Materials Lifetime

SCIENCE & ENGINEERING

0 cycles
150 Cycles
600 Cycles

Grips
Sample

5 mm
Materials Lifetime
SCIENCE & ENGINEERING


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This book is a collection of papers presented at a symposium on "Materials Lifetime Science and Engineering" sponsored by the Mechanical Behavior of Materials Committee of The Minerals, Metals & Materials Society (TMS) and ASM International. The symposium took place at the 2003 TMS Annual Meeting, San Diego, California, March 2 – March 6. The objective of the symposium was to provide fundamental understanding and theoretical modeling of materials lifetime science and engineering of metals and alloys including advanced materials. Advanced materials include biomaterials, bulk metallic glasses, intermetallics, composites, superalloys, etc.

The most complex and often most damaging processes that control the lifetimes of structural materials are those that involve synergistic interactions between environmental and mechanical effects. Mechanistic understanding and modeling are needed to further develop materials lifetime science and engineering, and formulate predictive methodologies. Emphases are placed on mechanical/environmental interactions, damage evolution, and final failure. Some of the areas explored are as follows:

1. Lifetime Studies of Conventional Materials in Aqueous Environments
2. Lifetime Studies of Advanced Materials in Aqueous Environments
3. Lifetime Studies of Advanced Materials in High-Temperature Gaseous Environments
4. Lifetime Studies of Oxide Scales in High-Temperature Gaseous Environments.

The symposium attracted scientists and engineers from universities, industries, and government agencies worldwide. We were very much encouraged by the turnout of the participants with strong interest in the research and application of materials lifetime science and engineering. The symposium was highlighted by thoughtful discussions and technical interchanges among the participants.

We would like to thank all of the participants for the success of the symposium, and the authors for their excellent contributions to the book. We are confident that this book will provide invaluable reference information for the research on "Materials Lifetime Science and Engineering."

It is our belief that it is only through vigorous research on and understanding of "Materials Lifetime Science and Engineering," the engineering applications of materials can then become a common practice.

The symposium organizers were Peter K. Liaw and Raymond A. Buchanan of the University of Tennessee, Dwaine L. Klarstrom of Haynes International, Inc., Robert P. Wei and D. Gary Harlow of Lehigh University, and Peter F. Tortorelli of Oak Ridge National Laboratory.
The organizers would like to thank the National Science Foundation for the financial support of the Integrative Graduate Education and Research Training (IGERT) Program on “Materials Lifetime Science and Engineering (DGE-9987548)” with Drs. W. Jennings and L. Goldberg as program monitors.

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Session

Materials Lifetime Science and Engineering
LINKAGE BETWEEN SAFE-LIFE AND CRACK GROWTH APPROACHES FOR FATIGUE LIFE PREDICTION

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Abstract

Current methods for predicting fatigue lives are based upon the safe-life and/or the crack growth approaches, both of which are empirically based. They do not adequately reflect long-term operating conditions, or identify the sources and extent of their contributions to variability. A linkage between these two approaches is established and demonstrated. Through this linkage, variability in S-N response can be related to key random variables that are more readily identified in the crack growth models. The identification and quantification of the role of these variables are paramount for predicting fatigue damage evolution and service lives. The effectiveness of this approach is shown through analysis of an extensive set of S-N data for 2024-T4 aluminum alloy from the literature. Variability associated with manufacturing and material variables are considered. The adoption of this demonstrated linkage to put life prediction on a sound scientific and probabilistic basis is recommended.

Introduction

The standard procedures for predicting fatigue lives embody either the safe-life or the crack growth approaches, or a combination of the two. Both approaches, however, are empirically based, in that they are developed primarily from experimental data. These data reflect conditions imposed by external variables, such as, applied stress, temperature, and environmental conditions. They do not effectively reflect the influence of internal variables, like microstructure and its interactions with the environment. Consequently, these approaches are only appropriate for interpolations over the available data, most of which are produced through accelerated testing and may not adequately reflect long-term operating conditions. Also the sources and extent of the variables that contribute to variability are seldom identified. A connection between these two approaches is established and demonstrated herein, so that the S-N response and the associated variability in fatigue lives can be related to the key random variables that are more readily identified in the crack growth models. The identification and quantification of these variables are vital for predicting fatigue damage evolution (or fatigue lives) and its distribution. The effectiveness of this approach is shown through analysis of an extensive set of S-N data for 2024-T4 aluminum alloy from the literature. Variability associated with manufacturing and material variables are considered, and the use of this linkage to put life prediction on a sounder scientific and probabilistic foundation is discussed.

A Model for Fatigue Crack Growth

A simplified fatigue crack growth model that captures the effect of localized damage in an open hole in an alloy is proposed. The model includes the pertinent material properties and damage mechanisms. The dominant damage is considered to be a semi-circular surface crack that transitions into a through-the-thickness crack. The quantity of interest is the fatigue life $N_f$, as a
function of the material properties, loading conditions, and subsequent damage evolution. Since the damage process is divided into two regimes, $N_f$ must be considered for each. In order to capture the observed scatter in fatigue lives, randomness associated with the initial damage size and material properties are explicitly represented in the model.

The fatigue crack growth rate $da/dN$, for the crack length $a$, is assumed to have the generalized power law form given by

$$\frac{da}{dN} = C(\Delta K - \Delta K_{th})^n,$$

where $n$, taken to be deterministic, represents the mechanistic dependence, specifically the functional dependence of the crack growth rate on the driving force $\Delta K$ and the threshold $\Delta K_{th}$. For the 2024-T4 aluminum alloy considered below, $n = 3.33$. The coefficient $C$ and the threshold $\Delta K_{th}$ are assumed to be random variables (rvs) that characterize the variability in the microstructural properties of the material and environmental influences.

The driving force $\Delta K$ is considered to be of two different forms according to whether the crack is a surface crack (sc) or a through-the-thickness crack (tc). For simplicity and computational expediency, it is assumed that $\Delta K$ for a surface crack is given by

$$\Delta K_{sc} = \frac{2.24}{\pi} k_t \Delta \sigma \sqrt{\pi a},$$

where $\Delta \sigma$ is the stress amplitude, $2.24/\pi$ is for a semi-circular crack in an infinite plate, and $k_t$ is the stress concentration factor. Subsequently, the specific data to be considered was generated from rectangular specimens containing a center cut circular hole, consequently $k_t = 2.8$. When the crack is a through-the-thickness crack, $\Delta K$ is assumed to be equal to the following:

$$\Delta K_{tc} = F_{tc}(a/r_0) \Delta \sigma \sqrt{\pi a},$$

where $r_0$ is the radius of the hole. Numerical values for $F_{tc}(a/r_0)$ for an infinite plate under uniaxial tension containing a circular hole with a single through crack emanating from the hole perpendicular to the loading axis can be fit empirically, to within graphical resolution, by the following function:

$$F_{tc}(a/r_0) = \frac{0.865}{(a/r_0) + 0.324} + 0.681,$$

which is quite suitable for analytical computations; see [1].

Let $N_{sc}$ and $N_{tc}$ represent the number of cycles required for surface crack and through-the-thickness crack growth, respectively. Then, the fatigue life $N_f$ is

$$N_f = N_{sc} + N_{tc} = \int_{a_o}^{a_{tc}} \frac{da}{C(\Delta K_{sc} - \Delta K_{th})^n} + \int_{a_{tc}}^{a_f} \frac{da}{C(\Delta K_{tc} - \Delta K_{th})^n},$$
where \( a_0 \) is the initial damage size, \( a_{tc} \) is the crack size at which the surface crack transitions into a through-the-thickness crack, and \( a_f \) is the final crack size. The initial damage size \( a_0 \) is assumed to be a rv that captures the scatter in the starting crack size. The first integral can be explicitly integrated when \( \Delta K_{tc} \) is given by equation (2); however, the second must be integrated numerically when \( \Delta K_{tc} \) is taken to be equation (3). Transition is assumed to occur at the crack size \( a_{tc} \) that is the solution of

\[
F_{tc} \left( \frac{a_{tc}}{r_0} \right) = \frac{2.24}{\pi} k_t,
\]

which is easily found to be

\[
a_{tc} = r_0 \left[ \frac{0.865}{(2.24 k_t / \pi) - 0.681} - 0.324 \right].
\]

### Characterization of the Random Variables

Statistical variability is assumed to be captured through \( a_0, C, \) and \( \Delta K_{th} \), which are assumed to be mechanistically and statistically independent of time. Scatter in material properties and resistance to fatigue crack growth are reflected in \( C \) and \( \Delta K_{th} \). Hole quality is depicted by the distribution for \( a_0 \).

The three-parameter Weibull cumulative distribution function (cdf), given by

\[
F(x) = 1 - \exp \left[ - \left( \frac{x - \gamma}{\beta} \right)^\alpha \right], \quad x \geq \gamma
\]

has been found to characterize each rv adequately. The parameter values for \( \Delta K_{th} \) are characteristic of data contained in [2]. Although the data are more limited, \( C \) was estimated from data in [3]. The parameters for \( a_0 \), for the following illustrations, were taken to be consistent with surface damage induced by machining processes. Table I contains the Weibull cdf parameters \( \alpha, \beta, \) and \( \gamma \), as well as the mean \( \mu \) of the cdf used for the computations. It should be noted that these parameters were developed from 2024-T3 aluminum alloy data, but it is assumed that these variables do not significantly differ from the 2024-T4 aluminum alloy from which the data considered below were generated.

<table>
<thead>
<tr>
<th>RV</th>
<th>( \alpha )</th>
<th>( \beta )</th>
<th>( \gamma )</th>
<th>( \mu )</th>
</tr>
</thead>
<tbody>
<tr>
<td>initial damage size ( a_0 ) (m)</td>
<td>1.5</td>
<td>5.5E-6</td>
<td>35.0E-6</td>
<td>40.0E-6</td>
</tr>
<tr>
<td>fatigue coefficient ( C ) (m/cyc)/(MPa√m)(^{3.33} )</td>
<td>20.0</td>
<td>3.16E-11</td>
<td>5.0E-13</td>
<td>3.13E-11</td>
</tr>
<tr>
<td>threshold driving force ( \Delta K_{th} ) (MPa√m)</td>
<td>20.0</td>
<td>1.35</td>
<td>1.01</td>
<td>2.32</td>
</tr>
</tbody>
</table>

### S-N Data for 2024-T4 Aluminum Alloy

Shimokawa and Hamaguchi [4] developed a rather extensive S-N data base, using about 1000 specimens, for 2024-T4 aluminum alloy. Their main concern was the effect of notch
configurations on fatigue life; however, they conducted a series of fatigue tests for rectangular specimens (110 mm long, 52 mm wide, and 1 mm thick) with a center cut circular hole of radius 5 mm. These data are the focus of this investigation. The holes were made by turning operations on a lathe, and burrs were removed by light stoning. The fatigue testing was conducted in a typical laboratory environment (temperatures of 295-297 K and relative humidities of 50-56%). All of the constant amplitude tests were performed at a frequency of 30 Hz on a single machine with a single operator. Thus, the experimental error was minimized.

A statistical summary of the fatigue life data produced by Shimokawa and Hamaguchi [4] is given in Table II. Note that the sample sizes are quite large. Testing 222 total specimens is an extensive undertaking. For stress amplitudes $\Delta\sigma$ of 157 MPa or greater the sample coefficients of variation are essentially the same, and they are reasonably small. For 137 MPa, however, the scatter increases significantly. Indeed, this is the dominant challenge for life prediction. As stress is lowered to reflect normal operating conditions, fatigue lives become very large and their scatter drastically increases. Consequently, fatigue life predictions cannot be based solely on empirical S-N data analyses. Mechanistically and physically based modeling is warranted.

<table>
<thead>
<tr>
<th>Stress Amplitude, $\Delta\sigma$ (MPa)</th>
<th>Sample Size, $m$</th>
<th>Median Life</th>
<th>Sample Average, $\bar{x}$</th>
<th>Sample Standard Deviation, $s$</th>
<th>Sample Coefficient of Variation (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>255</td>
<td>21</td>
<td>18,500</td>
<td>18,200</td>
<td>1760</td>
<td>9.6</td>
</tr>
<tr>
<td>235</td>
<td>30</td>
<td>29,100</td>
<td>28,700</td>
<td>2500</td>
<td>8.7</td>
</tr>
<tr>
<td>206</td>
<td>30</td>
<td>59,300</td>
<td>59,400</td>
<td>4230</td>
<td>7.1</td>
</tr>
<tr>
<td>177</td>
<td>30</td>
<td>144,200</td>
<td>146,000</td>
<td>12,600</td>
<td>8.6</td>
</tr>
<tr>
<td>157</td>
<td>30</td>
<td>251,700</td>
<td>264,000</td>
<td>22,600</td>
<td>8.6</td>
</tr>
<tr>
<td>137</td>
<td>30</td>
<td>469,100</td>
<td>519,000</td>
<td>96,200</td>
<td>18.5</td>
</tr>
<tr>
<td>127</td>
<td>30</td>
<td>1,424,700</td>
<td>1,710,000</td>
<td>1,090,000</td>
<td>63.8</td>
</tr>
<tr>
<td>123</td>
<td>21</td>
<td>4,401,800</td>
<td>4,530,000</td>
<td>2,660,000</td>
<td>58.7</td>
</tr>
</tbody>
</table>

In order to account for the scatter in data, given $\Delta\sigma$, the typical practice is to statistically fit a log-normal or a Weibull cdf to the data. In fact, Shimokawa and Hamaguchi [4] use a log-normal cdf. The characteristic S-N curve is usually determined from the medians or means of the estimated cdfs. The primary difficulty with this analysis is that it is empirical and its accuracy is entirely dependent on the quality and quantity of the available data. Again, an alternative approach is required for accurate life prediction.

**Computations and Analyses**

Computations from the proposed probabilistic fatigue crack growth model were made to evaluate its validity for the S-N data for 2024-T4 aluminum alloy produced by Shimokawa and Hamaguchi [4]. Figure 1 contains the experimental data as well as the computed cdfs from the model. Several observations are warranted. It should be noted that this database is excellent as far as the quantity of life testing data. The data are plotted on Weibull probability paper for convenience and to accentuate the lower tail behavior, which is, in turn, the region of high reliability and primary interest in design considerations. If the data are linear, then a two-parameter Weibull cdf would be acceptable as an empirically derived statistical model. Such is the case for $\Delta\sigma \geq 157$ MPa; however, the data are clearly not linear for the remaining values of $\Delta\sigma$. Possibly using a three-parameter Weibull cdf would be satisfactory, but again, the decision would be based solely on empiricism. Thus, this approach is not desirable. Statistically modeling the data with a log-normal cdf, as suggested by Shimokawa and Hamaguchi [4], does
not completely alleviate the problem. On the other hand, the proposed approach models damage evolution from which cdfs are computed and subsequently compared to the data for validation. None of the curves shown on Fig. 1 are linear, even though they appear to be nearly so for \( \Delta \sigma > 157 \) MPa. The reason for the linear appearance is attributable to the small amount of scatter in the data. As the statistical scatter significantly increases with decreasing \( \Delta \sigma \), the nonlinearity becomes pronounced.

The cdfs computed from the fatigue crack model fit the S-N data extremely well, except for \( \Delta \sigma = 137 \) and 157 MPa. In fact, the worst Kolmogorov-Smirnov (KS) statistic for the KS goodness-of-fit test is 0.24 when \( \Delta \sigma = 137 \), which implies that the fit is acceptable with a confidence exceeding 95%. The other cdfs fit the data, using the KS test with a confidence exceeding 90%. Thus, the cdfs computed from the proposed fatigue crack growth model are very accurate representations for these data. It should be reiterated that these cdfs depend explicitly on the microstructure and material properties, as well as, the loading conditions.

The other representation of the data that is of importance is the S-N plot. Figure 2 shows the data in Shimokawa and Hamaguchi [4] on a traditional linear versus logarithm S-N plot. Again, as the stress amplitude is reduced, the increased scatter is apparent. The lines on Fig. 2 are percentile lines abstracted from the computed cdfs from the proposed fatigue crack growth model. The solid line is the median, i.e., 50\(^{th}\) percentile, which is the common functional representation for S-N data. Indeed, it passes through the middle of the data at each \( \Delta \sigma \). The shortcoming of using the median for S-N behavior is that it does not capture the effect of scatter in the data. If the amount of scatter is not constant for every \( \Delta \sigma \), as is the case for the Shimokawa and Hamaguchi data, then the median is insufficient for characterizing the S-N behavior. The long-dashed lines are the 99\% confidence bounds, consisting of the 0.5 and 99.5 percentiles computed from the fatigue crack growth model. Similarly, the dashed-dotted lines are the 95\% confidence bounds. Clearly, the 99\% bounds capture the data entirely. Except for a few data, the 95\% confidence bounds encompass the data as well. Thus, the proposed model not only depicts the median behavior accurately, it also models the scatter in the data at each \( \Delta \sigma \). The importance of the model is that accurate predictions can be made for the fatigue life including the variability in life.

Figure 1: S-N data for 2024-T4 aluminum alloy (Shimokawa and Hamaguchi [4]) along with cdfs computed from the proposed fatigue crack growth model.
Conclusions

The primary purpose of this effort has been to demonstrate that there is a connection between the traditional S-N and the fatigue crack growth approaches, and that the linkage is provided through appropriate modeling of the processes of damage evolution. The modeling requires an accurate representation of all of the key variables, including internal and external variables, in the model as well as their probability distributions. The demonstration centered upon data provided by Shimokawa and Hamaguchi for 2024-T4 aluminum alloy. It was shown that the data for fatigue lives for fixed stress amplitude fit very accurately to cdfs computed from the proposed fatigue crack growth model. Furthermore, the S-N behavior of the data is entirely captured by the 99% confidence bounds computed from the fatigue crack growth model. Thus, the model accurately describes the median and scatter in the data. These findings establish the unity between the S-N and crack growth approaches. The approach demonstrated herein should be developed further and adopted for fatigue life prediction in design applications.

References


A MECHANISTIC BASED STUDY OF FATIGUE CRACK PROPAGATION IN THE SINGLE CRYSTAL NICKEL BASE SUPERALLOY CMSX-2

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Abstract

Creep considerations have been a driving force for the introduction of single crystal nickel base superalloy components into gas turbine engines. Nonetheless, fatigue considerations necessitate accurate fatigue crack propagation (FCP) modelling to prevent costly failures. FCP in single crystals is characterized by non-self-similar crack growth and for nickel base superalloys the appearance of multiple distinct fracture morphologies. Therefore, traditional FCP rate modelling using $\Delta K$ as a correlating factor for crack growth rates is questionable. FCP rate tests were conducted using CMSX-2 at room temperature and 973K in air and vacuum with two different crystallographic orientations. It was found that all cracking was associated with two distinct morphologies on the fracture surface; one in which the $\gamma$ precipitates were sheared and one in which they were avoided. Transmission electron microscopy (TEM) confirmed that two different mechanisms were operating; one in which the dislocations sheared the precipitates and one in which they bypassed the precipitates. A finite element analysis (FEA) was performed to predict the stresses and strains ahead of the crack tip for each tested condition. The combination of FEA and TEM in conjunction with scanning electron microscopy (SEM) observations suggests that the non-self-similar crack growth and distinct morphologies are controlled by the state of stress. High resolved shear stresses in the direction of the Burger's vector along with low normal stresses on the cube faces of the $\gamma$ precipitates produced crystallographic crack growth while low resolved shear stresses and high normal stresses produced $\gamma$ avoidance crack growth. This fundamental understanding of the crack growth mechanisms will assist in more accurately modelling FCP rates in single crystal nickel base superalloys.

Introduction

The turbine section of jet engines operates under hostile conditions which necessitate the extensive use of nickel base superalloys. Frequently, the combustion temperature can exceed the melting temperature of the material used in the turbine blades. The relatively constant rotational speed produces a uni-directional force on all the rotating components which in combination with the high operating temperature produces a creep loading environment. In addition to the creep loading, each engine cycle
introduces one major fatigue cycle between engine startup and shut down. Minor fatigue cycles can also result from changing thrust settings during the course of a flight.

Although nickel-base superalloys possess the unusual property of increasing yield strength with increasing temperature, the mechanical properties, particularly the creep related properties, suffer from the deleterious effects of oxidation. Transverse grain boundaries are usually the initiation sites for creep rupture when void coalescence and oxidation are the primary creep rupture mechanisms. Thus, a way to improve the creep rupture properties of turbine blades is to eliminate or reduce the number of grain boundaries as was achieved by directional solidification [1] and more recently by use of single crystal components. There has been considerable interest in the fatigue behavior and damage mechanisms of Ni-base superalloy single crystals [2,3,4,5,6,7].

Although creep properties are improved by the use of single crystal components, they make FCP rate predictions difficult. Fatigue crack growth prediction models have traditionally been based on global parameters, such as the stress intensity factor, \( K_t \), which characterize the stress and strain fields ahead of a crack tip. The mechanisms of damage accumulation in single crystal and directionally solidified alloys are significantly different than from those of isotropic polycrystalline materials. The lack of grain boundaries changes the dynamics of oxidation and void coalescence as well as removing a set of obstacles to dislocation movement. Additionally, the fixed orientation of single crystal components leads to a greater localization in dislocation induced damage than polycrystalline materials. Fundamentally sound life prediction models must address these peculiar characteristics of single crystal materials.

The specific goals of the research reported in this paper were:

- to experimentally characterize the fatigue crack propagation rates and mechanisms of CMSX-2 single crystal material as function of temperature, environment and crystallographic orientation
- to use the results of 3D finite element analysis (FEA) simulation of the observed damage modes in test specimens to study the crack tip stress distribution ahead of the crack and also across the thickness of the specimen.

Material

CMSX-2 in single crystal form was used in this study. Its physical metallurgy of CMSX-2 has been well characterized [1,8,9,10,11,12,13]. As with many other single crystal alloys, CMSX-2 also has a dendritic macrostructure. The size of the \( \gamma \) precipitates is a function of the radial position within the dendrite. At the core of the dendrite, the \( \gamma \) size is approximately 0.35 μm along the edge while in the interdendritic region its size is approximately 0.65 μm on side. In addition to the \( \gamma \) size variations in the radial direction, composition variations in the matrix have been observed from dendritic to interdendritic regions. The specimens used in this study were given a heat treatment developed by ONERA (Office Nationale Études et de Recherches Aérospatiales) and detailed elsewhere [1]. This heat treatment is designed primarily to form uniform cubes of \( \gamma \) precipitates for improved creep resistance. The specimens were made of a heat of material cast under industrial conditions in a single heat with a thermal gradient of 40°C/cm.
Experimental and Analytic Procedures

Fatigue crack growth testing was carried out using compact type (CT) specimens that closely, but not exactly, followed the geometry specified by ASTM standard E399 and E647[14,15]. The difference was found in the pin hole height location. The differences in the K-calibration between the specimen used and the standard specimen were verified to be negligible by a comparison between values of the stress intensity factor given by the well known ASTM equation and a FEA of the actual specimen geometry with a straight crack. Differences between the values determined by FEA and the ASTM equation were less than 1.5% for identical loads and crack lengths. A series of 16 specimens was tested at two temperatures in two different environments and two different crystallographic orientations. Duplicate specimens were tested for each condition. Test temperatures of 298K and 973K and were chosen because previous studies [16,17] demonstrated that these temperatures produced very different fracture surface morphologies for similar material systems. The higher temperature corresponds to the temperature at which the $\gamma$ precipitates have maximum strength while room temperature corresponds to a temperature for which the $\gamma$ precipitates have a significantly lower strength. The two test environments were laboratory air and ultra-high vacuum (i.e. $10^{-8}$ torr). The two crystallographic orientations which were tested are shown in Figure 1.

![Diagram of CT specimen orientations](image)

Figure 1. CT specimen orientations used in this study

Both orientations shared a common [001] loading direction but differed in the direction of the intended crack front movement [010] and [110] with a common projected crack plane of (100).

Crack Length Measurement

Crack length was measured for all tests using the electro-potential drop system. The leads were made of thin gage pure nickel wire and attached to the specimen using a spot welder. This process is known to damage the specimen and in some cases to introduce a polycrystalline structure at the attachment points. However, no cracks were observed to initiate at the attachment points for the lead wires nor were the
attachment points near any cracks. Crack length was computed using the well known Johnson equation [18,19]. Previous researchers have used this technique to measure crack length and have reported that the predictions correspond to the crack length as projected onto the (100) planes in single crystals[17]. Due to the primary loading direction, the projected crack length was always equivalent to the crack length as projected on the plane normal to the loading direction (i.e. the crystallographic {100} plane was normal to the loading direction). Post test comparison of predicted crack length and measured crack length validated this equivalency.

Electron Microscopy

Scanning Electron Microscopy (SEM) with an Hitachi S-800 was used to investigate the fracture surfaces. Both qualitative as well as quantitative measurements were obtained from these investigations. Images were formed using secondary electron emission with an accelerating voltage of 20KV.

Finite Element Analysis

FEA was used to calculate the stress and displacement fields ahead of the crack tip and along the specimen thickness for a variety of crack geometries and orientations. A three dimensional analysis was performed using the ABAQUS code with 20 node iso-parametric elements and singularity elements at the crack tip. Anisotropic elastic constitutive laws were used for the three dimensional analysis.

Specimen Geometries

In order to accurately predict the stress, strain and displacement fields, the specimen had to be modeled with a mesh representing the actual specimen. Although all specimens had the same exterior geometry, the geometry of the crack changed considerably with environment, temperature and orientation; therefore, a separate mesh was developed for each condition. The only differences between the meshes are found in the crack tip region which are shown in greater detail elsewhere[20].

Results

Fatigue Crack Growth Rates

The crack growth rates were calculated using the sliding seven point polynomial method outlined in ASTM standard E647[15]. While the applicability of $\Delta K_1$ as a unique correlating factor for crack growth rates in single crystal specimens is uncertain, it will be used here for comparing crack growth rates of identical specimens tested at identical loading ranges. Figure 2 compares crack growth rates for specimens tested under approximately the same load ranges of $P_{\max} = 2224$ N (500 lbs) and $P_{\min} = 222.4$ N (50 lbs).
The conditions tested are listed in order of ascending crack growth rate as follows:

- 298K, environment: air, orientation: <010>
- 298K, environment: air, orientation: <110>
- 973K, environment: vacuum, orientation: <010>
- 973K, environment: air, orientation: <010>
- 973K, environment: vacuum, orientation: <110>

One trend that emerges is that crack growth rates at elevated temperatures are higher than at room temperatures. Also, it is noted that specimens with <010> direction of nominal crack propagation have lower propagation rates compared to those with <110> orientations at both 298K in air and 973K in vacuum. At 973K in air, the crack propagation rate trends for the two orientations were reversed; specimens with <110> orientations have lower crack propagation rates than those with <100> orientations.

Crack Surface Morphologies

The fracture surfaces obtained from the fatigue tests were examined using SEM. When the fracture surfaces of all specimens were compared, two distinctive types of fracture features were observed on each crack surface: (1) precipitate avoidance and (2) “cleavage”. “Precipitate avoidance” type fractures were macroscopically flat in appearance. These surfaces were nominally perpendicular to the loading direction and were most commonly observed in the middle of the specimen. The “cleavage” type fractures were observed near the specimen surfaces and the crack plane coincided with a crystallographic plane, {111}, and were further inclined at a high angle (~45°) to the loading direction. The relatively flat fracture surface produced by the γ avoidance mechanism can be modeled as self-similar Mode I type fatigue crack growth. However, the presence of cleavage type fracture in the specimens leading to “flutes” cause the crack growth to become “non self-similar” thus violating the
The tenets of linear elastic fracture mechanics. In this context, "cleavage" is not used to indicate single cycle failure. This nomenclature refers to fatigue crack growth along a crystallographic plane. While all fracture surfaces contain both morphologies, the distribution and relative fractions differ with differing test conditions. Specimens with a nominal <110> direction of crack propagation showed the greatest range in apparent crack advance mechanisms ranging from 100% crystallographic (or cleavage) crack growth when tested in air at room temperature to almost 100% γ avoidance when tested at 973K in air. Specimens with this orientation tested at 973K in vacuum showed a combination of the two mechanisms of crack growth. Therefore, this orientation tested at room temperature in air and 973K in vacuum are examined in further detail.

<110>, Air, 298K Fracture Surfaces

Specimens tested at room temperature were tested only in an air environment because environmental effects were expected to be minimal. Near the machined notch tip, the fatigue fracture surfaces were predominantly on a macroscopic {100} plane normal to the loading direction. A closer examination of this region showed a series of wavelike formations growing orthogonally in <100> directions with a "wavelength" comparable to that of residual dendrites. These structures were subsequently identified as residual dendrites. When examined at high magnifications, the fracture surfaces in this region also showed a series of cubes approximately 0.5μ on side which appeared to be oriented with their faces parallel to {100} planes. This is shown in Figure 3.

Figure 3. Visible γ precipitates on fracture surface

This micrographs shows that the facets of the fracture surface are occurring on {100} planes. Comparisons with the measured size for γ strongly suggest that they are in fact γ precipitates. As expected, these γ precipitates did not appear to have corrosion products on their faces. Moving on the crack surface along the nominal direction of propagation, the non-crystallographic crack growth is no longer observed and is replaced with fracture surfaces which are highly crystallographic in nature. These planes, or 'flutes', are of {111} type that intersect in a common <110> direction parallel to side planes of the specimen. Examination of these flutes shows that they are not microscopically smooth as one
would expect if they were composed of a single \{111\} plane but rather appear to be composed of multiple parallel \{111\} planes. This is shown in Figure 4.

![Image](image.jpg)

**Figure 4.** High magnification micrograph of "cleavage" fracture.

Moving along the crack surface in the direction of nominal crack growth, complete transformation to cracking on \{111\} planes is observed. However, as one moves further, the crack surface morphology reverts to an appearance that is the same as the one near the machined notch. This transition appears very abruptly. The line delineating the crack surface morphology change is very straight and occurs along a line perpendicular to the width direction of the specimen.

**<110>, Vacuum, 973K Fracture Surfaces**

The fracture surfaces from specimens tested in vacuum at 973K with a secondary orientation of <110> showed some similarity to those tested at the same temperature in air with the same orientation but also some important differences. As with the previous specimens, the fatigue precrack region near the machine notch showed the two characteristic types of fractures with the cleavage faces appearing next to the free surfaces of the specimen. However, in contrast to the specimen tested in air, the cleavage faces in vacuum continued to form remote from the pre-crack. Again, one could clearly see the "waves" in the middle of the specimen thickness where the fracture morphology was $\gamma'$ avoidance but cleavage faces near the surface. Moving along the crack surface in the nominal crack propagation direction, the relative amounts of the two characteristic regions changes with the fraction of area covered by cleavage type faces increasing. These cleavage type faces appeared to grow towards the center of the specimen. Thus, the regions of $\gamma'$ avoidance were always at the center of the specimens. Moving further along the crack, the number of \{111\} type planes that the cleavage surface traverses decreases and that at the end of the crack, there are only two or three active \{111\} planes.

**Deformation Mechanisms**

TEM was used to characterize the deformation mechanisms in the crack tip region. Because there were two distinctly different fracture morphologies observed on each crack surface, it was considered likely that there are two distinct deformation mechanisms as well. In order to test this hypothesis, TEM foils were made from regions ahead of the crack tip which displayed each type of fracture morphology.
Within the CT specimens, the extent of deformation was extremely inhomogeneous with most deformation occurring near the crack tip as expected. The specimens which exhibited the greatest amount of $\gamma$ avoidance fracture were those tested in vacuum at 973K with a secondary orientation of $<110>$. Therefore, these specimens were used to investigate the deformation mechanisms associated with $\gamma$ avoidance. TEM foils taken from these specimens show two primary operating mechanisms; dislocation shearing of $\gamma$ precipitates and dislocation looping around $\gamma$ precipitates. Those foils which showed the greatest amount of precipitate avoidance by the dislocations came from the areas of $\gamma$ avoidance on the fracture surface. However, these foils also revealed some $\gamma$ shearing. Foils taken from regions of crystallographic crack growth exhibited the greatest amount of $\gamma$ shearing. For those foils which showed $\gamma$ avoidance, the majority dislocation line length is found in the $\gamma$ channel between adjacent $\gamma$ precipitates. From these observations one could conclude that dislocation avoidance results in a macroscopically flat fracture morphology while precipitate shearing is associated with crystallographic (or cleavage) fracture surfaces.

**Finite Element Results**

The results from FEA were used primarily in support of the dislocation arguments; therefore, except when specifically noted, a coordinate system was chosen such that the stresses appropriate to crystal plasticity are conveniently examined. (i.e. shear stresses in the direction of the Burgers vector and normal stresses to the plane containing the Burgers vector and the dislocation) In order to achieve this goal, the $x_3$-axis was chosen so that it coincides with the direction of the Burgers vector while the $x_2$-axis was normal to the slip plane containing that Burgers vector. Thus, $\tau_{x_1x_3}$ represents the resolved shear stress on the slip plane.

**High Temperature Specimens with $<110>$ Orientation**

These specimens were modeled with a flat subsection in the center of the specimen thickness corresponding to crack growth in which $\gamma$ precipitates were avoided and "flutes" on the edges that corresponded to crystallographic crack growth on {111} planes. Since these features are found on specimens from both air and vacuum, this configuration models both environments.

The normal stress on the plane containing the Burgers vector varies as a function of the thickness position. These stresses are the highest of the values found on any of the four slip planes. These stresses appear to be fairly constant across the center subsection in which the fracture was normal to the loading direction. There is a perturbation in stress associated with the "flute" in which the stress first decreases and then increases compared to the value in the mid-section of the specimen. The maximum value for the resolved shear stress also varies with position along the thickness direction. The absolute value of the stress is lowest in the center and increases as one approaches the edge with a perturbation associated with the "flute". Even with the perturbation, the trend is clearly for an increasing value of resolved shear stress as the are approached. Finally, the normal stresses acting on the faces of the $\gamma$ precipitates in the direction of the load are highest in the center and reduce in magnitude towards the specimen edges with a significant perturbation associated with the "flute".

**Room Temperature Specimens with $<110>$ Orientation**

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