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**CRACK INITIATION AND SHORT CRACK GROWTH CHARACTERISTICS
OF NOTCHED MERL 76 SPECIMENS**

by

Daniel M. Fuleki, B.Eng.

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

Master of Engineering

Department of Mechanical and Aerospace Engineering

**Ottawa-Carleton Institute for
Mechanical and Aerospace Engineering**

**Carleton University
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**CRACK INITIATION AND SHORT CRACK GROWTH CHARACTERISTICS
OF NOTCHED MERL 76 SPECIMENS**

submitted by Daniel M. Fuleki

in partial fulfilment of
the requirements for the degree of
Master of Engineering

Thesis Supervisor

Chair, Department of Mechanical and Aerospace Engineering

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ABSTRACT

The initiation and growth of a semi-elliptical fatigue crack from a notched Merl 76 specimen was monitored using the alternating current potential drop (ACPD) technique. This technique was capable of detecting and sizing cracks in the range of 200 μm in depth at room temperature and 480°C. Newly initiated cracks were semi-elliptical in shape and their aspect ratios (half surface length by crack depth) decreased with increasing crack length indicating that crack growth into the specimen was faster than the crack growth along the notch root. The crack growth rate as a function of crack length decreased at a crack depth of about half the monotonic notch root plastic zone length. Describing the crack growth rate within the notch root plastic field using the linear elastic fracture mechanics (LEFM) based Newman's model resulted in an overestimation of ΔK . A modification to Newman's model is proposed where the Neuber rule is implemented to account for the effects of plasticity on the stress concentration factor (SCF). The modified Newman's model brought the data into agreement with the long crack data.

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NOMENCLATURE

(Alphabetical Order)

γ	FCC matrix structure in nickel base superalloys
γ'	ordered cubic precipitate found in nickel base superalloys
Δ	ACPD probe spacing
δ	AC skin depth
ρ	notch root radius
ρ	conductivity
σ_e	endurance limit, maximum stress at which a crack will not grow until failure
a	half the surface crack length
ACPD	alternating current potential drop
AR	as received
c	crack depth
CCG	creep crack growth
CCGR	creep crack growth rate
CPI	crack propagation interval
d_1	measured ACPD crack length
d_2	corrected ACPD crack length
DEN	double edge notch
DTD	damage tolerant design

FCGR	fatigue crack growth rate
HAZ	heat affected zone
HCF	high cycle fatigue
HIP	hot isostatic press
IAR	institute for aerospace research
IPB	interphase boundary
K_{eff}	effective stress intensity factor
K_F	fatigue concentration factor
K_J	J-integral elastic-plastic stress intensity factor
K_p	crack tip plastic zone corrected stress intensity factor
K_{th}	threshold stress intensity factor
LCF	low cycle fatigue
LCG	long crack growth
LCGR	long crack growth rate
LEFM	linear elastic fracture mechanics
P/M	powder metallurgy
PD	potential drop
P_{max}	maximum load in fatigue load cycle
P_{min}	minimum load in fatigue load cycle
PPB	prior particle boundaries
PSB	persistent slip band
PSD	phase sensitive detector

PWC	Pratt and Whitney Canada
R	load ratio, ($P_{\max}/P_{\min}$)
RT	room temperature
SCG	short crack growth
SCGR	short crack growth rate
sd	standard deviation
SEN	single edge notch
SII	safe inspection interval
SML	structures and materials laboratory
T_M	melting temperature
VR	voltage ratio, ($PD_{\text{notch}}/PD_{\text{ref}}$)

1.0 INTRODUCTION

The application of cyclic loads to engineering materials can lead to initiation and propagation of fatigue cracks. Prior to the development of safe life and fracture mechanics principles, structural designers applied large safety factors to guard against fatigue failures [1]. This ensured that the critical stresses seen by the component were sufficiently low to prevent the initiation of fatigue cracks over the designated life time or the maximum number of load cycles.

In the case of gas turbine disks, low cycle fatigue (LCF) was identified as the major life limiting factor in the sixties and by 1975, a safe life methodology was applied to over 75% of the major structural components in advanced aeroengines [2]. This life cycle management methodology was statistically based on the LCF failure life of laboratory specimens where 1 in every 1000 retired components was expected to develop an LCF crack. This is obviously an inefficient use of component life since the remaining 999 were removed from service in a crack free condition thus having some fatigue life remaining.

Fracture mechanics has provided a better understanding of fatigue crack growth (FCG) which allows the prediction of a safe amount of crack propagation for a component without causing catastrophic failure during service. The application of these fracture mechanics principles to design form the basis of the damage tolerant design (DTD) methodology. This methodology was adopted by the USAF in 1984 which stated,

"critical rotating components must be able to tolerate propagating cracks without failure" [3]. Since the DTD methodology allows a crack to exist, the components that would be retired using the safe life methodology could remain in service thus increasing their service life. In DTD, it is assumed a flaw exists in a new component, the size of which is assumed to be the minimum detectable crack size, a_i , of the nondestructive evaluation (NDE) technique used to inspect the component. Using the fatigue crack growth rate (FCGR) properties of the material and the loading spectrum seen by the component in service, the crack propagation interval (CPI) can be predicted. The CPI can be defined as the number of loading cycles required to propagate the crack from a_i to a critical crack length, a_c , and by applying a safety factor, a safe inspection interval (SII) can be established. At the end of each SII, the component is removed from service and reinspected. If a crack is found, the component is retired, otherwise, it is returned to service. This inspection routine is shown graphically in Figure 1.1. DTD allows the life of a component to be fully utilized based on the understanding the material's FCGR properties. A critical parameter in DTD is a_i , where a decrease in a_i can increase the SII significantly thus reducing the cost of down time due to inspections. This signifies the importance of using sensitive NDE techniques for improving the cost effectiveness of DTD.

In an effort to detect smaller cracks in order to decrease a_i , improved NDE techniques have been able to detect cracks in the short crack growth (SCG) regime [1,4]. An example is the detection of a 40 μ m crack in a turbine disk and a 30 μ m crack in a

titanium bolt using an alternating current potential drop (ACPD) NDE technique [5]. Having an initial detectable crack length, a_i , in this region can greatly increase the SII, however, in order to describe the growth of such a small crack, the short crack growth rate (SCGR) characteristics of the material must be known. To determine the SCGR characteristics, one requires a crack monitoring technique sensitive enough to reliably size such small cracks. Techniques typically used are in-situ SEM, replication, DCPD and ACPD. In-situ SEM tests are restrictive since the fatigue specimen and loading attachments must fit inside a scanning electron microscope (SEM) chamber, as well, the testing can only be performed in vacuum. The replication technique is labour intensive since the test must be stopped and replicas must be made at regular intervals which also makes it impractical to apply to aggressive environments [6]. Both the ACPD and DCPD techniques allow automated testing and are applicable to aggressive environments. However, the AC technique is being viewed as the most promising technique for SCG testing (which will be discussed in later sections) and this is why it was chosen for the present study.

1.1 Anomalous Behaviour of Short Cracks

A common conclusion reached by researchers studying SCG is that short cracks grow differently than long cracks. This section will illustrate some general observations, however, a description as to why these anomalies occur will be provided in later sections. Summing up the anomalous behaviour of SCG, many researchers have noted a higher and more irregular FCGR of short cracks compared to long cracks [7-10] and in addition, the

occurrence of SCG below the long crack (LC) threshold [11,12] and the endurance limit [13]. Suresh and Ritchie provide some explanations for these differences between long and short crack growth behaviour and these include [9];

- i) The lack of similitude between SCG and LCG using LEFM due to an inappropriate explanation of the SC driving force. When the plastic zone size is in the order of the crack length, the higher order terms ignored by LEFM are no longer negligible in the description of the stress field about the crack tip,
- ii) the microstructural features which are insignificant at long crack lengths are influential to SCG,
- iii) the different crack extension mechanisms,
- iv) the differences in closure effects and,
- v) the differences in crack tip environment, ie. the stress state at the crack tip.

In an effort to understand SCG behaviour to explain these anomalies, researchers have concentrated on three different areas as described by Lankford et al [11];

- i) LEFM is assumed to be applicable to both LCG and SCG. An example is adjustment of the short crack length to give a larger effective length which compensates for its higher FCGR,
- ii) The second area uses LEFM, however assumes closure to be the source of discrepancy where ΔK_{eff} is assumed to be the proper term and,
- iii) The final hypothesis is that LEFM is not applicable thus requiring other methods to describe SCG.

In this study, the application of LEFM to the SCGR will be examined using a stress intensity factor (SIF) which is applicable to a semi-elliptical crack shape.

1.2 Objectives

The two main objectives of this thesis are;

- 1) To implement and determine the capabilities of the ACPD technique for monitoring crack initiation and SCG using current focusing technology.
- 2) To use the ACPD system to examine the initiation and SCG characteristics of the extruded and isothermally forged (Gatorized) powder metallurgy Merl 76 turbine disk material under constant amplitude fatigue loading conditions.

2.0 LITERATURE REVIEW

2.1 Fatigue Crack Initiation

A crack can either initiate within a grain or along the grain boundaries and these are respectively known as transgranular and intergranular initiation sites. It is generally accepted that the formation of a transgranular crack and the initial stages of short crack growth occur due to residual slip developed during cyclic loading [14-16]. The number of slip planes that form are proportional to the magnitude of the alternating stress; increasing the alternating stress increases the number of active slip planes [17]. It has been observed that transgranular crack initiation usually occurs in a slip plane oriented 45° with respect to the loading axis, the plane of maximum shear [14-16]. An additional factor in the initiation of a surface crack along this plane is the fact that it is in a plane stress state. Since the crack is close to the surface, the stress perpendicular to the surface is zero thereby increasing the maximum shear stress along the plane of maximum shear. It has been found that a crack can initiate in this manner below the endurance limit, however, such a crack eventually stops propagating [13¹,18,19]. Several authors have observed this maximum non-propagating crack length to coincide with the transition crack length from Stage I to Stage II growth.

The formation of a crack occurs by the movement of dislocations along a slip plane to form a persistent slip band (PSB) which elongates in the direction of the slip vector and

¹ Blom et al observed arrested cracks in specimens which did not fail after 10⁸ and 10⁹ loading cycles

becomes a potential initiation site for a microcrack [15]. The movement of dislocations along the slip plane typically ends at a grain boundary at which point there is an incoherency in the lattice structure with the next grain. This incoherency impedes the advance of the dislocations causing a dislocation pile-up which in turn, slows down the advance of the crack [20,21], Figure 2.1. This indicates that the crack growth is affected by the interaction of the advancing dislocations with the grain boundary and not due to interaction of the crack tip with the grain boundary. This phenomenon was observed by Zhang et al who monitored the plastic zone of cracks one or two grain diameters (about 50-100 μm) in length in 7075 aluminum using electron diffraction [20]. When the tip of the plastic zone reached the grain boundary, the FCGR slowed. The concept of growth by slip was further reinforced by their observation of a long and thin plastic zone at the crack tip running in the direction of the crack, Figure 2.2.

While intergranular crack initiation can also occur at room temperature, it is typically associated with elevated temperature FCG [10,22]. Similar to transgranular crack initiation, intergranular cracks typically form on a plane inclined at 45° with respect to the loading axis [10,22]. This being the plane of maximum shear stress may indicate that the bonds along the grain boundaries are broken because of shearing where the grain boundary dislocations move preferentially allowing a crack to initiate.

Wang et al observed intergranular initiation in Waspaloy at room temperature, however, the crack moved to transgranular growth after a crack length of less than a grain [10].

Arthuthnot observed both inter and transgranular initiation but found the mode of initiation to be a function of strain [17]. Fatigue testing of Nimonic PE16, a γ precipitate strengthened alloy, showed that at low strains ($<2.2\%$), deformation occurred only locally by the formation of slip bands from which microcracks initiated. At higher strains ($>2.2\%$), deformation occurred both along the slip bands and in the bulk material. This resulted in a more homogeneous deformation which created intergranular crack initiation. Zhang et al observed three distinct initiation sites with their four-point bend testing of Waspaloy [22]. In descending order of abundance, transgranular initiation sites at PSB's, intergranular initiation, and transgranular initiation where no visible PSB's existed, Figure 2.3. An explanation offered for the lack of visible PSB's was the reduced amount of deformation of the slip bands due to strain hardening which made them less distinguishable [17]. At all three sites, cracks typically initiated in larger grains at an angle of 45° with respect to the loading axis. Initiation at an angle of 45° is logical as explained previously, however, the preference of initiation in larger grains is less obvious. Tanaka et al explains that larger grains generate more dislocations under a given applied stress which results in a higher likelihood of PSB formation and an increase in the potential for transgranular initiation [24]. It is important to realize that this mechanism occurs in more than just one grain resulting in multiple crack initiation sites [25]. Eventually, however, one crack dominates and grows until failure [14].

Fatigue cracks can also initiate at the surface due to scratches, inclusions or casting pores and at a triple point [26-28]. Miller states that initiation will typically occur at a surface

triple point unless one or both of the following conditions are satisfied [14];

- 1) A slip system within a grain or a grain boundary is favourably oriented in the direction of maximum shear,
- 2) A surface defect exists to create a stress raiser.

Defects such as inclusions, pores and particles may also exist below the material surface which can give rise to sub-surface crack initiation [29]. This is especially true if the inclusion or particle is brittle which is surrounded by a ductile matrix. Under stress, the particle will resist the strain exerted by the matrix causing the particle to eventually fracture resulting in a crack initiation site [30]. Swain and Newman found this to be the case for reversed loading conditions in 2024-T3 DEN specimens where initiation occurred mainly at inclusions near the surface along the notch centre [31]. The MC precipitates found in many superalloys, have also been shown to reduce the crack initiation life because of their tendency to crack during processing [3]. In general, it can be deduced that a crack will initiate from a discontinuity in the material whether it be microscopic, macroscopic, at or below the surface.

2.1.1 Temperature Effects

Runkle and Pelloux examined the crack initiation characteristics in four different microstructures of Astroloy at different temperatures [32]. Their results show that at temperatures below $0.4T_M$, transgranular initiation was predominant, typically occurring along PSB's due to the transport of oxygen along slip bands to the crack tip [32]. At temperatures above $0.6T_M$, intergranular initiation became the predominant initiation

mode. In between this range of temperatures, a mixed mode initiation occurs.

A crack will initiate in the weakest region, either intergranularly if the grain boundary is the weakest site or transgranularly if the slip plane is the weakest site. As discussed in the previous section, room temperature (RT) initiation is typically transgranular demonstrating that at lower temperatures, the slip planes are more susceptible to initiation than the grain boundaries. Raising the temperature increases the number of intergranular initiation sites indicating a weakening of the grain boundaries relative to the slip planes. This weakening occurs because of the grain boundary's susceptibility to environmental interaction due to the incoherency of the lattice structure at the grain boundaries. Atoms in this region have a lower packing factor which increases their exposure to environmental effects. The severity of the interaction between the grain boundary and the environment increases with increasing temperature which results in a higher degree of intergranular initiation.

2.1.2 Notch Effects

Smith and Miller recognized that the localized plastic deformation at the notch root can be significant enough to initiate a crack, however, the crack will only continue to grow after leaving the notch's plastic zone if the crack tip plastic field is large enough to drive the crack [33]. This shows that at a sharp notch, a crack can initiate at a low bulk stress, however, it will probably not be able to grow after leaving the stress field of the notch due to insufficient crack tip plastic deformation [33]. The minimum alternating stress

required for crack initiation at a notch is given by [33],

$$\sigma_{\min} = \frac{\sigma_e}{K_f} \quad (2.1)$$

where σ_e is the fatigue limit and K_f is the fatigue concentration factor given by,

$$K_f = \left[1 + 7.69 \sqrt{\frac{D}{\rho}} \right]^{0.5} \quad (2.2)$$

where D is the notch depth and ρ is the notch root radius [34], Figure 2.4.

2.2 Short Crack Growth

The short crack regime is where the FCGR is anomalously high compared with long crack growth at a given ΔK . That is, the region where similitude using ΔK breaks down due to the small size of the crack. One such condition is crack growth below the long crack threshold. In an attempt to establish a length associated with SCG behaviour, the transition from short to long crack growth was assumed to be the point at which the LEFM parameter ΔK_{th} is applicable [35].

Kitagawa and Takahashi determined this transition crack length to be,

$$l = \frac{1}{\pi} \left(\frac{\Delta K_{th}}{Y \left(\frac{2}{3} \Delta \sigma_{cy} \right)} \right)^2 \quad (2.3)$$

where ΔK_{th} is the long crack threshold SIF, $\Delta\sigma_y$ is the cyclic yield stress at $R=-1$ and Y is the shape factor for the given specimen and crack geometry, Figure 2.5. Crack lengths below l are within the SCGR regime, however, crack growth beyond this length may still exhibit an anomalous FCGR indicating that the short crack regime can extend beyond this length.

2.2.1 Short Crack Growth Regions

SCG can be divided into two regions, Figure 2.5 [18];

- i) A crack length which is small relative to the microstructure typically displaying irregular faceted Stage I growth [4,9] and,
- ii) a physically small crack where propagation is due to Stage II growth while still displaying anomalous growth behaviour [12,13,32,36].

2.2.1.1 Stage I Growth

The first region is one in which traditional theoretical descriptions of crack growth are not applicable (eg. LEFM, Paris Law). The difficulties arise due to the lack of material continuity with respect to the crack. This results in a continuum mechanics limitation since the surrounding material no longer resembles a continuum but rather, a discrete network of microstructural features. In addition to the lack of continuity, LEFM is typically invalid because of the requirement for the crack to be much larger than the crack tip plastic zone [9]. The maximum allowable crack tip plastic zone size to crack length ratios range from $1/15^{th}$ [9] to $1/50^{th}$ [14]. At short crack lengths, the crack tip plastic zone size does not fall within these boundaries, Figure 2.2, thereby making LEFM

inapplicable.

After a crack has initiated in a grain, the crack growth will slow down or arrest when its plastic zone either reaches a grain boundary as discussed in Section 2.1 or interacts with a tougher particle [27,37]. Elongation of the crack occurs when a PSB is formed in an adjacent grain which allows the crack to extend into the next grain [13,20,29,36]. Since there is an incoherency in the lattice structure across the grain boundary, the activated slip plane will have a different orientation thus causing a deflection of the crack [36], Figure 2.6. Navarro analytically determined the activation of the slip system in the next grain to occur when the increasing stress concentration due to the piling up of the dislocations reaches the critical shear stress in the next grain [38]. The ease of propagation into the next grain is dependant on the orientation of that grain. If it is not favourably oriented, it will be more difficult for the slip band to extend, however, if it is favourably oriented, the grain boundary will not significantly inhibit crack extension [11].

Upon approaching a grain boundary or other microstructural feature, many authors have observed the SCGR to decrease followed by an acceleration in growth rate after overcoming the obstacle [13,21,28]. The combination of crack arrest and deflection at the grain boundaries results in the irregular growth rate and the faceted fracture surface associated with Stage I SCG as observed by many authors [13,20,22,24,29,36]. As the crack length increases, microstructural effects become less pronounced due to larger crack tip deformation [13,28]. This agrees with the observation made by several authors who

have recognized that the first grain boundary encountered has the most significant retarding effects on crack propagation [10,22,24].

As the crack moves from Stage I to Stage II growth, the loading mode changes from Mode II to Mode I. Zhang et al observed Stage I growth within the first several grains produced by local Mode II loading which formed a long and narrow plastic zone in the direction of the crack, Figure 2.2 [20]. After reaching several grain diameters in length, the amount of Mode II loading increased resulting in a semicircular plastic zone [9,20].

The crack shape tends to be somewhat irregular during Stage I growth since the crack takes the shape of the grain in which it exists [39]. Eventually, the crack develops a semi-elliptical shape with the majority of the growth initially along the notch root, Figure 2.7.

Specifically for γ/γ' nickel base superalloys, Stage I growth is extensive because of the heterogeneous slip associated with this type of material [32]. Nickel base superalloys are prone to high degrees of planar slip because of their low stacking fault energy and these partial dislocations make cross-slip difficult [2]. Larger amounts of Stage I growth tends to decrease the fatigue life compared to homogeneous slip since most of the strain occurs along a specific plane.

2.2.1.2 Stage II Growth

With Stage II commencing when the crack tip plastic zone size is equivalent to the grain size, homogeneous deformation occurs which results in the activation of more than one slip plane. This changes the growth mechanism from that of Stage I to Stage II with predominantly Mode I loading. The crack tip deformation is sufficient such that the amount of retardation caused by microstructural features decreases resulting in a smoother FCGR curve [10,22], Figure 2.6. Unlike that of Stage I growth, a crack undergoing Stage II growth will not arrest under constant amplitude loading [19].

Stage II growth for a SC is similar to that of LC with the difference being an anomalously high FCGR. Two conditions that would result in a higher growth rate for a Stage II SC are less severe closure effects and the effects of the notch root stress plastic field which are discussed in the later sections.

A short crack in Stage II growth will have a typical plastic zone shape [20] and is usually semi-elliptical in shape. It can be seen that crack propagation is initially predominant along the surface resulting in a very low surface length to crack depth ratio, however, as the crack grows longer, growth into the specimen dominates causing a sharp increase in the crack depth to surface length ratio tending to $a/c \approx 1.1$, Figure 2.7. Futhulla et al observed that semi-elliptical cracks will tend to a semi-circular shape regardless of its initial shape and growth mechanism [39]. Wanhill and Schra established an equation describing crack shape as a function of crack length for testing aluminum SEN specimens

given by [41],

$$\frac{c}{a} = 0.9 - 0.25 \left(\frac{a}{t} \right)^2 \quad (2.4)$$

where t is half the specimen thickness, a is half the surface crack length and c is the crack depth, Figure 2.7. This equation was shown to be applicable for $(a/t) > 0.04$, below this value, variation of the crack shape with surface crack length follows a sigmoidal shape.

2.2.2 Transition from Stage I to Stage II FCG

The boundary defines the progression from irregular faceted Stage I growth to smoother Stage II growth [42]. Reported crack sizes at Stage I to Stage II transition is given in Table 1 for different materials.

For crack lengths about the same size as the crack tip plastic zone size, Lankford et al found that the crack tip plastic zone did not significantly deform the grain in which it was located showing a heterogeneous type of deformation [11]. As the crack length increased, the crack tip deformation increased, changing the deformation from a heterogeneous to homogeneous mode. It was observed that Stage II growth commenced when slip moved from heterogeneous planar slip to homogeneous dispersed slip [32]. This transition was coincidental with the crack tip plastic zone reaching the same size as the average grain diameter [21,26,32,36]. Transition to Stage II growth can also be determined by examining the FCGR rather than the crack tip deformation [13]. At the transition, the irregular crack growth rate associated with Stage I cracks becomes smoother Stage II

growth (ie. the da/dN vs. a FCGR curve smooths out).

2.2.2.1 Stress Effects

The transition discussed above occurs because of the increase in crack tip deformation since the SIF increases with crack growth. The deformation at the crack tip is eventually large enough to lead to a homogenous stress distribution which results in dispersed slip.

Crack tip deformation is also a function of the applied stress. An increase in the applied stress results in more crack tip deformation for a given crack length resulting in an earlier transition to Stage II growth. This was observed by Laird and Smith who found that an increase in the alternating stress reduced the amount of Stage I fatigue crack growth.

2.2.2.2 Temperature Effects

As the temperature increases, the yield strength of a material drops [32]. If this drop in yield strength is due to ease of dislocation motion within the grains, larger plastic deformation within the grain will result for a given crack length and load. Therefore, the transition to Stage II growth will occur at smaller crack lengths in a given material as the temperature is increased. Examples of this effect can be seen in Table 2 where an increase in temperature changes the FCG from Stage I to Stage II for P/M Astroloy and Waspaloy [32].

2.2.2.3 Grain Size Effects

Decreasing the grain size will reduce the crack length at which the crack tip plastic zone is the same size as the grain size. This shows the significance of grain size on the transition crack length; smaller grains will yield shorter Stage I crack lengths. This phenomenon was observed by Brown et al when fatigue testing different grain structures of Astroloy [12]. The transition occurred at a $\Delta K \approx 17 \text{ MPa}\sqrt{\text{m}}$ for a grain diameter $\approx 13 \mu\text{m}$ and at a $\Delta K \approx 32 \text{ MPa}\sqrt{\text{m}}$ for a grain diameter $\approx 50 \mu\text{m}$.

2.2.3 Factors Affecting Short Crack Growth

2.2.3.1 Crack Closure

Closure effects are caused by the crack faces contacting before the crack opening load drops to zero. This implies that there is a minimum stress required for the crack to open, σ_{op} , Figure 2.8. Although there is some controversy with the traditional assumptions associated with crack closure from an energy stand point [43], for this study, it will be assumed that crack growth only occurs during the opening of the crack. Accounting for this effect gives an effective alternating SIF, ΔK_{eff} , which can be calculated by,

$$\Delta K_{eff} = K_{max} - K_{op}$$

where K_{max} is the SIF corresponding to the maximum stress and K_{op} is the SIF corresponding to the stress required to open the crack.

One cause of crack closure is the plastic wake created by the plasticity at the crack tip, Figure 2.9 [40]. The deformation at the crack tip causes a permanent expansion of the material within the crack tip plastic zone and as the crack grows through this region, this

expanded material meets behind the crack tip thus contacting before the load is zero, Figure 2.10. Oxides, debris from corrosion and rough, faceted fracture surfaces also prevent the crack from closing completely which are further sources of closure [7,9,13,44].

Closure is one of the most common explanations for the anomalously high SCGR [9,40,44,45]. A short crack, compared to a long crack, will have a smaller plastic wake since it has a smaller SIF, therefore, it will have less significant plastically induced closure effects, Figure 2.9. This results in a higher ΔK_{eff} for a short crack thus a higher FCGR. A newly initiated crack cannot experience closure due to plastic deformation since it has not grown into its residual plastic wake. Blom et al found closure effects did not prevail until a crack length just under the transition length from Stage I to II was reached [13]. Soniak and Remy, during their fatigue tests on Astroloy, observed that the material 0.5-1mm behind the crack tip caused crack closure which implies that closure can commence only after a minimum crack length [8]. The lack of closure due to plastic deformation within Stage I growth is logical when considering the plastic zone shape at the crack tip is long and narrow [20], Figure 2.2. This shows that only a small amount of material at the crack tip experiences plastic deformation thereby producing only a small plastic wake. Compensating for the more significant amount of closure associated with LCG may account for the anomalously high crack growth rates associated with short cracks [8,9].

Variation of closure with crack length shows ΔK_{eff} to initially decrease followed by an increase with the closure effects eventually reaching an equilibrium [31]. A schematic representation is given in Figure 2.11 which describes the SCG curve shown in Figure 2.12. Newman analytically examined crack initiation at a defect and observed a scenario where during the first stages of SCG, the crack opening stress (σ_{op}) increased rapidly which had the effect of decreasing ΔK_{eff} . Simultaneously, the increasing crack length had the effect of increasing ΔK_{eff} , however, the change in crack opening stress was larger thereby having a net effect of decreasing the SIF. Eventually, the crack opening stress levelled off as the crack length increased which eventually caused an increase in ΔK_{eff} [11,31]. This presents an additional reason (the other reason being impediment by microstructural features) why the SCGR may decrease to a minimum and then accelerate again or even arrest, Figure 2.12.

Newman also recognized that closure will vary with the amount of plastic deformation at the crack tip [31]. As closure effects increase, ΔK_{eff} decreases thus decreasing the cyclic plastic zone size where the fully reversed cyclic plastic zone size corrected for closure effects is given by,

$$\omega = \left(1 - \frac{\sigma_{op}}{\sigma_{max}} \right)^2 \frac{\rho}{4} \quad (2.6)$$

where ρ is the monotonic plastic zone size calculated using the maximum applied stress, σ_{op} is the stress at which the crack is opened and σ_{max} is the stress at maximum load. As the closure effects increase, σ_{op} increases which results in a reduction in the crack tip plastic zone size.

It is important to realize that closure effects provide one explanation of the anomalous SC behaviour, however, this is not the only factor affecting SCG. This is shown by the results of Lankford et al during the testing of an aluminum alloy where compensating for closure effects did not result in good agreement between the SCGR and LCGR [11]. In addition to closure effects, there may be differences in growth mechanisms causing a further lack of agreement between LCG and SCG.

2.2.3.2 Temperature

The mechanism for interaction between the environment and the crack front is analogous to that described in Section 2.1.1 for crack initiation. In summary, the environmental attack occurs preferentially along the grain boundaries which renders them weakened sites for crack initiation and growth [25]. Increasing the temperature will increase the amount of intergranular crack growth due to this interaction, Table 2 [32]. Although less abundant, transgranular growth can also occur at elevated temperatures where environmental effects involve the transportation of oxygen along slip bands ahead of the crack tip [32].

The magnitude of the environmental effects depends upon the severity of the environment interaction at the crack tip. This interaction is affected by the distance from the crack tip to the external environment (crack length), the amount of opening that occurs near the crack tip (crack closure) and the amount of time for the environment to interact with the crack tip (crack opening rate or hold time). Crack tip interaction with the environment decreases as the crack tip moves away from the specimen surface since it is more difficult for the environment to reach the crack tip [42]. Crack tip exposure is also affected by crack closure which will cause the crack faces behind the crack tip to contact before the load is zero. When the faces are in contact, interaction with the external environment will be reduced thus reducing its effects on the FCGR. This shows that environmental effects are magnified for short cracks due to their smaller closure effects and crack length. An additional parameter that effects both long and short cracks equally is the loading rate. If a crack opens slowly, there is more time for the external environment to enter the region near the crack tip compared to when the crack is opened and closed quickly [25,42]. Therefore, slower cyclic frequencies promote environmental effects causing a higher FCGR for a given ΔK [32].

An additional effect of higher temperatures is the increased capability of cross-slip to occur creating more homogeneous slip [2,46]. When a Stage I crack reaches an incoherency in the lattice structure such as a grain boundary, it must continue to grow along a new slip plane. This deviation from the original crack path will impede or arrest propagation of the crack. Increasing the number of slip systems that the crack can

propagate along increases the likelihood of the crack finding an adequate path to continue which reduces the impeding effect of microstructural features thus resulting in a higher crack growth rate. This was observed in Waspaloy tested in vacuum where the FCGR was appreciably higher at 500°C compared to 25°C resulting in a 23% decrease in fatigue life for crack growth from 20-400µm [46]. These results clearly show the effect of temperature on slip behaviour since the oxidation effect is completely removed.

Specifically for nickel base superalloys, an additional effect of elevated temperatures is the preferential oxidation of γ' precipitates [3,25]. Dai [25] found that Al, Cr and Ni atoms oxidized to Al_2O_3 and Cr_2O_3 with particles of $Ni(Cr,Al)_2O_4$. Formation of these oxides helped enhance the oxidation resistance of the material, however, this process preferentially reduced the γ' content because three out of the four elements that form the γ' phase, $Ni_3(Cr,Al,Ti)$, appear in the above noted oxides. This created a γ' depleted zone which resulted in a reduction in strength and fatigue resistance [25].

2.2.3.3 Microstructure

Critical in the fatigue life of a component is the amount of SCG that occurs where it has been shown that a reduction in Stage I growth will substantially increase a component's LCF life [32]. Controlling the SCGR properties of a material is critical in improving its fatigue life. One variable that can be modified to control the SCGR properties of a material is its microstructure.

Decreasing the grain size will increase yield strength and decrease SCGR because of the resultant increase in shear strength [36,38]. On a microscopic level, a decrease in the grain size will cause an increase in the number of grain boundaries and triple points which have a retarding effect on Stage I SCG. Therefore, an increase in the number of grains results in a slower SCGR. In addition, decreasing the grain size will mean that the crack tip plastic zone will reach the size of the grain sooner thereby decreasing the amount of Stage I growth [36]. There is, however, a trade off with a small grain size because smaller grains have a higher LCGR and CCGR [3]. The increase in the LCGR is likely due to a decrease in closure effects caused by the higher yield strength. A higher yield strength will reduce the amount of plastic deformation at the crack tip thereby reducing the residual plastic wake. This has the effect of increasing ΔK_{eff} thus increasing the LCGR. A decrease in closure effects will also decrease the threshold SIF associated with LCG [12,36]. The increase in the CCGR is due to the larger amount of grain boundary area associated with the smaller grain size [3].

In a study of the room temperature FCGR behaviour of three different microstructures of Astroloy, Brown et al found a necklace grain structure to be most resistant to Stage I growth, fine grain second and coarse grain third, Figure 2.13 [12]. Although small grains caused more grain boundary arrests, the large grains in the necklaced structure caused large crack deflections which increased closure effects due to fracture surface roughness, [12]. An additional effect of large deflections in the crack path is a decrease in the crack driving force due to the deviation from the path perpendicular to the load axis [2,47].

This combination of large and small grains optimized the resistance to SCG as well as reduced the negative effects of a strictly fine grain structure on the LCGR and CCGR properties.

In addition to the grain size, precipitation size and morphology have an important effect on the SCGR properties of a material. In nickel base superalloys, γ' precipitates impede the advance of the crack, which in some cases, requires the crack to loop around the precipitate which resulting in fatigue hardening [2]. This hardening causes slip to occur on other planes which produces a more homogenous deformation which slows the crack growth. Eventually the stress increases to a level at which precipitate cutting commences resulting in fatigue softening [2]. This effect is shown in Table 3 for several different nickel base superalloys. When these precipitates are present in the matrix, they impede slip band extension within the grain itself. Runkle et al found that fine $0.05\mu\text{m}$ γ' precipitates found in Astroloy significantly reduced the propagation rate of Stage I cracks with similar results obtained in Waspaloy where decreasing the γ' size decreased the Stage I FCGR [2]. Along with matrix precipitates, grain boundary precipitates can slow Stage I growth by blocking slip band extension into the next grain [25,32]. In Astroloy, the γ' morphology was found to affect the Stage I SCGR of fine and necklaced grain structures, however, it had no effect on the SCGR in the coarse grained material [12]. The effect of grain boundary precipitates on the LCGR was not significant, however, it has been postulated that precipitates can deflect or slow the LCGR as well [3].

In addition to transgranular growth, grain boundary precipitates can affect intergranular crack initiation and growth by suppressing grain boundary sliding [3,25]. Such was the case with Astroloy at temperatures above $0.55T_M$ where the growth rate was controlled by the grain boundary structure [32]. The wavy boundaries impeded intergranular cracking thus increasing the fatigue life.

2.2.3.4 Notches

Under LCF conditions, plasticity will typically be present at the point of crack initiation at geometrical discontinuities such as bolt holes or rim serrations in turbine disks. The notch root plasticity has been found to increase the SCGR compared to elastic conditions, Figure 2.14, which is one of the reasons for the high Stage II growth rates. Several authors have observed abnormally high growth rates during Stage II growth and have concluded that the effects were due to the notch stress field [25,48]. This can be seen from Figure 2.15 where the crack growth rate decreases after leaving the region affected by the notch.

Ewalds states that the higher growth rate within the notch plastic field is due to the large plastic strain range within this region [40]. This strain range rapidly drops off as the crack passes through the notch plastic zone which results in a rapid decrease of the SCGR. The variation of growth rate due to a notch, Figure 2.15, can be related to the change in closure effects due to the notch stress field [48,49], Figure 2.16. Similar to the growth from a free surface, Figure 2.11, however, the effect of the notch is to increase

K_{max} in the notch plastic field.

After passing through the stress field of the notch, further propagation of the crack is dependant upon the crack tip plasticity [34]. If the crack tip plasticity is too small to drive the crack, it will arrest, Figure 2.15. This can also be seen in Figure 2.4 where cracks have initiated in reverse loaded mild steel because of a large SCF at the notch ($K_t > 6$), but eventually arrested after leaving the notch plastic field. Smith and Miller [34] define the minimum nominal stress required for a crack to initiate and propagate until failure at small notches under fully reversed loading as,

$$\Delta\sigma = \frac{0.5\Delta K_{th}}{\sqrt{D}} \quad (2.7)$$

where ΔK_{th} is the threshold intensity factor and D is the notch depth. This equation was in good agreement with the data obtained by testing notched mild steel specimens, Figure 2.4.

2.2.3.5 Crack Tip Blunting

Several explanations exist describing the anomalous behaviour of short cracks (eg. closure, notch plasticity), however, an explanation not presented in the investigated literature is that of crack tip blunting. This is a suspected retardation effect of long cracks caused by the activation of more than one slip system, Figure 2.17. With slip occurring along both planes at 45° to the crack direction, the crack tip blunts which is said to result in a reduction in the stress concentration at the crack tip [50] Crack tip

blunting would not occur in Stage I growth since only one slip plane is activated resulting in a different crack extension mechanism. This would eliminate the retarding effect of crack tip blunting for a Stage I crack which would result in a higher FCGR at a given SIF.

2.2.3.6 Prior Particle Boundaries (PPB)

PPB's have been observed in Merl 76 and although their effect on the propagation of short cracks has not been reported in the examined literature, they cause an increase in LCGR's for high stress tests. A PPB in Merl 76 is created by hafnium rich oxide particles forming around some of the larger powder particles [56]. These oxides form during the powder production process, preferentially on the surface of larger particles because these particles remain hot for longer periods which promotes oxidation. These oxides inhibit consolidation which causes the particle to maintain its original size and shape. PPB's were observed by Gayda and Miner during their testing of Gatorized Merl 76 specimens [56]. High stress tests caused crack propagation to occur along the PPB's which increased the crack growth rate compared to lower stress tests at a given ΔK [56]. The PPB's have also been found to decrease stress-rupture life and ductility at temperatures above 700°C, however, they had no significant effect on the yield or ultimate strength [58].

2.2.4 Short Crack Growth Models

In fine grained materials such as Merl 76, the amount of Stage I growth is small.

Assuming the maximum Stage I growth may be confined to about 10 grain diameters, this yields a maximum Stage I crack length of about 70 μ m. For most P/M turbine disk materials, this crack length lies below the detection limits of ACPD or DCPD system, therefore, models which assume Stage II crack growth are examined in this study.

In some cases, the SIF has been successfully applied to describe the crack tip stress field during Stage II SCG [12,59], and due to this success and its simplicity, this approach appears most attractive. When a crack initiates in the notch root, it will initially grow semi-elliptically in shape before becoming a through crack. Most models consider the growth of a through thickness crack although Newman has derived a model applicable to semi-elliptical cracks [31], Appendix C. Beres and Koul have examined Newman's SIF model under elastic conditions for several crack sizes and compared it to solutions obtained using the J-integral and displacement approximation methods [68]. This analysis indicated that the Newman model accurately described the crack tip stress state for notched specimens.

3.0 EXPERIMENTAL MATERIAL AND METHODS

3.1 Material

The material used in this study was in the form of an unfinished Gatorized Merl 76 turbine disk forging, Figure 3.1. From a chemical composition point of view, Merl 76 is a modified version of IN100, with a reduced carbon content and an increased columbian and hafnium content to improve ductility and carbide control [51,52]. The vanadium content is completely eliminated to improve Fe corrosion resistance [52]. Its composition along with that of IN100 is given in Table 4. Merl 76 was developed in 1976 for P/M technology for higher load and temperature capabilities [52]. It was introduced into commercial service in 1983 [53] and was used in JT9D-7R4 high pressure turbine parts [54].

As a nickel base superalloy, its microstructure is comprises of FCC matrix, γ , containing a large number of alloying elements for solid solution hardening and corrosion resistance [2]. The dominant precipitate found, γ' , is an ordered FCC phase which is coherent with the matrix and typically composed of $\text{Ni}_3(\text{Al,Ti})$ [2]. The tensile strength of the alloy remains relatively constant until about 700°C which allows for high temperature applications [2]. The addition of Fe, Cr, and Co in Merl 76 decreases the solubility of Al and Ti into the matrix thereby promoting the formation of γ' precipitates [2] resulting in its high γ' volume fraction of 60%. The tensile properties at room temperature and 700°C are shown in Figure 3.2.

3.1.1 Gatorizing

The as received Merl 76 turbine disk was fabricated using a proprietary process developed by InterTechnologies called Gatorizing. There was little information available on actual Gatorizing parameters since it is proprietary, however, Coyne et al described it as an isothermal forge of consolidated powder metal [55]. Gayda and Miner describe the process to consist of the hot compacting of powder using extrusion of powder cans and then isothermally forging the slices of the extruded stock at a low strain rate [56].

Allen et al suggested that this process required less forging material compared to that of traditional techniques because it can produce near net shapes which also reduces the amount of finish machining required [57]. An example application cited was the 15 lb F100 9-11-13 compressor disk forging where an initial 126 lbs of material was required using traditional methods compared to only 60 lb of material using the Gatorizing process [57].

3.1.2 Grain Size Measurement

Before the specimens were cut from the disk, the average grain size at different locations in the disk cross-section was examined. A micrograph was taken at 13 different locations on the disk cross-section and the average grain size was determined using the mean linear intercept method as described by ASTM Standard E112-88. The matrix γ grains and the γ' precipitate grains (γ' grains with interphase boundaries) were both used to calculate the average grain size since the γ' with interphase boundaries (IPB's) existed outside the

matrix grains and were comparable in size. Figure 3.3 shows the grain size to be relatively uniform over the disk cross-section with an overall average grain size of $6.3\mu\text{m}$ with a standard deviation of $\pm 0.7\mu\text{m}$. A micrograph showing the general grain structure is given in Figure 3.4. This grain size is much finer than that reported for HIP'ed P/M Merl 76 compacts by other authors who determined the average grain size to be in the range of $15\text{-}20\mu\text{m}$ [52,56]. The fine grain size of the disk material can be attributed to the Gatorizing process.

3.1.3 γ' Precipitate Determination

Apart from the γ' with IPB's $1\text{-}6\mu\text{m}$ in diameter, the matrix γ' precipitates were cuboidal $\approx 0.2\mu\text{m}$ in size and spherical $\approx 0.05\mu\text{m}$ in size, Figure 3.5. Similar observations of γ' size and shape variations were made by other authors [52,56]. Evans and Eng found the fine γ' precipitates ($\approx 0.05\mu\text{m}$) to vary with section thickness [52]. The γ' precipitates were found to be more abundant in thinner sections due to the higher cooling rate from the solutioning temperature [52].

3.2 Double Edge Notch (DEN) Specimen

The DEN specimens were used for crack initiation and propagation testing and were selected for this study because they are geometrically similar to the bolt holes of a turbine disk, Figure 3.6, and have a similar SCF of 2.7, Appendix A. These specimens are smaller than those typically used in the AGARD test program in order to allow extraction of specimens from the thinner section of the turbine disk where the bolt holes would be

located. This DEN specimen has been successfully used for fatigue crack initiation and propagation studies [59]. The axis of all specimens were oriented radially within the disk and were cut from the thin part of the cross-section as shown in Figure 3.7. To ensure consistent surface roughness between specimens, the notch was ground to give a uniform 400 grit finish.

3.3 The ACPD Technique

To examine the ACPD technique, the flow of an alternating current (AC) through a conductor must first be understood. When AC is passed through a conductor, the current density will not be uniform across the conductor's cross-section as with direct current (DC). A high current density exists at the surface of the conductor and decays towards the middle, this phenomenon is known as the skin effect. This effect, as described by Gilbert for a cylindrical conductor, is due to a larger amount of induction occurring in the middle of the conductor compared to that of the outer edges [60]. Consider the cross-section of a cylindrical conductor with two concentric circles equal in area. When the magnetic field collapses as the current changes direction, a larger amount of induction will occur within the inner area because more magnetic lines of flux collapse through it. This results in the middle region having a higher impedance. This will shift current flow from the middle area to the outer area, thus causing the skin effect. As the frequency is increased, the inductance as well as the impedance in the inner area increases thus magnifying the skin effect.

The skin effect results in the significant portion of the current being confined to the surface making it highly sensitive to surface flaws as well as requiring smaller current supplies relative to DCPD systems. This makes the ACPD technique an ideal method for short crack detection. For this reason, this is the technique being used for crack detection for this project.

A further advantage of the ACPD system is its capability to amplify signals with the same frequency of the generated current at a specific phase. This type of lock-in amplifier cuts out signals that are not at the same frequency as the generated current which assists in eliminating stray voltages [61] as well as, eliminating the effect of thermal EMF's since these are DC in nature. Further control of the signal can be done by the use of a PSD. With this type of circuitry, the incoming PD signal can be adjusted such that only the real component of the signal is amplified thus eliminating the imaginary (induced) component.

3.3.1 Skin Depth

To quantify the skin effect, the term 'skin depth' was defined as the region within which the current flow is substantial compared to the rest of the conductor, Figure 3.8 [62]. It is important to realize that even within this skin, the current density is not constant but increases towards the conductor surface. The skin depth for a cylindrical conductor, δ , is given by,

$$\delta = \left(\frac{\rho}{\pi \mu_r \mu_0 f} \right)^{1/2} \quad (3.1)$$

where ρ is the resistivity of the material, μ_r is magnetic permeability of the conductor relative to free space, μ_0 is the magnetic permeability of free space and f is the AC frequency.

This relationship demonstrates that for a given material, the skin depth decreases as frequency increases resulting in better resolution, Figure 3.9. Smaller skin depths will increase the surface current density resulting in higher sensitivity, however, there are more inductive effects associated with higher frequencies as will be discussed in a subsequent section.

The above equation assumes a cylindrical conductor with remote current input. The DEN specimen is not cylindrical and the current is input on the specimen edge, however, since there were no relationships in the examined literature which described the skin depth for such a geometry, Equation (3.1) was used. For Merl 76, the skin depth as a function of AC frequency is,

$$\delta = 0.6 \sqrt{\frac{1}{f}} \quad (3.2)$$

where f is frequency in Hz. The derivation of this equation is given in Appendix B. Therefore, at an AC frequency of 30 kHz, the skin depth is about 3.5mm.

3.3.2 Application of VR for Crack Detection

Two PD signals were used to monitor crack growth and these were the PD across the

crack and the PD from an uncracked reference. This gave a non-dimensional voltage ratio given by,

$$VR = \frac{PD_{crack}}{PD_{ref}} \quad (3.3)$$

where PD_{crack} is the PD across the crack and PD_{ref} is the PD across the reference. This normalized parameter was used to compensate for changes in material properties, current and instrument drift. Variations in either will affect each PD reading equally thereby leaving the voltage ratio unchanged [62]. Since crack growth increases PD_{crack} , VR will also increase with increasing crack length.

The VR was obtained by the ratio of the two notch signals. The use of a notch signal as a reference was necessary to account for the signal decay that occurred due to cyclic softening only at the notch (discussed in further sections). As the crack grew in the one notch, the other notch signal would be used as a reference, however, the effect of the crack from the opposite notch would eventually affect this reference signal. This effect occurred at a crack depth of about 200 μ m at which point the reference was set to a constant value. Eliminating the drift at this point had little effect since the change in PD_{crack} was many times faster than the drift of PD_{ref} . The VR would not be applicable if a crack were to initiate in both notches, however, this occurred only once out of 21 specimens tested.

3.4 The ACPD Lead Configuration

The basic lead configuration for the ACPD method involved the attachment of a pair of current supply leads to the specimen with a set of PD signal leads from the specimen to the digital voltmeter.

Both the current and PD lead wires were insulated by placing a fibreglass sleeving over each wire before twisting, with both the wire and sleeving being capable of operating at elevated temperatures. Care was taken to ensure that all the leads were similar in position on the specimen and loop area formed with the specimen. This was done to maximize repeatability between specimens.

In an effort to minimize any induction [63] or capacitance [64] effects, the PD and current leads were made as short as possible by placing the PD and current terminal blocks as close to the specimen as possible while allowing clearance for the furnace, Figure 3.10. To further reduce the induction, the current leads approached the specimen from below and the PD leads approached the specimen from the side, Figure 3.11b. This maintained a distance between the two sets of leads thus reducing the interaction of any radiated magnetic field from the current leads with the PD leads.

3.4.1 Lead Material

The criteria used to select a material for the PD and current leads was high temperature capability along with low resistance. High temperature capabilities were required so that

the FCGR testing could be conducted at temperatures up to 600°C and low resistivity was required for the ACPD system (Section 3.5) to operate optimally.

Two different types of high temperature PD lead materials examined were Chromel and fine silver. Chromel was examined because it is a low resistance nickel base alloy used for high temperature applications and silver because of its low resistance and its high melting temperature.

In order to check the applicability of the two different materials at temperature, silver and Chromel PD leads were welded to the same specimen edge with a silver current wire. The results showed that over a 70 hour period at 600°C, the PD signal from the silver lead had a significant amount of drift and noise compared with the Chromel lead. Therefore, the silver PD leads were best suited to room temperature tests because of their lower resistance, however, at elevated temperatures, silver PD leads could not be used for Merl 76 specimens. The stability of the Chromel PD signal showed that the use of silver for the current leads does not adversely effect the PD signal. This stability was likely due to the circuitry in the ACPD unit (Section 3.5) which was designed to maintain a constant current flow regardless of changes in the current path.

3.4.2 Lead Attachment

Spot welding was the easiest method of lead attachment, however, this technique was not suitable for the PD leads which had to be butt welded. Furthermore, spot welding silver

wire was only possible by first welding a piece of Chromel ribbon to the silver wire which made for a complex procedure. For these reasons, laser welding was chosen as the method for PD and current lead attachment.

There was little difficulty in laser welding the Chromel wire to the specimen, however, laser welding silver was difficult due to its high reflectivity. This required the use of a high power density which resulted in residual cracks in the weld. The cracks were not a problem for the current leads because the welds were in a low stress region, however, the higher stress at the PD lead welds caused the cracks to propagate resulting in failure at the PD weld rather than at the notch root. This was overcome by shaping the laser pulse to provide pre-heating and post heating which reduced the amount of energy input into the weld.

3.4.3 PD Leads

In order to adequately butt weld the PD leads to the specimen, the specimen edge and the PD lead ends had to be flat and parallel. The PD leads were filed to give a flat smooth surface and the specimen edge was sanded down using a 240 grit wet sanding disk. This procedure yielded good mating surfaces for welding.

The PD leads were butt laser welded on opposite corners of the notch along the specimen edge, Figure 3.11. This location maximized the region within the notch root over which the potential drop could be detected [64]. In addition, the corners provided a consistent

locating point from specimen to specimen for welding thus minimizing any error associated with variations in lead placement [62]. This is especially critical for thick skin situations where slight variations in lead position will greatly affect the PD readings [62].

When using silver initially, 0.025"φ wire was used for the PD leads which allowed for accurate placement of the leads at the notch corners, however, this small diameter was barely stiff enough to support its own weight thus making the leads too delicate. This required the use of 0.040"φ wire in order to have leads robust enough to handle. For the Chromel leads, 0.030"φ wire was used since this provided the best balance between conductivity and stiffness. Larger diameters would have provided better conductivity, however, the stiffness proved to be too difficult to handle.

3.4.3.1 Lead Induction

Minimizing induction in the PD leads required twisting of the leads as close to the specimen surface as possible which minimized the loop area [63]. There was, however, a minimum amount of lead required for the welding procedure. The PD lead loop was made by starting the twist 5 mm from the ends of the lead wires with 3 mm of the wire running perpendicular from the surface while the other 2 mm ran into the twisted pairs. The two PD wires were then bent and positioned such that each of the two wires were connected to an opposite corner of the notch. Using these established dimensions resulted in constant loop areas between leads thus making the induction equal between specimens.

Induction was measured by adjusting the PSD such that the incoming PD signal being measured was in phase with the specimen current where the specimen current phase was taken from the current focusing lead. The resultant magnitude was that of the real signal. The total PD signal (the vector addition of the real and imaginary components) was obtained by switching the PSD to 'auto phase' which automatically adjusted the phase to maximize the PD signal. The imaginary signal was calculated by,

$$PD_{Im} = \sqrt{PD_{Tot}^2 - PD_{Re}^2} \quad (3.4)$$

where PD_{Im} and PD_{Re} are the imaginary and real components of the PD signal respectively and PD_{Tot} is the magnitude of the vector sum of PD_{Im} and PD_{Re} . From these values, the phase angle ϕ , can also be calculated from the measured quantities,

$$\phi = \cos^{-1} \left(\frac{PD_{Re}}{PD_{Im}} \right) \quad (3.5)$$

With a PD signal greater than one volt, there is a $\pm 0.5\text{mV}$ reading error on the CGM5 (Section 3.5) display which results in an error of about $\pm 1.5^\circ$ on a signal of 3.000 volts.

3.4.4 Current Leads

The current leads were welded as close to the threaded ends of the specimen as possible to ensure that the HAZ was well away from the stress concentration caused by the radius of the reduced section, Figure 3.11. One current lead travelled perpendicularly from the specimen edge while the other travelled along the specimen edge and through the notch until it reached the other wire at which point the two leads were tightly twisted together.

3.4.4.1 Lead Induction and Current Focusing

A factor affecting the magnitude of the radiated signal from the current leads is the loop area formed by the current leads with the specimen [65]. The placement of the current carrying wire along the surface minimized this loop area thus minimized the radiated signal causing induction [65]. In addition to reducing the radiated signal, the interaction of the magnetic field of the current wire placed along the surface resulted in an increase in the surface current density [65]. This 'current focusing' can be applied to maximize the sensitivity while reducing the radiated signal. For the current focusing to be optimized, the focusing wire along the surface was connected to the shield wire of the coaxial current output cable from the ACPD unit (Section 3.5).

Initially, the current focusing wire was placed straight along the edge of the specimen. This configuration, however, resulted in the current focusing lead being more than 1mm above the notch root surface. Initial attempts to bend the wire into the notch proved unsuccessful since the notch radius was very small, 1.19mm.

An alternative was to lay the wire through the notch perpendicular to the specimen axis, Figure 3.11. This configuration allowed the current wire to lay directly on the notch root surface thus maximizing the interaction between the specimen surface and the magnetic field of the current lead. This through notch current focusing required the wire to be bent sharply at the notch in order to keep the wire close to the specimen surface. To ensure repeatability, a fixture was made to place two 90° bends in the wire and when the current

leads were attached to the specimen, it was ensured that the wire running through the notch was pressed into the root.

3.5 The ACPD Unit

The equipment used for crack initiation and growth monitoring was a CGM5 multi-frequency ACPD unit produced by Matelect. This system consisted of an adjustable AC current generator capable of outputting a 0-2.1 amp current at frequencies of 0.3, 1, 3, 10, 30, and 100 Khz. The PD signal from the specimen was fed into a pre-amp which fed into the CGM5 main unit. The gain was variable in increments of 10 dB from 50 to 90 dB. An output channel was available which converted the ACPD signal from the specimen to a DC signal which was sent to the data acquisition computer where it was recorded.

There were several limitations with the CGM5 which required special adaptations in order to maximize the sensitivity of the system.

- 1) The unit had a maximum output of 3.5 VDC which meant that the analog-digital conversion board in the data acquisition system could not be fully utilized over its 0-10 VDC range. To maximize this range, thus increasing the sensitivity, the signal from the CGM5 was connected to a conditioner channel where it was amplified 1.989 times. With a gain of 70dB on the CGM5 and 1.989 on the strain channel, the resultant amplification was 6289 times on the PD signal from the

specimen.

- 2) The CGM5 had a low amplifier input resistance of about 30 ohms which meant signal loss would occur if the PD lead resistance was significantly high. For example, if the PD lead resistance was 3 ohms, roughly 10% of the signal would be lost. For this reason, the PD lead material must possess the lowest possible resistance.

- 3) The current generator had a maximum voltage drop it could generate, therefore, if the current leads were of a sufficiently high resistance, then the maximum attainable current would be much less than its rated maximum of 2.1 amps. For example, current leads with a resistance of about 4 ohms limited the current to about 0.5 amps. As with the PD leads, to maximize the sensitivity, the wire used for the current leads had to be low resistant material capable of function at 600°C.

3.6 Fatigue Test and ACPD Data Acquisition System

The ACPD system was integrated with the MTS closed-loop servo-hydraulic loading frame as shown in Figure 3.10. The pre-amp was positioned as close to the specimen as possible while allowing enough space for the furnace to close. This minimized the length of the PD leads which minimized the capacitance in the wires which occurs at high frequencies [64]. The ACPD parameters used for testing was $i=2.1$ amps, $f=30$ kHz and a gain of 6289 times unless otherwise specified.

The CGM5 used in this investigation had two PD input channels, however, there were problems with the stability of the PD signal when the secondary channel was used. Since only the main channel could be used, a double pole, double throw switch was required to switch the two PD signals from the notches into the one channel with the switching trigger signal sent from the data acquisition computer, Figure 3.12.

A similar switching configuration was required for the current input into the specimen. In order to maintain the effects of current focusing, the current lead must travel along the edge over which the PD is measured. Since there were two edges to the specimen, two separate current leads were required, however, the CGM5 had only one current output. This required the current to be switched to the side of the specimen at which the notch PD signal was being read. Care was taken to ensure that the shield wire of the CGM5 co-axial current cable was switched to the current focusing lead running along the specimen edge. A photo of the actual test configuration is shown in Figure 3.10.

The specimen was threaded into the load line which was electrically insulated from the load frame to prevent electrical leaks or stray electrical inputs. To ensure only axial loads, a universal joint and a double knife edge were attached into the load line above and below the specimen respectively.

All electrical equipment were fed from the same power bar inside the MTS data acquisition system to ensure a common ground.

3.6.1 Fatigue Loading and Data Acquisition Cycle

A triangular loading waveform with $R=0.1$ was used at a frequency of 0.5Hz for room temperature tests and 0.1Hz for high temperature test. The lower frequency is more representative of the actual loading seen in service, however, frequency effects are not significant for room temperature tests therefore a higher frequency was used to reduce the test time.

The fatigue loads were determined such that the cycles to initiation were in the range of 30000-60000 cycles [67]. This required $P_{\max}=15.5$ to 18 kN with $R=0.1$ which resulted in a maximum stress at the notch root of $\sigma_{\max}=1.7\sigma_y$ to $2.0\sigma_y$, respectively.

The data acquisition cycle consisted of fatigue loading for 'N' cycles, typically 10 cycles, after which the load would be brought to the maximum cyclic load followed by a quarter second delay to allow stabilization of the PD signal. Ten measurements of both the PD across one of the notches and the temperature were taken and averaged in order to reduce the effects of scatter. The time, cycle count, maximum and minimum cyclic loads before the PD measurements and the holding load at which the PD measurements were made were also taken. All these measured values for one side of the specimen were then saved as data onto the computer's hard drive. The current and PD switches were then activated to apply the current to the opposite edge and read the PD across the opposite notch. The cycling was restarted for another 'N' cycles after which the same set of readings would be taken. A set of readings required the load to be held at maximum for about 1.5

seconds. A graphic representation of this sequence can be seen in Figure 3.13.

This data acquisition sequence was adequate for crack calibration, however, the holding load could affect the crack growth rate. With data being acquired every ten cycles, the effects of this hold load could be significant compared to a continuous triangular wave, especially at elevated temperatures. To avoid the hold effects of the previous data acquisition cycle, a new program was written which allowed continuous cycling thus eliminating the effect of the data acquisition hold load on the FCGR. The data acquisition cycle consisted of loading for 'N' cycles, (once again, typically 10 cycles), after which two consecutive peak and valley measurements of PD and load were made. The PD values were averaged and saved on the computer's hard disk along with the temperature, cycle count and time. The switches were then activated and the same measurements were made for the other side after 'N' cycles with a schematic of this sequence given in Figure 3.14.

3.7 Crack Calibration

With the use of ACPD crack monitoring, the crack length can typically be analytically determined when either a thick or thin skin situation exists. Lugg defined thin skin to occur when both δ/Δ and δ/d_1 are less than 0.1 and thick skin theory to be applicable when δ/d_1 is greater than two with no limitations given for the δ/Δ ratio where δ is the skin depth, Δ is the probe spacing and d_1 is the crack depth, Figure 3.8 [66]. For short cracks, δ/d_1 would be in the thick skin regime since the crack depth would be much

smaller than the skin thickness for the Merl 76 specimens. The through thickness crack depth then becomes,

$$d_1 = \frac{\Delta}{2} \left(\frac{V_{ref}}{V_{crack}} \right)^{1/2} \left(\frac{V_{crack}}{V_{ref}} - 1 \right) \quad (3.6)$$

When a crack first initiates, it will be semi-elliptical and not a through thickness crack as required by Equation 3.6. This can be accommodated for and the true crack depth, d_2 , measured along the crack centreline can be estimated by,

$$d_2 = d_1 M_o \quad (3.7)$$

where M_o is a compensation multiplier. There are however, two limiting aspects to using this analytical relationship. First, the determination of M_o requires knowledge of the surface crack length $2a$. This may be difficult to determine if the crack is very small or is inaccessible. Secondly, M_o is only valid for $2a/\Delta > 1$. For a crack with a surface length of say, 500 μ m, the probes would have to be less than 1mm apart. This limitation on probe spacing makes it difficult to use this relationship for small cracks especially considering the probes for the DEN specimen must be welded outside of the notch which fixes their spacing to about 4.4 mm along the notch surface.

Therefore, in order to correlate the ACPD signal with crack size, calibration testing was required. This was done by monitoring VR during the fatigue loading of a specimen.

When VR increased by a predetermined amount, ΔVR , the fatigue loading was stopped and the specimen held at maximum load at an elevated temperature for about 30 minutes. This heat tinting procedure resulted in oxidation of the crack faces thus increasing the visibility of the cracked surface for more accurate crack measurements.

The specimen was then pulled apart and the crack depth, surface length and crack area were determined by the use of a travelling microscope with a $1\mu m$ resolution. The increase in VR was determined by subtracting the initial VR² from the final VR before the heat tint. This was carried out for several different crack sizes producing a correlation between crack size and ΔVR .

To verify the applicability of the room temperature calibration curve to FCGR tests at 480°C, some calibration specimens were tested at 480°C. The same methodology was applied to the high temperature calibration specimen as was used for the room temperature calibration specimens. Summarizing, a crack was propagated until a certain increase in VR was attained at which point, the load was held at P_{max} for about 30 minutes to heat tint the crack surface. The fatigue loading was restarted and the heat tinting procedure invoked after another specified increase in VR. In order to minimize the number of specimens required, it was necessary to obtain several calibration points (crack size versus ΔVR) per specimen. However, since the crack surfaces continue to

² The initial VR was set at the point when VR started to increase due to crack growth and not at the beginning of the test.

oxidize during growth, the initial heat tints loose some of their contrast thus decreasing the resolution the crack front. This in turn would reduce the accuracy of crack length measurements.

3.7.1 Crack Area Measuring Techniques

The most accurate method for crack area determination was to measure the crack depth at different points along the crack front and fit a polynomial curve to these points. By integrating the equation of the fitted curve, the area of the crack could be determined. The second method used was to photograph the fracture surface and trace and fill in the crack area. A rectangular frame was drawn around the traced crack area with a known area, A_{frame} . A Beuhler Image Analyzer was used for this procedure which had a variable 'window' which was positioned over the rectangular frame drawn around the crack. The area of this window, $\%A_{window}$ was determined as a percentage of the total screen area. Using the contrast adjustment, the crack area, $\%A_{crack}$ was determined as a percentage of the total screen area. Knowing the magnification of the photograph, the actual crack area can be calculated by,

$$A_{crack} = \left(\frac{\%A_{crack}}{\%A_{window}} \right) \frac{A_{frame}}{magnification^2} \quad (3.8)$$

The third method was applicable to only semi-elliptical cracks. Assuming a perfect semi-elliptical crack front, the area can be calculated by,

where $2a$ is the surface crack length and c is the crack depth.

$$A_{crack} = \frac{\pi}{4}(2a)(c) \quad (3.9)$$

In comparing these methods, only semi-elliptical cracks were used since the third method was only applicable to this crack shape.

3.8 Crack Growth Rate Testing

3.8.1 Short Crack SIF Calculations

The SIF was calculated using the model proposed by Swain and Newman [31] with the most recent version of this model given in Appendix C. This SIF model was developed for a semi-elliptical crack initiating in the centre or corner of a semi-circular notch subjected to a uniform remote tensile load, Figure 3.15 [31]. The major advantage of this model is its applicability to a semi-elliptical crack shape compared to other models which assume a through crack.

Although r/W is a variable in the SIF model, Newman described it as a constant of $1/16$. Special consideration was required in applying this model since our DEN specimens had $r/W=1/8$. A 3-D FEM analysis was conducted to compare the SIF along the crack front for different crack lengths of our specimen and these results were compared to those predicted by Newman's model [68]. There was good agreement with the Newman model showing that the model was applicable to our DEN specimens using a different r/W ratio.

At small crack lengths, the LEFM assumptions can be invalid due to the large crack tip

plastic zone size relative to the crack length. Newman accounted for the crack tip plasticity by adding $\frac{1}{4}$ of the crack tip plastic zone size to the crack length [31]. This extended the application of the LEFM SIF into part of the elastic-plastic region of short crack growth.

The cyclic crack tip plastic zone size, ω shown in Figure 3.16, can be calculated by [40],

$$\omega = 0.033 \left(\frac{(1-R)K_{\max}}{\sigma_o} \right)^2 \quad (3.10)$$

where R is the stress ratio, $K_{\max}$ is the crack tip SIF at maximum load and σ_o is the flow stress. Due to the large deformation at the notch root, the stress is above yield but below the ultimate strength. The flow stress was estimated to be the average of the 0.2% offset yield and ultimate strength. From Figure 3.2, this yields a notch root flow stress of, $\sigma_o^{RT} = (1082 + 1590)/2 = 1336$ MPa.

To examine the effect of crack tip plasticity, a crack 150 μ m in depth loaded at $P_{\max} = 18$ kN was considered. This is the smallest crack size detectable by the ACPD system and the highest load tested. Therefore, this presents a situation in which the ratio of the crack tip plastic zone size to crack depth would be a maximum. The alternating SIF at this crack depth would be 26.2 MPa $\sqrt{m}$ and from Equation (3.10), this would result in an 11.4 μ m cyclic plastic zone size. Adding one quarter of this to the original crack length, the new elastic-plastic alternating SIF becomes 26.4 MPa $\sqrt{m}$. This shows that compensating for the crack tip plastic zone is seen to have very little effect at this crack length and would

have even less of an effect at larger crack lengths. For this reason, the crack tip plastic zone correction was not applied to the data.

3.8.2 Elevated Temperatures

The design temperature provided by Pratt and Whitney Canada (PWC) for the Merl 76 turbine disk bolts holes was in the order of 800-900°F [67]. For conservativeness, the higher temperature was selected as the FCGR test temperature; converting to degrees celsius gives 480°C.

Due to the sensitivity of the ACPD system, thermocouples could not be placed on the specimen during testing. This required the calibration of the specimen temperature with reference thermocouples which in this case, were the furnace control thermocouples. With this calibration, the specimen temperature could be estimated using the control thermocouples.

3.9 Fatigue Crack Growth Rate Determination

Using the correlations determined during the calibration tests, the raw test data in the form of VR versus cycles was converted to crack length versus cycles. From this information, the crack growth rate was established for a given ΔK . Typically, the FCGR is determined using ASTM's seven point polynomial fit. During the first detectable stages of crack growth, the growth rate was small compared to the data acquisition interval of 20 cycles. This meant that over seven data points (ie. 140 cycles), the increase

in VR due to crack advance would be small compared to the signal scatter. Therefore, fitting a curve to seven data points in this low growth rate region would not provide an accurate indication of the true growth rate.

This problem was noted by Larsen et al who recommended that a polynomial be fit to data over a change in crack length, Δa_{reg} , rather than a specific number of points [69], Figure 3.17. In this technique, given the scatter in the measuring technique, the interval by which the regression region would be advanced, Δa_{int} , can be set to be a multiple of the standard deviation. With Δa_{int} set, the regression region can then be calculated by $\Delta a_{reg} = 6\Delta a_{int}$ [69].

The regression was preformed using the same algorithm as recommended by ASTM Standard E-647 [70], however, the second order polynomial was fit to 'n' data points rather than the recommended seven. As the growth rate increases, the number of data points that lie within Δa_{reg} will decrease. To ensure an adequate number of points are used for the regression, a check was built into the software which would extend Δa_{reg} to encompass a minimum of seven points at the higher growth rates.

4.0 RESULTS

4.1 The ACPD Crack Monitoring System

Before fatigue testing was commenced, several system parameters were examined. The variation of the PD signal with current and gain was shown to be linear thus functioning properly, Figure 4.1 and Figure 4.2 respectively. The variation of PD with frequency, f and $1/\sqrt{f}$ is shown in Figure 4.3. Examining the skin depth relationship given in Equation (3.1), the variation of PD with frequency should be proportion to $1/\sqrt{f}$, however, this was only true for frequencies under 3kHz. Higher frequencies caused the PD to increase at a much faster rate than predicted by the skin depth equation. This may be a result of the current focusing caused by the interaction of the magnetic field of the current carrying conductor with the specimen surface. At higher frequencies, the magnetic field about the current carrying conductor would increase thus having a greater effect on the current distribution.

The stability of the system was examined by shorting the inputs at the pre-amp and the respective left and right PD signals were measured for a duration of about 16 hours. Figure 4.4 shows the system to have reasonable stability with a scatter of about $\pm 0.8\text{mV}$ and a drift of about 0.5mV over the 16 hours.

When a specimen was first connected to the ACPD system, the initial PD values at room temperature with $i=2.1$ amps, $f=30$ kHz and Gain=6289 were in the order of 6 volts, however, this value varied from notch to notch, Table 5 with similar observations noted

by another author [71].

4.1.1 Temperature Effects

The effect of temperature on the PD signal was observed by monitoring the PD while the specimen furnace cooled at an initial rate of about $3^{\circ}\text{C}/\text{min}$. The fluctuation of the PD signal was nearly linear with temperature with a slope of about $0.9\text{ mV}/^{\circ}\text{C}$, Figure 4.5.

4.1.2 Stress Effects

An increase in stress caused an increase in the PD signal, Figure 4.6. The linear region of Figure 4.6 had a slope of about $3.4\text{ mV}/\text{kN}$ for an AC current of 2.1 A and 30 kHz with a gain of 6289 times. This linearity exists for all but the first half cycle where the PD signal varied linearly with load until yielding at the notch root occurred. This increase due to plasticity caused an offset to the PD signal, Figure 4.6.

With a crack of $c \approx 200\mu\text{m}$ present, three distinct slopes are visible and were fitted with a linear regression. For $P < 4\text{ kN}$, the signal increased at $3.7\text{ mV}/\text{kN}$; similar to the rate in Figure 4.7a where no crack existed. As the load increased, a second region became visible where the slope increased dramatically to $8.1\text{ mV}/\text{kN}$ for $4\text{ kN} < P < 9\text{ kN}$. The slope then decreased resulting in a third region with a slope of $4.1\text{ mV}/\text{kN}$ for $P > 10\text{ kN}$.

At this crack length of $c \approx 200\mu\text{m}$, the test was stopped and the crack was heat tinted by holding at maximum load at an elevated temperature. Cycling after heat tinting gave a

constant slope of about 4.2 mV/kN, Figure 4.7c, which was similar in magnitude to the final region before the heat tint. This slope is calculated for the entire loading range, however, upon close examination one can see that the first part of the slope ($P < 3\text{kN}$) is slightly lower than the rest. Unlike the other PD versus load results, there is a significantly larger amount of scatter with these results.

As the crack grew larger ($c \approx 500\mu\text{m}$), the low slope associated with the first region disappeared leaving only two distinct regions, Figure 4.7d. The first with a high slope of 7.4 mV/kN for $P < 5\text{kN}$ and the second with a slope of 4.6 mV/kN for $P > 6\text{kN}$.

Just before fracture, the crack growth for each cycle becomes apparent, Figure 4.7e. At this larger crack size (through crack over 1mm in depth), there are no obvious regions of PD variation with load with an average slope of 6.75mV/kN. At maximum load, there is a jump in the PD signal resulting in an offset for the subsequent cycles.

4.1.3 Effect of Cyclic Loading

As the specimen was cycled, the PD decreased in both notches at roughly the same rate with the rate of decay decreasing with increased number of cycles, Figure 4.8. The decay was only a function of cycles since a change in the frequency did not vary the rate of decay which showed that the drift was related to the cyclic behaviour of the material rather than instrument drift which would be expected to be time dependant. The rate of decay was not affected by an increase in the temperature to 480°C , however, increasing

the magnitude of the cyclic load resulted in a faster decay rate, Figure 4.9.

4.2 Crack Calibration

To calibrate VR to crack size, several parameters representing crack size were investigated; maximum crack depth, surface crack length along the notch root, and crack area. Crack area provided the best calibration as would be expected since the reduction in the cross-sectional area would be the controlling parameter in the change of the PD signal.

Figure 4.10 shows the variation of crack area with ΔVR at room temperature and is described by,

$$CrackArea = 3.45(\Delta VR)^{0.5} + 11.13(\Delta VR) - 23.07(\Delta VR)^2, \quad r = 0.997 \quad (4.1)$$

This equation was forced through the origin since at zero crack length, there should be no increase in VR.

Figure 4.10 shows that initially, crack area varies non linearly with VR, however, the curve tends to a straight line as the crack size increases. This variation in crack size with ACPD signal is in good agreement when compared to the results of another author, Figure 3.9 [63].

To provide a better physical interpretation of the calibration, the crack area can be converted to crack depth by substituting Equation (3.9) into Equation (4.1) with $a/c=1.3$,

Figure 4.11. This shows that changes in crack length at small crack sizes ($c < 150\mu\text{m}$) does not affect the VR significantly.

4.2.1 Elevated Temperatures

Two calibration specimens were tested at 480°C and were superimposed on the room temperature calibration curve, Figure 4.10. Both points lie slightly to the right of the average curve with the error being more pronounced for the longer crack. This indicates that the high temperature specimens were more sensitive to crack growth, however, because there are only two data points at 480°C , it was not possible to accurately determine the magnitude of the effect of temperature. Therefore, the room temperature calibration curve was applied to the high temperature specimens.

Only one point per specimen was produced because the final heat tint was the only heat tint that could be resolved. Heat tints at smaller crack lengths were obscured by the continued exposure to the environment during FCG until the final crack length.

4.2.2 Resolution of the ACPD Technique

As discussed above, changes in crack size at small crack lengths cause little change to the ACPD signal. This is due to the low resolution at small crack lengths as can be seen by the initially high slope of the curve shown in Figure 4.10 and Figure 4.11, however, as the crack grows, the resolution improves.

The resolution at different VR's can be calculated by differentiating Equation 4.1 which gives,

$$\frac{d(\text{Area})}{d(\Delta VR)} = 11.13 + 1.73(\Delta VR)^{-0.5} - 46.1(\Delta VR) \quad (4.2)$$

From Table 6 it can be seen that the resolution improves with crack depth where a change in the crack depth of a factor of four increases the resolution by a factor of six.

4.2.3 Crack Area Determination

An example of a semi-elliptical crack used for calibration after heat tinting and the curve fit method used to determine its area is shown in Figure 4.12. The error calculated for the Curve Fit Method was derived by applying a reading error of $\pm 3.2\mu\text{m}$ to the crack depth and surface length measurements. The areas determined by this method were assumed to be the most accurate possible with the facilities available.

With the assumption of a semi-elliptical crack shape, the area was calculated using Equation (3.9). Table 7 shows agreement with the curve fit method was moderate for the smaller cracks, however, as the crack size increased, agreement improved. For crack sizes greater than $\approx 0.18 \text{ mm}^2$, the measurements had less than a 5% error.

Agreement with the contrast method was poor for all crack sizes, Table 7. For this reason, it was not used for the last two specimens, DEN15 and DEN2.

4.2.4 Crack Shape Variation

The crack aspect ratio of half surface length, a , to crack depth, c , was shown to decrease with crack growth, Figure 4.13. This shows the crack growth rate into the specimen was faster than that along the notch root surface for $a/t < 0.69$.

In order to convert the estimated crack area from Equation (4.1), into crack depth and surface crack length data, a calibration equation was fit to the aspect ratio data shown in Figure 4.14 giving,

$$\frac{a}{c} = 1.334 - 0.215(Area)^{0.5}, \quad Area < 2.15 \text{ mm}^2 \quad (4.3)$$

where a is half of the surface length and c is the crack depth. The exponent of one half was used since this was the form used by Wanhill and Schra [41]. The data from one crack was dismissed (one of the two cracks found in DEN21) since the measurements were in poor agreement with the other data.

Measured values of crack aspect ratio were not obtained for $a/t > 0.69$ since one side of the crack would typically grow to the end of the notch root making it a partially through thickness crack. Since it is not possible to account for this effect, Equation (4.3) is used with the assumption that the crack will grow in to a through thickness crack when the surface crack length along the notch root reaches the specimen thickness which occurs at a crack area of 2.15 mm^2 . Once this condition is reached, the crack is a through crack and Newman's SIF model is not applicable.

4.3 Fatigue Crack Initiation Behaviour

The initiation life was determined by the number of cycles required to cause VR to increase by 0.2% which translated into a crack depth of about 300µm. This amount of signal change provided a reliable crack size between specimens.

Within the stress ranges tested, room temperature initiation ranged from about 19000 cycles for the high stress range to 50000 cycles for the low stress range, Figure 4.15. This figure shows that the initiation life decreases roughly linearly with the log of cycles which is in agreement with typical S-N data with the equation of this line given by,

$$\Delta S = 10^C N^b \quad (4.4)$$

where ΔS is the remote stress range, N is the number of cycles to initiation, the SCF $K_T=2.7$ with the material constants determined to be: $C=3.41$ and $b=-0.106$.

Increasing the temperature to 480°C increased the resistance to fatigue crack initiation where at $\Delta S=680$ MPa there was a slight improvement in the initiation life compared to the room temperature, however, at $\Delta S=655$ MPa the initiation life was significantly longer, Figure 4.15. The significant difference in initiation lives at 480°C was due to different crack initiation features as is discussed in the next section.

4.3.1 Crack Initiation Sites

The room temperature crack initiation sites were dominated by transgranular, Stage I

growth identified by the faceted fatigue surface. The crack length at which the transition from Stage I to Stage II growth occurred was dependant upon the magnitude of the loading where higher loads resulted in a shorter Stage I growth region, Figure 4.16. For $P_{\max}=16.3$ kN, Stage II growth commenced at a crack depth of 55-90 μm , Figure 4.16a, whereas for $P_{\max}=18.0$ kN, Stage I growth was almos. nonexistent, Figure 4.16b.

The faceted surface of Stage I growth can be seen from the crack profile of Figure 4.17. The initial crack path follows a tortuous route with many of the crack deflections occurring at the grain boundaries. At longer crack lengths, the fatigue surface smooths out indicative of the transition to Stage II growth.

At 480°C, two types of crack initiation sites were observed with both showing distinct features of a particle just below the surface of the notch root, Figure 4.18. For the specimen with the longer initiation life, the initiation site was highly textured and about 10 μm in diameter with only a slight amount of Stage I growth visible, Figure 4.18a. For the specimen with the shorter initiation life, the initiation site was a void on the fatigue surface with a diameter of about 15 μm . There was no visible Stage I growth and the crack propagation mode around the void appeared predominantly intergranular, Figure 4.18b. The surfaces of the grains within the void were pitted indicating that small particles may have existed along the boundary of the particle that was located in the present void. In both cases, further propagation beyond the initiation particle was predominantly transgranular with some secondary intergranular cracks.

4.4 Fatigue Crack Growth Characteristics

After processing the data through the crack size and shape calibration equations, a plot of crack length versus cycles was produced as shown in Figure 4.19. Crack length data for this specimen was not calculated for $N < 25600$ cycles because the calibration of crack length was not valid for that region. The calibration was determined for an increase in VR relative to the point at which crack growth was initially detected (when VR began to increase) which for this case was at $N = 25600$ cycles.

Figure 4.19 reveals that there was a larger amount of scatter associated with the measurements made at smaller crack lengths with this scatter decreasing as the crack grew. The magnitude of scatter was calculated by determining the standard deviation for five consecutive readings which decreased from $\pm 17\%$ for $c = 152\mu\text{m}$ to $\pm 1\%$ for $c = 200\mu\text{m}$, Table 8. This shows that the ACPD system can detect a crack $150\mu\text{m}$ in depth, however, due to the system noise, it is difficult to accurately size. It is not until $c \approx 200\mu\text{m}$ that the noise decreased to acceptable levels. For tests conducted at 480°C , there was an increase in noise to about $\pm 3\%$ at $c = 200\mu\text{m}$, however, this level of noise did not significantly affect crack sizing.

4.4.1 Fatigue Crack Growth Rate Data

To calculate the crack growth rate, the effect of the scatter on the crack length was minimized by determining an adequate value for the regression interval, Δa_{int} , Figure 3.17. Δa_{int} had to be large enough to encompass enough data to minimize the effect of the

scatter while being small enough not to average out any trends. Examining the first 1000 cycles in Figure 4.19 before any significant crack growth had occurred gave a standard deviation on the crack length of 0.028mm. Since this was a maximum standard deviation associated with the first detectable crack, it would be an excessively large interval for larger crack lengths, therefore, two values of Δa_{int} (0.030mm and 0.015mm) were applied to examine its effect.

Increasing the regression interval from 15 μ m to 30 μ m had no significant effect on the variation of the FCGR with ΔK other than to smooth out perturbations, Figure 4.20. This shows that $\Delta a_{int}=0.030$ mm minimized the effects of noise without altering the data, therefore, this was the value used for subsequent FCGR data analyses.

The FCGR results for room temperature tests for the crack depth are shown in Figures 4.21 and 4.22. For specimens 18 and 19, there was a decrease in the growth rate at $\Delta K \approx 30 \text{ MPa}\sqrt{\text{m}}$ ($c \approx 250\text{-}280\mu\text{m}$), however, no such growth rate decrease occurred for specimen 16. There was also a sharp decline in the growth rate at $\Delta K \approx 32\text{-}35 \text{ MPa}\sqrt{\text{m}}$ ($c \approx 500\mu\text{m}$) for specimens 16 and 19 with a slight decrease in the slope for specimen 18.

The FCGR results for testing at 480°C are given in Figures 4.23 and 4.24 which show that there was less scatter between specimens compared to the room temperature data. Both specimens 33 and 36 showed a slight increase in the FCGR at $\Delta K \approx 35 \text{ MPa}\sqrt{\text{m}}$ ($c \approx 500\mu\text{m}$) followed by a decrease, however, there was no effect on specimen 34.

The crack depth growth rate for $\Delta K > 30 \text{ MPa}\sqrt{\text{m}}$ follows a relatively straight line associated with the Paris region and was fit with a linear regression to give the Paris constants: $C = 2.25 \times 10^{-18} \text{ MPa}\sqrt{\text{m}}$ and $m = 8.72$ for room temperature and $C = 4.98 \times 10^{-15} \text{ MPa}\sqrt{\text{m}}$ and $m = 7.07$ for 480°C , Figure 4.25.

5.0 DISCUSSION

In general, the ACPD system proved to be a satisfactory technique for obtaining fatigue crack initiation and FCGR data. However, its sensitivity was not significantly better than the DCPD system implemented for previous FCGR testing of DEN specimens at the IAR [59] where the detectable crack depth was $c \approx 240 \mu\text{m}$ compared to $c \approx 200 \mu\text{m}$ for the ACPD system used in this study. The advantage of the ACPD technique over its DC counter part lies in the localized nature of the current flow, the skin effect, which requires less current input for the same sensitivity. Although the sensitivities were similar, a significantly lower current was used with the ACPD technique (2.1 AC amps compared to 30 DC amps) demonstrating the effectiveness of the AC skin effect. Improved sensitivity of the ACPD system could be obtained by increasing the current flow. This will be possible for future testing using the CGM6 which has a maximum current output rating of 5 amps.

In addition, an increase in specimen size while maintaining the same ACPD system parameters would not significantly change the sensitivity of the ACPD system. The sensitivity of the DCPD system on the other hand, would drop with an increase in specimen size. However, for small specimen cross-sections such as those used in this study, the current density of a DCPD system is high which allows it to obtain sensitivities similar to that of an ACPD system. For larger specimens, the DC current would have to be increased proportionally to the increase in specimen cross-sectional area to maintain roughly the same sensitivity. In contrast, the skin effect associated with the ACPD

technique would provide roughly the same sensitivity without the need to vary the AC parameters. For this reason, the ACPD system is more versatile which makes it a preferred choice for larger specimens.

The variation of the initial PD signals presented in Table 5 may have been due to slight variations from one specimen to another. Properties that affect the PD signal and that may have varied slightly from specimen to specimen is material resistivity, level of initial plasticity or the resistance of the PD welds. These three parameters would affect the initial PD signal, however, it is difficult to accurately quantify them.

5.1 Temperature Effects on the PD Signal

The variation of the PD signal can be explained by the variation in resistance with temperature. Although no published data were found on the resistivity of Merl 76 as a function of temperature, Figure 5.1 shows this relationship for Incoloy 909, another nickel base superalloy. Comparing Figure 4.5 and Figure 5.1 shows similar trends in the changes in resistivity and the PD signal over the given temperature range. This increase in resistivity with temperature also affects the phase angle. As the resistivity increases with temperature, so does the real component of the PD signal with the imaginary component remaining constant which results in a change in the phase angle. For Merl 76, an increase of about 455°C (room temperature to 480°C) reduced the phase angle from about 3° to 0°, Table 9. With a reading error of about $\pm 1.5^\circ$ (Section 3.4.3.1), it can be seen that induction is small at room temperature and negligible at 480°C. It is

important to realize, however, that any phase adjustment to the data acquisition system must be made after temperature has stabilized.

5.2 Stress Effects on the PD Signal

The variation of the PD signal with load is due to a change in the magnetic permeability of the material when it is stressed [72]. Dover et al [72] explains that each atom in a given substance has its own magnetic field, in a randomly distributed fashion they cancel out (unlike in a magnetic material where the orientations give an additive effect which produces a net magnetic field). When the material is stressed, the orientation of the atoms changes which also changes the magnetic field conduction paths thereby changing its magnetic permeability. This will cause a variation in the skin depth which will influence the PD along the stressed region. Although this phenomenon occurs with all metals, the severity of the stress effect varies with different materials, with magnetic materials being most sensitive.

During loading for the first cycle, the increasing magnitude of the PD signal with increasing load is related to plastic deformation, Figure 4.6, where a rate of increase in PD occurs at the load which causes plasticity at the notch root. In addition, this increase in the PD signal becomes permanent further indicating it to be related to the plastic deformation. As explained above, changing the magnetic orientation of the atoms will change the PD. Plastic deformation would cause a permanent change in this orientation thus causing a permanent change in the PD. However, plastic deformation would also

affect the resistivity. When the material deforms plastically, dislocations move and pile-up eventually resulting in slip band formation. This deformation will reduce the path through which the current can travel thus increasing the resistance. Since this deformation is permanent, the increase to the PD signal becomes permanent. This observation indicates that the ACPD system can provide an indication of plastic deformation and if calibrated, it could possibly quantify it.

5.2.1 Effect of a Crack on the PD Signal

In general, the PD will increase with load because of the stress at the notch root, the stress at the crack tip and the separation of the crack faces. Each of these effects may dominate at a particular point during loading, although, it is important to recognize that the overall PD signal is a superposition of all of these effects.

Three distinct regions were observed from the PD versus load curve at a crack depth of about 200 μ m, Figure 4.7b. Each of these regions vary as the crack length increases as will be discussed herein.

Region I identified by the low slope ($P < 4\text{kN}$) in Figure 4.7b, is related to the notch stress field effect because the slope of this region was similar to that observed this specimen in an uncracked state, Figure 4.7a. Since the crack contributed nothing to the variation of the PD signal, it can be concluded that the crack was not open at a load of less than 4 kN. This occurs because the crack size ($c \approx 0.2\text{mm}$) is small relative to the specimen size

(thickness=2.4mm). Upon loading, the material surrounding the crack could resist the crack opening due to the notch root plasticity, ie. closure effects. This implies that a certain amount of load would be required to overcome the compressive stresses for the crack to open. As the crack length increased, this low slope region decreased in size indicating that the crack started to open at lower loads until this region completely disappeared, Figure 4.7d.

Region II is characterized by a high slope, Figure 4.7b, caused by the crack opening as reported by Dai [73]. The separation of the crack surfaces reduced the conducting area dramatically thus causing a significant increase in the PD signal. For longer cracks, the Region II slope decreases indicating that the crack faces were separating slower relative to the smaller crack lengths. As well, there is a smaller change in PD indicating less crack area is separating, Figure 4.7d.

The decreasing amount of crack area separation can be explained by the fact that some of the crack surfaces remained open at minimum load. The presence of Region I indicates that the crack is fully closed until a certain load. The eventual disappearance of Region I with crack growth indicates that the crack is partially open at the minimum load. This is further proved by the increase of the PD at minimum load from 5080 mV from Figure 4.7b, to 6030 mV from Figure 4.7d. With part of the crack faces separated at minimum load, this will decrease the total area that will be separated during loading thus reducing the change in PD during crack opening.

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At the end of Region II, the crack is assumed to be fully opened. If this is the case then the load at this point is P_{op} which can be used to calculate ΔK_{eff} . This shows that the ACPD technique may be able to account for closure effects, however, further investigation and calibration are required to substantiate this hypothesis.

In Region III, the slope decreases after the crack has been fully opened, however, this slope is higher than that of Region I, Figure 4.7b. Since the crack is now open, the crack tip stress field becomes more significant and its effect is added to that of the notch root stress field thus causing a faster rate of change of PD with load compared to Region I. As the crack grows longer, the crack tip stress increases causing the PD to increase at a higher rate with load. This is shown by the higher slope of Region III in Figure 4.7d compared to Figure 4.7b. This region increases with crack growth, eventually dominating at longer crack lengths, Figure 4.7e.

5.2.2 Effect of Oxidization on the PD Signal

Heat tinting oxidized the crack faces which prevented current flow between the two crack faces when they were in contact. This removed the effect of crack opening on the PD signal, Figure 4.7c, where the variation in the PD signal with load is similar to that of Figure 4.7b less the effects of crack opening. This shows that during FCGR testing in corrosive environments, the ACPD system cannot be used to determine the crack opening load.

Although it is not clear why there is an increased scatter after heat tinting, a possible explanation is that the elevated temperatures may have caused some stress relieving during the heat tinting procedure. This means that the measurements during the first few cycles after heat tinting may go through a transient period before the material and the PD attains an equilibrium state.

5.3 Cyclic Loading Effects on the PD Signal

The first cycle caused yielding at the notch resulting in an increase in the PD signal, Figure 4.6, after which, the PD signal decreased with consecutive cycles, Figure 4.9. Since the PD signal increased with plastic deformation, the decrease of the PD signal with cycling could be a result of the relieving of plasticity, ie. cyclic softening. Cyclic softening was observed during strain controlled LCF tests of the same as received Merl 76 disk material [74].

As described by Hertzberg, cyclic softening occurs by the rearrangement of dislocations such that they offer less resistance to deformation [75]. The rate of cyclic softening is more significant for a harder material with the rate of softening decreasing with cycling, eventually levelling off to reach an equilibrium [75]. The pattern of the PD decay is comparable to the pattern that cyclic softening would experience, Figure 4.9. The higher decay rate observed with higher loads was due to the larger volume of material experiencing a stress range sufficient to cause softening.

To further confirm that the decay in the PD signal was due to cyclic softening, the loading frequency was decreased during testing. Since cycle softening is a function of cycles, a change in the loading frequency should not change in the decay pattern of the PD signal with respect to the cycle count. Such is the case as shown in Figure 4.8 where a decrease in the cyclic load frequency showed no significant change to the decay pattern of the PD signal. This demonstrates the rate of decay of the PD signal to be cycle dependant and not time dependant thus eliminating an alternative explanation, that of instrument drift.

It is important to realize that the PD signal decayed to a level well below its original value after the first cycle. This indicates that the majority of cyclic softening was the rearranging of dislocation that initially existed in the material and not those induced by the loading.

5.4 Crack Calibration

The initial steep slope of the calibration curve, Figure 4.10, may indicate lower resolution at the smaller crack lengths as will be discussed in the next section. However, the scatter associated with the cracks in this region is an additional factor affecting the shape of the curve. This occurs as a result of errors that are insignificant at large crack lengths that can become significant at smaller crack sizes. These errors add to the difficulty of accurately sizing small cracks and explain why there can be a larger degree of scatter associated with calibrating cracks in this region.

One such error is associated with crack size measuring. Ten measurements of the same crack depth gave one standard deviation of $\pm 3.2\mu\text{m}$ on an average measurement of $195.2\mu\text{m}$. This error is insignificant for large crack sizes, however, it becomes more significant for smaller cracks, Table 7.

Another source of error are cracks which grow on planes other than the one visible during crack size measurements. Cracks will initiate at different points along the notch root, parallel and perpendicular to the loading axis. Each crack will contribute to an increase in the PD signal, but typically, only a few of these cracks coalesce to form a dominant crack while the others arrest. When the specimen is fractured to measure the crack area, only the area of the cracks which exist in the plane of the dominant crack can be measured since these are the only ones visible. When calibrating the smaller crack areas, these arrested cracks may exist and be comparable in size to the visible crack but cannot be accounted for since it is not possible to measure their areas or to quantify their effect on the PD signal. An example of this situation was specimen DEN21 where surface cracks such as that shown in Figure 5.3, were detected along the notch root below the dominant crack. These arrested cracks were of sufficient size to affect the PD signal in the very early stages of crack growth.

The combination of these error sources along with the small crack size relative to the skin depth (low resolution, discussed in the next section) may explain why there is only a slight trend visible with the smaller crack sizes of the calibration curve, Figure 4.10.

The lack of agreement between the high temperature calibration points and the room temperature calibration curve may be due to the initial PD at the notch at 480°C being lower than expected. With only two high temperature calibration points, it was difficult to accurately quantify the error over the entire calibration range. It is clear however, that there is a discrepancy between the room temperature and 480°C calibration points. As the temperature increased, so did the resistance of the material which caused the PD to be more sensitive to crack growth. This increase in sensitivity however, should not affect VR since the reference voltage (same as the initial notch voltage) would increase proportionally. This did not occur as seen from Table 5 where the higher temperature specimens had roughly the same initial PD voltages as the room temperature ones. This may have been due to the difference in resistance of the PD leads where the high temperature Chromel leads had a higher resistance than the silver leads. This may have caused the decrease in the initial PD reading before cracking thereby increasing the sensitivity of VR to crack growth.

5.4.1 Resolution of the ACPD Technique

From Figure 4.10, it can be seen that the resolution, (change in the ACPD signal with respect to crack length), increases with crack length. The lower resolution at the smaller crack lengths may be due to the small crack size relative to the skin depth.

At an AC frequency of 30 kHz, the skin depth was estimated to be about 3.5mm, Appendix B. For the smaller cracks in Figure 4.10, the crack depth was in the order of

several hundred microns, less than one tenth the size of the skin depth. For cracks small relative to the skin depth, their effect will be less significant on the current flow than would a larger crack thus causing a less significant change in VR for a given change in crack area. This is indicated by the change in the slope of the calibration curve, Figure 4.10, where larger crack lengths lead to a higher resolution.

The large skin depth would also explain the lack of linearity between the PD signal and the crack area. As the crack length increases, it approaches a thin skin situation in which the PD signal and crack length should be linear. This can be seen in the calibration curve, Figure 4.10, where linearity increases at longer crack lengths.

5.4.2 Crack Area Measurement Techniques

The assumption of a semi-elliptical crack shape and the use of Equation (3.9) to calculate the crack area was applicable for crack areas greater than $\approx 0.18 \text{ mm}^2$, Table 7. The larger errors associated with the smaller cracks was likely due to perturbations along the crack front. At small crack sizes, these perturbations would significantly affect the shape of the crack. However, as the crack grows, these perturbations become less significant relative to the crack size resulting in a smoother semi-elliptical crack shape.

The Contrast Method showed very poor agreement with the curve fit method. The lack of accuracy of this method was attributed to the reduction in fracture surface resolution with a photograph combined with the subjectivity involved in the operation of the image

analyzer.

5.4.3 Crack Shape Changes

The crack aspect ratio was shown to decrease with crack length, Figure 4.13, which shows that the surface crack length was growing at a slower rate relative to the crack depth. These results were similar in magnitude with those shown in Figure 2.7 with $a/c \approx 1$, however, the variation in crack shape with growth is opposite. This shows that the change in crack shape with crack size is not universal and may be dependant on material, specimen geometry and loading condition. The slower growth rate of the surface crack along the notch root for the Merl 76 DEN specimens may be due to the large amount of plasticity in that region causing closure. The data in Figure 2.7 were produced using Aluminum SEN specimens with $\sigma_{\max} \approx 0.72\sigma_y$, compared with the DEN specimens used in this study where $\sigma_{\max}$ ranged from $1.7\sigma_y$ to $2.0\sigma_y$. With this loading difference, the Merl 76 DEN specimens would have had plastic deformation at the notch root unlike the Aluminum SEN specimens. This plasticity could increase the closure effects for the surface crack growing along the notch root thus reducing the surface growth rate, $2a$, compared to a crack growing along a notch root subjected to elastic stresses.

The notch root plasticity also creates a discrepancy in the variation of the crack shape when compared with what is expected by Newman's SIF model. Calculating the variation of $K_{\phi=90^\circ}/K_{\phi=0^\circ}$ with crack depth using the Newman model, Appendix C, shows that $K_{\phi=90^\circ}$ increases faster than $K_{\phi=0^\circ}$, Figure 5.4. Therefore, the Newman model predicts that

the surface crack length should grow faster than the crack depth, which is the opposite to what was observed in Merl 76 specimens. Because the suspected cause of this discrepancy was the plasticity at the notch root, it was assumed that the SIF at the crack depth was more accurate therefore ΔK results were produce for crack depth only.

An anomalous occurrence with a crack in the DEN21 specimen was an increase in its crack aspect ratio with growth compared to the decrease observed with the other specimens. This behaviour was attributed to the coalescence of the dominant crack with a smaller one Figure 5.3b. The effect of the smaller crack coalescing with the larger one was to cause an increase in the surface crack length but not crack depth. This would result in an abnormally high aspect ratio as well as causing the SIF of the surface crack to be much higher than that of the crack depth. With a higher SIF at the surface, the growth rate would be faster thus causing the aspect ratio to increase further as seen in, Figure 4.13. For this reason, these data points were not included in the determination of the crack shape calibration equation.

5.5 Fatigue Crack Initiation Life

The room temperature fatigue crack initiation data follows a straight line showing a pattern typical of fatigue crack initiation data. Increasing the test temperature to 480°C resulted in an increase in fatigue life, Figure 4.15. The improved fatigue life was likely due to an increased ductility with temperature which resulted in a reduction of the notch root stresses [75]. In addition, the fine matrix γ' precipitates, whose strength increases

with temperature, would impede the formation of long planar slip bands. Their interference with dislocation motion would cause the deformation to move to another plane rather than trying to cut the precipitate. This would result in homogeneous slip which would allow the material to absorb more plastic deformation.

The room temperature fatigue lives determined in this study were compared to proprietary crack initiation data for smooth fatigue specimens tested at 900°F with $R=0$ obtained from PWC. At a given initiation life, the notch root stress range used in this study was almost twice that used by PWC. Although part of this discrepancy may be related to differences in the notch geometries, such a significant difference indicates that the initiation life of the Merl 76 tested in this study is superior to that used to produce the PWC initiation data.

5.5.1 Initiation Sites

The transition from Stage I to Stage II growth is said to correspond to the crack length at which the crack tip plastic zone is equivalent to the grain size. This theory would indicate that larger loads would reduce the amount of Stage I growth which agrees with the observations made with the DEN specimen results, Figure 4.16. To calculate the expected transition crack length from Stage I to Stage II growth, the crack tip plastic zone size was equated to the average grain size for $P_{max}=16.3$ and 18.0 kN with an example calculation in Appendix D. For these load levels, the transition crack depth was estimated to be 70 μ m and 55 μ m respectively, Table 10.

The estimated transition length for $P_{\max}=16.3$ kN is in good agreement with the initiation site, Figure 4.16a, however, for $P_{\max}=18$ kN, the agreement was poor. The expected transition crack depth was $55\mu\text{m}$ but Figure 4.16b shows there to be very little Stage I growth present. The lack of agreement may be due to the increased amount of deformation at the notch root at the higher load where $\sigma_{\max}=2.0\sigma_y$. The theory states that Stage II growth will commence when the crack tip deformation becomes more homogenous. It is possible that at the higher loads, there is sufficient deformation at the notch root to cause Stage II growth to commence earlier.

There were two types of initiation sites observed for specimens tested at 480°C , Figure 4.18. The site which had a significantly longer initiation life contained a highly textured fracture feature about $10\mu\text{m}$ in diameter, larger than the average grain size, displaying transgranular initiation, Figure 4.18a. This site was likely an undeformed powder particle as shown in Figure 3.4 which were typically larger than the average grain size. The feature is highly textured due to the dendritic structure of the prior powder particle. Its lower resistance to initiation relative to the surrounding material was likely due to the inhomogeneity of the particle which would create a stress concentration. The second initiation site with a lower fatigue initiation life was an intergranular initiation around a particle about $15\mu\text{m}$ in diameter. The void was larger than the average grain size indicating that this may have been the location of a larger than average powder particle. In addition, there was a significant amount of pitting visible on the grain surfaces within the void indicating that the particle was surrounded by a significant amount of fine grain

boundary precipitates. These observations indicate that this void was likely a site of a PPB. The PPB's have been shown to reduce the FCGR resistance of Merl 76 under high stress conditions at elevated temperatures [56] therefore, it is logical that a PPB would reduce the initiation life when located within the high stress region at the notch root at 480°C.

There was a small amount of Stage I growth observed around the dendritic initiation site, however, none was observed around the PPB site. This was likely due to the difference in cycles to initiation where there was an insufficient number of loading cycles at the PPB initiation site to cause slip bands to form. Rather, the initiation at the PPB occurred around the particle boundary, Figure 4.18a.

In both cases, propagation beyond the initiation site was predominantly transgranular with some secondary cracks which appeared to be intergranular. This shows an increase in intergranular growth due to temperature, however, at 480°C, the environmental effects are not sufficient to cause this to be the dominant mode of propagation. This agrees with observations of other authors on Merl 76 at 650°C who reported that crack growth was predominantly transgranular [56].

5.6 Fatigue Crack Growth Behaviour

The variation of crack length with cycles shows there to be more scatter associated with the smaller crack lengths, Figure 4.19. This was not totally unexpected upon examination

of the calibration curve which has a very steep slope at smaller crack lengths, Figure 4.10, making it very sensitive to noise from the VR signal. At longer crack lengths, the slope of the calibration curve decreases thereby decreasing the sensitivity to noise resulting in a smoother curve of crack length versus cycles. The precision is further increased because of the higher growth rates at longer crack lengths which minimizes the effects of any instrument drift.

5.6.1 Fatigue Crack Growth Rate Data

Only the crack depth growth rates were used since the surface crack growth rate was deemed less accurate due to notch root plasticity. Along the notch root surface, there is more significant deformation which is believed to have resulted in the lack of agreement between the observed crack shape development compared to $K_{\phi=90^\circ}/K_{\phi=0^\circ}$ and Figure 2.7 as discussed in Section 5.4.3. This lack of agreement was associated with the larger amount of plasticity within the notch root where the surface crack length was measured, therefore, it is assumed that ΔK was most applicable to the crack depth measurements.

The FCGR at room temperature dropped at a crack length of $c_1 \approx 250\text{-}280\mu\text{m}$ and at $c_2 \approx 500\mu\text{m}$, Figure 4.22. Drops in the FCGR for a crack growing from a notch have been related to the effects of the notch root plastic field, therefore, a finite element analysis was conducted to determine the extent of the plane stress notch root plastic region [78]. The monotonic, plane stress notch root plastic zone length for this specimen with $P_{\text{max}} = 17\text{-}18\text{kN}$ is $\rho \approx 1.1\text{-}1.25\text{mm}$ with the cyclic plastic zone size being about one quarter that of

the monotonic zone which yields $w \approx 300 \mu\text{m}$, Figure 5.2. From this estimation of the notch root plastic zone size, it can be seen that c_1 corresponds to the length of the cyclic notch root plastic zone. This may lead to the conclusion that the drop in FCGR was caused by the crack tip growing beyond the cyclic plastic zone. However, there was no such effect observed on the FCGR for specimen 16 at this crack length. This lack of agreement between the specimens may indicate that rather than a physical phenomenon creating the drop in the FCGR, it may be due to some other cause such as signal noise. For crack depths less than c_1 , there is much scatter associated with the data. Although Δa_{int} was selected to minimize the effect of scatter, it cannot eliminate it, therefore, the FCGR data reduction in this region may be less accurate. Until further specimens are tested to establish this as an actual physical occurrence, no concrete conclusions can be drawn.

Examining the drop in the FCGR at c_2 shows that this crack length corresponds to about half the length of the notch root monotonic plastic zone. Although the effect on specimens 16 and 19 were more severe than that of specimen 18, the data showed some perturbation in the FCGR at this crack depth. This is in agreement with the FCGR results obtained by Dai using nickel base superalloy SEN specimens where the growth rate dropped at a crack length of half the notch root plastic zone [73]. This indicates that the notch root plastic region elevates the growth rate until a crack depth of about half the monotonic plastic zone size.

Perturbations in the FCGR at 480°C occurs at $c=500\mu\text{m}$ as well, however, this occurs for specimens 33 and 36 but not for specimen 34. Although there seems to be a perturbation in the FCGR due to the notch root plastic zone for both room temperature and 480°C FCGR tests, the magnitude of the perturbation is different from one specimen to another. This indicates that there may be factors other than notch root plasticity affecting perturbations in the FCGR results.

The calibration curve is applied to crack growth using the assumption that there is one crack that has initiated at the center of the notch root. This crack is further assumed to remain semi-elliptical in shape until the surface crack length is equal to the specimen thickness. In reality however, usually more than one crack initiates which will cause variation in the crack growth rate when coalescence occurs. In addition, one side of the crack will typically growth through the thickness before the other thus having a partial corner crack geometry compared to the semi-elliptical geometry that was assumed. Neither of these actual occurrences can be accounted for but they can cause variation in the FCGR data from one specimen to another.

A linear regression was fit to the room temperature and 480°C FCGR data which resulted in the Paris constants of $C_{RT}=2.25 \times 10^{-18} \text{ MPa}\sqrt{\text{m}}$, $m_{RT}=8.72$ and $C_{480^\circ\text{C}}=4.98 \times 10^{-15} \text{ MPa}\sqrt{\text{m}}$, $m_{480^\circ\text{C}}=7.07$, Figure 4.25. The growth rate of the specimens tested at 480°C was higher due to the environment acting as a catalyst to FCG by weakening the material as discussed in Section 2.1.1. There was, however, no FCGR data available in the examined

literature for comparison either at room temperature or 480°C. Therefore, it was necessary to obtain some proprietary long crack data from PWC. Unfortunately, the PWC data cannot be displayed in this document but only discussed in relative terms with respect to the data produced in this study. The lowest temperature data available from PWC was at $\approx 280^{\circ}\text{C}$ (500°F). The FCGR data produced from this study was expected to have a higher growth rate than the long crack data, however, this was not the case. The room temperature growth rate was about an order of magnitude slower than the long crack data with the largest discrepancy at lower ΔK 's. A similar situation existed for the 480°C data where the growth rate was below that of the long crack data at lower ΔK values. However, the growth rate eventually exceeded the long crack data at higher values of ΔK . An additional discrepancy was the slope, where the long crack data had a significantly lower slope compared to both the room temperature and 480°C data. This discrepancy between the two data sources is considered in greater detail in the following section.

5.6.1.1 Notch Root Plasticity Effects

The growth rate of a crack through the notch root plastic field was expected to be faster than that of long crack data because overall work required to cause fracture is reduced. The above comparison, on the other hand, shows that the observations were directly in contrast with the expected trend. These atypical results most likely indicate that LEFM based SIF calculations using Newman's model do not accurately describe the stress state at the crack tip. Since LEFM does not consider large scale yielding of the material, the

estimated notch root stress could exceed the fracture strength under these conditions. This would cause an overestimation of the stress at the crack tip and result in a higher value of ΔK than actually existed. This in effect would cause the FCGR vs. ΔK curve to move right relative to its actual position. Correcting in the opposite direction would reduce the ΔK at a given crack growth rate thus bringing the data in better agreement with the long crack data. The higher slope of the DEN specimen data perhaps indicates that the discrepancy in ΔK calculations was more significant at smaller crack depths. As the crack tip moves further away from the notch root, the notch root plasticity effect will diminish thereby decreasing the error associated with ΔK calculation. Therefore, for short crack depths relative to the notch root plastic length, the overestimation of ΔK would be greater than for larger crack depths which would explain the slope being larger than expected. Furthermore, this means that the agreement with the long crack data should improve at large crack depths which was indeed the case.

Although the LEFM SIF has been successfully used to describe the growth of cracks from a notch root, Figure 2.13 [12], its applicability may be limited to cases where only a limited amount of notch root plasticity is present. A small amount of plasticity would not cause a significant error between the actual crack tip stress field and the one predicted by LEFM. Even when plasticity occurs at the notch root and LEFM over predicts the actual stress, this overestimation may be offset by the fact that the crack would grow faster through the plastic region. In order for a crack to propagate, a certain amount of plastic work is required for dislocations to move and form slip bands. Since the notch root

plasticity provides some of this work, less work is required to cause slip band formation at the crack tip thereby reducing the net amount of crack tip plasticity required to propagate the crack. The effects of the over prediction of the crack tip stress by LEFM along with the greater ease with which the crack could grow through a region of plasticity could offset each other resulting in an apparently accurate description of crack growth rate as a function of ΔK . With a larger amount of plasticity, as was the case for the DEN specimens used in this study, the overestimation of ΔK using LEFM may overcompensate for the higher growth rate within the plastic region resulting in poor agreement with the long crack data.

Considering that the overestimation of the stress occurs due to the assumption of linear elastic conditions, the plasticity must be considered in order to improve agreement with long crack data. The parameter in Newman's model that can be adjusted for plasticity is the SCF. The actual SCF when plasticity exists is lower than the elastic SCF and can be estimated using the Neuber equation [75],

$$K_t^2 = K_e K_\epsilon \quad (5.1)$$

where K_e is the elastic SCF, $K_\sigma = \sigma/S$ is the actual SCF, and $K_\epsilon = \epsilon/e$ is the actual strain concentration factor. Rearranging Equation 5.1 to put it in terms of the true stress and true strain gives,

$$\sigma \cdot \epsilon = \frac{(S \cdot K_t)^2}{E} \quad (5.2)$$

Equation 5.2 and the true stress-strain curve of the material can be plotted on a set of stress versus strain axes with the intersection of the two denoting the actual stress and strain experienced in the region of the stress concentration. For Merl 76, an equation was fit to the monotonic, room temperature true stress-strain curve obtained from Figure 3.2. Ideally, under fatigue loading conditions, the cyclic stress-strain curve would be preferable, however, such data were not available. For a given remote stress, Equation 5.2 was plotted with the true stress strain data and the point of intersection was graphically determined, Figure 5.5. The ratio of the actual notch root stress over the applied remote stress gives K_o with values given in Table 12.

Thus, K_o then depicts the true stress concentration factor operative at the notch root. However, when the crack moves out of the notch root plastic zone, it is suspected to obey LEFM conditions where K_i then becomes applicable. El Haddad et al stated that the crack length at which the Neuber SCF and the elastic SCF are roughly equivalent ($K_i = K_o$) "corresponds to the size estimated for the general yielded zone in the absence of a crack" [49]. Therefore, the SCF starts at a value of K_o at the notch root and reaches a value of K_i a distance of ρ_{mono} away from the notch root. Applying a linear interpolation between K_o and K_i over the monotonic notch root plastic zone length gives a SCF corrected for plasticity, K_{pl} ,

$$\begin{aligned} K_{pl} &= K_o + \frac{c}{\rho_{mono}} (K_i - K_o), & c \leq \rho_{mono} \\ K_{pl} &= K_i, & c > \rho_{mono} \end{aligned} \quad (5.3)$$

where ρ_{mono} is the monotonic notch root plastic zone length in the direction of crack growth and c is the crack depth. K_{pl} is used to replace K_I in the Newman model (see modification factor g_4 in Appendix C) which reduces the SIF significantly at small crack depths with its effect diminishing at longer crack depths. A schematic of the effect of K_{pl} on the stress distribution along the notch is given in Figure 5.6. Applying K_{pl} to the data from Figure 4.25 reduced the slope, Figure 5.7, which resulted in good agreement with the long crack data where the 480°C regression line had a slightly lower slope and the RT regression had a slightly higher slope. As would be expected, the FCGR data at 480°C lies slightly above the 280°C PWC FCGR data and the room temperature data lie slightly below it. This agreement indicates that the use of K_{pl} was successful in correlating FCG within a notch root plastic region with long crack data for this material and specimen.

Using the IAR-DEN specimen restricted the maximum crack depth that a semi-elliptical crack could exist to about 1mm. This meant that the discontinuity of Equation 5.3 at $c=\rho_{\text{mono}}$ was never reached. Extrapolating for $c>\rho_{\text{mono}}$ using Equation 5.3 reveals that the FCGR curve would steepen since K_{pl} would no longer increase. This general trend is reasonable when compared to typical SCGR/LCGR plots as show in Figure 2.12. However, this indicates that the slope of dc/dN vs. ΔK beyond this crack depth would be higher than the long crack data. This may be due to the inapplicability of the Newman SIF to a crack this large since growth beyond ρ_{mono} would cause a through thickness crack. This indicates that further tests are necessary to examine the continuous FCGR

within and beyond the notch root plastic zone to better establish the effectiveness and applicability of Equation 5.3.

In order to verify the applicability of the plasticity correction, its effectiveness on Al 7075-T6 SCGR data was examined. A collaborative test program between the National Aeronautical and Space Administration (NASA) and the Chinese Aerospace Establishment (CAE) examined the propagation of semi-elliptical cracks from a notch using Newman's SIF model [79]. The SCGR data was in good agreement with their predicted results except for specimens tested at $R=0.5$ where plasticity existed at the notch, Figure 5.8. In order to apply K_{pl} to their data, several parameters were determined as described in Appendix E and rather than correcting all the data points, the error was applied to the predicted curve. This reduced the number of data points requiring adjustment thus simplifying the analysis. The result was a movement of the predicted curve to the right with the movement being most pronounced at lower ΔK 's with zero movement at $\Delta K=7.6$ $\text{MPa}\sqrt{\text{m}}$. This correction shifted the predicted curve which led to better agreement with the SCGR data, Figure 5.8. The plasticity correction is applied until the end of the notch root plastic zone which corresponds to $\Delta K=7.6$ $\text{MPa}\sqrt{\text{m}}$. Beyond this, the curve remains in its original position lying slightly above the SCGR data points. This may indicate that K_{pl} should be extended further than ρ_{mono} . Even after the crack depth, c , has moved beyond ρ_{mono} , the majority of the crack front is still within the monotonic plastic field indicating its effect on the crack growth rate may continue for $c > \rho_{\text{mono}}$. Regardless, it is clear from this analysis that the application of K_{pl} over ρ_{mono} improved the agreement

between the predicted and the test results for the SEN Al 7075-T6 specimens experiencing notch root plasticity [79].

6.0 CONCLUSIONS

The ACPD technique along with current focusing technology was effectively used for the acquisition of fatigue crack initiation and FCGR data on DEN specimens. With a current of 2.1 amps at 30 kHz, it had a detectable crack depth of $c \approx 200 \mu\text{m}$ for the IAR-DEN specimen which was slightly better than that of the DCPD system with a current of 30 amps. However, the PD and current lead requirements were much more complex than that of the DC system. For this reason, the 30 amp DCPD system would be more simple to use for the DEN-IAR specimen than the 2.1 amp, 30 kHz ACPD system. For larger specimens, however, the ACPD system would be preferable since it would be able to maintain about the same sensitivity without adjusting the current magnitude or frequency, but in contrast, the DC system would require much greater current than 30 amps.

6.1 The ACPD System

- Increasing the temperature caused the PD signal to increase due to an increase in the resistance of the material with a change of about $0.9 \text{ mV}/^\circ\text{C}$ for Merl 76 with ACPD parameters: $f=30 \text{ kHz}$, $i=2.1 \text{ A}$ and $\text{Gain}=6289$. The increase in temperature also resulted in a reduction in the phase angle to a value too low to be measured (below 1.5°). This reduction in angle occurs due to the increase in resistivity which only affects the real portion of the PD signal.
- The PD signal increases with load and this occurs as a result of the superposition of the effects of the notch root stress, the crack tip stress and the separation of the

crack surfaces due to crack opening. Examining this information on a PD versus load plot can provide an indication of the load at which the crack is fully opened. Knowing this load, crack closure effects can be estimated and the effective SIF calculated. This is not possible, however, if the crack faces are insulated from each other as was the case with elevated temperature tests where the crack faces oxidized. To determine the accuracy of this approach, the crack opening loads determined using the PD versus load plot should be verified or calibrated with known techniques.

- Room temperature cyclic loading of Merl 76 resulted in a decay of the PD signal likely caused by cyclic softening or the relieving of residual stresses.
- The resolution of the ACPD system improved with crack growth; eg. 156mV/mm for a crack area of 0.45 mm² to 279mV/mm for A=2.5 mm². This increase in resolution was likely due to the increase in crack size with respect to the skin depth.
- With a current of 2.1 amps at a frequency of 30 kHz, the ACPD system was capable of detecting a crack $\approx 150\mu\text{m}$ in depth, however, there was much scatter at this crack size (1sd= $\pm 11\%$) making it difficult to accurately size. Improved resolution with crack growth allowed for more accurate crack sizing at a crack depth of about 200 μm (1sd= $\pm 1\%$).

6.2 Crack Shape

- The variation of crack aspect ratio for $a/t < 0.69$ showed that the crack depth grew faster than the surface crack length thus causing the aspect ratio, a/c , to decrease,

$$\frac{a}{c} = 1.48 - 0.460 \left(\frac{a}{t} \right)^{0.5} \quad a/t < 0.54 \quad (6.1)$$

where a is half the surface length, c is the crack depth and t is half the specimen thickness. This variation of crack shape was opposite to that observed by other authors with the suspected reason for the discrepancy being closure effects resulting from the large amount of plasticity at the notch root.

- The assumption of a semi-elliptical crack was effectively applied to calculate the crack area for cracks greater than 0.18mm^2 with less than a 5% error with respect to the actual crack area. Applying the assumption of a semi-elliptical crack to smaller crack areas resulted in larger errors in the order of 10-20%.

6.3 Fatigue Crack Initiation

- The room temperature initiation life ($c \approx 300\mu\text{m}$) of Merl 76 DEN specimens ($K_t = 2.7$) with respect to the remote stress range followed a typical S-N failure pattern given by,

$$\Delta S = 10^c N^b \quad (6.2)$$

where ΔS is the remote stress range, N is the number of cycles to initiation with the material constants determined to be: $C=3.31$ and $b=-0.108$.

- The crack initiation region at room temperature revealed predominantly Stage I growth identified by the faceted fatigue surface. Increasing the fatigue load reduced the amount of Stage I growth until there it was almost totally eliminated at $\sigma_{max} \approx 2\sigma_y$, $R=0.1$. Growth beyond the Stage I region was transgranular Stage II.
- Increasing the temperature to 480°C improved the resistance to fatigue crack initiation resulting in an increase in the initiation life, however, the initiation life was largely dependant upon the existence of PPB's near the notch root. If a PPB was located near the notch root, it would serve as an intergranular initiation site which resulted in a slightly better initiation life compared to room temperature. If no PPB's were located near the notch root, the initiation life increased where the initiation site (for DEN36) was suspected to be an undeformed powder particle located close to the notch root which resulted in transgranular initiation. Growth beyond the initiation site with the predominant mode of crack growth being transgranular with some secondary cracks which appeared to be intergranular.

6.4 Fatigue Crack Growth Rate

- The FCGR is accelerated by the notch root plastic field as shown by the decrease

in the FCGR when the crack depth reached about half the monotonic notch root plastic zone length for specimens at both room temperature and 480°C. In addition to plasticity, variations of the FCGR with crack depth are suspected to be affected by random occurrences such as crack coalescence or growth through to one side of the specimen.

- The crack growth rate was lower than expected compared to long crack data which was likely due to the inability of the LEFM parameter, ΔK , to accurately describe the crack tip stress field. With the significant yielding at the notch root ($\rho_{\text{monotonic}} \approx 1\text{mm}$), LEFM would over-predict the actual crack tip stress resulting in an artificially high ΔK for a given crack growth rate. The application of the Neuber rule and modification to Newman's model was effective in correcting the overestimation of ΔK which brought the crack growth data into good agreement with long crack data. Its effectiveness was verified when it was applied to crack growth within the plastic field of a notch for ΔL 7075-T6 SEN specimens from another study [79]. The new Paris constants determined for the Merl 76 DEN specimens after the correction for plasticity were: Room Temperature: $C_p = 1.42 \times 10^{-12} \text{ MPa}\sqrt{\text{m}}$, $m_p = 5.45$; 480°C: $C_p = 2.36 \times 10^{-9} \text{ MPa}\sqrt{\text{m}}$, $m_p = 3.64$.

7.0 RECOMMENDATIONS FOR FUTURE WORK

- Closure effects are said to contribute substantially to the anomalous behaviour of short cracks and this is also used to explain different growth rates at different load ratios. Accounting for closure can better describe what is actually occurring at the crack tip allowing for more accurate analyses. Examining the variation of the PD with load may indicate the crack opening load, P_{op} , thus allowing ΔK_{eff} to be determined. Verification of this, however, requires further investigation to determine how accurately the PD signal indicates P_{op} . If applicable, then closure effects and crack size can be established during a fatigue test by using the ACPD technique.
- The application of a SCF corrected for notch root plasticity, K_{pl} , using the Neuber equation proved effective in correlating FCGR data within a notch plastic field to long crack data. However, due to the specimen size, all the crack growth was within the notch plastic field. Further tests, which can continuously examine the FCGR within and beyond the notch root plastic zone, are required to better establish the effectiveness and applicability of Equation 5.3. In addition, verification that $c=\rho_{mono}$ is the point at which LEFM is applicable would also be beneficial in establishing more accurate limits on K_{pl} .
- By improving the sensitivity of the ACPD system, more data could be obtained at smaller crack lengths thus providing a better understanding of the earlier stages

of crack growth. Upgrading the CGM5 unit used in this study to a CGM6 unit would allow the maximum applied current to increase from 2 amps to 5 amps. This should more than double the sensitivity which would push the detectable crack depth under 100 μ m. This improvement would allow for a more thorough understanding of crack shape development at smaller crack sizes as well as the variation of crack growth within the notch plastic field.

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Table 1 Points of transition from Stage I to Stage II crack growth

Material	Transition to Stage II growth			Reference
	Grain Diameters	Length (μm)	Other	
Al 2024	---	100	---	[13]
Al 7010	5-20	100	---	[7]
Al 2024-T3	3	---	---	[31]
0.4%C Steel	---	200-400	---	[37]
304 Stainless	10	---	---	[36]
Low Carbon Steel	10	---	---	[36]
Waspaloy	9	600	---	[22]
Astroloy	---	---	$\Delta K=30$ $\text{Mpa}\sqrt{\text{m}}$	[12]

Table 2 Crack initiation and growth modes for different temperatures and loading frequencies [76]

Temperature, K (°F)	Frequen- cy, Hz	Alloy	Mode of Initiation	Initial Mode of Propagation ^a
RT ^b	0.04	Waspaloy	I	I
RT	0.05	L/C ^c Astroloy (I)	I	I
		(II)	I	I
		(III)	I	I
		(IV)	I	I
700 (800)	0.05	Waspaloy	I	II
		L/C Astroloy (I)	I	II
		(II)	I	I
		(III)	I	II
		(IV)	I	II
811 (1000)	0.05	L/C Astroloy (I)	I	II
		(II)	I	I
		(III)	I	I + I
		(IV)	I	II
922 (1200)	0.05	Waspaloy	I	II
		L/C Astroloy (I)	I	II
		(II)	I + IX	I + II
		(III)	IX	IX
		(IV)	I	II + I
	0.005	L/C Astroloy (I)	IX	II + 50%IX
	0.5	L/C Astroloy (I)	I	II
	0.5	L/C Astroloy (II)	I	I
1033 (1400)	0.05	Waspaloy	I	II + 20%IX
		L/C Astroloy (I)	IX	II (after less than 1 grain diameter)
		L/C Astroloy (II)	IX	IX
		L/C Astroloy (III)	IX	IX
		L/C Astroloy (IV)	IX	II

^a KEY: I = Stage I; II = Stage II; IX = intergranular.

^b Room temperature.

^c Low carbon.

Table 3 Deformation behaviour of γ' precipitates found in nickel base superalloys

Alloy	Ref	Precipitates and sizes	Fatigue hardening/softening		Precipitate dislocation interactions
			Temp. °C	Effect	
718	13	γ' spherical 20 nm	20	Softening	Shearing
718	11	γ'' discs 30 x 5 nm thick γ'' discs 20-30 x 5-9 nm thick	550	Softening	Shearing
			20	Slight hardening then softening	Shearing
			538	Softening	Shearing
Waspaloy	13	γ' spherical bimodal 30 and 150 nm	20	Hardening then softening	Looping to shearing
Waspaloy	11	γ' spherical bimodal 30-40 and 200-250 nm	650	More hardening than at 20°C	More looping/bypassing
			20	Hardening then softening	Looping to shearing
	11	γ' spherical 40-50 nm	538	Hardening, some softening	More looping/bypassing
			20	Less hardening, more rapid softening	Looping to shearing
901	11	γ' spherical 25-30 nm	20	Hardening then softening	Looping to shearing
			538	Hardening then softening	Looping to shearing

Table 4 Composition of Merl 76 and IN100

Composition, wt. %	
Merl 76 HIP'ed and Gatorized TM	IN 100
C 0.02	C 0.07
Cr 12.4	Cr 12.5
Co 18.5	Co 18.5
Mo 3.2	Mo 3.2
W ---	W ---
Hf 0.4	Hf ---
Nb 1.4	Nb ---
Al 5.0	Al 5.0
Ti 4.3	Ti 4.3
V ---	V 0.75
B 0.02	B 0.02
Zr 0.06	Zr 0.04
Ni rem	Ni rem

Table 5 Initial ACPD voltages before crack growth at room temperature (Ag PD leads) and 480°C (Chromel PD leads), $i=2.1$ amps, $f=30$ Khz, $G=6289$.

	Specimen	Initial specimen PD signal (V)	
		Left Notch	Right Notch
Room Temperature	DEN 21	6.05	6.02
	DEN 22	5.98	5.85
	DEN 2	6.46	6.05
	DEN 13	6.33	5.85
	DEN 9	6.08	6.08
	DEN 16	5.96	5.96
	DEN 18	6.22	6.16
	Avg. $\pm 1s.d.$	6.15 ± 0.17	6.00 ± 0.11
480°C	DEN 33	6.19	5.93
	DEN 34	6.02	6.12
	DEN 36	5.89	5.97
	Avg. $\pm 1s.d.$	6.03 ± 0.15	6.01 ± 0.10

Table 6 Variation of Crack resolution with crack size; $a/c=1.2$

Crack Area (mm ²)	Crack Depth (mm)	ΔVR	$d(\Delta VR)/d(\text{Area})$ (mm ⁻²)
0.052	0.251	0.0005	0.011
0.45	0.586	0.01	0.036
1.45	1.05	0.07	0.069

Table 7 Comparison of methods used to determine the area of a semi-elliptical crack relative to the curve fit method.

Specimen	Crack Area (mm ²)					
	Curve Fit ¹	% Error on Crack Area $\pm 1\sigma$	Semi-Elliptical Crack ²	% Difference	Contrast ³	% Difference
DEN10	0.073	2.10	0.089	21.92	0.128	75.34
DEN15	0.145	1.55	0.159	9.66	----	----
DEN4	0.150	1.5	0.168	12.00	0.223	48.67
DEN13	0.187	1.48	0.189	1.07	0.259	38.50
DEN22	0.233	1.34	0.233	0.00	0.164	-29.61
DEN15	0.406	0.93	0.423	4.19	----	----
DEN2	0.474	0.91	0.500	5.49	----	----

- 1 Crack depth was measured at incremental positions along the crack front and a polynomial was fit to these points. The area was calculated by integrating this equation.
- 2 The crack shape was assumed to be semi-elliptical giving $A = \pi/4(2a)(c)$.
- 3 The crack shape was traced from a photograph and darkened in. This was placed under an image analyzer and the area calculated. This area was divided by the square of the magnification to give the actual crack area.

Table 8 Standard deviation associated with different crack lengths

Average Crack Depth (μm)	± 1 Standard Deviation	
	(μm)	(%)
152	0.017	11
173	0.004	2.3
200	0.002	1.0

Table 9 Examination of induction for the DEN specimen at room temperature and 480°C; ACPD: i=2.1 A, G=6289 and f=30kHz.

Temperature (°C)	Potential Drop Signal (V)			Phase Angle, (degrees)
	Real	Total	Imaginary, $\sqrt{(PD_{Tot}^2 - PD_{Rc}^2)}$	
25	2.951	2.954	0.1718	3.3
480	3.176	3.176	0.0	0.0

Table 10 Crack depth at which the crack tip plastic zone size is equivalent to the average grain size from Appendix E; a/c=1.4, $K_t^{bulk}=2.7$, plane strain

P_{max} (kN) (Load ratio 0.1)	Maximum Remote Stress (MPa)	Crack Depth at Transition (mm)
18	805	0.055
17	729	0.070
15.5	693	0.078

Table 11 Magnetic permeability of various nickel base superalloys. Measurements taken at room temperature and $H=200$ oersteds unless otherwise specified.

Material	Relative* Magnetic Permeability
In 600	1.0-1.1
In 718	1.0011-1.0013
Incoloy 901	1.008
Hastelloy X	<1.002
In 625	1.0006
In 601	1.003 1.004@-104°F 1.016@-320°F
Nimonic 105	1.000715 200<H<2000 oersteds
Rene 41	<1.002
Waspaloy	1.011 1.035@25 oersteds

* relative to air (ie. $\mu_{\text{material}}/\mu_{\text{air}}$) where $\mu_{\text{air}}=1.257 \times 10^{-6}$ henry/m

Table 12 Values of the actual SCF, K_{σ} , and the plane stress monotonic notch root plastic zone length, ρ at different loads

Load, P_{max} (kN)	Actual SCF, $K_{\sigma}=\sigma/S$	Notch root Plastic zone length, ρ (mm)
16.3	1.50	1.08
17.0	1.45	1.16
18.0	1.38	1.32

Figure 1.1 Schematic of damage tolerant methodology showing the application of the safe inspection interval

Figure 2.1 Formation of a slip band in a grain showing dislocation pile-up at the grain boundary [21]

Figure 2.2 Hatched region denotes the crack tip plastic zone for a Stage I crack. The crack is slowed by the interaction of the plastic zone with a grain boundary [20]

Figure 2.3 Micrographs of transgranular crack initiation sites; a) initiation along a persistent slip band and b) initiation with no visible slip band [22]
(Note: original micrographs of poor quality)

Figure 2.4 The effects of stress concentration factor on initiating non-propagating cracks in fully reversed loading of mild steel [34]

Figure 2.5 Schematic of the Kitagawa-Takahashi curve showing the two regions of short crack growth where σ_{cy} is the cyclic yield stress in a reversed stress test, d_1 , d_2 , d_3 represent sizes of microstructural elements and l represents the crack length at which LEFM principles are valid [18]

Figure 2.6 Crack path and growth rate behaviour showing the effect of grain boundary interactions [10]

Figure 2.7 Variation of crack shape with crack length for a single edge notch specimen [41]

Figure 2.8 Schematic showing how crack closure causes a reduction in the applied SIF range, ΔK [40]

Figure 2.9 Differing plasticity in the wakes of short and long cracks with the same ΔK . σ_{op} -stress required to open crack tip; $\Delta K_{eff} = \Delta K_{max} - \Delta K_{op}$ [40]

Figure 2.10 Schematic showing how plastically induced closure occurs [40]

Figure 2.11 Schematic of $K_{\max}$, K_{closure} , and $K_{\min}$ for a crack emanating from a free surface or notch subjected to constant applied load range showing the variations in closure effects with crack length, $R < 0.5$ [44]

Figure 2.12 Schematic of small and large fatigue crack propagation behaviour at low load ratio ($R < 0.5$). The growth rates correspond to the variation of closure shown in Figure 2.11 [44]

Figure 2.13 Room temperature fatigue crack growth rates for different microstructures of Astroloy; FG-fine grains, N-necklaced, CG-coarse grains. a) Long crack growth, b) Short crack growth [12]

Figure 2.14 Schematic showing the increased CGR for a crack propagating within a notch root plastic zone [40]

Figure 2.15 Test data showing the effect of a notch on the short crack growth rate behaviour [48]

Figure 2.16 Variation of stress intensity factor with crack length for short cracks growing from a notch at $R=0$, $\rho=0.16\text{mm}$ [48]

Figure 3.1 As received Gatorized Merl 76 turbine disk, diameter=52cm.

Figure 3.2 Engineering stress-strain properties of as received Merl 76 turbine disk material at room temperature and 700°C.

Location	1	2	3	4	5	6	7	8
Average Grain Size (μm)	6.4	6.0	5.1	6.0	6.7	6.1	6.2	7.0

Location	9	10	11	12	13
Average Grain Size (μm)	6.2	5.6	7.4	7.6	6.1

Overall Average of 13 measurements: $D_{avg}=6.3\pm0.7 \mu\text{m}$

Figure 3.3 Average grain size at different locations along the P/M Merl 76 turbine disk cross-section determined using ASTM Standard E112-88 lineal intercept method

Figure 3.4 General structure of as received Merl 76 with the γ' phase removed by etching. The dendritic structure indicates a powder particle that was not deformed during the Gatorizing process.

a) large grain boundary precipitates 1-6 μm ;

b) cuboidal matrix precipitates $\approx 0.2\mu\text{m}$ and fine matrix precipitates $\approx 0.05\mu\text{m}$

Figure 3.5 Micrographs showing γ' size distribution in Merl 76

Figure 3.6 Double edge notch specimen; All dimensions in millimetres unless otherwise specified

Figure 3.7 Scaled drawing showing the location in disk cross-section from which the DEN specimens were cut

Figure 3.8 AC field distribution and probe locations used for theoretical determination of crack depth [62]

Figure 3.9 Increase in sensitivity to crack growth with increase in AC frequency [63]

Figure 3.10 Configuration for the initiation and short crack growth test showing the fatigue test frame with ACPD system and specimen.

a) Front view: current leads placed along the specimen to minimize loop area

b) Side view: leads are bent for installation into test frame maintaining distance between the PD and current leads

Figure 3.11 Current and PD lead configuration for the double edge notch specimen.

Figure 3.12 Schematic of the ACPD system

Figure 3.13 Schematic of fatigue loading and data acquisition cycle for crack calibration

Figure 3.14 Schematic of improved fatigue loading and data acquisition cycle for FCGR testing

Figure 3.15 Parameters used for the Newman SIF applied to a semi-elliptical crack at a notch [31]

Figure 3.16 Schematic of the cyclic, plane strain crack tip plastic zone size, ω

Figure 3.17 Schematic representation of the modified incremental polynomial method for reducing crack growth data; $\Delta a_{\text{reg}} = 6\Delta a_{\text{int}}$ [69]

Figure 4.1 Variation of the PD signal with current

Figure 4.2 Variation of the PD signal with gain

Figure 4.3 Variation of the PD signal with frequency, f and $1/\sqrt{f}$ (inset)

Figure 4.4 Stability test for ACPD data acquisition system with PD inputs shorted at the pre-amp

Figure 4.5 Variation of PD with temperature for Merl 76. ACPD: $f=30\text{kHz}$, $i=2.1\text{A}$, $G=6289$

Figure 4.6 Effects of load on the PD signal; ACPD: $f=30\text{kHz}$, $i=2.1\text{A}$, $G=6289$

a) Effect of notch stress field, no crack

b) $\approx 200 \mu\text{m}$ crack depth

Figure 4.7 Variation of PD signal with load at different crack sizes (a)-(e), ACPD:
 $f=30\text{kHz}$, $i=2.1\text{A}$, $G=6289$

c) After heat tinting 200 μm crack in (b) where the insulation of the crack faces due to oxidization altered the PD vs. load curve.

d) $\approx 500\mu\text{m}$ crack depth

Figure 4.7 (Continued)

e) $\approx 1\text{mm}$ crack depth.

Figure 4.7 (Continued)

Figure 4.8 Decay of PD signal with cycles at different frequencies

Figure 4.9 Effect of cyclic load level on the decay rate of PD signal

Figure 4.10 Calibration of ΔVR to crack area

Figure 4.11 Variation of crack depth, c , with ΔVR

a) Photograph of semi-elliptical crack, (500 μm graduations)

b) Curve fit to measured points along crack front used to calculate crack area

Figure 4.12 Example of a semi-elliptical crack used for calibration of the ACPD system

Figure 4.13 Variation of crack shape with surface length, $t=1.19\text{mm}$. Note: ◆ denotes the same crack in DEN21 at different lengths. The increase in a/c was due to coalescence with a smaller crack along the notch root.

Figure 4.14 Variation of crack shape with crack area. Note: ◆ denotes the same crack in DEN21 at different lengths. The increase in a/c was due to coalescence with a smaller crack along the notch root.

Figure 4.15 Room and elevated temperature fatigue crack initiation data for DEN Merl 76 specimens; $K_t=2.1$, $R=0.1$, triangular wave; the error bars consist of data from three specimens; Note: Initiation established when $\Delta VR=0.002$, $c \approx 300 \mu\text{m}$

crack
growth
↓

a) $P_{\max}=16.3$ kN with $R=0.1$, (1000x)

crack
growth
↓

b) $P_{\max}=18.0$ kN with $R=0.1$, (1000x)

Figure 4.16 Room temperature initiation regions showing decrease in faceted, Stage I growth with load

Figure 4.17 Etched crack profile showing faceted Stage I growth followed by smoother Stage II growth. Crack deflection caused by grain boundaries is apparent for about the first 20 μm .

a) Highly textured feature at point of initiation; likely caused by the dendritic structure of an undeformed powder particle

b) Void located at point of initiation; likely the former site of a PPB

Figure 4.18 Initiation region for Merl 76 DEN specimens at 480°C a) $N_{init}=118600$ cycles, $P_{max}=16.3$ kN; b) $N_{init}=38000$ cycles, $P_{max}=17.0$ kN.

Figure 4.19 Variation of crack depth and surface crack length with cycles

Figure 4.20 Effect of regression interval on data reduction

Figure 4.21 FCGR data for Merl 76 double edge notch specimens, crack depth; Triangular wave, $f=0.5\text{Hz}$, $R=0.1$, Room Temperature

Figure 4.22 Room Temperature FCGR as a function of crack depth; Triangular wave, $f=0.5\text{Hz}$, $R=0.1$

Figure 4.23 FCGR data for Merl 76 double edge notch specimens, crack depth; Triangular wave, $f=0.1\text{Hz}$, $R=0.1$, $T=480^\circ\text{C}$

Figure 4.24 FCGR as a function of crack depth; Triangular wave, $f=0.1\text{Hz}$, $R=0.1$, $T=480^\circ\text{C}$

Figure 4.25 FCGR versus ΔK for Merl 76 DEN specimens at room temperature and 480°C before the effect of notch root plasticity is considered.

Figure 5.1 Variation of resistivity with temperature for Incoloy 909 [77]

Figure 5.2 Variation of monotonic notch root plastic zone size with load for the IAR-DEN specimen

a) Initiation site at surface flaw, $2a \approx 115\mu\text{m}$. Shows jagged crack path associated with transgranular, Stage I growth.

b) Secondary crack coalesced with dominant crack, $2a \approx 400\mu\text{m}$

Figure 5.3 Micrographs of cracks which arrested along the notch root on one side of the dominant crack

Figure 5.4 Variation of SIF ratio with crack length and crack aspect ratio; $K_{\phi=0}$ -SIF at crack depth, $K_{\phi=90}$ -at surface crack

Figure 5.5 Plot of true monotonic Merl 76 stress-strain data and the Neuber equation. The intersection point provides an estimate of the actual notch root stress and strain at a given load; $K_t=2.7$, $S=807$ MPa, $E=207$ GPa

• - Linear Interpolation, $K_{pl} \cdot S$

Figure 5.6 Schematic of stress distribution along crack path using the K_{pl} correction

Figure 5.7 FCGR versus ΔK for Merl 76 DEN specimens at room temperature and 480°C showing the improvement to the data when the SCF is corrected for plasticity

a) Original data [79]

b) K_{pl} applied to prediction curve

Figure 5.8 Experimental ($S_{max} = 195$ to 220 MPa) and predicted small surface crack growth rates against ΔK in 7075-T6 aluminum alloy SENT specimens at $R = 0.5$ [79].

Appendix A

Calculation of stress concentration factor of the DEN specimen

The DEN specimen from Figure 3.6, $r/d=0.17$, $D/d=1.33$

From Figure 2-8¹, $K_T^{red}=2.1$

where K_T^{red} is based upon the reduced section stress.

Converting the SCF relative to the remote stress gives,

$$\begin{aligned} K_T^{bulk} &= K_T^{red} \frac{\sigma_{red}}{\sigma_{bulk}} \\ &= K_T^{red} \frac{W_{red}}{W_{bulk}} \end{aligned}$$

where $W_{bulk}=9.4\text{mm}$ and $W_{red}=7.3\text{mm}$ after grinding.

Therefore $K_T^{bulk}=2.7$

¹Spotts, M.F., Design of Machine Elements, Fourth Ed., Prentice-Hall Publishers, 1971.

Appendix B

Estimation of AC skin depth

Recall, the skin depth for a cylindrical conductor, δ , is given by,

$$\delta = \left(\frac{\rho}{\pi \mu_r \mu_o f} \right)^{1/2}$$

where ρ is the resistivity of the material, μ_r is magnetic permeability of the conductor relative to free space, $\mu_{air}=1.257 \times 10^{-6}$ henry/m is the magnetic permeability of free space and f is the AC frequency.

The resistivity and magnetic permeability of Merl 76 were not available in the examined literature, therefore, this data was obtained from materials as similar to Merl 76 as possible.

Resistivity: The resistivity was measured using a Hocking, Autosigma 2000 conductivity measuring device; conductivity measured 1.23% IACS, therefore,

$$\underline{\rho = 140 \mu\Omega \cdot \text{cm}}$$

Magnetic Permeability: Table 10 shows the magnetic permeability for several different

nickel base superalloys with all of them being close to unity. Since Waspaloy was closest in composition, its permeability was used.

$$\underline{\mu_r = 1.011}$$

$$\text{therefore, } \underline{\delta = 0.6\sqrt{1/f}}$$

$$\text{For } f = 30 \text{ kHz, } \delta = 3.5 \text{ mm}$$

Description of stress intensity factor for a semi-elliptical crack growing from the centre of a semi-circular notch [31]

$$K-S\sqrt{\frac{\pi a}{Q}}F_m\left(\frac{a}{c}, \frac{a}{t}, \frac{c}{r}, \frac{c}{w}, \frac{r}{t}, \frac{r}{w}, \phi\right)$$

$$Q = 1 + 1.464(a/c)^{1.65} \quad \text{for } a/c \leq 1$$

$$Q = 1 + 1.464(c/a)^{1.65} \quad \text{for } a/c > 1$$

$$F_m = [M_1 + M_2(a/t)^2 + M_3(a/c)^4]g_1g_2g_3g_4g_5f_\phi f_w$$

for $0.2 < a/c < 2$, $a/t < 1$, $1 < r/t < 3.5$, $(r+c)/w < 0.5$, $r/w = 1/16$, and $-\pi/2 < \phi < \pi/2$. (Note that here t is defined as one-half of the full sheet thickness.) For $a/c \leq 1$

$$M_1 = 1$$

$$M_2 = 0.05/[0.11 + (a/c)^{1.2}]$$

$$M_3 = 0.29/[0.23 + (a/c)^{1.2}]$$

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

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$$g_2 = [1 + 0.358\lambda + 1.425\lambda^2 - 1.578\lambda^3 + 2.156\lambda^4]/(1 + 0.08\lambda^2)$$

$$\lambda = 1/[1 + (c/r) \cos(0.9\phi)]$$

$$g_3 = 1 + 0.1(1 - \cos\phi)^2(1 - a/t)^{10}$$

$$g_4 = K_T[0.36 - 0.032/(1 + c/r)^{1/2}]$$

$$g_5 = 1 + (a/c)^{1/2}\{0.003(r/t)^2 + 0.035(r/t)(1 - \cos\phi)^3 \\ - 0.35(a/t)^2(1 - 0.5a/c)^3 \cos\phi$$

The finite-width correction f_w was

$$f_w = 1 - 0.2\gamma + 9.4\gamma^2 - 19.4\gamma^3 + 27.1\gamma^4$$

for uniform stress and

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

for uniform displacement with a specimen-length-to-width (L/w) ratio of 1.5 (L is measured from the crack plane to the grip line on the specimen) where

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

The function f_ϕ is given by

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

For $a/c > 1$

$$M_1 = (c/a)^{1/2}(1.04 - 0.04c/a)$$

and f_ϕ is given by

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

Appendix D

Transition crack length from Stage I to II FCG at room temperature

- 1) The transition from Stage I to Stage II growth is said to occur when the crack tip plastic zone size attains the grain size.

With the plane strain cyclic crack tip plastic zone size defined by, Figure 3.12 is given by [40],

$$\omega = 0.033 \left(\frac{(1-R)K_{\max}}{\sigma_o} \right)^2$$

where R is the stress ratio, $K_{\max}$ is the crack tip SIF at maximum load and σ_o is the flow stress.

Solving for $K_{\max}$ gives,

$$K_{\max} = \left(\frac{\omega}{0.033} \right)^{1/2} \frac{\sigma_o}{1-R}$$

With large plastic deformation at the notch root, the flow stress would be above the yield and below the ultimate strength, therefore, the flow stress was estimated by the average

of the two,

$$\sigma_o = \frac{\sigma_y + \sigma_{ult}}{2}$$

Determine K_{max} required to make the crack tip plastic zone size equivalent to the grain size.

i) Assuming only one lobe of the crack tip plastic zone must engulf a grain, let ω , Figure 3.12, equal the average grain diameter;

$$\omega = D_{avg} = 6\mu\text{m},$$

and the flow stress at room temperature taken as,

$$\begin{aligned}\sigma_o &= (S_y + S_{ult})/2 \\ &= (1082 + 1590)/2 \\ &= 1336 \text{ MPa}\end{aligned}$$

this gives,

$$K_{max} = \left(\frac{6 \times 10^{-6}}{0.033} \right)^{1/2} \frac{1336 \times 10^6}{(1 - 0.1)}$$

Therefore, $K_{max} = 20.0 \text{ MPa}\sqrt{\text{m}}$ at the transition from Stage I to II growth under plane strain

conditions where $\omega = D_{avg}$

2) Determine crack depth c , corresponding to $K_{max} = 20.0 \text{ MPa}\sqrt{\text{m}}$ for $P_{max} = 16.3 \text{ kN}$.

The Newman SIF described in Appendix C was used to determine the SIF corresponding the crack depth using a program written in Turbo Pascal

Program inputs:

- i) remote stress = 729 MPa @ $P_{max} = 16.3 \text{ kN}$
- ii) $K_I = 2.7$ (based on remote stress)
- iii) $a/c = 1.4$

Therefore, $c_{transition} = 0.070 \text{ mm}$ at the transition to Stage II growth for $P_{max} = 16.3 \text{ kN}$

Appendix E

Application of K_{pt} to the SCGR data of 7075-T6, Figure 5.8 [79]

- 1) The monotonic, plane stress notch root plastic length was determined by using Glinka and Newports' equations describing the stress distribution at a notch [80]. These equations gave the stress in the direction parallel (σ_{yy}) and perpendicular to (σ_{xx}) the loading direction with $\sigma_{zz}=0$, plane stress. With the assumption of these being the principle stresses, ρ_{mono} was established as the distance away from the notch root at which the von Mises stress was equal to the material yield stress, ie. $\sigma_y = \sqrt{(\sigma_{yy}^2 + \sigma_{xx}^2)}$, Table E1.

- 2) The Neuber SCF was determined using the method described in Section 5.6.1.1 where the intersection between the stress strain curve of 7075-T6¹ and the Neuber equation could be used to calculate K_σ , Table E1.

- 3) Values of da/dN and ΔK were read off the graph and the crack length at a given ΔK was calculated using the Newman's SIF model, Table E2, given the following information;

¹Aerospace Structures Metals Handbook, Department of Defence, Mechanical Properties Data Center, 1991.

$$\frac{a}{c} = 40 \left(\frac{a}{t} \right) + 0.2 \quad \frac{a}{t} < 0.2$$

$$\frac{a}{c} = 1 + 0.1 \left(\frac{a}{t} \right) \quad \frac{a}{t} \geq 0.2$$

$$K_t = 3.15, \quad t = 1.15 \text{ mm}, \quad W = 50 \text{ mm}, \quad r = 3.2 \text{ mm}, \quad S = 200 \text{ MPa}$$

4) Given the crack depth, c , K_{pl} could be determined from Equation 5.3,

$$K_{pl} = 2.45 + \frac{c}{0.365} (3.15 - 2.45), \quad c \leq 0.365 \text{ mm}$$

5) Rather than applying K_{pl} to the data points and reducing their ΔK , it was much simpler to apply it to the predicted curve to increase its ΔK . Since the SIF is proportional to the SCF, ΔK_{new} can be determined by,

$$\Delta K_{new} = \frac{K_t}{K_{pl}} \Delta K, \quad c \leq 0.365 \text{ mm}$$

This caused the curve to shift to the right until a crack length of $c = 0.365 \text{ mm}$ ($\Delta K < 7.6 \text{ MPa}\sqrt{\text{m}}$) at which point it joined with the original curve, Table E2.

Table E1 Determination of parameters required to apply K_{pl} to Al 7075-T6 FCGR data [79]

Bulk Stress, S (MPa)	Notch Root Plastic Length*, (mm)	$K_{\sigma} = \frac{\sigma_{max}}{S}$
195	0.311	2.49
200	0.365	2.45
220	0.589	2.28

Table E2 Determination and application of K_{pl} to 7075-T6 SCGR data from ref [79]

Readings from FCGR Graph, Figure 5.8a		a (μm)	c (μm)	K_{pl} ($\text{MPa}\sqrt{\text{m}}$)	ΔK_{new} ($\text{MPa}\sqrt{\text{m}}$)
da/dN (mm/cycle)	ΔK ($\text{MPa}\sqrt{\text{m}}$)				
10^{-6}	1.27 (knee)	4.7	13	2.47	1.62
$10^{-5.71}$	1.5	7	16	2.48	1.91
$10^{-5.52}$	2	17	22	2.49	2.53
$10^{-5.14}$	3	49	49	2.54	3.72
$10^{-4.71}$	4	90	89	2.62	4.81
$10^{-4.51}$	5	150	148	2.73	5.77
$10^{-4.28}$	6	220	216	2.86	6.61
$10^{-4.11}$	7	310	302	3.03	7.28
$10^{-3.67}$	10	640	606	3.15	10

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