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DAMAGE ACCUMULATION BY CRACK GROWTH UNDER COMBINED CREEP AND FATIGUE

DISSERTATION

Presented in Partial Fulfillment of the Requirements for the Degree Doctor of Philosophy in the Graduate School of The Ohio State University

By

Carl Edward Jaske, B.S., B.S., M.S.

* * * * *

The Ohio State University

1984

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INTRODUCTION

Engineering equipment that operates at elevated temperatures typically is subjected to a combination of both steady (creep) and cyclic (fatigue) loading. Examples of such equipment are steam turbines, steam generators, gas turbines, nuclear power plant structural components, rocket engines, and systems for chemical and petroleum processing. This equipment is usually fabricated from metallic materials, and its structural integrity depends on the ability of the material to resist damage from the loadings that it experiences during service operations.

In a generic sense, damage may be defined as any phenomenon that degrades a material and its load-carrying capabilities and/or decreases its remaining operating life. Creep damage is usually associated with steady or slowly varying loadings and has a time-dependent nature. In commonly used engineering alloys, such damage is evidenced by the formation and growth of voids, cavities, and fissures along grain boundaries and results in intergranular fractures. Although other types of creep damage, such as irreversible slip, do occur, the types cited above are the ones frequently encountered at the stress levels and temperatures of practical interest.
In contrast to creep damage, fatigue damage normally is associated with rapidly varying cyclic loadings. Fatigue-crack initiation and growth usually follow a transgranular path, although they may sometimes follow an intergranular path. Transgranular initiation and early stages of growth are at and along persistent slip bands. The later stages of fatigue-crack propagation and those of main engineering interest follow a transgranular path controlled by the direction and nature of the stress field.

In general high-temperature operations, it is likely that both creep and fatigue damage will occur. Under creep-fatigue loading conditions, a synergistic interaction between creep and fatigue damage may take place and degrade the material much more rapidly than might otherwise be anticipated. As pointed out in an extensive survey report\(^1\), creep-fatigue damage interaction has been observed in many laboratory studies. Most past creep-fatigue work has been concerned with the prediction of crack initiation or failure in small, unnotched specimens subjected to fully reversed axial loading. Little work has been directed toward understanding and dealing with the problem of crack growth explicitly under combined creep-fatigue.

Because typical engineering structural components often contain small flaws or cracks when they are put into service, the development of procedures for quantifying creep-fatigue-crack propagation is an important item. Dealing directly with crack growth is a powerful tool than enables service life to be computed from some initial to some final crack size by integration of the appropriate crack-growth-rate relationships. This type of approach involves the
use of fracture mechanics. In general, the deformations associated with creep-fatigue are inelastic; inelastic fracture mechanics (IFM) rather than linear elastic fracture mechanics (LEFM) should be applied to the creep-fatigue-crack growth problem.

The objective of this study was to evaluate the application of IFM methods to the problem of creep-fatigue-crack growth. To achieve this objective, a combined experimental/analytical research program was undertaken. Creep-fatigue-crack-growth experiments were performed on specimens of Type 316 stainless steel in air at 593 and 649 C. IFM analyses were carried out using path-independent line integrals, and a crack-tip-zone interaction model was developed for creep-fatigue-crack growth.
I. METHODS OF CREEP-FATIGUE-DAMAGE ASSESSMENT

An extensive amount of literature, such as the 441-page volume published by Oak Ridge National Laboratory\(^{(1)}\), has been devoted to the subject of creep-fatigue in recent years. It is beyond the scope of the current study to review exhaustively all of this past work in detail. It is instructive, however, to discuss briefly some well-known approaches of treating creep-fatigue damage. This discussion will provide background and perspective for the research undertaken in this study. Readers interested in more detailed discussions of creep-fatigue-damage assessment should refer to one of the recent reviews of this subject\(^{(1-9)}\).

ENGINEERING DESIGN

Interest in developing methods of creep-damage assessment has been motivated by the needs of designers of equipment for elevated-temperature operation. Linear summation of life fractions is the procedure most widely used to account for creep-fatigue-damage accumulation in engineering design. For example, this procedure is used in the design of certain nuclear reactor structural components for elevated-temperature service following the rules of Case N-47 of the American Society of Mechanical Engineers (ASME) Boiler and Pressure Vessel Code.

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The linear life-fraction rule is simply a combination of Miner's\(^{(10)}\) cycle-fraction rule for fatigue and Robinson's\(^{(11)}\) time-fraction rule for creep, as originally suggested by Taira\(^{(12)}\). It has been shown, for example by Jaske et al\(^{(13)}\), that it is often inadequate and suffers many shortcomings. For this reason, there has been a great deal of research devoted to developing and evaluating alternative creep-fatigue-damage accumulation models.

An early approach to incorporating the effect of creep on fatigue into life-prediction methods was to add simple modifying factors to the empirical relationships between strain range and fatigue life. Manson's 10 percent rule\(^{(14)}\) lowered the normally-anticipated total strain range-versus-fatigue life curve by a factor of 10 on cyclic life to account for the effects of elevated temperature. Coffin\(^{(15)}\) added a frequency-dependent multiplier to the inelastic strain range-versus-fatigue life relationship to incorporate the effects of cyclic frequency and hold-time variations on fatigue life. These approaches worked only for simple loading histories that introduced small amounts of time-dependent damage.

The concept of strain-range partitioning was introduced by Manson et al\(^{(16)}\) to account for differences between damage induced by tensile and compressive straining. As reviewed in a recent paper by Halford and Saltsman\(^{(17)}\), strain-range partitioning has been widely evaluated by research investigators and applied to some engineering design problems. With a similar motivation to take into account the behavioral differences induced by tension going and compression going straining, Coffin\(^{(18)}\) developed the frequency separation
approach for creep-fatigue-damage prediction.

The approaches described above were developed to satisfy engineering design needs. They all have a phenomenological basis and employ various manipulations of strain range and/or stress versus-life relationships developed in laboratory studies.

CONTINUUM MODELS

More recently, creep-fatigue-damage accumulation models have been developed using the principles of continuum mechanics. These models have evolved from Kachanov's brittle rupture theory(19), where damage is a scalar internal variable that represents the state of partial cracking of the material. Lemaitre and Plumtree(20-22) have employed this approach to account for nonlinear creep-fatigue-damage accumulation in OFHC copper, Alloy 800, and a proprietary austenitic stainless steel. Continuum models deal with macroscopic quantities that can be measured experimentally, but they make no attempt to relate these quantities to observed metallurgical mechanisms of creep and fatigue damage.

MECHANISTIC MODELS

Mechanistic creep-fatigue-damage approaches attempt to model the observed metallurgical mechanisms of damage evolution, such as the growth of voids and cavities and the propagation of cracks. The most popular of these methods is the damage-rate approach of Majumdar and Maiya(23-27). It assumes that there exists some size of initial crack, which may correspond to an inherent microstructural defect.
Under combined creep and fatigue loading, the initial crack then grows to a final critical size associated with failure. The time-dependent-crack-growth rate is assumed to follow a power-law function of the inelastic strain and the inelastic strain rate with different proportionality constants for tensile and compressive loading. The growth and shrinkage of grain boundary cavities is incorporated in their approach using a similar type of power law.

In their review of life-prediction methods for creep-fatigue, Lloyd and Wareing\(^{(9)}\) concluded that none of the existing approaches are adequate for general design applications. They point out that a major obstacle to comprehensive evaluation of these methods is the lack of a sufficient set of materials data. However, they do feel that crack-propagation approaches show the most promise and deserve further investigation.

In applying the damage-rate approach to creep-fatigue-life prediction, Majumdar and Maiya do not explicitly treat crack growth. Instead, they integrate the damage-rate equations for various types of loading waveforms and then evaluate the integration and material constants to provide proper correlation with the total life of specimens tested in the laboratory. If the damage-rate equations were developed by means of independent observations of the damage evolution process, they could be integrated to predict total life.
II. FATIGUE- AND CREEP-CRACK GROWTH

Fracture mechanics has been applied to the problem of characterizing fatigue- and creep-crack-growth behavior independently. Before treating the more complex problem of combined creep-fatigue-crack-growth, methods of dealing with either fatigue- or creep-crack-growth will be discussed. First, the application of LEFM will be reviewed briefly. Second, the application of IFM will be addressed.

LINEAR ELASTIC FRACTURE MECHANICS

Paris et al\(^{(28,29)}\) first introduced the application of LEFM to characterize subcritical crack growth. They demonstrated that the cyclic-fatigue-crack-growth rate, \(\frac{da}{dN}\), could be correlated with the range of the linear elastic stress intensity factor, \(\Delta K\), as long as the plastic deformation associated with fatigue cycling was confined to a small zone at the crack tip relative to the crack size and the specimen dimensions. Today, use of \(\Delta K\) versus \(\frac{da}{dN}\) relationships in fatigue-life calculations is a commonly accepted engineering practice.

The first applications of LEFM to fatigue-crack propagation were for experimental data on metals at room temperature. Later, it was found that the same LEFM approach could be applied to fatigue-crack-propagation behavior at elevated temperature as long as the same
restrictions regarding plastic deformation were maintained. Many of the elevated-temperature crack-growth studies have been concerned with austenitic stainless steels, as discussed in the extensive review paper of James (30). Other good reviews of LEFM applications to high-temperature fatigue-crack growth have been written by Speidel (31) and Sadananda and Shahinian (32).

Time-dependent effects have been investigated by measuring da/dN as a function of ΔK for various load-time waveforms. The most commonly used waveforms are trapezoidal (rapid loading and unloading with hold times at maximum and/or minimum load) and sawtooth (slow loading followed by rapid unloading or vice versa). Detrimental time-dependent effects are found to increase da/dN values for given ΔK values compared with rapid continuous fatigue cycling. No well-accepted model has been developed to account for these effects. The usual design approach is to use ΔK versus da/dN data for the waveform that simulates service loadings. That method has difficulty in handling complex load histories, requires development of a large materials database, and does not provide a satisfactory framework for interpolation and extrapolation.

The next logical historical step was to attempt to apply LEFM to creep-crack-growth behavior. Ellison and Harper (33), Fu (34), Sadananda and Shahinian (35), and Christian et al (36) have reviewed the application of fracture mechanics, including LEFM, to creep-crack growth. LEFM is found to be not generally applicable under creep conditions, but it does characterize creep-crack-growth behavior in some cases. (Analytical reasons for the validity of applying LEFM
to creep are discussed in the next section of this chapter.) Typically, the linear elastic stress intensity factor, $K$, characterizes creep-crack-growth rate, $da/dt$, for materials with low creep ductility and pronounced environmental sensitivity. 

**INELASTIC FRACTURE MECHANICS**

Realizing the shortcomings of LEFM for applications where significant plastic deformation occurs, investigators sought to apply IFM to low-cycle fatigue-crack propagation. The applicability of the J integral to this problem was evaluated. As illustrated in Figure 1, J is defined as a path-independent line integral taken counterclockwise about an arbitrary contour, $\Gamma$, in a cracked body.

\[
J = \int_{\Gamma} W \, dy - \mathbf{r} \left( \frac{\partial u}{\partial x} \right) ds ,
\]

**FIGURE 1.** COORDINATE SYSTEM AND ARBITRARY LINE CONTOUR AT CRACK TIP
where the strain energy density, $W$, is

$$W = \int_{0}^{r_{mn}} \sigma_{ij} \epsilon_{ij} \, ds$$  \hspace{1cm} (2)

The stress and strain tensors are $\sigma_{ij}$ and $\epsilon_{ij}$, respectively. The directions parallel and perpendicular to the crack are $x$ and $y$, respectively, with the crack tip at the origin. The arc length along $\Gamma$ is $s$, and $T$ is the traction vector defined by the outward normal vector, $\hat{n}$ ($T_i = \sigma_{ij} n_j$). The displacement vector is $\hat{u}$. As reviewed by Begley and Landes(38), the Hutchinson(39) and Rice and Rosengren(40) (commonly referred to as HRR) crack-tip stress and strain singularities are functions of $J$ for a power-law stress-plastic strain relationship.

Dowling and Begley(39) used an operational definition of the range of the $J$-integral, $\Delta J$, to characterize fatigue-crack growth in compact-type (CT) specimens of A533B steel at room temperature under conditions of gross cyclic plastic deformation. Dowling later extended this work to other specimen configurations and another steel(42,43) and to the growth of small cracks in low-cycle fatigue (LCF) specimens(44). Mowbray(45) showed that the $J$-integral approach could be used to derive a relationship quite similar to functions used to characterize low-cycle fatigue resistance. El Haddad et al(46) showed that $\Delta J$ could be used to characterize the growth behavior of short cracks in low-cycle fatigue. Thus, $\Delta J$ can be used to explicitly treat fatigue-crack growth, and the results of its use
are consistent with engineering fatigue-life approaches.

Jaske and Begley\(^\text{(47)}\) demonstrated that \(\Delta J\) also could be applied to elevated-temperature low-cycle fatigue of Type 316 stainless steel. Sadananda and Shahinian\(^\text{(48)}\) and Taira et al\(^\text{(49)}\) also used \(\Delta J\) to characterize the elevated-temperature fatigue-crack growth of this alloy. Recently, Sehitoglu and Morrow\(^\text{(50)}\) used the cyclic range of crack-tip opening displacement (CTOD) to characterize thermal-fatigue crack growth and to correlate the thermal-fatigue results with room-temperature (isothermal) data for AISI 1070 carbon steel. Since \(J\) generally is interpreted as being equivalent to CTOD\(^\text{(51)}\), \(\Delta J\) also should be a good characterization parameter for such thermal-fatigue-crack-growth data.

Paralleling the application of \(J\) to fatigue-crack growth, Landes and Begley\(^\text{(52)}\) proposed a \(C^*\) integral for creep.

\[
C^* = \int_{\Gamma} W^* \, dy - \frac{1}{T} \left( \frac{dU}{dx} \right) ds , \tag{3}
\]

where the strain energy rate density, \(W^*\), is

\[
W^* = \int_0 \sigma_{ij} \dot{\varepsilon}_{ij} \, ds . \tag{4}
\]

Equations (3) and (4) are the same as Equations (1) and (2) except that the displacement vector is replaced by the displacement rate vector, \(\dot{\mathbf{u}}\), and strain is replaced by the strain rate, \(\dot{\varepsilon}_{ij}\). Goldman and Hutchinson\(^\text{(53)}\) have shown that the HRR crack-tip stress and strain-rate singularities are functions of \(C^*\) for power-law steady-state creep.
Landes and Begley\(^{(52)}\) performed creep-crack-growth experiments on Discaloy\(^{+}\) at 649 C and found that the C\(^{*}\) parameter could be used to correlate creep-crack-growth rates successfully. In contrast, they found that K or nominal stress provided no correlation. As highlighted in recent review papers\(^{(33-36)}\), a number of other investigators have experimentally evaluated the application of C\(^{*}\) to creep-crack growth. The consensus of these reviews is that the C\(^{*}\) parameter is of more general use than K or nominal stress, especially for materials exhibiting ductile creep deformation response. Thus, there is strong experimental evidence that C\(^{*}\) is a useful parameter for characterizing creep-crack-growth behavior.

Riedel and coworkers\(^{(54-59)}\) have analyzed the stress and strain fields near cracks in creeping materials. For stationary tensile cracks, they evaluated these fields for elastic plus power-law creep and elastic plus power-law, strain-hardening primary creep constitutive response. At short times, the near-tip fields are characterized by K, and it is the appropriate parameter to correlate in the creep-crack-growth rate. At long times, these near-tip fields are of the HRR type and are characterized by C\(^{*}\), and it is the appropriate parameter to correlate with creep-crack-growth rate. The transition time, \(t_{1}\), from K to C\(^{*}\) dominated fields under plane-strain conditions for an elastic-nonlinear viscous material is given by\(^{(54)}\)

\[
t_{1} = \frac{K^{2}(1-v^{2})}{(n+1) E C^{*}} ,
\]

where \(n\) is the power-law-creep exponent, \(E\) is elastic modulus, and \(v\)

\(^{+}\) A 12Cr-24Ni-2.5Mo-1.45Ti-0.8 C (max) iron-base austenitic alloy.
is elastic Poisson's ratio. For plane stress, the \( (1-\nu^2) \) term goes to unity. For a strain-hardening material, the corresponding relation for \( t \) is implicit\(^{(55)}\).

Kubo\(^{(60,61)}\) incorporated the effects of strain hardening and creep recovery by using Robinson's\(^{(62)}\) constitutive model for creep, where strain rate is proportional to an effective stress power law. The effective stress is the difference between an applied stress and an internal stress. He showed that HRR-type singularities characterized the crack-tip stress and strain-rate fields and that for long-times \( C^* \) can be used to estimate the intensity of these fields. Bassani\(^{(63)}\) has used a hyperbolic-sine creep law for antiplane shear (Mode III) conditions and showed that \( C^* \) is an appropriate characterization parameter for long-time creep. Thus, \( C^* \) has been found to be an applicable parameter for characterizing the HRR-type stress and strain-rate fields for stationary creep cracks with a variety of creep constitutive relations.

Hui and Riedel\(^{(56)}\) analyzed the stress and strain fields near a slowly growing crack in a elastic-nonlinear viscous material and found that the nature of these fields changed sharply at \( n = 3 \). For \( n < 3 \), the singularity follows an inverse square root relation and the elastic strain rates dominate. For \( n > 3 \), a new type of singular field is developed that has a stress and strain dependence of the same form as the stress dependence of the HRR-type field. In this case, the field is dependent only upon the crack-growth rate and material parameters.
Riedel and Wagner\cite{57} extended the work of Hui and Riedel\cite{56} to investigate the conditions under which $K$- or $C^*$-controlled crack growth may be expected for slowly growing cracks, in addition to the previously discussed conditions for stationary cracks. They defined two characteristic crack-tip zone lengths, $r_1$ and $r_2$, within which the singular field of the growing crack is dominant

\begin{equation}
\begin{aligned}
r_1 &= \left[ \frac{A}{E} \frac{K^{n-1}}{\frac{2}{(n-3)}} \right], \\
\end{aligned}
\end{equation}

and

\begin{equation}
\begin{aligned}
r_2 &= \left[ \frac{\frac{da/dt}{A}}{\left( \frac{n+1}{2} \right)} \right] \left[ \frac{A}{E} \frac{C^*}{\frac{2}{(n-1)}} \right], \\
\end{aligned}
\end{equation}

where $A/E$ is the coefficient in the power-hardening creep law\cite{64}. For the short-time or small-scale yielding case where $K$ is the appropriate parameter, $r_1$ applies. For the long-time, extensive creep case where $C^*$ is the appropriate parameter, $r_2$ applies. Values of $r_1$ and $r_2$ must be small compared with the crack length and uncracked ligament size for $K$ and $C^*$ to be valid characterization parameters.

In addition to the constraints on $r_1$ and $r_2$, the amount of crack growth must be small compared with the crack length and uncracked ligament size\cite{57}. The reason for this restriction is that the crack growth rate is history dependent based on a critical-strain criterion for crack advance. As the crack grows and creep ahead of the crack tip becomes more extensive, the uncracked ligament becomes more damaged and dependence of $da/dt$ upon crack length as well as upon $K$ or $C^*$ is predicted. Kubo et al\cite{65} also have shown such a history
dependence in their analysis of creep-crack growth. In the extensive creep case, crack growth is expected to become unstable if $\frac{da}{dt}$ exceeds some critical value ($57$).

Recently, Stonesifer and Atluri$^{(66,67)}$ have evaluated the application of the path-independent integral $(A\Delta T)_{C}$ to creep-crack growth. They showed that the time-rate of the first term of that parameter, $(\dot{T}_{1})_{C}$, characterized tensile crack-tip fields for both steady and nonsteady creep. Finite-element analyses were performed for both CT specimens and infinite-strip geometries$^{(67)}$. At steady state, the value of $(\dot{T}_{1})_{C}$ was found to be essentially the same as that for plane strain and only 4 percent different for plane stress. In a creep-crack-growth simulation for Type 304 stainless steel, they found that the crack-tip field was at essentially steady-state conditions. From the above results, either $(\dot{T}_{1})_{C}$ or $C^{*}$ would be expected to characterize creep-crack-growth behavior in austenitic stainless steels. From the arguments presented by Stonesifer and Atluri$^{(66,67)}$, $(A\Delta T)_{C}$ may have broader applicability than $C^{*}$.

This brief review of selected IFM parameters shows that $\Delta J$ and $C^{*}$ are useful parameters for characterizing low-cycle fatigue-crack growth and creep-crack growth, respectively. Thus, it is reasonable to evaluate their potential application to creep-fatigue-crack growth.
III. APPROACH FOR CREEP-FATIGUE-CRACK GROWTH

As part of this study, Jaske and Begley\(^{(47)}\) developed an approach for assessing creep-fatigue-crack propagation. In the ensuing sections of this chapter, the basic assumptions upon which the approach is based, the methodology for implementing the approach, the modeling of creep-fatigue interaction with the approach, and the advantages and limitations of the approach are discussed.

**BASIC ASSUMPTIONS**

In the general case, creep-fatigue-crack growth is an extremely complex problem. It may involve large inelastic deformations in regions both near and far from the crack tip. The stress state is generally multiaxial with nonproportional loadings. Cyclic deformation is time independent and usually associated with reversed slip processes, whereas creep deformation is time dependent and typically associated with mechanisms such as dislocation climb, grain boundary cavitation, and grain boundary sliding. Synergistic interactions between these two modes of deformation can markedly influence the cumulative deformation behavior and, in turn, the crack-growth process. Temperature often varies, rather than remaining constant, resulting in considerations of thermal creep-fatigue. The loading history is often a complicated variable amplitude one rather than a simple constant amplitude one. Environmental factors related to
corrosion and oxidation can affect crack-growth behavior. In view of these many complexities, development of a completely general approach for assessing creep-fatigue-crack growth would be an immense task.

Because of the complexities described above, it is necessary to make a number of simplifying assumptions in order to make a first step toward dealing with the problem of creep-fatigue-crack growth. With the assumptions described below, the approach must be limited to situations where they are reasonable. The temperature is assumed to be constant; thus, thermal fatigue problems are not treated and only isothermal ones are considered. Environmental factors are assumed to be negligible or to be incorporated in the baseline data through proper simulation of anticipated service conditions. For simplicity in identifying damage events, the initial development of this approach is limited to constant amplitude cycling. In other words, an increment of crack growth can be easily related to each cycle of loading. The three-dimensional crack is assumed to be characterized by a two-dimensional approximation so that crack size is characterized by a single measure of length. Finally, the damage or material cracking is assumed to take place within a small process zone near the crack tip. Large deformations may occur within this process zone, but it must be surrounded and contained by a region that is limited to small-scale, proportional inelastic deformations.

The final point above is key to the approach. It is analogous to restricting inelastic deformation to a process zone near the crack tip in LEFM. With small-scale inelastic deformation in the bulk material, deformation theory of plasticity can be employed. Then,
for two-dimensional problems the J and C* integrals can be applied.

The irreversibility of inelastic flow in metals restricts the application of J and C* to monotonically increasing loading of a cracked body without any crack extension in the strictest sense. However, analysis by Hutchinson and Paris (68) indicates that J-controlled crack extension will occur when (1) the uncracked ligament is much larger than the increment of crack advance, and (2) the rate of change in J with respect to crack length times the ratio of the uncracked ligament length to the value of J is much larger than unity. Although quantitative values for these two requirements are not yet available, it seems reasonable to expect that they may often be satisfied during a cycle of fatigue-crack growth. In a similar fashion, the work of Riedel and Wagner (57) discussed earlier provides guidelines for estimating the conditions under which C*-controlled crack extension is anticipated.

Another restriction must be placed on the crack size in order to make the application of fracture mechanics reasonable. Namely, the crack length should be much larger than the process zone size. For, if the crack size is too small, the dimensions of the process zone will be significant in comparison with crack length. In such a case, the above field parameters no longer provide a reasonable measure of stress and strain (or strain rates) in the near crack-tip zone.

**METHODOLOGY**

Within the limitations of the above assumptions, J or K is used to correlate the fatigue (cyclic) portion crack growth and C* or K.
is used to correlate the creep (time-dependent) portion crack growth in the current work. Thus, it is assumed that cyclic crack growth takes place by a time-independent mechanism related to plastic straining and that creep-crack growth takes place by a separate time-dependent mechanism related to creep straining. On this basis, \( \frac{da}{dN} \) and \( \frac{da}{dt} \) can be expressed as separate functions of \( J \) and \( C^* \) respectively,

\[
\frac{da}{dN} = f_1(J) \text{ or } f_1(K),
\]

and

\[
\frac{da}{dt} = f_2(C^*) \text{ or } f_2(K),
\]

Above the threshold region of crack growth, \( f_1 \) and \( f_2 \) often can be approximated well by simple power functions \((41,42,52)\)

\[
\frac{da}{dN} = C_1 \Delta J^\gamma
\]

and

\[
\frac{da}{dt} = C_2 C^* \gamma'
\]

where \( C_1 \), \( C_2 \), \( \gamma \), and \( \gamma' \) are constants. If \( K \) is the valid parameter, \( \Delta K \) is substituted in Equation (10) and \( K \) is substituted in Equation (11) with appropriate changes in the values of the constants.

Manson \((69)\) has suggested four partitioned cyclic crack-growth relationships for creep-fatigue. Equation (8) corresponds to his suggestion for reversed plasticity (p-p). However, instead of three
cyclic relations for creep-crack growth, only one time-based relation for crack growth is proposed herein, Equation (9). It is reasoned that creep-crack growth will take place only under effective tensile stressing and that the differences in growth rate associated with those three types of cycling can be accounted for in terms of the closure stress associated with the previous compressive going cyclic excursion. For example, consider the stress-strain response in a uniaxially loaded specimen under fully reversed straining with a tension hold time, as illustrated in Figure 2. Both $\Delta J$ and $C^*$ must be computed with consideration of the closure stress level, $\sigma_c$.

Cumulative creep-fatigue-crack growth is treated by linearly summing increments of crack advance associated with critical events in the loading history in the same sequence that these events take place. These histories are presently limited to constant amplitude

† Creep reversed by plasticity (c-p), plasticity reversed by creep (p-c), and creep reversed by creep (c-c).
cycling with a stable stress-strain response so that each cycle is clearly defined. Under a rapid stress or strain reversal, damage is assumed to be time-independent (i.e., negligible creep) and the crack-growth increment is evaluated using $\Delta J$. During a very slow strain reversal or hold period, damage is treated as being time-dependent (i.e., negligible fatigue) and the crack-advance increment is evaluated using $C^*$. At intermediate strain rates, both $\Delta J$ and $C^*$ are employed. In this case, the fatigue-crack advance per cycle is linearly added to the creep-crack advance per cycle, computed by integrating the $da/dt$ versus $C^*$ relation for the cycle.

**Computation of $\Delta J$**

To apply $\Delta J$ to fatigue-crack propagation, one must be able to compute $\Delta J$ values for the specimen or component configuration of interest. Dowling and Begley\(^{41}\) and Dowling\(^{42,44}\) developed methods for computing $\Delta J$ using operational definitions of J-value estimates for compact-type (CT), center-cracked-tension (CCT), and low-cycle-fatigue (LCF) specimen configurations. In the current work, cyclic loadings of the CT and CCT specimens were within the regime where $\Delta K$ was valid, so there was no need to compute $\Delta J$ values for these two configurations. The $\Delta K$ values were computed using the well-known relations given in ASTM Designation: E647-81, Standard Test Method for Constant-Load-Amplitude Fatigue Crack Rates Above $10^{-8}$ m/cycle.

For the current experiments using LCF specimens, $\Delta K$ was never the valid parameter and $\Delta J$ values were computed following a modification
of the procedure given by Dowling\textsuperscript{44}. Basically, $\Delta J$ is approximated by the relation,

$$\Delta J = (G_1 W_e + G_2 W_p) a,$$  \hspace{1cm} (12)

where $W_e$ and $W_p$ are the elastic and plastic portions of the strain energy density, respectively, and $G_1$ and $G_2$ are geometrical factors.

The value of $G_1$ is the product of the elliptical shape-correction factor and the surface-crack correction factor,

$$G_1 = (1.12)^2/Q$$  \hspace{1cm} (13)

The value of $G_2$ is similar to that of $G_1$, but it is multiplied by a factor, $f_3(n)$, to account of strain hardening during plastic flow,

$$G_2 = (1.12)^2 f_3(n)/Q$$  \hspace{1cm} (14)

For a crack in an infinite sheet under-plane stress conditions, Shih and Hutchinson\textsuperscript{70}, showed that

$$f_3(n) = \left[ 3.85 \sqrt{n} (1-1/n) + \pi/n \right] \left[ 1+1/n \right].$$  \hspace{1cm} (15)

The above procedure is the same as that used by Dowling\textsuperscript{44} except that instead of assuming a semi-circular crack shape an elliptical crack shape was used.

Figure 3 shows a plot of $(1.12)^2/Q$ versus $a/2c$ that was used in the $\Delta J$ calculations of this study, where $a$ corresponds to the minor axis of the ellipse and $c$ corresponds to the major axis of the ellipse. For LCF specimens, the effective stress range was assumed
FIGURE 3. CORRECTION FACTOR FOR ELLIPTICALLY SHAPED SURFACE CRACKS
to be equal to the nominal stress range because the stress-strain hysteresis loops showed no cusps, indicative of significant crack closure, during compressive loading.

**Estimation of C***

To use C* in evaluating creep-crack-growth behavior, it is desirable to have simple methods of computing values of C* for specimen configurations used in the laboratory. Since C* is similar in mathematical form to J, procedures for estimating C* were patterned after those used to estimate values of J\(^{(71)}\). Considering the problem of a cracked infinite plate for which Rice\(^{(37)}\) evaluated J and the problem of a LCF specimen with a small surface crack, Jaske and Begley\(^{(47)}\) evaluated C* for a power hardening creep law

\[
\dot{\varepsilon}_c = \frac{(A/E)}{\sigma^n} t^m, \quad (16)
\]

and showed that

\[
C^* = \left(\frac{n}{n+1}\right) \dot{J}, \quad (17)
\]

where \(\dot{J}\) is the derivative of J with respect to time assuming that the inelastic strain is creep strain.

It then was argued that values of C* could be estimated by differentiating known expressions developed for J and multiplying the result by \(n/(n+1)\). This approach is used to estimate values of C* in the ensuing discussion.
C* for LCF Specimen

As pointed out in the earlier discussion of Equation (12), the plastic component of J, J_p, for an LCF specimen with a small embedded surface crack is of the form,

$$J_p = \frac{(1.12)^2}{Q} f_3(n) W^* a$$  \hspace{1cm} (18)

From the similarity of J and C*, it is reasonable to assume that

$$C^* = \frac{(1.12)^2}{Q} f_3(n) W^* a$$  \hspace{1cm} (19)

Equation (19) was used to compute C* values for LCF specimens in the current study.

**Constant Stress.** For power-law creep under constant stress conditions, the value of W* in Equation (19) can be evaluated by integrating Equation (4) using Equation (16) to yield

$$C^* = \left[\frac{n}{n+1}\right] \frac{(1.12)^2}{Q} f_3(n) \sigma_c a$$  \hspace{1cm} (20)

for \( \sigma = \) constant value, as shown by Jaske and Begley\(^{(47)}\).

**Constant Strain.** Similar to the constant-stress case, the value of W* can be evaluated for constant-strain conditions where stress relaxation occurs\(^{(47)}\). Integrating Equation (4) using Equation (16) gives

$$W^* = \left[\frac{n}{n+1}\right] \frac{A}{E} \sigma^{n+1} t^m$$  \hspace{1cm} (21)
The value of $\sigma$ in Equation (21) should be the range associated with crack closure (see Figure 2) just as was the case for fatigue-crack growth using $\Delta J$ earlier. The tensile-stress, $\sigma_t$, relaxation response is usually characterized in terms of the initial stress, $\sigma_0$, at time zero. Thus, the appropriate value of $\sigma$ is equal to $\sigma_t - \sigma_c$, where $\sigma_c$ is the stress level at closure.

For constant-strain conditions,

$$\frac{d \sigma_t}{dt} = -E \dot{\varepsilon}_c .$$ (22)

Using Equation (16) to write Equation (22) in terms of $\sigma$ and $t$, separating variables, and integrating in accordance with a time hardening rule (72) gives

$$\ln \left( \frac{\sigma_c}{\sigma_0} \right) = \left[ \frac{A}{m+1} \right] t^{m+1}, \quad n = 1$$ (23)

and

$$\sigma_c^{1-n} - \sigma_t^{1-n} = \frac{A(1-n)}{m+1} t^{m+1}, \quad n > 1$$ (24)

As reviewed by Laflen and Jaske (64), good approximations of stress relaxation response can be obtained by fitting data to either Equation (23) or (24), depending on the appropriate value of $n$.

Solving Equations (23) and (24) for $\sigma_t$, subtracting $\sigma_c$, and substituting in Equation (21) yields

$$W^* = \left( \frac{A}{2E} \right) t^m \left[ \sigma_0 \exp \left( \frac{-A}{m+1} t^{m+1} \right) - \sigma_c \right]^2, \quad n = 1$$ (25)

and

$$W^* = \left( \frac{n}{n+1} \right) \left( \frac{A}{E} \right) t^m \left[ \left( \sigma_0^{1-n} - \frac{A(1-n)}{m+1} t^{m+1} \right)^{1/1-n} - \sigma_c \right]^{n+1}, \quad n > 1$$ (26)
For the LCF specimens, $\sigma_c$ was taken to be equal to the maximum compressive stress just as in the calculations of $\Delta J$. Combining either Equation (25) or (26) with Equation (19) provides the desired estimate of $C^*$ for the constant-strain case.

$C^*$ for CT Specimen

Harper and Ellison (73) estimated $C^*$ for the CT specimen using limit-load analysis,

$$C^* = \left( \frac{n}{n+1} \right) \frac{P \delta_c}{Bw} \left[ - \frac{1}{m'} \frac{dm'}{d(a/w)} \right] ,$$

(27)

where $P$ is load, $\delta_c$ is load-line creep displacement rate, $B$ is specimen thickness, $w$ is specimen width, and $m'$ is the ratio of tensile limit load of a cracked specimen to that of an uncracked specimen.

From the work of Rice et al. (74), with a subsequent correction factor (75), $J$ for the CT specimen is given approximately by the formula,

$$J = \frac{2.3}{B(w - a)} \int_0^\delta P \, d\delta ,$$

(28)

where $\delta$ is the load-line deflection. Assuming a power hardening relation between $P$ and $\delta$, integrating Equation (28) and then differentiating and cancelling terms, it can be shown that

$$J = \frac{2.3 P \delta_c}{B(w - a)} .$$

(29)

Employing Equation (17) to estimate $C^*$ from $J$ gives -
Equation (30) differs from (27) by only the term in brackets. These two terms are plotted as a function of $a/w$ in Figure 4, and the agreement between them is quite good, especially for $0.4 \leq a/w \leq 0.6$. Since Harper and Ellison \(^{73}\) obtained good agreement with experimental values from Landes and Begley \(^{52}\), it is concluded that Equation (30) provides a concise simple method of calculating estimates of $C^*$ for a CT specimen.
C* for CCT Specimen

For the CCT-specimen configuration,

\[ J = \frac{k^2}{\varepsilon} + \frac{1}{B(\omega-a)} \left[ -\frac{P\delta}{2} + \int_{0}^{\delta_0} P \, d\delta \right] , \quad (31) \]

where

\[ K = \frac{P}{2Bw} \sqrt{\phi(a)} = \frac{P}{2Bw} \sqrt{\frac{ma}{2\omega}} . \]

Assuming that \( P = C_o \delta^{1/n} \),

\[ J = \left( \frac{C_o}{2Bw} \right)^2 \frac{\delta^{2/n} \phi(a)}{E} + \frac{1}{B(\omega-a)} \left[ \frac{C_o \delta^{1/n+1}}{(1/n+1)} - \frac{C_o \delta^{1/n+1}}{2} \right] . \quad (33) \]

Differentiating the above expression with respect to time and eliminating \( C_o \) by substituting Equation (32) gives

\[ J = \left( \frac{P}{Bw} \right)^2 \frac{\delta_c \phi(a)}{2nE \delta_c} + \frac{P \delta_c}{B(\omega-a)} \left[ 1 - \frac{1+n}{2n} \right] . \quad (34) \]

Then, it follows that

\[ C^* = \left( \frac{P}{2Bw} \right)^2 \frac{\delta_c \phi(a)}{2(n+1)E \delta_c} + \frac{P \delta_c}{B(\omega-a)} \left[ \frac{n-1}{2(n+1)} \right] . \quad (35) \]

If \( n = 1 \), the second term vanishes and

\[ C^* = \left( \frac{P}{2Bw} \right)^2 \frac{\delta_c \phi(a)}{E \delta_c} \quad (n=1) \quad (36) \]

If \( n \gg 1 \), then \((n-1)/(n+1) \geq 1\) and
\[ C^* = \left( \frac{P}{Bw} \right)^2 \frac{\dot{\delta}_c}{(n+1)} E \frac{\phi(a)}{c} + \frac{P \dot{\delta}_c}{2(W-a)B} \]  

(37)

\textbf{C* for 3-Point-Bend Specimen}

For the 3-point-bend-specimen configuration,

\[ J = \frac{2 \ P \ \dot{\delta}}{B(w-a)} , \]  

(38)

so it follows that

\[ \dot{J} = \frac{2 \ P \ \dot{\delta}}{B(w-a)} \]  

(39)

Thus,

\[ C^* = \left( \frac{n}{n+1} \right) \frac{2 \ P \ \dot{\delta}_c}{B(w-a)} \]  

(40)

Nibkin et al\(^{(76)}\) derived the following expression:

\[ C^* = \left( \frac{n}{n+1} \right) \frac{P \ \dot{\delta}_c}{Ba} \]  

(41)

This equation incorrectly says that \(C^*\) decreases as the crack length, \(a\), increases at constant load. Instead, \(C^*\) should increase as the crack length increases at constant load, as correctly predicted by Equation (40).

\textbf{CRACK-TIP-ZONE INTERACTION MODEL}

In the previous discussion of "Methodology", it was pointed out that creep and fatigue crack-growth are linearly summed to obtain total crack advance. This approach works well when interaction...
between the creep and fatigue modes of crack growth are negligible. However, there are cases when fatigue cycling occurs after a period of creep or vice-versa, where an interaction between the creep and fatigue modes of cracking may occur. To take into account creep-fatigue interaction effects, a crack-tip-zone interaction model\(^{(77)}\) was developed.

**Description of Model**

An idealized depiction of the crack-tip-zone model is shown in Figure 5. As illustrated in Figure 5a, the crack-tip process zone is assumed to be similar to the simple Irwin\(^{(78)}\) model of a crack-tip plastic zone. The crack length is \(a\) and the zone size is \(r\). Deformation in the bulk material is limited to the small-scale yielding (SSY) case, and the stress levels in the crack-tip process zone approach the true fracture stress, \(\sigma_f\), of the material. The term \(\sigma_y\) indicates the nominal yield stress of the bulk material.

It is assumed that prior creep-crack growth will influence subsequent fatigue cracking significantly only in the process zone, but not in the surrounding bulk material. In other words, the creep damage is restricted primarily to the process zone and is negligible in the material ahead of the growing crack. Similarly, it is assumed that prior fatigue-crack growth will influence subsequent creep-crack growth significantly only in the process zone, but not in the surrounding bulk material. Thus, creep-fatigue interaction will occur only when a fatigue (or creep) crack is growing through a process zone generated by a prior creep- (or fatigue-) crack growth.
a. CRACK-TIP-ZONE MODEL

b. CREEP ZONE LARGER THAN FATIGUE ZONE

c. FATIGUE ZONE LARGER THAN CREEP ZONE

FIGURE 5. IDEALIZED DEPICTION OF CRACK-TIP-ZONE MODEL
The size of the creep crack-tip process zone is designated as \( r_c \), and the size of the fatigue crack-tip process zone is designated as \( r_f \). As illustrated in Parts b and c of Figure 5, the creep zone may be either larger or smaller than the fatigue zone. Depending upon which case exists, the interaction between creep- and fatigue-crack growth will differ. For a creep-crack-tip process zone to exist the time, \( t \), involved in creep cracking must be greater than some incubation period, \( t_0 \).

First, consider the case where fatigue cracking occurs following a period of creep-crack growth: (1) if \( r_f \) is less than \( r_c \), the cyclic growth rate may be increased or decreased due to prior creep cracking. (2) If \( r_f \) is similar to or greater than \( r_c \), the fatigue-crack-growth rate will not be increased after creep cracking and will be similar to that typically expected for these nominal loading conditions.

Second, consider the case where creep-crack growth occurs after a period of fatigue or cyclic crack growth: (1) if \( r_c \) is greater than or similar to \( r_f \), then the creep-crack growth rate will be that usually expected for these creep conditions with no apparent influence of the prior fatigue cracking. (2) If \( r_c \) is less than \( r_f \), then the creep-crack-growth rate after fatigue cycling may be decreased or increased for some period of time until the crack grows through the zone associated with the prior fatigue crack. This model then provides a framework for assessing history effects in creep-fatigue-crack growth.
Estimates of Zone Size

From the results of Riedel and coworkers (55-59), it is reasonable to assume that, for small-scale yielding in the bulk elastic-plastic material, the size of the creep-crack-tip process zone is determined by the value of the maximum stress intensity factor, $K_{\text{max}}$, during the creep period. As long as deformation in the bulk material is small, the elastic stress solution characterized by $K$ provides a first-order approximation of the stress distribution near the crack tip, and it can be used to estimate the extent of the process zone. Hence, a simple first-order estimate of the process-zone size can be obtained using Irwin's plastic-zone relation.

There are two types of process zones associated with creep cracking. For a creep mode of cracking followed by a fatigue mode of cracking at a reasonably high cyclic frequency, the zone size is designated as $r_{c_f}$. The Irwin-type model for plane strain gives the following first-order estimate:

$$r_{c_f} = \frac{1}{3\pi} \left( \frac{K_{\text{max}}}{\sigma_f} \right)^2 .$$

(42)

For plane stress, $r_{c_f}$ would be about three times the above value. This formulation assumes that only the area near the crack tip where the stresses approach the level of the true fracture stress, $\sigma_f$, ahead of the growing creep crack will have an influence on the subsequent fatigue-crack-growth process where fatigue-crack growth occurs in a transgranular mode.
When the creep cracking is followed by slow fatigue cycling or creep at a lower stress, the crack will continue to grow in a creep-damage mode. The size of this zone is $r_c$, and it can be related to the stresses that are at the level that will be associated with stress rupture in the crack-tip-zone area in the time for which the increment of crack will grow through the zone at the steady-state rate at which it is propagating. For example, let the time to rupture, $t_r$, in a simple tensile stress-rupture experiment be given by the following relationship:

$$
\sigma_r = D t_r^d \quad , \quad (43)
$$

where $\sigma_r$ is the value of the rupture stress and $D$ and $d$ are material constants. As a simple first approximation, it can be assumed that the time-to-rupture of the crack-tip plastic zone increases inversely proportional to the creep-crack-growth rate and is given by the following expression:

$$
t_r = r_c / (da/dt) \quad . \quad (44)
$$

In a similar fashion to Equation (42) described earlier, the value of $r_c$ is given by the relationship,

$$
r_c = \frac{1}{3\pi} \left( \frac{K_{\text{max}}}{\sigma_r} \right)^2 \quad , \quad (45)
$$

for plane strain. Substituting Equations (43) and (44) into Equation (45) and solving for the value of $r_c$ gives the following:
This relation applies to plane-strain conditions, for plane stress, \( r_{c,2d+1} \) would be three times this value. Equation (46) then gives an estimate of the size of the crack-tip process zone where a significant amount of intergranular damage associated with creep cracking will be expected.

The cyclic plastic zone associated with fatigue cracking is expected to be about 1/4 the size of a comparable plastic zone for monotonic loading, so the fatigue crack-tip plastic zone size, \( r_f \), is given approximately by the relationship,

\[
    r_f = \frac{1}{12\pi} \left( \frac{\Delta K_{\text{eff}}}{\sigma_f} \right)^2 .
\]

Equation (47) is also for plane-strain conditions, and \( \Delta K_{\text{eff}} \) is the range of effective stress intensity factor associated with the fatigue-crack-growth process. This means that residual stress and/or crack-closure effects associated with fatigue cracking must be taken into account in calculating the value of \( \Delta K_{\text{eff}} \). The value of \( r_f \) for plane stress would be three times that given by Equation (47).

**ADVANTAGES AND LIMITATIONS OF APPROACH**

The primary advantage of the current approach is that it provides a simple, straightforward method of measuring creep-fatigue damage in terms of rates and increments of macroscopic crack growth. In contrast, other procedures have been used to provide measures of creep-fatigue damage in terms of cyclic life. They have not dealt directly
with rates and increments of crack growth even though some of them do have some features similar to those of the current approach.

Cycle-dependent and time-dependent crack growth are treated as two separate processes so that allowance can be made for the different types of mechanisms associated with each process. Also, the crack-tip-zone model permits the evaluation of creep-fatigue interaction effects during crack growth. At present, the model is largely qualitative, and it is quantified only by simple, first-order estimates.

The limited amount of experimental data developed in support of this approach restricts its current use to research studies. However, the ease with which $J$ and $C^*$ can be calculated using appropriate structural analysis methods and constitutive relationships gives the technique future potential uses in design and safety assessment of structures. As currently postulated, the approach requires a minimal amount of input data at each temperature, basically the two crack-growth relations. This is desirable in that it reduces the need for development of data under a wide variety of loading histories.

The major shortcoming of the approach is that it is based on a number of simplifying assumptions. Thus, it is only an initial step towards treatment of a complex problem. Since $J$ and $C^*$ are field parameters, just as the linear elastic stress intensity factor, their use must be limited to situations where a dominant process zone exists in a region near the crack tip. The surrounding material in which this region is imbedded must follow the rules of deformation theory of plasticity to a reasonable engineering approximation.
Therefore, cases of highly nonproportional loading could not be handled with the approach, as currently formulated. For example, large inelastic strains and/or redistribution of stresses in areas significantly removed from the crack tip would violate the theoretical basis of the model. Also, both J and C* are based on two-dimensional models where crack size is reasonably well-characterized by a single measure of length; whereas, in reality the crack is three-dimensional. Where this three-dimensional nature plays an important role, it is likely that these parameters will not provide adequate measures of damage.

Variable-amplitude loading histories are important in many practical applications. The closure-stress level provides a possible means of accounting for history effects on a damage-event basis (e.g., cycle-by-cycle basis). However, neither experimental nor analytical determination of this stress level is a simple, straightforward procedure in most cases. Thus, the model as currently presented may be difficult to apply to complex, variable-amplitude loading histories.

Even though this approach has many theoretical limitations, the subsequent experimental work shows that it has a great deal of promise. Thus, the approach merits further evaluation and development as an engineering tool for assessing creep-fatigue-crack growth.
IV. EXPERIMENTAL STUDIES

The experimental studies concentrated on the creep-fatigue-crack growth behavior of Type 316 stainless steel in air at 593 and 649 C. First, the sample preparation and material characterization are described. Second, the experimental procedures are described. Third, the experimental results are presented. Fourth, results of fractographic and metallographic examinations of tested specimens are discussed.

SAMPLE PREPARATION AND MATERIAL CHARACTERIZATION

The material used in this study was a 0.61-m by 0.61-m by 12.7-mm thick piece of solution-annealed Type 316 stainless steel, supplied by Oak Ridge National Laboratory (ORNL). This piece was taken from Plate 2AB (1.22m by 2.44m), Lot 391 of Reference Heat 8092297. This plate was certified to meet the requirements of ASME Specification SA 240. This heat was procured by ORNL to use in experimental studies, and its processing history and properties have been extensively documented and characterized as reported by Beaver and Martin(79).

Small (about 38 by 22 mm) samples were cut from the plate for metallographic examination and chemical analysis. The measured chemical composition is listed in Table 1. It is in good agreement with
## TABLE 1
CHEMICAL COMPOSITION OF TYPE 316 STAINLESS PLATE USED IN PRESENT STUDY

<table>
<thead>
<tr>
<th>Element</th>
<th>ASME Specification</th>
<th>Certification Record</th>
<th>Check Analysis of ORNL(^{(a)})</th>
<th>Present Analysis</th>
</tr>
</thead>
<tbody>
<tr>
<td>Fe</td>
<td>Balance</td>
<td>Balance</td>
<td>Balance</td>
<td>63.53</td>
</tr>
<tr>
<td>C</td>
<td>0.08(^{(b)})</td>
<td>0.057</td>
<td>0.060</td>
<td>0.059</td>
</tr>
<tr>
<td>Mn</td>
<td>2.00(^{(b)})</td>
<td>1.86</td>
<td>1.88</td>
<td>2.15</td>
</tr>
<tr>
<td>P</td>
<td>0.045(^{(b)})</td>
<td>0.024</td>
<td>0.024</td>
<td>0.020</td>
</tr>
<tr>
<td>S</td>
<td>0.030(^{(b)})</td>
<td>0.019</td>
<td>0.020</td>
<td>0.021</td>
</tr>
<tr>
<td>Si</td>
<td>1.00(^{(b)})</td>
<td>0.58</td>
<td>0.61</td>
<td>0.58</td>
</tr>
<tr>
<td>Cr</td>
<td>16.00-18.00</td>
<td>17.25</td>
<td>17.30</td>
<td>17.2</td>
</tr>
<tr>
<td>Ni</td>
<td>10.00-14.00</td>
<td>13.48</td>
<td>13.4</td>
<td>13.8</td>
</tr>
<tr>
<td>Mo</td>
<td>2.00-3.00</td>
<td>2.34</td>
<td>2.34</td>
<td>2.45</td>
</tr>
<tr>
<td>N</td>
<td>0.10(^{(b)})</td>
<td>0.030</td>
<td>0.032</td>
<td>-</td>
</tr>
<tr>
<td>Cu</td>
<td>-</td>
<td>0.10</td>
<td>0.06</td>
<td>0.08</td>
</tr>
<tr>
<td>Ti</td>
<td>-</td>
<td>0.02</td>
<td>&lt;0.01</td>
<td>0.001</td>
</tr>
<tr>
<td>Co</td>
<td>-</td>
<td>0.02</td>
<td>0.02</td>
<td>0.04</td>
</tr>
<tr>
<td>B</td>
<td>-</td>
<td>0.0005</td>
<td>0.0005</td>
<td>0.0002</td>
</tr>
<tr>
<td>Nb+Ta</td>
<td>-</td>
<td>0.00</td>
<td>&lt;0.01</td>
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<td>Sn</td>
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</tr>
<tr>
<td>Al</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>0.009</td>
</tr>
<tr>
<td>V</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>0.03</td>
</tr>
<tr>
<td>Zr</td>
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<td>0.002</td>
</tr>
<tr>
<td>W</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>0.00</td>
</tr>
</tbody>
</table>

\(^{(a)}\) Values reported by Beaver and Martin\(^{(79)}\).

\(^{(b)}\) These are maximum specified values.
the compositions of the material certification records and the check analyses of Beaver and Martin(79).

The metallographic sample was given a laboratory solution anneal of 1/2 hour at 1066 C, followed by rapid quenching to room temperature. The hardness then was measured and found to have an average value of 71 Rockwell B. Next, the metallographic sample was mounted, polished, and etched with 97HCl-3HNO₃-1/2g CuCl₂ solution. The microstructure was examined optically and found to be typical of that of such austenitic steels as illustrated by Figure 6. Using the comparison method of ASTM Designation: E112, both the longitudinal and transverse sections were found to have an ASTM micro-grain size number of 3. The chemistry, hardness, and microstructure were in good agreement with the information reported by ORNL(79).

Four types of specimens were machined from the plate — (1) tensile creep, (2) low-cycle fatigue (LCF), (3) compact type (CT), and (4) center-cracked tension (CCT). The configurations of these specimens are shown in Figures 7, 8, 9, and 10, respectively. All specimens were oriented such that the loading axis was parallel to the final rolling direction of the plate. Thus, the direction of crack propagation was perpendicular to the final rolling direction.

After completion of all machining operations, the specimens were given a laboratory solution annealing heat treatment. First, they were heated to 1066 C and held at temperature for 0.5 hour, using an argon gas environment to protect the specimen surface from oxidation. After the period at 1066 C, the specimens were rapidly quenched (at a rate greater than 82 C per minute) to room temperature.
FIGURE 6. MICROSTRUCTURE OF AS-RECEIVED MATERIAL, 97HCl-3HNO₃-1/2g CuCl₂ ETCHANT
44° V-groove 0.381 deep on both shoulders for attaching extensometers

6.35 dia

3.175 r

\( \frac{1}{2} \) -13 UNC threads

12.7

3.175

3.175

31.75

15.9

76.2

FIGURE 7. CONFIGURATION OF CREEP SPECIMEN,
ALL DIMENSIONS ARE IN mm
* Average (rms) roughness of indicated surface

a. Overall Configuration of Notched and Unnotched Specimens

b. Detail A-A for Notched Specimens

FIGURE 8. CONFIGURATION OF LOW-CYCLE FATIGUE SPECIMEN, ALL DIMENSIONS ARE IN mm
6-32 UNC threads to 4.76 depth, 2 places

Location for potential output leads, 2 places

* Average (rms) roughness of indicated surface

FIGURE 9. CONFIGURATION OF COMPACT-TYPE SPECIMEN, ALL DIMENSIONS ARE IN mm
Location for potential output leads, 2 places.

6-32 UNC threads, 2 places.

Sawcut plus EDM notch with ≤ 0.254 tip radius.

6.35 dia, 12.7 dia

114

25.4

4.78

22.2

30.5

22.9

53.3

0.4 \mu m^*

6.35 r

6.35

12.7

0.8 \mu m^*

* Average (rms) roughness of indicated surface

FIGURE 10. CONFIGURATION OF CENTER-CRACKED-TENSION SPECIMEN, ALL DIMENSIONS ARE IN mm
Four creep-rupture experiments were performed as summarized in Table 2. The temperatures and stress levels were selected for comparison with past work on this same heat of material reported by Sikka et al(80).

<table>
<thead>
<tr>
<th>Specimen Number</th>
<th>Temperature, °C (°F)</th>
<th>Stress, MPa (ksi)</th>
<th>Time to Rupture, hr</th>
<th>Minimum Creep Rate, percent/hr</th>
<th>Elongation, percent</th>
<th>Reduction of Area, percent</th>
</tr>
</thead>
<tbody>
<tr>
<td>T4</td>
<td>593 (1100)</td>
<td>276 (40)</td>
<td>66.8</td>
<td>5.0 x 10^-2</td>
<td>14.0</td>
<td>18.3</td>
</tr>
<tr>
<td>T1</td>
<td>593 (1100)</td>
<td>241 (35)</td>
<td>267.2</td>
<td>9.9 x 10^-3</td>
<td>13.1</td>
<td>19.3</td>
</tr>
<tr>
<td>T3</td>
<td>649 (1200)</td>
<td>207 (30)</td>
<td>78.1</td>
<td>1.5 x 10^-1</td>
<td>52.0</td>
<td>45.3</td>
</tr>
<tr>
<td>T2</td>
<td>649 (1200)</td>
<td>172 (25)</td>
<td>293.9</td>
<td>3.7 x 10^-2</td>
<td>41.4</td>
<td>40.3</td>
</tr>
</tbody>
</table>

The stress-rupture data for this heat of material are shown in Figure 11. The four circular symbols represent the present results, which are in good agreement with those of Sikka et al(80). The solid curves represent an average prediction based upon ultimate tensile (UTS) at temperature. These fall above the data points, especially at 593 C. The dashed curves give a good approximation of the average creep-rupture strength of the material used in this study.

Figure 12 shows the minimum or steady-state-creep-rate data for this heat of Type 316 stainless steel. Again, the present results (four circular symbols) are in good agreement with the other available data. The slope of the lines in Figure 12 give the power-law creep exponent, which is 11.2 at 593 C and 8.96 at 649 C. These are the values of n that were used in computing estimates of C*.
FIGURE 11. STRESS-RUPTURE OF TYPE 316 STAINLESS STEEL FROM REFERENCE HEAT 8092297

Predicted by UTS model of Sikka et al
Adjusted to data from present study

13-mm plate, reannealed, present study
16 to 51-mm plate, reannealed
25 to 64-mm bar, reannealed
10 to 13-mm pipe, reannealed
10 to 13-mm plate, mill-annealed
16-mm plate, mill-annealed
FIGURE 12. MINIMUM CREEP RATE FOR TYPE 316 STAINLESS STEEL FROM REFERENCE HEAT 8092297
Experiments on LCF and CT specimens were performed using a closed-loop electrohydraulic system, while those on CCT specimens were conducted using a constant-load, lever-arm creep unit. The CT and CCT specimens were heated to elevated temperature using a standard laboratory resistance heating element furnace. The LCF specimens were heated to temperature using an induction heating coil. Temperature was maintained within ±2°C of the nominal test temperature. A standard strain-gaged load cell was used to measure the load applied to the LCF and CT specimens by the hydraulic actuator. The CCT specimens were loaded by applying weights to the creep unit.

Load-line displacement was measured on the CT and CCT specimens using a high-temperature extensometer equipped with a linear variable differential transformer (LVDT). The arms of this extensometer were attached to the loading pins. Axial strain was measured over a 12.7-mm gage length on the LCF specimens using high-temperature extensometer equipped with a LVDT.

During creep periods on the CT specimens, the load-line-displacement rate was controlled to a desired constant value. Fatigue cycling of the CT specimens was performed under constant amplitude load control at a stress ratio (ratio of minimum to maximum stress) of $R = 0.05$ and at a frequency of about 1 Hz (unless otherwise noted). The CCT specimens were tested only under constant-load conditions using a creep unit. The LCF specimens were cycled under a fully-reversed waveform at a constant strain rate of either $4 \times 10^{-3}$ or
3.33 \times 10^{-4} \, \text{sec}^{-1}. \, A \, \text{hold period of} \, 0, \, 0.25, \, \text{or} \, 1.0 \, \text{hour was incorporated at the peak tensile strain during each cycle.} \, \text{The hold-time length was the same for every cycle of a given experiment. When the hold-time is} \, 0, \, \text{the aforementioned strain rates correspond to cyclic frequencies of} \, 0.20 \, \text{and} \, 0.0167 \, \text{Hz, respectively.}

\textbf{Crack-Growth Measurement}

Methods commonly used to measure crack growth include direct optical measurement, use of a gage applied to the surface ahead of the crack, measurement of crack-mouth opening displacement (CMOD), and measurement of electric potential drop. In creep-crack-growth experiments, the specimen is enclosed in a furnace, loaded for a long period of time (100 to 1000 hours or more), and often subjected to inelastic deformation in the net section ahead of the growing crack. Optical measurements are difficult, time-consuming, tedious, and costly to perform under such conditions. At temperatures above 300 C, the bonding of gages is difficult and unreliable. Additionally, optical and gage methods measure only the surface-crack length. CMOD and electric potential drop measure average changes in crack length, but CMOD is difficult to calibrate for conditions where net-section inelastic deformation exists. Thus, electric potential drop was chosen as the best overall method for this study.
Use of DC Potential Drop With CT and CCT Specimens

A number of investigators have discussed the application of electric potential drop to measurement of crack growth \(^{(81-102)}\). For creep-crack growth, the DC potential-drop method gave good results \(^{(52, 84, 91, 99)}\), so it was chosen for use in this work. Because of the simplicity of the experimental setup, the apparatus used was patterned after that employed by Vosikovsky \(^{(86)}\). The data reduction and calibration procedures were adopted from those reported by Saxena \(^{(91)}\).

A constant DC current of 10 amps was supplied to the CT specimens and 20 amps to the CCT specimens using a stabilized power supply and 2.0-mm diameter nickel-chrome alloy lead wires attached directly to the specimens. The potential change was monitored using a commercial nanovoltmeter with 0.5-mm diameter nickel-chrome alloy wires spot welded to the specimen near the crack. This system was capable of detecting crack-size changes with a sensitivity of 25 \(\mu\text{m}\) at an accuracy \(\pm 50 \mu\text{m}\). The change in DC potential was continuously monitored and recorded on a standard laboratory strip-chart recorder.

For CT specimens, the potential-drop measurements were calibrated using fractographic "bench marks" at various points on the specimen fracture from changes in the loading history. The crack-front profiles were always well-defined on the fracture surface. The average crack length was determined from a seven-point average of the length across the profile. For CCT specimens, calibration was achieved by periodically interrupting the experiment, cooling the specimen to room temperature, and optically measuring the surface-crack length using a
traversing microscope.

The change in crack length versus potential drop calibration data are presented in Figures 13 and 14 for the CT and CCT specimens, respectively. For the CT specimens, the normalized crack length change is 
\[ \frac{\Delta a}{a_0 + 0.25 w} \]
where \( \Delta a \) is the change in crack length, \( a_0 \) is the initial crack length, and \( w \) is the specimen width\(^{(91)} \). For the CCT specimen, \( \frac{\Delta a}{a_0} \) is the normalized change in crack length. In both cases, \( \frac{\Delta V}{V_0} \) is the normalized change in DC potential, where \( \Delta V \) is the potential change that is continuously measured throughout each test and \( V_0 \) is the initial reference potential, given by

\[ V_0 = V_T - V_{th} \] \( (48) \)

The initial potential is \( V_T \) and the thermal emf is \( V_{th} \), which are measured at the start of a test.

As shown in Figures 13 and 14, the results are quite consistent and in good agreement with Saxena's\(^{(91)} \) data on Type 304 stainless steel. For \( \frac{\Delta V}{V_0} \) less than 0.4, the analytical solution of Johnson\(^{(81)} \) agrees well with the data for CCT specimens. In one case (Specimen CT-5 in Figure 13), an error of undetermined origin was observed in the later stages of the experiment. The potential change measurements still could be used to interpolate between the measured values of crack advance. In all other cases, the final crack length for which data were obtained had a \( \frac{\Delta V}{V_0} \) value falling within the band of calibration curve data.

In computing crack length, the calibration for each individual specimen was used rather than the average calibration curve. Thus,
the electric-potential measurements provide a calibrated means of interpolating between the measured amounts of crack advance. This approach reduces errors that may be introduced by specimen-to-specimen variations. It is consistent with the recommendation of Wilson (102). His finite-element analyses of electric-potential fields in CT-specimen configurations showed that the parameter \( (\Delta a/a_o + 0.25w) \) does not truly normalize the change in crack length.
FIGURE 14. CRACK-LENGTH CHANGE VERSUS POTENTIAL CHANGE FOR CENTER-CRACKED-TENSION (CCT) SPECIMENS OF TYPE 316 STAINLESS STEEL.
Procedures for LCF Specimens

For LCF specimens, electric-potential drop was not used because of the electrical noise produced by the induction heating system used to heat the specimens. The smooth LCF specimen had no crack initially. As in the early work of Jaske and Begley (47), it was assumed that a crack equal to about 1-1/2 to 2 grain diameters must be developed before the crack growth can be described using the approach described earlier. This assumption is consistent with the observations of Maiya (103, 104) who found that initiation of a consistently definable crack in Type 304 stainless steel corresponded to the number of cycles to develop a crack 1 to 2 grain diameters deep.

The cyclic stress amplitude history was used to estimate when crack initiation occurred in the smooth LCF specimens. Figure 15 illustrates such a history. Initial cyclic hardening is followed by a stable stress amplitude until crack initiation occurs at $N_0$ cycles. The value of $N_0$ is determined from such a data record and is estimated to correspond to initiation of a 1-1/2-grain-diameter crack. If $\sigma_a$ represents the stable tensile stress amplitude, then $N_5$ is the number of cycles where the stress amplitude decreases to a value 5 percent lower than $\sigma_a$ and is estimated to correspond to a crack depth of about 1.02 mm (47). Thus, $N_5 - N_0$ represents a period of crack growth where the approach of this study should apply.

The notched LCF specimens had a surface flaw that was 0.178-mm deep (except one specimen where it was only 0.127-mm deep) with a 0.076-mm root radius (see Figure 8). The notches were produced by
electrodischarge machining. The notch depth was chosen to correspond to approximately 1-1/2 average grain diameters in this material. At the high strain amplitudes (0.25 percent or greater) typically used in low-cycle fatigue studies, cracking will initiate at this type of notch within several cycles. For this reason, it is reasonable to assume that $N_0 > 0$ for the notched LCF specimens.

Instead of estimating the final crack size from the stress amplitude history, the present LCF experiments usually were stopped before complete specimen failure so that the final crack size could be measured. During strain cycling, the tensile stress amplitude was continuously monitored. When it dropped to a value of about 10 percent below the stable stress amplitude, cycling was stopped. The specimen then was cooled to room temperature and cycled until fracture.
occurred. In this way the extent of the high-temperature cracking was clearly evident as an oxidized region on the fracture surface. The crack-fronts were approximately elliptical in shape. Changes in shape were accounted for using the elliptic flaw-shape parameter, as described earlier. Typically, the final crack length was about 1 to 2 mm.

The fatigue-crack-growth rate on these LCF specimens was measured after the experiment by fractographic examination of the area that was cracked at high temperature using a scanning electron microscope. Various regions of the high-temperature cracked areas were examined and striation-spacing measurements were made in these regions to determine the crack-growth rate in those areas. In cases where the mode of cracking was intergranular, no such measurements could be made and only the initial and final crack lengths were available.

EXPERIMENTAL RESULTS

The experimental results obtained in this study include fatigue-crack-growth data from long (≥ 27 mm) cracks in CT specimens and short (0.15 to 2 mm) cracks in LCF specimens, creep-crack-growth data for long (≥ 27 mm) cracks in CT and CCT specimens, and creep-fatigue-crack-growth interaction effects in CT and LCF specimens. Table 3 summarizes the control conditions used for tests of CT and CCT specimens, and Table 4 summarizes the control conditions and measured data for tests of LCF specimens. The total strain range, $\Delta \varepsilon_t$, was controlled, and the plastic strain range, $\Delta \varepsilon_{pp}$, and the creep reversed by plastic strain range, $\Delta \varepsilon_{cp}$, were measured from stress-strain hysteresis.
<table>
<thead>
<tr>
<th>Temperature, °C</th>
<th>Specimen Number</th>
<th>Type of Loading</th>
<th>Displacement Rate, m/sec</th>
<th>Maximum Load, N</th>
<th>Load Ratio (a)</th>
<th>Number of Cycles</th>
<th>Cyclic Frequency, Hz</th>
<th>Duration, hours</th>
</tr>
</thead>
<tbody>
<tr>
<td>649</td>
<td>CT4</td>
<td>Cyclic</td>
<td>-</td>
<td>7,342</td>
<td>0.05</td>
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<td>1</td>
<td>0.417</td>
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<tr>
<td></td>
<td></td>
<td>Cyclic</td>
<td>-</td>
<td>5,712</td>
<td>0.05</td>
<td>1,793</td>
<td>1</td>
<td>0.499</td>
</tr>
<tr>
<td></td>
<td></td>
<td>Cyclic</td>
<td>-</td>
<td>4,317</td>
<td>0.05</td>
<td>9,655</td>
<td>1</td>
<td>2.63</td>
</tr>
<tr>
<td></td>
<td>CT1</td>
<td>Cyclic</td>
<td>2.34 x 10^{-8}</td>
<td>8,541</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>19.8</td>
</tr>
<tr>
<td></td>
<td></td>
<td>LLDR(b)</td>
<td>2.35 x 10^{-9}</td>
<td>4,857</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>120.0</td>
</tr>
<tr>
<td></td>
<td></td>
<td>Cyclic</td>
<td>8.27 x 10^{-9}</td>
<td>4,502</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>100.0</td>
</tr>
<tr>
<td>593</td>
<td>CT3</td>
<td>Cyclic</td>
<td>9.24 x 10^{-9}</td>
<td>6,530</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>49.1</td>
</tr>
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<td></td>
<td></td>
<td>Cyclic</td>
<td>4,083</td>
<td>8,028</td>
<td>1</td>
<td>1,0 hr HT(c)</td>
<td>70.0</td>
<td>0.977</td>
</tr>
<tr>
<td></td>
<td></td>
<td>Cyclic</td>
<td>3,147</td>
<td>70</td>
<td>1.0 hr HT(c)</td>
<td>70.0</td>
<td>70.0</td>
<td>0.977</td>
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<td></td>
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<td>Cyclic</td>
<td>2,997</td>
<td>3,518</td>
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<td>207.1</td>
<td>207.1</td>
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<tr>
<td></td>
<td></td>
<td>Cyclic</td>
<td>3,090</td>
<td>207</td>
<td>1.0 hr HT(c)</td>
<td>207.1</td>
<td>207.1</td>
<td>207.1</td>
</tr>
<tr>
<td></td>
<td>CT5</td>
<td>Cyclic</td>
<td>5,338</td>
<td>12,570</td>
<td>1</td>
<td>3.49</td>
<td>3.49</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>Cyclic</td>
<td>5,338</td>
<td>1,654</td>
<td>0.2</td>
<td>2.30</td>
<td>2.30</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>Cyclic</td>
<td>7,866</td>
<td>928</td>
<td>0.2</td>
<td>1.29</td>
<td>1.29</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>LLDR(b)</td>
<td>3.46 x 10^{-9}</td>
<td>6,984</td>
<td>-</td>
<td>-</td>
<td>89.6</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>Cyclic</td>
<td>2,669</td>
<td>750</td>
<td>0.6</td>
<td>3.47</td>
<td>3.47</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>LLDR(b)</td>
<td>9.67 x 10^{-9}</td>
<td>4,849</td>
<td>-</td>
<td>-</td>
<td>67.7</td>
<td></td>
</tr>
<tr>
<td></td>
<td>CT6</td>
<td>Cyclic</td>
<td>5,061</td>
<td>505</td>
<td>1.0 hr HT(c)</td>
<td>505.1</td>
<td>505.1</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>Cyclic</td>
<td>2,903</td>
<td>22,600</td>
<td>0.78</td>
<td>7.98</td>
<td>7.98</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>Cyclic</td>
<td>3,488</td>
<td>37,600</td>
<td>0.78</td>
<td>13.4</td>
<td>13.4</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>Cyclic</td>
<td>4,167</td>
<td>36,312</td>
<td>0.78</td>
<td>12.9</td>
<td>12.9</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>Cyclic</td>
<td>4,682</td>
<td>1,810</td>
<td>0.25 hr HT(c)</td>
<td>453.1</td>
<td>453.1</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>Cyclic</td>
<td>2,503</td>
<td>77,600</td>
<td>0.81</td>
<td>26.5</td>
<td>26.5</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>Cyclic</td>
<td>3,488</td>
<td>7,800</td>
<td>1.08</td>
<td>2.01</td>
<td>2.01</td>
<td></td>
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<tr>
<td>649</td>
<td>CC1</td>
<td>Constant</td>
<td>25.29 x 10^{-10}</td>
<td>22,320</td>
<td>-</td>
<td>-</td>
<td>732.5</td>
<td></td>
</tr>
<tr>
<td>593</td>
<td>CC6</td>
<td>Constant</td>
<td>25.29 x 10^{-9}</td>
<td>34,120</td>
<td>-</td>
<td>-</td>
<td>23.6</td>
<td></td>
</tr>
<tr>
<td></td>
<td>CC4</td>
<td>Constant</td>
<td>21.58 x 10^{-9}</td>
<td>27,660</td>
<td>-</td>
<td>-</td>
<td>50.1</td>
<td></td>
</tr>
<tr>
<td></td>
<td>CC3</td>
<td>Constant</td>
<td>21.01 x 10^{-11}</td>
<td>22,600</td>
<td>-</td>
<td>-</td>
<td>&gt;3000</td>
<td></td>
</tr>
</tbody>
</table>

(a) Ratio of minimum to maximum load.
(b) LLDR means control of load-line displacement rate.
(c) Tension hold time (HT) of indicated length was introduced in each cycle with the frequency of the remaining portion of the waveform the same as in the subsequent continuous cycling.
<table>
<thead>
<tr>
<th>Specimen Number</th>
<th>Specimen Type</th>
<th>Hold Time, hr</th>
<th>Strain Range, percent</th>
<th>Stress, MPa</th>
<th>Number of Cycles</th>
<th>Final Crack Length, mm</th>
<th>$\sigma_{Lt}/\sigma_t$</th>
</tr>
</thead>
<tbody>
<tr>
<td>F9</td>
<td>Smooth</td>
<td>0</td>
<td>1.00 0.64</td>
<td>284 291</td>
<td>1085</td>
<td>0.91</td>
<td>0.950</td>
</tr>
<tr>
<td>F10</td>
<td>Smooth</td>
<td>1.0</td>
<td>1.02 0.72 0.042</td>
<td>221 237</td>
<td>459</td>
<td>1.45</td>
<td>0.751</td>
</tr>
<tr>
<td>F2</td>
<td>Notched</td>
<td>0</td>
<td>1.02 0.68</td>
<td>278 283</td>
<td>369</td>
<td>1.92</td>
<td>0.903</td>
</tr>
<tr>
<td>F3</td>
<td>Notched</td>
<td>0.25</td>
<td>1.00 0.72 0.032</td>
<td>220 235</td>
<td>273</td>
<td>2.85</td>
<td>0.773</td>
</tr>
<tr>
<td>F1</td>
<td>Notched (a)</td>
<td>1.0</td>
<td>1.01 0.71 0.037</td>
<td>256 264</td>
<td>339</td>
<td>1.12</td>
<td>0.902</td>
</tr>
<tr>
<td>F7 (c)</td>
<td>Notched</td>
<td>0</td>
<td>1.00 0.69</td>
<td>259 263</td>
<td>249 (c)</td>
<td>2.03</td>
<td>0.896</td>
</tr>
</tbody>
</table>

At 593 C

<table>
<thead>
<tr>
<th>Specimen Number</th>
<th>Specimen Type</th>
<th>Hold Time, hr</th>
<th>Strain Range, percent</th>
<th>Stress, MPa</th>
<th>Number of Cycles</th>
<th>Final Crack Length, mm</th>
<th>$\sigma_{Lt}/\sigma_t$</th>
</tr>
</thead>
<tbody>
<tr>
<td>F8</td>
<td>Smooth</td>
<td>0</td>
<td>1.11 0.69</td>
<td>317 323</td>
<td>1225</td>
<td>1.09 (b)</td>
<td>0.921</td>
</tr>
<tr>
<td>F4 (c)</td>
<td>Notched</td>
<td>0</td>
<td>1.01 0.61</td>
<td>320 338</td>
<td>574</td>
<td>1.72</td>
<td>0.900</td>
</tr>
<tr>
<td>F5 (c)</td>
<td>Notched</td>
<td>0</td>
<td>1.01 0.61</td>
<td>334 334</td>
<td>312 (c)</td>
<td>1.70 (d)</td>
<td>0.900</td>
</tr>
<tr>
<td>F6</td>
<td>Notched</td>
<td>0.25</td>
<td>0.98 0.59 0.021</td>
<td>295 309</td>
<td>125</td>
<td>1.70 (d)</td>
<td>0.900</td>
</tr>
</tbody>
</table>

(a) Notch was 0.127 mm deep instead of 0.178 mm deep.
(b) A second crack 0.46 mm deep was also present.
(c) Frequency was 0.0167 Hz instead of 0.20 Hz.
(d) Estimated crack length at $\sigma_{Lt}/\sigma_t = 0.90$ because specimen cycled to complete failure.
recordings as illustrated in Figure 2 (page 21). The results obtained from the experiments summarized in Tables 3 and 4 are explained in the ensuing subsections.

**Fatigue-Crack Propagation**

The fatigue-crack-propagation data obtained from experiments on the CT specimens are summarized in Figure 16 in terms of a logarithmic plot of cyclic crack growth rate, $\frac{da}{dN}$, versus the range of stress intensity factor, $\Delta K$. All of the plotted symbols represent experimental data obtained during the present study. The open symbols are based upon DC potential measurements of the crack advance. The solid symbols are based upon fractographic examination and measurements of striation spacings and are in good agreement with the macroscopic measurements. The half-filled symbols represent maximum values of calculated cyclic crack-growth rate after a period of creep based upon integration of the appropriate creep-crack-growth rate, $\frac{da}{dt}$, versus $C^*$ relationship. These calculated values are somewhat higher but in reasonably good agreement with the cyclic crack-growth-rate data for tests where hold times were present. The open triangular symbols in Figure 16 represent data from experiments with hold times at maximum tensile load during each cycle. The numbers beside the data symbols represent the length of hold time in hours.

The solid lines in Figure 16 represent best-fit curves to the present data for continuous cycling at approximately 1 Hz at 593 and 649 °C. The dashed lines represent data developed by workers at the Naval Research Laboratory (105,106). Those data at 593 °C are for
continuous cycling at 0.167 Hz and \( R = 0 \) while those at 649 C are for tests with one-minute hold times at maximum load and \( R = 0 \). In the latter case, the author stated that the one-minute hold times did not significantly increase the crack-growth rate above that for continuous cycling, but no data were reported for continuous cycling.

FIGURE 16. FATIGUE-CRACK-PROPAGATION BEHAVIOR OF TYPE 316 STAINLESS STEEL IN AIR AT 593 AND 649 C IN TERMS OF THE ELASTIC STRESS INTENSITY FACTOR
In both cases, the results are in very good agreement with the results of the present study at crack-growth rates above $10^{-7}$ m/cycle.

For hold times of 0.1 to 1.0 hours at 649 C and 1 hour at 593 C, the best fit to the crack-growth-rate data is represented by a chain-dashed line in Figure 16. For hold times of 0.25 hours at 593 C, the cyclic crack-growth rate fell between the data for continuous cycling and those data for hold times of 1 hour. The computed values of cyclic crack-growth rate after a period of creep fall along a similar trend as the hold-time data, with slightly higher crack-growth rates. The cyclic crack-growth rates for the hold-times cases were about 10 times greater than those for continuous cycling under comparable conditions.

The results and calculations indicate that very little of the acceleration in crack-growth rate occurs during the actual hold period at maximum load for relatively short holds. The major reason for the increase in cyclic crack-growth rate is an increase in the rate of crack advance during the loading portion of the cycle. The major role of the hold period is to allow time for a larger crack-tip process zone to develop under creep conditions. Then, during the loading cycle following the hold period, the crack can advance more rapidly through the process zone of damaged material. For this reason, a slower rate of loading after the hold period will produce a higher cyclic crack-growth rate. Thus, the half-filled diamond-shaped point in Figure 16 falls well above the other half-filled data points. In the case of the half-filled diamond point, the loading rate corresponded to a cyclic frequency of 0.06 Hz, whereas in the other
cases the loading rate corresponded to a cyclic frequency of about 1.0 Hz. This trend in behavior is also in agreement with that reported by James (10); he reported that the cyclic-crack-growth rate for a sawtooth waveform was more rapid than that for a test with a hold period of the same overall waveform frequency. James' work was carried out on Type 304 stainless steel.

For cases where $\Delta K$ is no longer the valid parameter to characterize the fatigue-crack-growth rate, values of the range of the $J$ integral, $\Delta J$, was used to characterize the fatigue-crack-growth rate. Figure 17 is a plot of $da/dN$ versus $\Delta J$ for Type 316 stainless steel tested in air. All of the symbols in Figure 17 represent data for fully reversed strain controlled fatigue cycling of low-cycle fatigue specimens with small surface flaws. The open circles represent data reporting by Wareing (108) for Type 316 stainless steel specimens with initial crack lengths from 0.05 to 0.88 mm tested in air at 625 C. Wareing did not report $J$ values in his work. These were computed in the present study using average cyclic stress-strain properties for Type 316 stainless steel (64). The other open symbols represent data for similar low-cycle fatigue experiments conducted at 649 C in the present study. There is very good agreement between the two sets of data and reasonably good agreement with the other curves shown in the figure. The solid symbols represent data from experiments conducted at 593 C in the present study. In this case, the crack-growth rate was slightly below that observed at 649 C and close to that at room temperature (half-filled symbols). The points at room temperature are
in good agreement with the room-temperature data reported by Shahinian et al\(^{105}\).

For other data, only trend lines are shown in Figure 17. Where data were taken from cases where $\Delta K$ was the valid characterization parameter, the information is plotted as $(\Delta K)^2$ divided by the elastic modulus, $E$. The chain-dashed lines indicate the scatter band of data reported by Taira et al\(^{49}\) for low-cycle fatigue-crack growth of Type 316 stainless steel in air at 650°C. The solid line falling approximately parallel and between these two chain-dashed lines
represents an estimate of the $\Delta J$ versus $da/dN$ relationship made earlier by Jaske and Begley\textsuperscript{(47)}. This estimate is in good agreement with the data reported by Taira et al. The lowest dashed line in Figure 17 represents the fatigue-crack-growth behavior of Type 316 stainless steel at room temperature as reported by Shahinian et al\textsuperscript{(105)}. The other lines represent data and curves taken from Figure 16.

The results shown in Figure 17 indicate that $\Delta J$ is a reasonable inelastic fracture mechanics parameter for treating the low-cycle fatigue-crack growth of Type 316 stainless steel at elevated temperatures. At cyclic crack-growth rates near $2 \times 10^{-6}$ m/cycle and above, the data for the specimens with small cracks (symbols) fell below those for specimens with larger cracks and were close to the crack-growth behavior at room temperature. This indicates that for the small cracks at elevated temperature the influence of environment in accelerating the crack-growth rate is small because there is not sufficient time for environmental attack at these higher growth rates. At lower growth rates (i.e., below $10^{-6}$ m/cycle), the crack-growth rates for the smaller cracks were generally in agreement with those for the specimens with larger cracks. For the small cracks, no crack-growth rate data were obtained for cyclic rates much lower than $10^{-7}$ m/cycle. In this low crack-growth-rate regime, it is not known if the short cracks will grow as slowly as the longer ones did at 593 C where the slope of the crack-growth-rate curve becomes quite steep, approaching a threshold at very low crack-growth rates.

Overall, the $\Delta J$ parameter appears to be a reasonable one for describing low-cycle fatigue-crack-growth rate under inelastic cycling
conditions at crack-growth rates above $10^{-7}$ m/cycle. The major question in dealing with short cracks is whether or not the cyclic crack-growth-rate behavior at low $\Delta K$ or low-crack-growth rates below $10^{-7}$ m/cycle will be similar to that for long cracks. Other studies of short-crack behavior\(^{(46,109)}\) have found that short cracks grow at a much higher rate than long cracks at low levels of $\Delta K$. In that other work, the fatigue-crack-growth rate near the threshold for small cracks has been modeled by adding a material constant term to the actual physical crack length. It was not found necessary to use such a material constant in the present study, even for initial crack lengths as small as 0.05 mm.

**Creep-Crack Growth**

Creep-crack-growth data obtained in the present study are shown in Figure 18 where the time-dependent crack-growth rate, $da/dt$, is plotted versus the $C^*$ integral. These data were obtained from both CT and CCT specimens as indicated, and the values of $C^*$ were computed using the estimation procedures described earlier. In all cases, the value of $r_1$ (Equation 6) was of the order of the crack length and uncracked ligament size while the value of $r_2$ (Equation 7) was much smaller than the crack length and uncracked ligament size. Also, the amount of crack advance was smaller than the crack length and uncracked ligament size. Thus, $C^*$ is the applicable parameter. All of the symbols, except the triangles, represent data from the present study, with the circles being data obtained at 649 °C and the squares and diamonds representing data obtained at 593 °C.
FIGURE 18. CREEP-CRACK GROWTH-RATE BEHAVIOR FOR TYPE 316 STAINLESS STEEL
The open circles and diamonds are for data from CT specimens, while the filled circles, diamonds, and squares are for data from CCT specimens. The chain-dashed lines present the scatter band of data for Type 316 stainless steel at 600 and 650°C as reported by Taira et al. The scatter band shown by solid lines represents data of Saxena for Type 304 stainless steel at 593°C, and the solid triangle symbols represent similar data reported by Saxena at 649°C. The data developed by Taira et al are for very high crack-growth rates of greater than $10^{-8}$ m/sec. When the current results are extrapolated to lower values of $C^*$, the two sets of data are in fairly good agreement. Those data also agree with Saxena's except when his results at 593°C are extrapolated to values of $C^*$ above 2 watts/m², they fall at lower crack-growth rates.

The data for 593°C (squares and diamonds) fall slightly above those for 649°C (circles), indicating a small influence of test temperature. Considering the typical scatter observed in crack-growth data, the data for CCT specimens fell close to those for CT specimens under comparable conditions. For $C^*$ to be a useful parameter, those results should be independent of specimen configuration.

For most engineering problems, crack-growth rates of $10^{-9}$ m/sec or less are of practical interest. This is the regime where little data presently are available. For values of $C^*$ near or less than 1 watt/m², all three sets of data are in fairly good agreement. By performing a constant-load experiment in a standard creep unit for a period of more than 3000 hours, it was shown that creep-crack-growth rates approaching $10^{-11}$ m/sec or 0.315 mm/year could be measured.
In contrast, $10^{-8}$ m/sec is equal to 0.315 m/year which is a very high crack-growth rate for most engineering applications.

The data near 0.5 watts/m$^2$ at 649 C (open circular symbols) fell at a slower crack-growth rate than expected. The reason for that behavior may be that the specimen was previously at temperature for a considerable period of time during hold-time fatigue-crack-growth-rate testing. Such aging at 649 C is known to produce significant precipitation of $\text{M}_{23}\text{C}_6$ carbides in this material, which could reduce the rate of subsequent creep-crack growth. At the present time, however, this explanation is speculative, and further experimental work needs to be conducted on specimens that have not been aged and on specimens that have been aged before creep-crack-growth rate testing to see if there is a significant difference in their behavior at this temperature. It was concluded, that, in general, the $C^*$ parameter is useful for characterizing the creep-crack-growth behavior of this material under the conditions investigated.

**FRACTOGRAPHIC APPEARANCE OF FAILED SPECIMENS**

The fracture surfaces of the failed specimens showed evidence of two basic modes of cracking - (1) intergranular cracking associated with creep- or time-dependent-crack growth, and (2) transgranular cracking associated with fatigue- or cycle-dependent-crack growth. In a few cases, there were mixtures of the two modes of crack growth, but in most cases the mode of crack growth was predominantly one or the other.
Figure 19 shows the typical appearance of the fracture surface of an LCF specimen (F2) after completion of the experiment. The starter notch is shown at the bottom of the figure. The darker semicircular area represents the oxidized region, marking the extent of crack growth that occurred at 649°C. The lighter region resulted from room-temperature fatigue-crack growth, with the shear-lip portion (top of figure) being the area of final fracture.

![Figure 19. Appearance of Fracture Surface of LCF Specimen F2](image)

Table 5 summarizes observations made of the fracture mode for the LCF specimens using scanning electron microscopy. A typical region of low-cycle fatigue-crack growth at elevated temperature is shown in Figure 20. Ductile striations are clearly evident. Crack-growth rates were determined for the fatigue regions of these low-cycle
TABLE 5
SUMMARY OF TYPE OF FRACTURE MODES OBSERVED DURING EXAMINATION OF LCF SPECIMEN USING SCANNING ELECTRON MICROSCOPY

<table>
<thead>
<tr>
<th>Specimen Number</th>
<th>Observations (a)</th>
</tr>
</thead>
<tbody>
<tr>
<td>F9</td>
<td>All transgranular with ductile striations</td>
</tr>
<tr>
<td>F10(b)</td>
<td>All intergranular, even at room temperature</td>
</tr>
<tr>
<td>F2</td>
<td>All transgranular with ductile striations</td>
</tr>
<tr>
<td>F3(b)</td>
<td>Mixture of transgranular and intergranular at 649 C, transgranular with ductile striations at room temperature</td>
</tr>
<tr>
<td>F1(b)</td>
<td>All transgranular with ductile striations</td>
</tr>
<tr>
<td>F7(b)</td>
<td>Mixture of transgranular and intergranular at 649 C, transgranular with ductile striations at room temperature</td>
</tr>
<tr>
<td>F8</td>
<td>All transgranular with ductile striations</td>
</tr>
<tr>
<td>F4</td>
<td>All transgranular with ductile striations</td>
</tr>
<tr>
<td>F5</td>
<td>All transgranular with ductile striations</td>
</tr>
<tr>
<td>F6(b)</td>
<td>Mostly intergranular at 593 C, transgranular with ductile striations at room temperature</td>
</tr>
</tbody>
</table>

(a) Excluding final shear failure.
(b) Evidence of significant intergranular cracking on outer surface of test section.

Fatigue specimens using measurements of such striation spacing, as mentioned previously. A typical region of intergranular crack-growth (Specimen F10) is shown in Figure 21. Finally, Figure 22 shows intergranular cracking that was observed on the surface of Specimen F10.

The transgranular fatigue cracking in Specimen F1 was unexpected because it was cycled with a 1-hour tensile hold time. The crack growth was transgranular as shown in Figure 23, even though the
FIGURE 20. TYPICAL APPEARANCE OF DUCTILE FATIGUE STRIATIONS OBSERVED ON FRACTURE SURFACE OF LCF SPECIMENS, SPECIMEN F4 IN AIR AT 593 C

FIGURE 21. TYPICAL APPEARANCE OF INTERGRANULAR CRACK GROWTH, SPECIMEN F10
FIGURE 22. APPEARANCE OF INTERGRANULAR CRACKING ON SURFACE OF SPECIMEN F10

FIGURE 23. TYPICAL APPEARANCE OF CRACK PROPAGATION AT 649 C IN SPECIMEN F1
surface suffered significant intergranular cracking as shown in Figure 24. Because notched specimens with shorter hold times (F3 and F6) and a smooth specimen with the same length of hold time (F10) all exhibited mixed or intergranular crack growth, this observed behavior was probably caused by dynamic aging of material in the process zone at the crack tip. If intergranular damage in the bulk material ahead of the crack becomes extensive enough, even the crack growth at room temperature will be intergranular as was found for Specimen F10 and shown in Figure 25.

Figure 26 shows the fracture surface of one of the compact-type specimens (CT-2) after testing at 649 C. Initially, the specimen was precracked from the chevron-shaped starter notch at room temperature.
Then, there was a region of intergranular time-dependent crack growth during hold-time cycling at elevated temperature. This region was followed by a period of fatigue-crack growth, which was, in turn, followed by a period of creep-crack growth where some degree of tunneling was noted. After the creep-crack-growth period, another region of fatigue cracking was evident. Finally, there was a small band of intergranular creep cracking associated with crack growth during a one-hour hold-time test. The final light-appearing area of the fracture represents fatigue cracking at room temperature at the end of the test.
Fatigue at room temperature

One hour hold

Fatigue

Creep

Fatigue

0.25 hour hold

Precrack

Direction of crack growth

FIGURE 26. APPEARANCE OF FRACTURE SURFACE OF TYPICAL COMPACT TYPE SPECIMENS TESTED UNDER ALTERNATING PERIODS OF CREEP-FATIGUE LOADING AT ELEVATED TEMPERATURE
Typically, the creep-crack-growth region showed an intergranular-type fracture surface as illustrated in Figure 27a and 27b. Figure 27a shows the creep-crack fracture surface at low magnification. Region (a) is magnified in Figure 27b. Figures 28a and 28b show a region of transition from transgranular fatigue cracking to the intergranular creep cracking. Figure 28a shows this region at low magnification. Region (b) is shown at higher magnification in Figure 28b. The transition from transgranular to intergranular cracking was clearly evident. In examining the specimens in the scanning electron microscope, it was easy to follow along this boundary from one cracking mode to the other.

The observed modes of fatigue and creep cracking are consistent with the assumptions made in earlier modeling of the crack-tip-zone fracture process. These types of fracture modes are typical of those usually observed in high-temperature creep-fatigue-crack growth in austenitic steels. However, these are not necessarily representative of the types of crack-growth mechanisms observed in ferritic steels. Thus, the model of the crack-tip-interaction damage process developed in the current study is believed to be valid for the austenitic alloys, but no supporting evidence of its validity for the ferritic steels has been obtained to date.
(a) Overall Appearance at Low Magnification

(b) Appearance of Region (a) at Higher Magnification

FIGURE 27. APPEARANCE OF REGION OF CREEP-Crack
PROPAGATION IN SPECIMEN CT-1
FIGURE 28. APPEARANCE OF REGION OF TRANSITION FROM TRANSGRANULAR FATIGUE-CRACK PROPAGATION TO INTERGRANULAR CREEP-CRACK PROPAGATION IN SPECIMEN CT-1.
V. DISCUSSION OF EXPERIMENTAL EVIDENCE

A model for creep-fatigue-crack growth was developed and evaluated in this study. The experimental evidence in support of this model consists of both indirect information from published literature and direct information from this study. In evaluating this model, it is useful to employ the first-order estimates of crack-tip-zone sizes – $r_{cf}$, $r_c$, and $r_f$ – developed earlier and expressed by Equations 42, 46 and 47.

It is instructive to compare the values of $r_{cf}$ and $r_f$ for cycling with tensile hold times at two frequently used stress ratios, $R = 0$ and $R = -1$. At $R = 0$, $\Delta K = K_{\text{max}}$ and $r_f$ is only 1/4 the size of $r_{cf}$. At $R = -1$, $\Delta K = 2K_{\text{max}}$ and $r_f$ is the same size as $r_{cf}$. A creep-fatigue interaction is expected in the former case but not in the latter one. These and several other cases of creep-fatigue interaction effects in crack growth are discussed subsequently.

CREEP FOLLOWED BY FATIGUE-CRACK GROWTH

James (112) conducted experiments on Type 304 stainless steel at 538 C where samples were subjected to creep strains up to half the expected values at rupture. Specimens then were machined from the creep-damaged material, notched, and precracked for fatigue-crack-growth testing. The fatigue-crack-growth-rate behavior at 538 C was
shown to be unaffected by this prior creep damage. This result is in agreement with the proposed model that assumes that subsequent fatigue-crack growth is not significantly influenced by prior creep damage in the bulk material. Because the samples were not precracked during creep, \( r_{cf} \) and \( r_c \gtrsim 0 \) and \( r_f > r_{cf} \) and \( r_c \). The volume fraction of creep cavities was not large enough to alter the fatigue cracking mechanism. When the volume fraction of creep cavities is large enough to alter fatigue cracking, the model is not valid. In the framework of the model, only material in the near-tip region of a creep crack should contain these large amounts of creep damage and exhibit significantly altered subsequent fatigue-crack growth rates.

**CYCLING WITH TENSILE HOLD PERIODS**

Numerous creep-fatigue studies have been conducted with tensile hold periods at either maximum load or maximum strain\(^{(1)}\). As illustrated in Figure 29, load-hold periods usually are employed in crack-growth studies with cycling at stress ratios of \( R > 0 \). And as shown in Figure 30, strain holds typically are employed in low-cycle fatigue studies at a stress ratio of \( R = -1 \). For austenitic steels, such as Types 304 and 316, at temperatures below 700 C, creep effects are much greater than environmental ones\(^{(113)}\). Thus, tensile hold periods cause quite marked creep-fatigue interactions manifested by greatly increased cyclic crack-growth rates.
FIGURE 29. ILLUSTRATION OF LOAD-TIME WAVEFORM USED IN TENSION HOLD-TIME CYCLING WITH STRESS RATIO OF $R = 0$

$$R = 0, \ k_{\text{max}} = \Delta K, \ r_f = 1/4 \ r_{\text{cf}}$$

FIGURE 30. ILLUSTRATION OF STRAIN-TIME WAVEFORM USED IN TENSION HOLD-TIME CYCLING WITH A STRESS RATIO OF $R = -1$

$$R = -1, \ k_{\text{max}} = 1/2 \ \Delta K, \ r_f = r_{\text{cf}}$$
Crack Growth at $R \approx 0$

For load cycling at $R \approx 0$ with a tensile hold period (Figure 29) longer than the incubation period required to develop a crack-tip zone, $K_{\text{max}}$ is approximately equal to $\Delta K$. This implies that $r_f = 1/4 \ r_c$ so the model predicts an acceleration of crack-growth rate up to that for time-dependent creep cracking during each loading cycle. This prediction is in agreement with many observations reported in the literature as reviewed by James and with the data developed in this study (Figure 16).

Crack Growth at $R = -1$

For strain cycling at $R = -1$ with tensile hold periods at peak tensile strain, $K_{\text{max}} = 1/2 \Delta K$, which implies that $r_{cf} = r_f$. In this case, no interaction between fatigue and creep cracking is predicted. That is, during the strain cycle, the crack should advance at the normal fatigue rate and during the hold period it should advance at the time-dependent rate.

Assuming no crack-tip-zone interaction and using the procedures described previously (pages 19-31), creep-fatigue-crack-growth calculations were carried out for low-cycle-fatigue specimens of the Type 316 stainless steel tested under strain controlled conditions with hold times at peak tensile strain (114). As illustrated in Figure 31, this approach gave reasonable predictions of the number of cycles of crack growth, $N_s - N_0$ (Figure 15). The predictions were good even when average-cyclic-stress-strain and stress-relaxation data (64) (filled
points) were employed rather than the data from individual tests (open points). Thus, the approach used in this study can be applied to the problem of creep-fatigue-crack growth in low-cycle-fatigue specimens under conditions where hold periods are introduced into each cycle at the peak tensile strain.

FIGURE 31. CORRELATION BETWEEN ACTUAL AND CALCULATED CYCLES OF CRACK GROWTH FOR PEAK-HOLD-TIME LOW-CYCLE-FATIGUE EXPERIMENTS ON TYPE 316 STAINLESS STEEL
Using the average-cyclic-stress-strain and stress-relaxation data, calculations of \( N_5 - N_0 \) as a function of total strain range, length of hold time, and temperature, by the aforementioned approach were made. The results are illustrated in Figure 32. Increasing temperature and increasing length of hold time increased the amount of calculated crack growth. However, at long hold times, the influence of strain range is greatly diminished.

![Figure 32. Calculated effect of hold period at peak-tensile strain on low-cycle-fatigue crack-growth behavior of Type 316 stainless steel.](image)

Similar calculations were made for the data of the present study using the notch depth or 0.127 mm for the unnotched specimen (F10) and using the measured value of the final crack size, as described earlier. The results are presented in Table 6. Calculated values of
TABLE 6

COMPARISON OF MEASURED VALUES WITH CALCULATED VALUES OF CRACK GROWTH IN LCF SPECIMENS OF TYPE 316 STAINLESS STEEL (HEAT 8092297)

<table>
<thead>
<tr>
<th>Specimen Number</th>
<th>Cycles of Crack Growth</th>
<th>Ratio of Measured to Calculated</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Measured</td>
<td>Calculated</td>
</tr>
<tr>
<td>F10</td>
<td>84</td>
<td>56</td>
</tr>
<tr>
<td>F3</td>
<td>273</td>
<td>121</td>
</tr>
<tr>
<td>F1</td>
<td>339</td>
<td>35</td>
</tr>
<tr>
<td>F6</td>
<td>125</td>
<td>147</td>
</tr>
<tr>
<td>F7</td>
<td>249</td>
<td>249</td>
</tr>
<tr>
<td>F5</td>
<td>312</td>
<td>213</td>
</tr>
</tbody>
</table>

The amount of crack growth are in reasonably good agreement with measured values, except for Specimens F3 and F1. For F1 crack growth was all transgranular, and for F3 a significant portion of crack growth was transgranular. In both cases, the calculations significantly underpredicted the actual number of cycles.

As briefly indicated previously, the longer than expected cyclic lives of Specimens F3 and F1 are believed to be the result of an interaction between the small notch and dynamic strain aging of the material. When no notch is present (F10), the damage initiates at grain boundaries and thus continues to propagate intergranularly. The small notch promotes early transgranular cracking. If the crack-tip process zone size is smaller than some critical measure of intergranular cavity spacing the crack will continue to grow transgranularly. The long cracks in CT specimens had much larger process zones than the
short cracks in LCF specimens for similar loading conditions so they grew intergranularly.

According to the model of Lloyd and Wareing\(^{115}\), the intergranular mode of crack growth will dominate when the process-zone size is of the order of magnitude of \( \lambda - p \), where \( \lambda \) is the cavity spacing and \( p \) is the cavity size. The cavity nucleation and growth process is associated with precipitation of \( M_23C_6 \) carbides in Type 316 stainless steel. As aging progresses, the carbides grow and coalesce so that \( \lambda - p \) tends to increase and intergranular growth is thus inhibited until the crack grows long enough so that the process-zone size exceeds the critical value. This model qualitatively explains that observed behavior. Quantification of this postulated explanation was beyond the scope of the current study. It is a subject worth investigating in future research.

**CRACK GROWTH UNDER CONTINUOUS CYCLING**

For rapid continuous cycling where low-cycle fatigue-crack propagation occurs by a ductile-striation mechanism, the current approach of using \( \Delta J \) is consistent with other past studies\(^{8,108,116}\). For example, Wareing\(^{108}\) showed that for strain-controlled LCF cycling

\[
\frac{da}{dN} = a, \tag{49}
\]

where the proportionality constant is a function of strain range. Equation (49) is actually a special case of Equation (10). Combining Equations (10) and (12), it follows that
\[ \frac{da}{dN} = C_1 \left( G_{1\text{e}} + G_{2\text{p}} \right)^\gamma a^\gamma. \]  

When \( \gamma \approx 1 \), as is often the case, and the crack-shape remains constant, Equation (50) is the same as Equation (49).

Under continuous cycling, the model predicts a frequency effect in agreement with experimental observations on austenitic steels\(^{(1)}\). At high cyclic frequencies, \( r_c \sim 0 \) because there is not sufficient time for a creep-crack-tip zone to develop. In this case, completely cycle-dependent crack growth is predicted. At intermediate frequencies, \( r_c \sim r_f \) and a mixture of time- and cycle-dependent crack growth is expected. Finally, at very low cyclic frequencies, \( r_c > r_f \) and completely time-dependent crack growth is anticipated. These predictions agree qualitatively with experimental results.

Specimens F7 and F5 were cycled at low frequencies where a significant contribution of time-dependent crack growth was expected. The cycles of crack growth (Table 6) were calculated by using an "equivalent tensile hold period", which was equal to 1/4 of the cycle period, \( t_{1/4} \), for these two specimens. As shown in Figure 33, this period is a good approximation of the time where tensile stress is near or above the short-time yield strength during each strain cycle. At the midpoint of \( t_{1/4} \), the stress is \( \sigma_e \) which was used as the equivalent peak stress in the calculations. As seen from Table 6, this method gives a reasonable prediction of the influence of decreased cyclic frequency on the number of cycles of crack growth.
FIGURE 33. STRESS ($\sigma$) - STRAIN ($\varepsilon$) RESPONSE FOR SPECIMEN F7 DURING CYCLE 114.
Effect of Tensile Mean Stress

For continuous, rapid fatigue cycling at stress ratios greater than zero and in the absence of environmentally accelerated crack growth, the cyclic crack growth rate is expected to accelerate when the creep-crack-tip zone associated with the mean value of stress intensity factor, $K_{\text{mean}}$, is greater than that associated with $\Delta K$. Thus, increasing $R$ should increase $da/dN$, when for plane-strain conditions,

$$r_{\text{cf}} = \frac{1}{3\pi} \left( \frac{K_{\text{mean}}}{\sigma_f} \right)^2 > r_f = \frac{1}{12\pi} \left( \frac{\Delta K}{\sigma_f} \right)^2 . \quad (51)$$

Now, $K_{\text{mean}} = K_{\text{max}} - \Delta K/2$ and $\Delta K = K_{\text{max}} (1-R)$, so this inequality implies that

$$K_{\text{max}} - 1/2K_{\text{max}}(1-R)^2 > 1/2K_{\text{max}}(1-R)^2 \quad (52)$$

or

$$R + 1/4(1-R)^2 > 1/4(1-R)^2 . \quad (53)$$

This result holds true for $R > 0$, which implies that accelerated crack growth is expected when $R > 0$. Experimental data on effects of stress ratio typically show increasing values of $da/dN$ with increasing values of $R$ at constant values of $\Delta K$.

Effect of Waveform

James(107) has compared sawtooth with hold-time waveforms in a study of crack growth in Type 304 stainless steel in air at 538 C. Such waveforms are illustrated in Figure 34. Cyclic frequencies from
1.38 × 10^{-3} \text{ Hz} \text{ to } 6.67 \times 10^{-2} \text{ Hz} \ were \ employed \ by \ James. \ Only \ at \ the \ lowest \ cyclic \ frequency, \ which \ corresponded \ to \ a \ 0.18\text{-hour} \ hold \ time, \ was \ a \ creep-fatigue \ interaction \ observed. \ The \ present \ model \ predicts \ that \ this \ should \ be \ the \ case \ because \ an \ incubation \ time \ is \ needed \ to \ form \ the \ creep-crack-tip \ zone. \ For \ relatively \ short \ hold \ periods, \ such \ as \ 0.18 \ hour, \ little \ creep-crack \ growth \ is \ expected. \ The \ acceleration \ of \ cyclic \ crack \ growth \ occurs \ because \ the \ hold \ time \ is \ long
enough or the rise-time of the sawtooth wave is long enough to allow
\( r_{cf} \) to be greater than \( r_f \). In this case, most of the crack advance
occurs during the rise time and the sawtooth, having a longer rise
time than the hold time, is expected to cause faster overall cyclic
crack growth. Such a trend is observed in the data reported by
James (107).

PERIODS OF CREEP AND FATIGUE

Consider periods of creep intermixed with periods of fatigue
cycling, as illustrated in Figure 35; the case investigated in experi­
ments on CT specimens in this study. If \( \Delta K < K_{max} \) (Figure 35a),
\( r_f < r_{cf} \) and accelerated cyclic crack-growth rates are predicted. If
\( \Delta K \geq 2K_{max} \) (Figure 35b) \( r_f \geq r_{cf} \) and no creep-fatigue interaction is
predicted.

The plots in Figure 36 show cyclic or fatigue-crack growth imme­
diately after periods of creep. In each of these plots, \( da/dN \) is
plotted versus the change in crack length. This change in crack
length is relative to the length of crack that existed at the end of
the creep-crack-growth period. In each case, the open points repre­
sent the experimental data obtained in the present study, with the
points plotted at zero change in crack length being computed assuming
a time-dependent-growth-rate process during the loading portion of the
initial cycle. This computed crack-growth rate was based upon inte­
gration of the \( C^* \) versus \( da/dt \) relationship for the Type 316 stainless
steel. The solid symbols shown in these figures by comparison show
the expected average fatigue-crack-growth-rate behavior for undamaged
FIGURE 35. PERIOD OF CREEP FOLLOWED BY FATIGUE CYCLING

(a) Creep Load Higher Than Fatigue Loading

\[ \Delta K < K_{\text{max}}, \quad r_f < r_{\text{ef}} \]

(b) Fatigue Loading Higher Than Creep Load

\[ \Delta K \geq 2K_{\text{max}}, \quad r_f \geq r_{\text{c}} \]
Observed Behavior
- AK of 23 to 26 MPa m$^{1/2}$ after creep at C* of 0.092 Watts m$^2$
- AK of 21 to 22 MPa m$^{1/2}$ after creep at C* of 1.4 Watts m$^2$

Computed Rate For Creep Only Using Initial C* Value

Expected Average Behavior of Undamaged Material

(a) Two periods of creep followed by fatigue at 649 °C.
(b) One period of creep followed by fatigue at 649 °C.

FIGURE 36. INFLUENCE OF PRIOR CREEP-CRACK GROWTH ON SUBSEQUENT FATIGUE-CRACK GROWTH FOR TYPE 316 STAINLESS STEEL
Computed Rate for Creep Only Using Initial Value of C^*

Observed Behavior at AK of 14 to 16 MPa m\(^{1/2}\) \((f = 0.06 \text{ Hz})\) After Creep at C^* of 0.16 Watts/m\(^2\)

Note: Expected Average Crack Growth Rate at 1 Hz for Undamaged Material Was Less Than 10\(^{-8}\) m/Cycle

(c) One Period of Creep Followed by Fatigue at a Low-Cyclic Frequency Causing Intergranular Cyclic Crack Growth at 593 C

(d) One Period of Creep Followed by Fatigue at a \(\Delta K\) Level Greater Than the \(K_{\text{max}}\) Level in the Prior Creep Period.
material under the same loading history. These are calculated from the \( \frac{da}{dN} \) versus \( \Delta K \) relationship developed for this material.

Figures 36a and 36b illustrate cases where the value of \( \Delta K \) during the fatigue loading portion was much less than the value of \( K_{\text{max}} \) at the end of the creep portion of the experiment (as in Figure 35a). As was observed experimentally, the crack-tip-zone interaction model predicts that initially the crack-growth rate should be accelerated and then approach the value expected for undamaged or virgin material. In both of these cases, the cyclic frequency was 1 Hz and the mode of fatigue-crack growth was transgranular.

Figure 36c indicates a case where the fatigue cycling after the period of creep was at a slow frequency of 0.06 Hz, and the associated mode of fatigue cracking was actually time dependent by an intergranular crack-growth mechanism. In this case, the crack grew initially at a very high cyclic-growth rate and even the final growth rate at the end of the fatigue cycling was much greater than that expected for cycling undamaged or virgin material at 1 Hz where the cracking would be by a transgranular growth mechanism.

Figure 36d illustrates a case where \( \Delta K \) during the fatigue cycling portion was larger than \( K_{\text{max}} \) during the creep period (as in Figure 35b) such that the size of the crack-tip zone associated with fatigue cycling approached that of the creep-crack zone. In this case, it was expected that the fatigue-crack-growth rate would be similar to that for undamaged materials which was the type of observed behavior.
In each of the plots in Figure 36, two sizes of crack-tip zone are indicated — $r_{cf}$ and $r_c$. The zone sizes were estimated using Equations (42) and (46). In all cases, the value of $r_c$ was greater than the corresponding value of $r_{cf}$. When the fatigue cycling after creep was rapid and the growth occurred by a transgranular mode, then the extent of the damaged zone measured by a high cyclic crack-growth rate corresponded approximately to the calculated value of $r_{cf}$ as evident in Figures 36a and 36b.

When the rate of fatigue cycling after the creep period was at a slow frequency, the value of $r_c$ corresponded closely to that of the damaged region ahead of the crack tip where accelerated crack growth occurred as illustrated in Figure 36c. In this latter case, the crack growth under cyclic loading at a slow frequency was still by an intergranular mechanism so the complete region where significant intergranular damage was expected showed a higher than anticipated crack-growth rate. In Figure 36c, the fatigue-crack-growth rate stabilized to a value of about $5 \times 10^{-7} \text{m/cycle}$, which was considerably more than the cyclic crack-growth rate of less than $10^{-8} \text{m/cycle}$ expected for undamaged material tested at 1 Hz. The reason for this difference in cyclic-crack-growth-rate behavior is that the mode of crack advance at the slow frequency was intergranular, whereas at 1 Hz, the mode of crack advance was transgranular.

For the data shown in Figure 36d, the value of $r_f$ was about half that of $r_{cf}$ instead of being much smaller than the value of $r_{cf}$. Thus, the cyclic crack-growth rate associated with fatigue cracking after creep was close to that expected for undamaged material. This behavior
approached the case depicted in Figure 35b.
VI. SUMMARY

The results of this study of creep-fatigue-crack propagation in Type 316 stainless steel in air at 593 and 649 C indicate that inelastic fracture mechanics can be applied to the problem of creep-fatigue-crack growth. Fatigue-crack-propagation data between $10^{-7}$ and $10^{-5}$ m/cycle have been developed at 649 C and between $10^{-9}$ and $2 \times 10^{-5}$ m/cycle have been developed at 593 C. Also, cyclic-crack-propagation-rate data have been obtained for hold times at tensile load of up to 1 hour at both 593 and 649 C. For short surface cracks with lengths from about 0.05 mm to about 1.5 mm in low-cycle-fatigue specimens, the cyclic- or fatigue-crack-growth rate has been well characterized in terms of the $\Delta J$ integral at crack-growth rates between $10^{-7}$ and $10^{-5}$ m/cycle.

The time-dependent- or creep-crack-growth rate has been well characterized using the $C^\alpha$ integral. Results of this study compare favorably with past work reported in the literature for Type 316 stainless steel at high growth rates and for Type 304 stainless steel. Creep-crack-growth-rate behavior has been characterized for time-dependent rates as low as approximately $10^{-9}$ m/sec. Limited creep-crack-growth data at rates approaching $10^{-11}$ m/sec also have been generated.
A crack-tip-zone interaction model has been applied to the problem of assessing creep-fatigue interaction during high-temperature crack growth. This approach can be used to describe the crack-growth behavior during low-cycle-fatigue experiments with hold periods at the maximum tensile strain. It can be applied also to problems where large cracks (approximately 2.5 mm or more) are growing under intermixed periods of creep and fatigue. By monitoring the change in fatigue-crack-growth rate after a period of creep, the relative extent of the damage in crack-tip zones ahead of a growing creep crack have been measured quantitatively. These measurements are in good agreement with the model of creep-fatigue interaction that has been developed.
LIST OF REFERENCES


