If

```

```

If aspect > 1 Then

```

```

    M1 = Sqr(1 / aspect) * (1.04 - 0.04 * (1 / aspect))

```

```

    Fphi = (((1 / aspect) ^ 2) * ((Sin(phi)) ^ 2) + ((Cos(phi)) ^ 2)) ^ (1 / 4)

```

```

End If

```

```

Fsn = (M1 + M2 * ((cdivt) ^ 2) + M3 * ((cdivt) ^ 4)) * g1 * g2 * g3 * g4 * g5 * Fphi * Fwidth

```

```

delK = S * Sqr(PI * (surfacecrack / 1000000) / Qfactor) * Fsn

```

```

End Sub

```

```

Private Sub SysInfo1_DisplayChanged()

```

```

If Screen.ActiveForm.Width <> sysDetectOS.WorkAreaWidth Then

```

```

    Screen.ActiveForm.Left = sysDetectOS.WorkAreaLeft

```

```

    Screen.ActiveForm.Width = sysDetectOS.WorkAreaWidth

```

```

End If

```

```

If Screen.ActiveForm.Height <> sysDetectOS.WorkAreaHeight Then

```

```

    Screen.ActiveForm.Top = sysDetectOS.WorkAreaTop

```

```

    Screen.ActiveForm.Height = sysDetectOS.WorkAreaHeight

```

```

End If

```

```

End Sub

```

Form frmAbout

```

Private Sub sysDetectOS_DisplayChanged()

```

```

If Screen.ActiveForm.Width <> sysDetectOS.WorkAreaWidth Then

```

```

    Screen.ActiveForm.Left = sysDetectOS.WorkAreaLeft

```

```

    Screen.ActiveForm.Width = sysDetectOS.WorkAreaWidth

```

```

End If

```

```

If Screen.ActiveForm.Height <> sysDetectOS.WorkAreaHeight Then

```

```

    Screen.ActiveForm.Top = sysDetectOS.WorkAreaTop

```

```

    Screen.ActiveForm.Height = sysDetectOS.WorkAreaHeight

```

```

End If

```

```

End Sub

```

```

Private Sub cmdContinue_Click()

```

Unload Me

End Sub

Private Sub SysInfo1_DisplayChanged()

If Screen.ActiveForm.Width <> sysDetectOS.WorkAreaWidth Then

Screen.ActiveForm.Left = sysDetectOS.WorkAreaLeft

Screen.ActiveForm.Width = sysDetectOS.WorkAreaWidth

End If

If Screen.ActiveForm.Height <> sysDetectOS.WorkAreaHeight Then

Screen.ActiveForm.Top = sysDetectOS.WorkAreaTop

Screen.ActiveForm.Height = sysDetectOS.WorkAreaHeight

End If

End Sub

Form frmFCGModel

Private Sub sysDetectOS_DisplayChanged()

If Screen.ActiveForm.Width <> sysDetectOS.WorkAreaWidth Then

Screen.ActiveForm.Left = sysDetectOS.WorkAreaLeft

Screen.ActiveForm.Width = sysDetectOS.WorkAreaWidth

End If

If Screen.ActiveForm.Height <> sysDetectOS.WorkAreaHeight Then

Screen.ActiveForm.Top = sysDetectOS.WorkAreaTop

Screen.ActiveForm.Height = sysDetectOS.WorkAreaHeight

End If

End Sub

Private Sub cmdContinue_Click()

frmSelectFile.Show

End Sub

Private Sub SysInfo1_DisplayChanged()

If Screen.ActiveForm.Width <> sysDetectOS.WorkAreaWidth Then

Screen.ActiveForm.Left = sysDetectOS.WorkAreaLeft

Screen.ActiveForm.Width = sysDetectOS.WorkAreaWidth

End If

If Screen.ActiveForm.Height <> sysDetectOS.WorkAreaHeight Then

Screen.ActiveForm.Top = sysDetectOS.WorkAreaTop

Screen.ActiveForm.Height = sysDetectOS.WorkAreaHeight

End If

End Sub

Form frmInput Parameters

Dim G As Variant

Dim crackdepth As Single, aspect As Single

Private Sub cmdContinue_Click()

frmFCGModel.Show

End Sub

Private Sub SysInfo1_DisplayChanged()

If Screen.ActiveForm.Width <> sysDetectOS.WorkAreaWidth Then

Screen.ActiveForm.Left = sysDetectOS.WorkAreaLeft

Screen.ActiveForm.Width = sysDetectOS.WorkAreaWidth

End If

If Screen.ActiveForm.Height <> sysDetectOS.WorkAreaHeight Then

Screen.ActiveForm.Top = sysDetectOS.WorkAreaTop

Screen.ActiveForm.Height = sysDetectOS.WorkAreaHeight

End If

End Sub

Private Sub txtG_Change()

G = Val(txtG.Text)

End Sub

Dim A As Single, B As Single, Vo As Single

Dim D As Single, E As Single

Private Sub cmdContinue_Click()

frmInputParam2.Show

End Sub

Private Sub SysInfo1_DisplayChanged()

If Screen.ActiveForm.Width <> sysDetectOS.WorkAreaWidth Then

Screen.ActiveForm.Left = sysDetectOS.WorkAreaLeft

Screen.ActiveForm.Width = sysDetectOS.WorkAreaWidth

End If

If Screen.ActiveForm.Height <> sysDetectOS.WorkAreaHeight Then

Screen.ActiveForm.Top = sysDetectOS.WorkAreaTop

Screen.ActiveForm.Height = sysDetectOS.WorkAreaHeight

End If

End Sub

Private Sub txtA_Change()

A = txtA.Text

End Sub

Private Sub txtB_Change()

B = txtB.Text

End Sub

Private Sub txtD_Change()

D = txtD.Text

End Sub

Private Sub txtE_Change()

E = txtE.Text

End Sub

Private Sub txtVo_Change()

Vo = txtVo.Text

End Sub

Form frmIntro

Private Sub cmdAbout_Click()

frmAbout.Show

End Sub

Private Sub cmdContinue_Click()

Unload frmIntro

frmInputParameters.Show

End Sub

Private Sub cmdQuit_Click()

Unload frmIntro

End Sub

Private Sub SysInfo1_DisplayChanged()

If Screen.ActiveForm.Width <> sysDetectOS.WorkAreaWidth Then

Screen.ActiveForm.Left = sysDetectOS.WorkAreaLeft

Screen.ActiveForm.Width = sysDetectOS.WorkAreaWidth

End If

If Screen.ActiveForm.Height <> sysDetectOS.WorkAreaHeight Then

Screen.ActiveForm.Top = sysDetectOS.WorkAreaTop

Screen.ActiveForm.Height = sysDetectOS.WorkAreaHeight

End If

End Sub

Form SelectFile

Private Sub cmdFileNo_Click()

txtFilename = " Then go find the right file! "

End Sub

```

Private Sub cmdFileYes_Click()
Dim filename As String
    filename = Datafile
    frmCheck.Show
End Sub

Private Sub Dir1_Change()
    File1.Path = Dir1.Path    'sets file path
End Sub

Private Sub Drive1_Change()
    Dir1.Path = Drive1.Drive    'sets directory path
End Sub

Private Sub File1_Click()
    txtFilename = File1.Path + "\" + File1.filename
    txtNewFile = File1.Path + "new" + File1.filename
    txtFCGfile = File1.Path + "\FCG" + File1.filename
End Sub

Private Sub SysInfo1_DisplayChanged()
    If Screen.ActiveForm.Width <> sysDetectOS.WorkAreaWidth Then
        Screen.ActiveForm.Left = sysDetectOS.WorkAreaLeft
        Screen.ActiveForm.Width = sysDetectOS.WorkAreaWidth
    End If
    If Screen.ActiveForm.Height <> sysDetectOS.WorkAreaHeight Then
        Screen.ActiveForm.Top = sysDetectOS.WorkAreaTop
        Screen.ActiveForm.Height = sysDetectOS.WorkAreaHeight
    End If
End Sub

```

Figure 76: VB6.0 Introduction Form.

Figure 77: VB6.0 About Form.

☐ Input Parameters

In this form you will be prompted for the necessary values required for the conversion of probe voltage to crack depth size as well as crack area. Follow the instructions provided.

Probe Voltage to Crack Depth:

After performing numerous calibration tests, you should have determined experimentally the relationship between probe voltage and crack depth. In doing so, an equation of this approximation should have been generated and is of the form:

$$\Delta V = Ae^{B(a)}$$

where: a = crack depth (microns),
 ΔV is the change in probe voltage (V) from the onset of crack growth,
 A and B are the coefficients for the best fit exponential.

Since the probe voltage is supplied by the raw data file, the user must input only the coefficients, (A,B) to this equation and the initial voltage reading, V_0 , (stable voltage reading before crack begins to propagate) Enter the values below.

$\Delta V = \sqrt[1.9554971]{e^{\sqrt[0.0042552]{a}}}$

 and $V_0 = \sqrt{\hspace{1cm}}$ (Volts)

Probe Voltage to Crack Area:

In the same manner, a relationship between the crack area (mm^2) and probe voltage should also have been generated. The equation should be of the form:

$$\Delta V = D(\text{area})^2 + E(\text{area})$$

Enter the values for the coefficients below.

$\Delta V = \sqrt[9.304928]{A}^2 + \sqrt[40.393215]{A}$

Figure 78: VB6.0 Input Parameters Form.

In this form you will be asked to supply the parameters that are related to the crack aspect ratio, i.e. the surface crack length (2c) and the crack depth (a) relationship.

Crack Aspect Ratio

After performing numerous calibration tests, you should have determined experimentally the relationship between surface crack length and crack depth. In doing so, an equation of this approximation should have been generated and is of the form:

$$c = G \times (a)$$

where: a = crack depth (microns)
c = 1/2 the surface crack length
G is the coefficient to the above relationship

Since the crack depth has already been determined, the user must input only the coefficients, (G and M) to this equation. Enter the values below:

$$\frac{\text{half surface}}{\text{crack length}} = c = \sqrt{1.438725} \times (a)$$

Figure 79: VB6.0 Input Parameters-2 Form.

In this step you will be asked to verify the dimensions of the Single Edged Notch (SEN) or Double Edged Notch (DEN) specimen that you are using. If the dimensions or properties do not match to your specifications please make the appropriate changes in the textbox provided. The information is required for the Newman small crack growth model that is used to determine the stress intensity factor K .

$$K = S \sqrt{\frac{\pi c}{Q}} \cdot F_{SM} \left\{ \frac{c}{a}, \frac{c}{L}, \frac{a}{r}, \frac{a}{t}, \frac{r}{L}, \frac{r}{t}, \frac{L}{W}, \frac{t}{W}, \phi \right\}$$

Specimen dimensions and loading conditions

notch radius, r :	1.2446	mm
specimen width, w :	10.668	mm
half thickness, t :	2.667	mm
remote tensile stress, S :	562.36	MPa
stress conc. factor, K_t :	2.682	

Figure 80: VB6.0 FCGR Model Form.

Select the drive location of the raw data file that is to be processed:

d: [user disk]

Choose the directory under which the raw data file was stored. If necessary, you may open any subdirectories that follow. A list of files that are stored in each directory will be displayed below.

d:\

- My Documents
- Rick's Data Files
- Fatigue tests

Within this listbox, choose the raw data filename that you wish to process. Only the filenames ending with the extension *.dat will be displayed for the user.

CAL1.DAT
CAL2.DAT
CAL3.DAT
CAL4.DAT
CAL5.DAT
CAL6.DAT

In this the correct path and filename that you are looking for?

d:\My Documents\Rick's Data Filestestigug testst\CAL1.D\

The new data file name will be stored as:

d:\My Documents\Rick's Data Filestestigug testst\newCAL

An extra file for FCG parameters will be stored as:

d:\My Documents\Rick's Data Filestestigug testst\FCGCAL

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VB6.0 Test Parameters List Form

Below is a list of the parameters that you have entered so far ..... if any of these values are not correct please go back and change them now by clicking on the appropriate parameter in the listbox

raw data path and filename:
coefficient A :
coefficient B :
coefficient D :
coefficient E :
initial probe voltage (volts), Vo :
coefficient G :
notch radius (mm), r :
specimen width (mm), w :
half specimen thickness (mm), t :
remote tensile stress (MPa), S :
stress conc. factor, Kt :
raw data path and filename :
FCG model data path and filename :

d:\My Documents\Rick's Data Files\Fatigue tests\CAL1.DAT
1.9554971
0.0042552
9.304928
40.393215
1.98
1.438725
1.2446
10.668
2.667
562.36
2.682
d:\My Documents\Rick's Data Files\Fatigue tests\newCAL1.DAT
d:\My Documents\Rick's Data Files\Fatigue tests\FCGCAL1.DAT

If all these settings are correct, click the Process button to generate the new data file that contains crack depth, crack area, dN/dM, and data K information. The new data filenames are as stated in the last two lines of the above list.

Figure 82: VB6.0 Test Parameters List Form.