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Creep-Fatigue Interactions

Test Methods and Models

JAI Guest Editors:

Ashok Saxena

Bilal Dogan

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Foreword

THIS COMPILATION OF THE *JOURNAL OF ASTM INTERNATIONAL (JAI)*, STP1539, *Creep-Fatigue Interactions: Test Methods and Models*, contains only the papers published in JAI that were presented at a Symposium on Creep-Fatigue Interactions: Test Methods and Models held during November 17–19, 2010 in San Antonio, TX, USA. The Symposium was sponsored by ASTM International Committee E08 on Fatigue and Fracture in cooperation with ICF and EPRI.

Dr. Ashok Saxena, University of Arkansas, Fayetteville, AR, USA and Dr. Bilal Dogan, EPRI, Charlotte, NC, USA served as the Symposium Co-Chairmen and JAI Guest Editors.

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Overview

Creep-fatigue interaction as a degradation mechanism is a primary design concern in fossil power-plant and nuclear power-plant components, and in land, sea and air based gas turbine hot-section components. The energy conversion efficiency of these plants and turbines is strongly dependent on the operating temperatures and as these temperatures rise to boost efficiency, creep damage and its interaction with fatigue becomes an increasing design and operational concern. This book is intended to describe the latest advances in the understanding of creep-fatigue interaction mechanisms and become an important reference on the topic for several years.

The papers in this Special Technical Publication (STP) were presented in a two and half day ASTM International symposium on Creep-Fatigue Interactions: Test Methods and Models held in San Antonio, Texas, during November 17–19, 2010. The symposium was co-sponsored by the Electric Power Research Institute (EPRI) and was also billed as an Inter-quadrennial of the International Congress on Fracture, ICF. The symposium addressed the latest research in the area of creep-fatigue crack formation and crack growth in high temperature materials and structures. Thirty-four papers on recent developments in experimental techniques, models for representing the data and applications to structures were presented in a single session format. The presentations included six plenary presentations, four keynote presentations and contributed papers. Eventually 17 papers were submitted and accepted for publication in the STP.

Over the past few years, this area of research has seen some major advances in technology. In just the past two years, ASTM has published two standards for conducting creep-fatigue testing; the first E 2714-2009 addresses test methods for characterizing the creep-fatigue crack formation properties while the other E 2714-2010 addresses the creep-fatigue crack growth properties. This progress has happened largely with international level cooperation between several research groups from Europe, Asia, and America facilitated by EPRI. Therefore, the time was right to document this progress by holding an international symposium and collecting the papers into this book.

The first part of the STP addresses creep-fatigue interaction behavior of ferritic steels and in austenitic stainless steels. The emphasis among the ferritic steels was entirely on the relatively new class of materials with high chromium content designated by ASTM as P91 and P92 class of steels. These materials are being used extensively in advanced, high efficiency fossil power plants. The papers in this STP on this topic deal with properties, test methods and the latest models for applying the test data to components. The

austenitic stainless-steels continue to be the favored materials in nuclear power-plants. The topics of papers presented on austenitic stainless-steel cover the same range as the papers on the ferritics.

The second part of the STP covers creep-fatigue interactions in Nickel-base superalloys being considered for use or already in use in advanced nuclear plants and in gas turbines. The materials covered are IN 617, IN 718 and directionally solidified alloy 247. One paper also addresses creep-fatigue interaction in single crystal materials. A few papers also deal with thermal-mechanical fatigue and with the behavior of protective thermal barrier coatings that enhance the creep-fatigue performance of the components.

The Organizing Committee consisted of Ashok Saxena, University of Arkansas (Co-chair) and Bilal Dogan, EPRI, (Co-chair), Kamran Nikbin, Imperial College; Jeff Evans, University of Alabama at Huntsville; Andrew Rosenberger, Air Force Research Laboratory, Wright Patterson Air Force Base; Andre Pineau, Ecole de Mines; Peter Skelton, Consultant; Yukio Takahashi, CREIPI, Japan; Stuart Holdsworth, EMPA, Switzerland; A.T. Yokobori, Sendai University, Japan; S. Kalluri, OAI/NASA-GRC; R. Ainsworth, British Energy- part of EDF Energy; Laura Carroll, INEL, Idaho, S.D. Antolovich, Washington State University; Richard Neu, Georgia Tech; David Taplin, ICF; Alberto Carpinteri, Politecnico di Torino; F. Masuyama, Kyushu Institute of Technology, Jonathan Parker, Structural Integrity Associates, Karl Maile, MPA Stuttgart; Helmuth Klingelhoff, BAM, Berlin. Their invaluable contributions are very much appreciated. We also wish to acknowledge the contributions of all the authors and reviewers. The contributions of the ASTM Meetings staff and their publication Department staff are also gratefully acknowledged.

Ashok Saxena
University of Arkansas
Fayette, AR, USA

Bilal Dogan
EPRI – Charlotte
Charlotte, NC, USA

Symposium Co-Chairs and JAI Guest Editors

PART I

CREEP-FATIGUE INTERACTIONS IN FERRITIC AND AUSTENITIC STEELS

S. R. Holdsworth¹

Component Assessment Data Requirements from Creep-Fatigue Tests

ABSTRACT: Typically, the lifetime assessment of high temperature components, for which both creep and fatigue are principal damage mechanisms, involves (1) the determination of the stress/strain state at critical locations and (2) a numerical evaluation of the consequential damage condition. Until recently, there has not been a standard procedure specifically covering creep-fatigue testing practices. Creep-fatigue tests had been performed for many years, but the reliability and the form of the information generated was dependent on the expertise and specific interests of individual test laboratories. This situation is now resolved by the publication of a new ASTM standard based on an Electrical Power Research Institute (EPRI) led initiative. The way in which the guidance contained in this new standard specifically aims to provide creep-fatigue data requirements for defect-free component assessment is examined. Cyclic/hold tests, usually involving a simple cycle shape representative of the service transients experienced by the target component, are particularly informative in the provision of data to indicate the concurrent influences of cyclic loading on creep deformation characteristics and creep deformation on cyclic plastic response. The crack initiation endurances of such tests also provide the basis of creep-fatigue damage summation representations for a given material. Sequential creep-fatigue testing can be used to quantify the effects of prior cyclic deformation on creep rupture or prior creep deformation on fatigue endurance. The use of such test methods is also considered.

KEYWORDS: creep-fatigue interaction, testing, defect-free assessment

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¹EMPA: Swiss Federal Laboratories for Materials Science and Technology, Überlandstrasse 129, 8600 Dübendorf, Switzerland, e-mail: stuart.holdsworth@empa.ch

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Nomenclature

- D_C, D_F = creep damage fraction (Eqs 1 and 2), fatigue damage fraction (Eq 3)
 D_F, D_C = damage fraction coordinates at point of intersection in bi-linear damage locus (Eq 4)
 j = cycle type number
 LCF = low cycle fatigue
 N = number of cycles (to creep-fatigue crack initiation)
 N_i = number of cycles to crack initiation
 $PC(D)$ = prior creep (deformed/damaged)
 $PF(D)$ = prior fatigue (deformed/damaged)
 t = time
 t_h = duration of hold time or steady running period during operation
 t_R = time to creep rupture
 T = temperature
 T_i = indicated temperature
 Z = elastic follow-up factor [1], i.e., the ratio of the elastic strain recovered during the hold time to the creep strain accumulated during the hold time
 ε = strain
 $\varepsilon_C, \dot{\varepsilon}_C$ = creep strain, creep strain rate
 $\dot{\varepsilon}_{C,\min}$ = minimum creep strain rate
 ε_R = creep rupture strain (ductility)
 $\Delta\varepsilon_t$ = total strain range
 $\sigma, \Delta\sigma$ = stress, stress range
 σ_o = initial stress at start of hold time
 σ_P, σ_S = stresses due to primary (directly applied) and secondary (self equilibrating) loading
 σ_t = stress after time t in hold time

Introduction

Critical locations in high temperature power plant components may be subject to the combined accumulation of cyclic damage arising from strain transients generated during start-up and shut-down operation, and creep damage arising from primary (directly applied) and secondary (self equilibrating) stresses during operation. The development of creep-fatigue damage in most power plant steels depends on temperature, strain range, strain rate, hold time, and the creep strength and ductility of the material [2–4]. In the absence of a significant hold time (or at relatively high strain rates, or both), crack initiation and growth are fatigue dominated, even at high application temperatures (Fig. 1(a)). With increasing hold time (or decreasing strain rate, or both) at high temperatures, the creep damage condition within the structure becomes increasingly influential, to the limit beyond which crack development becomes fully creep dominated (Fig. 1(b)). At intermediate hold times and strain rates, fatigue cracking interacts with creep damage, developing “simultaneously” or “consequentially,” resulting in accelerated crack propagation (Fig. 1(c), 1(d)). The extent of any interaction increases with decreasing creep ductility [3]. Creep-fatigue

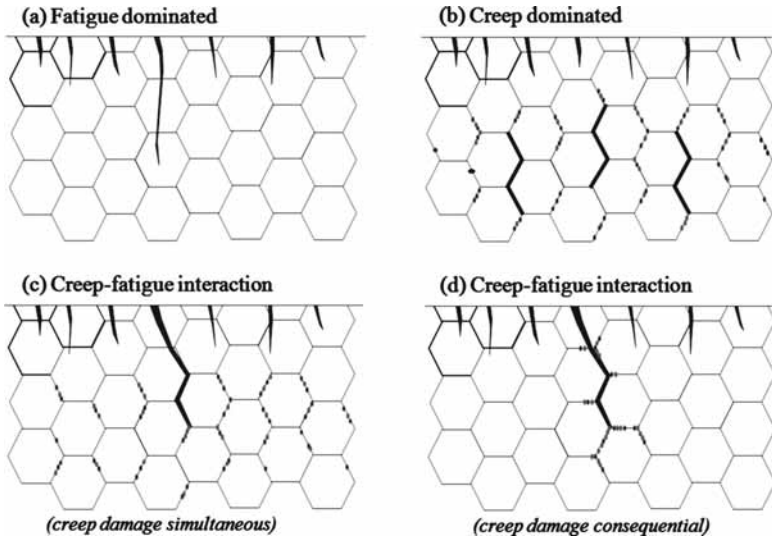


FIG. 1—Creep-fatigue cracking mechanisms: (a) Fatigue dominated, (b) creep dominated, (c) creep-fatigue interaction (due to “simultaneous” creep damage accumulation), and (d) creep-fatigue interaction (due to “consequential” creep damage accumulation).

deformation interactions are also influential, and in a dominant way for a number of alloys under certain conditions [5].

Depending on the component type and the purpose of the analysis, a defect-free or defect assessment procedure (or both) is undertaken to assess fitness for service at high temperatures. Defect-free assessment procedures are commonly used for design purposes whereas defect assessment procedures are used for inspection management and remaining life evaluations, although there are exceptions to this generality. For example, there can be circumstances when it is necessary to conduct defect assessment as part of a design strategy, such as for cast components and weld containing structures. Defect-free assessment procedures to assess the tolerance of high temperature components to thermo-mechanical loading rely on the availability of creep-fatigue test data for the material(s) of construction. These data may be the result of (a) concurrent or (b) sequential creep and fatigue loading campaigns. Examples of (a) are high temperature low-frequency triangular or saw-tooth waveform tests, or cyclic/hold tests (Fig. 2).

The focus of this paper is on those data requirements for the defect-free assessment of high temperature components, which can be determined from creep-fatigue tests.

Creep-Fatigue Assessment of Defect-Free Components

Overview

Various published and in-house procedures are adopted to assess the integrity of defect-free components subject to creep-fatigue loading, which can be

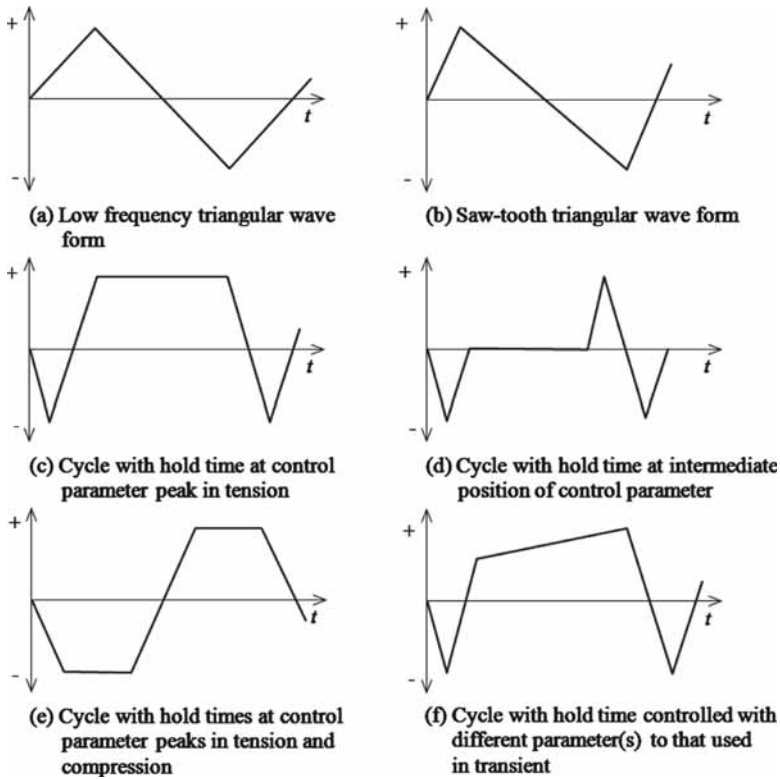


FIG. 2—Creep-fatigue test cycle shapes.

represented by the generic flow diagram shown in Fig. 3 [1,6–8]. Other approaches are possible (e.g., Refs 9–11), but are not so widely adopted. An important step in all creep-fatigue assessment procedures is the determination of the state of stress and strain at critical locations in the component. This requires knowledge of the external forces and thermal transients experienced by the component during service and representations of cyclic and creep deformation properties of the material(s) of construction. The magnitudes of stresses and strains generated during thermal transients depend largely on the physical properties of the material at the critical location, such as coefficient of expansion, thermal conductivity, and elastic modulus. Irrespective of whether the local stress/strain state is determined by approximate analytical solutions or finite element analysis, the test sources of material property data are generally the same [12]. Having established the stress/strain state history at critical locations, cyclic and creep damage fractions are determined by reference to the appropriate material endurance property data (Fig. 3). The D_F and D_C fractions are finally compared with, for example, the crack initiation locus in a creep-fatigue damage summation diagram, and the risk of cracking is evaluated.

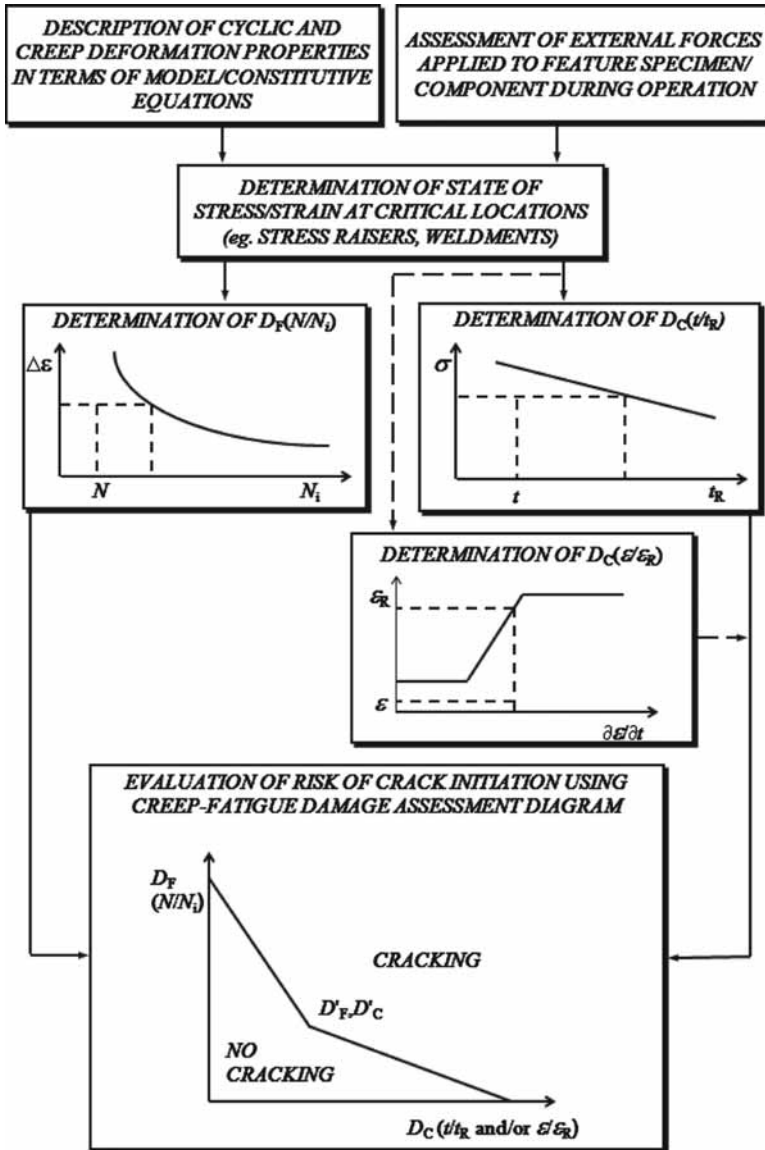


FIG. 3—Generic flow diagram representing analysis route adopted by a number of “defect-free” creep-fatigue assessment procedures (e.g., Refs 1, 6–8).

Determination of Local Stress/Strain State

Analysis options to determine the state of stress and strain at critical locations vary from simplified-elastic (for structures that can be shown to operate in shakedown [1]) to evolutionary full-inelastic, requiring a detailed description of

deformation as a function of temperature, strain rate, and any metallurgical changes leading to hardening or softening relative to monotonic behaviour (e.g., Fig. 4). At one analytical extreme, steady-state cyclic and creep deformation are assumed, with time independent plasticity and creep behaviour being clearly distinguished. However, alternative approaches range in increasing complexity to the adoption of fully unified state variable constitutive material models (e.g., Ref 13).

When global shakedown can be demonstrated, simplified models of analysis can be used. The cyclic stress-strain behaviour at areas of local plasticity may be determined using an elastic-plastic $\Delta\sigma(\Delta\epsilon_t)$ relationship for a given cycle number or for steady-state (mid-life) conditions (e.g., Ref 14), in conjunction with an appropriate strain concentration model (e.g., Ref 15) and an allowance for time dependent deformation. In such circumstances, periods of steady operation leading to load-controlled creep, strain-controlled stress relaxation, or stress relaxation under the influence of elastic follow-up may be assessed using forward creep $\epsilon_c(T, \sigma, t)$ model equations; e.g., Ref 16, with an appropriate hardening law and a correction for multi-axial loading (e.g., Ref 17). Alternatively, stress relaxation may be described directly, i.e., using empirical $\sigma_t(T, \sigma_o, t)$ expressions such as in Ref 18.

If shakedown cannot be demonstrated, a more detailed inelastic analysis may be necessary, which may be based on steady-state (mid-life) conditions or may represent evolutionary behaviour (e.g., Fig. 4). Irrespective of the level of complexity, it may be necessary to account for the effects of creep on cyclic stress-strain behaviour and vice versa, and this subject is examined further in later sections.

In an attempt to overcome some of the difficulties that can be associated with the preceding methods, models have been developed that use a state variable to describe the metallurgical condition of the material at any stage of the cyclic history (e.g., Fig. 4). These variables evolve according to the time-temperature-strain history of the component. State variable models fall into two broad categories, namely, non-unified and unified models. Non-unified models retain separate equation sets for plastic and creep deformation (e.g., Ref 19). In this approach, the state variable is the accumulated plastic strain, which is incorporated into the creep model to reflect the effect of cyclic softening on subsequent creep deformation and into the cyclic deformation model to reflect the effect of creep softening in the plasticity model. Fully unified representations include most notably the Chaboche model [13]. Their principal feature is that they describe the stress-strain-time-temperature response of a material subjected to any combination of monotonic, cyclic, and static loading. They are visco-plastic models based on internal variables evolving with time during the inelastic deformation process. Such models are unified in that no partition or distinction is made between plastic and creep inelastic strains.

Calculation of Creep-Fatigue Damage

Creep-fatigue assessment procedures are broadly differentiated by the way the components of creep and fatigue damage are determined. As a generality, they fall into two main categories, depending on how the creep damage due to

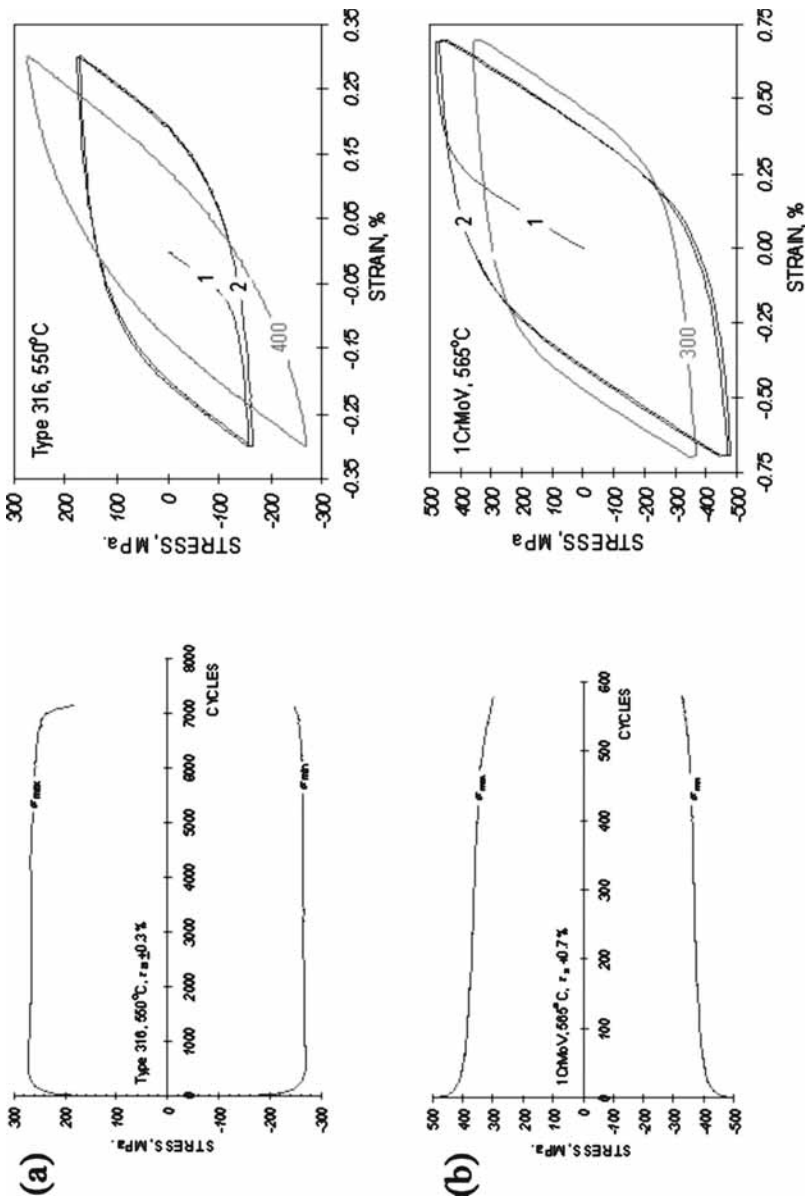


FIG. 4—Examples of (a) cyclic hardening and (b) cyclic softening behaviour in high temperature strain-controlled LCF tests.

secondary (self equilibrating) loading is calculated, i.e., by time-fraction or strain-fraction analysis methods. Time-fraction methods are currently the most widely used to determine the creep-fatigue damage accumulated (e.g., Refs 6–8). In any case, a time-fraction approach is always used to determine the creep usage arising from primary (directly applied) loading.

In time-fraction approaches, creep usage is determined as a function of the creep rupture time, $t_R(\sigma)$, i.e.,

$$D_C = \sum_j \left\{ N_j \left[\frac{t_h}{t_R(\sigma_P)} \right] \right\} + \sum_j \left\{ N_j \left[\int_0^{t_h} \frac{dt}{t_R(\sigma_S)} \right] \right\} \quad (1)$$

The two terms on the right side of Eq 1 respectively cover the damage accumulated due to primary and secondary loading. The creep rupture times $t_R(T, \sigma_P)$ and $t_R(T, \sigma_S)$ are determined using creep rupture strength data for the appropriate material and temperature (e.g., Ref 20).

The best known approach for the determination of creep damage in terms of strain-fraction (or ductility exhaustion) is that adopted in the R5 procedure [1], i.e.,

$$D_C = \sum_j \left\{ N_j \left[\frac{t_h}{t_R(\sigma_P)} \right] \right\} + \sum_j \left\{ N_j Z_j \left[\int_0^{t_h} \frac{\dot{\epsilon}_C}{\epsilon_R(\dot{\epsilon}_C)} \right] \right\} \quad (2)$$

where Z_j is the elastic follow-up factor for the j th cycle type and $\epsilon_R(\dot{\epsilon}_C)$ characteristics are determined from the results of creep rupture tests. The consideration of ductility exhaustion is a developing area, and it has already been demonstrated that a stress modified version of the denominator in the second term on the right of Eq 2 can provide a more effective indication of the state of ductility exhaustion (e.g., Ref 21).

For the future, there are emerging energy based models for the determination of creep damage fraction (e.g., Refs 22 and 23).

Fatigue damage fraction is typically determined from a model representation of smooth specimen $N_i(\Delta\epsilon_t)$, low cycle fatigue (LCF) crack initiation endurance data; e.g., Ref 24, and the expression

$$D_F = \sum_j \left\{ \frac{N_j}{N_i(\Delta\epsilon_t)} \right\} \quad (3)$$

Some analysts account for creep damage due to secondary loading as part of the fatigue damage fraction calculation by using a $N_i(\Delta\epsilon_t, t_h)$ endurance model based on observations from strain-controlled cyclic/hold tests in Eq 3; e.g., Refs 25 and 26, this being possible when the position in the cycle and the duration of the dwell period are representative of the service application. In such circumstances, creep damage is estimated using only the first term in Eq 1 (creep damage due to secondary loading now being contained in the D_F term). An example of the effect of hold time on cyclic/hold crack initiation endurance is shown in Fig. 5.

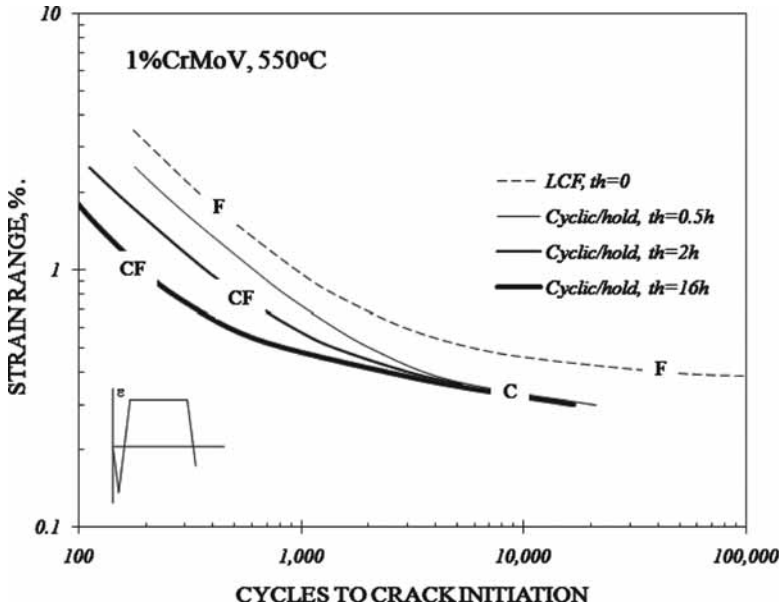


FIG. 5—Effect of increasing hold time on cyclic/hold creep-fatigue test endurance of 1 % CrMoV steel at 550°C 2.

Damage Summation

Creep and fatigue damage may be summed with reference to a creep-fatigue damage summation diagram (Fig. 3). Bi-linear damage loci as used in Refs 1, 6–8 may be modelled using

$$D_F = 1 - D_C(1 - D'_F)/D'_C \text{ for } D_C < D'_C$$

$$D_F = (1 - D_C) \cdot D'_F/(1 - D'_C) \text{ for } D_C > D'_C$$

where D'_F, D'_C defines the intersection coordinate [27]. For linear damage accumulation as in Refs 1 and 7, $D'_F, D'_C = 0.5$. An alternative to the bi-linear locus is the quadratic locus, which is said to represent the theoretical lower bound [28]. Apart from the quadratic representation, other damage summation options are determined on the basis of experimental observations from creep-fatigue tests, with D'_F, D'_C being dependent on material, and the adopted procedures for determining the stress/strain state, and creep and fatigue damage fractions [5].

Creep-Fatigue Tests

Creep-fatigue tests performed on smooth (usually uniaxial) specimens at appropriate temperatures, strain (load) ranges, and strain (loading) rates provide the

conditions for deformation and damage development due to creep-fatigue interaction (Fig. 1(c) and 1(d)). Such tests typically involve continuous lower strain rate triangular waveform (LCF) cycles (e.g., Fig. 2(a) and 2(b)) or cyclic/hold waveforms (e.g., Fig. 2(c)–2(f)). Until recently, there was no testing standard providing specific advice for the determination of deformation and crack initiation endurance properties under creep-fatigue loading conditions, and reliance was made upon existing guidance for LCF testing (e.g., Refs 29–31). While such publications provided the necessary information to conduct slow strain rate LCF type creep-fatigue tests, guidelines for cyclic/hold type creep-fatigue testing were not available until the publication of Ref 32.

The new ASTM standard E2714-09 [32] originated from an Electrical Power Research Institute (EPRI) led initiative [33]. It combines the strain measurement and control (including alignment) requirements adopted for LCF testing [31] and the temperature measurement and control requirements adopted for creep and stress relaxation testing [34], to determine creep-fatigue deformation and crack initiation endurance properties. In particular, the temperature in the specimen gauge section is required to be controlled to within $\pm \max[2^{\circ}\text{C}, 0.01 \cdot T_i]$ with due consideration to all sources of error. Specific emphasis is placed on the characterisation of damage development mechanisms through specimen post test examination.

Creep-fatigue tests may be conducted in load or strain control, although typically for the provision of data for power generation purposes they are conducted in strain control (e.g., Ref 31). Such tests are typically loaded to give crack initiation or failure endurance in the range ~ 100 to $< 100\,000$ cycles. At elevated temperatures, the influence of creep processes on deformation and damage response is increased by reducing strain rate(s) during strain transients, by increasing hold times usually (but not exclusively) at peak strain in tension, or both, as shown in Fig. 2. In addition to providing crack initiation or failure endurance, for the determination of fatigue damage condition, they also provide the basis for model representations of cyclic stress-strain response throughout life.

Typically, metals subject to cyclic plastic deformation harden or soften (or both) during the course of life, by an amount depending on the temperature and the total (plastic) strain amplitude. As a generality, cyclic hardening is due to the generation and interaction of dislocations, whereas softening occurs as a consequence of dynamic recovery of the dislocation sub-structure. Many austenitic steels such as Grade 316 cyclic harden at high temperatures (Fig. 4(a)) while exhibiting both hardening and softening during the course of life at lower temperatures. In contrast, low alloy and martensitic steels such as 1 % CrMoV and Grade 91 cyclic soften throughout life at temperatures within their application range (Fig. 4(b)). The consequent evolution in mechanical response associated with these microstructural changes can lead to a requirement for evolutionary stress-strain representations for some applications. These are underpinned by test data, and hence the requirement for a well defined deformation data collection strategy in Ref 32.

In addition to providing crack initiation endurance data to form the basis of damage summation relationships [2–4,35], the results of creep-fatigue tests also

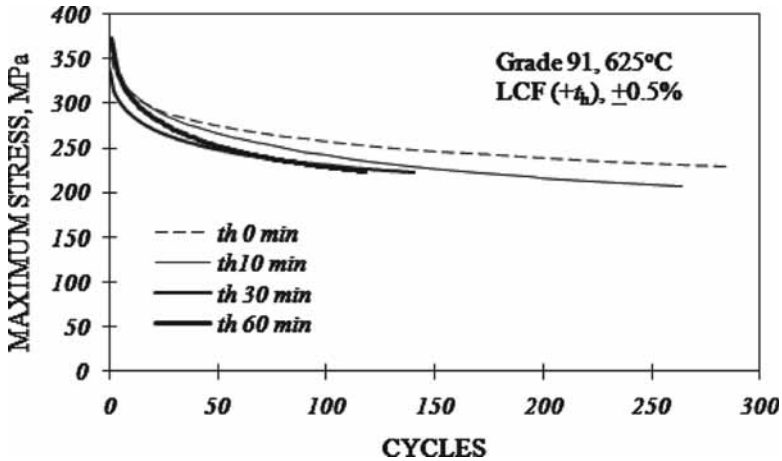


FIG. 6—Effect of increasing hold time on the extent of cyclic softening to $0.5 \cdot N_i$ cycles in Grade 91 steel at 625°C .

provide data to quantify the effects of creep deformation on cyclic plasticity, and of cyclic plasticity on creep deformation.

Cyclic Plastic Deformation

The extent of cyclic softening exhibited by low alloy and martensitic steels in standard LCF tests is enhanced in creep-fatigue tests. Increasing hold times or reducing the strain rate in continuous cycle tests can have a significant influence on cyclic plastic flow response, depending on the temperature and material. For example, for Grade 91 steel, increasing hold times at peak strain in tension up to ~ 20 min is responsible for enhanced softening, although further increases appear to have little additional effect (e.g., Fig. 6). Such saturated softening profiles can be used to form the basis of evolutionary creep-enhanced cyclic plasticity models.

It appears that low alloy steels are less prone to creep-enhanced cyclic softening than martensitic steels (Fig. 7, with data from Refs 36 and 37). It also appears that the extent of this creep enhancement is greater in lower frequency continuous cycling tests than in cyclic hold tests, in particular, for the martensitic steels (Fig. 7(c) and 7(d)).

Creep Deformation

Two important considerations during the evaluation of creep response as part of creep-fatigue assessment are (1) the extent of creep softening due to cyclic plastic loading and (2) whether or not the creep process should be re-initiated during analysis after each inelastic strain transient into compression (i.e., referred to here as primary creep persistency). Indications of the need and basis for considering both of these aspects may be determined from the results of creep-fatigue tests.

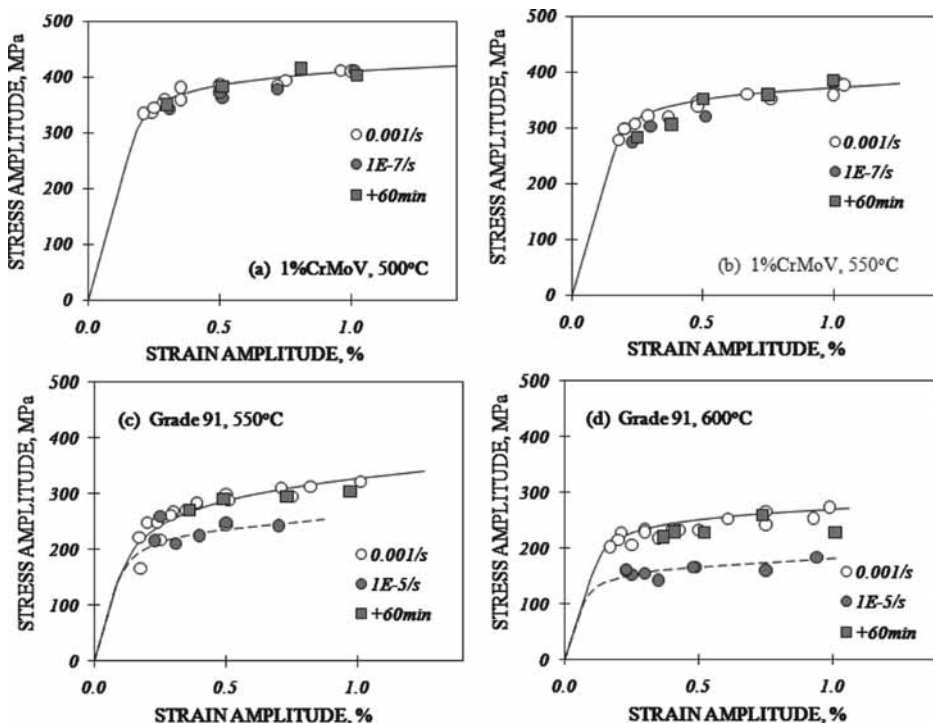


FIG. 7—Influence of reducing frequency or increasing hold time on mid-life cyclic stress-strain properties of: (a) 1 % CrMoV at 500°C, (b) 1 % CrMoV at 550°C, (c) Grade 91 at 550°C, and (d) Grade 91 at 600°C 36,37.

Creep Softening—Creep response is usually determined from constitutive model equations based on results from load-controlled creep rupture tests on the component material in the virgin (quality heat treated) condition. While this is invariably the only way to utilise the evidence from large data sets to realistically represent mean or minimum properties for the specified alloy (e.g., Ref 20), such model equations can only precisely represent creep during the first loading transient. Thereafter, creep response becomes dependent on cyclic history. The cyclic stress relaxation profiles from creep-fatigue tests may be used to determine $\dot{\epsilon}_C(\sigma)$ characteristics as a function of cycle number (or accumulated plastic strain). Example comparisons of cycle 1 and mid-life cycle $\dot{\epsilon}_C(\sigma)$ profiles for 1 % CrMoV and Grade 91 are shown in Fig. 8(a) and 8(b), respectively, for high application temperatures for the two steels. These illustrate the greater propensity of the martensitic steels to creep soften due to cyclic plasticity compared with low alloy ferritic steels such as 1 % CrMoV.

Primary Creep Persistency—Following inelastic deformation in compression, some materials restart creep strain accumulation from the very beginning of the primary stage. For such materials, the creep strain calculated per cycle

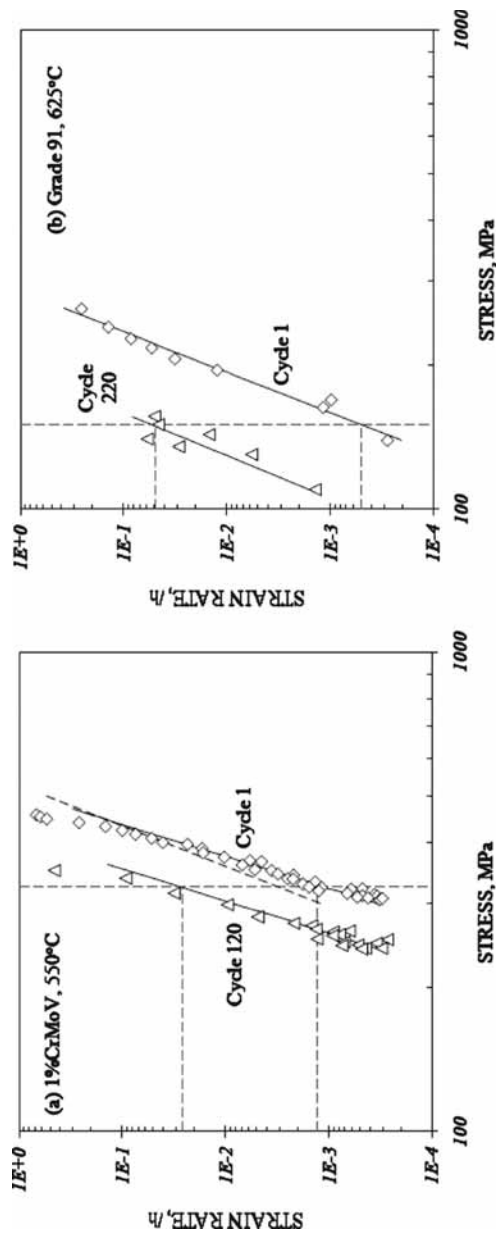


FIG. 8—Comparisons of cycle 1 and mid-life cycle $\dot{\epsilon}_c(\sigma)$ profiles for (a) 1 % CrMoV at 550°C and (b) Grade 91 at 625°C (for test conditions: $\epsilon_a \pm 0.5$ %, t_h , 30 min) 5 (the broken line in (a) is the $\dot{\epsilon}_{C_{min}}(\sigma)$ line for the same heat of 1 % CrMoV in the virgin condition determined from the results of load-controlled creep tests).

during creep-fatigue assessment can be significantly underestimated if this behaviour is not considered. One indication of whether a material responds in this way can be identified from the hold time $\dot{\epsilon}_C(\sigma)$ profiles of creep-fatigue tests (e.g., Fig. 8), although this approach is not conclusive.

For the 1 % CrMoV steel shown in Fig. 8(a), it is possible to demonstrate that after an initial primary creep controlled stress relaxation phase, the $\dot{\epsilon}_C(\sigma)$ characteristics determined during the hold time of the first cycle of a cyclic/hold creep-fatigue test are (as expected) almost identical to the $\dot{\epsilon}_{C,\min}(\sigma)$ properties determined from conventional load-controlled creep tests. A primary creep phase of stress relaxation is also detected at the start of the hold time of the mid-life cycle (Fig. 8(a)), and indicates the necessity to invoke primary creep persistency for this steel at this temperature. The evidence for Grade 91 in Fig. 8(b) is not so clear using this approach, and in such circumstances, it is necessary to perform cyclic creep tests involving inelastic transients in compression to clarify the position concerning primary creep persistency.

Sequential Creep-Fatigue Testing

An alternative type of creep-fatigue test is that involving the sequential application of block(s) of cyclic and steady loading to a specimen in any predefined order (e.g., Refs 38–41). They are most commonly employed to represent long duration service type conditions (e.g., Refs 38 and 40) or for assessing the effect of prior creep (or fatigue) deformation on fatigue (or creep) properties, e.g., Refs 39 and 41. It has been shown that $N_i(\Delta\epsilon_i)^{PC}$ prior creep deformed LCF crack initiation endurance data and $t_R(\sigma)^{PF}$ prior fatigue deformed creep rupture time data can be substituted, respectively, in Eq 3 or Eq 1 and used to construct a creep-fatigue damage summation diagram for 1 % CrMoV steel with coordinates that more closely relate to actual physical damage condition [5].

Prior Creep Effects

Evidence to indicate the influence of prior creep on cyclic crack initiation endurance properties for 1 % CrMoV steel was systematically collected by Shinya et al. [39], and subsequently re-evaluated by Binda [42] to show that creep may have a beneficial or detrimental influence on the fatigue crack initiation resistance for this steel, depending on the deformation regime and the mechanism of damage development. Prior to the formation of physical damage, either as void particle/matrix decohesion or grain boundary cavities, fatigue crack initiation resistance is enhanced in material that has accumulated primary or secondary creep deformation. The beneficial effect of creep deformation is reversed once physical creep damage is present (Fig. 9). The point at which this occurs depends on the stress and the temperature responsible for creep and the creep ductility of the steel [3]. At higher stresses or lower temperatures (or both), significant physical creep damage is only generated relatively late in life in the form of voids due to particle/matrix decohesion, and final fracture is transgranular. In these circumstances, the fatigue crack initiation resistance continues to be enhanced until late in the creep life. At lower stresses or higher

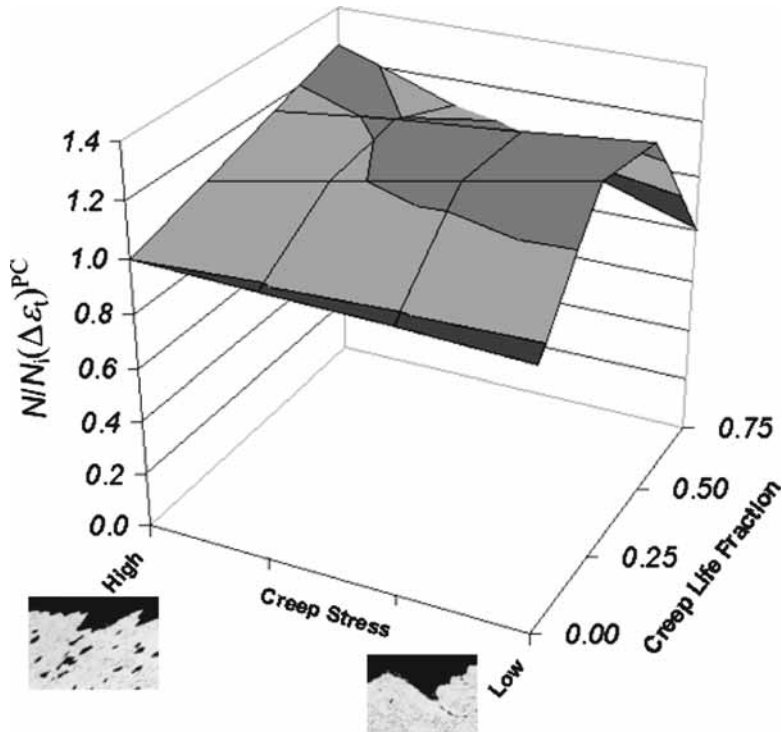


FIG. 9—Influence of stress responsible for prior creep condition on fatigue crack initiation endurance 42.

temperatures (or both), physical creep damage in the form of grain boundary cavities can occur relatively early in life when the rupture ductility is low, and later in life when the rupture ductility is higher. The development of physical creep damage at the grain boundaries has an increasingly detrimental effect on fatigue crack initiation resistance.

The numerical representation of the PCD functional surface shown in Fig. 9 enables the determination of $N_i(\Delta\epsilon_t)^{PC}$ and substitution for $N_i(\Delta\epsilon_t)$ in Eq 3 to determine an alternative quantity for fatigue damage fraction.

Prior Cyclic Deformation

Prior cyclic deformation is responsible for cyclic softening in 1 % CrMoV steels, which in turn influences the magnitude of the creep strength [41]. The outcome is a reduction in rupture times and an increase in creep strain rates, but with $\dot{\epsilon}_{C,min}t_R$ maintaining the same constant of proportionality as the virgin steel, Fig. 10(a) and 10(b) [41]. The magnitude by which the rupture time is reduced depends on the magnitude of the stress and the consequent rupture mechanism. At stress levels responsible for a ductile transgranular rupture mechanism (arising from void formation due to particle/matrix decohesion), the reduction in

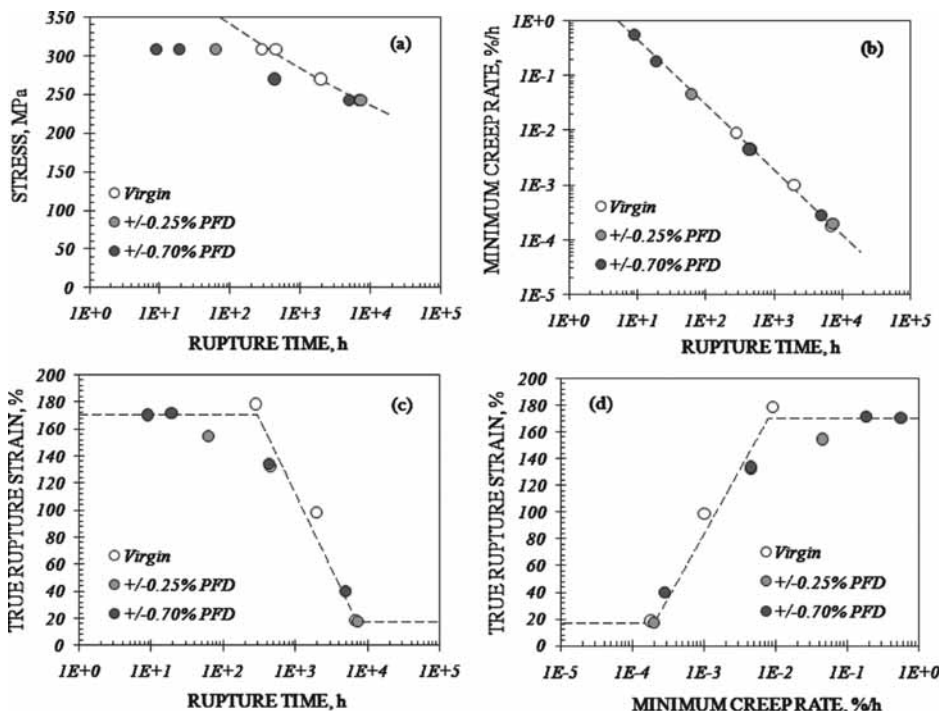


FIG. 10—Influence of prior fatigue deformation on creep rupture properties of 1 % CrMoV steel 41.

rupture time is significant. In contrast, at stress levels responsible for intergranular rupture (arising from grain boundary cavity formation), the reduction in rupture time is relatively small.

The effect of prior cyclic deformation on the rupture ductility of 1 % CrMoV steel also depends on stress and strain rate (rupture mechanism) [41]. At stresses (and strain rates) associated with lower and upper shelf rupture ductilities, the influence of prior cyclic deformation and the consequent increase in minimum creep rate is small (Fig. 10(c) and 10(d)). At stresses (and strain rates) associated with the transition from upper to lower shelf rupture ductilities, the influence of prior cyclic deformation can be responsible for a significant increase in rupture ductility.

Analytical representations of the data given in Fig. 10(a) and 10(d) enable the determination of $t_R(\sigma)^{PF}$ or $\varepsilon_R(\dot{\varepsilon}_C)^{PF}$ or both, for substitution into Eq 1 or Eq 2 to determine alternative quantities for creep damage fraction.

In contrast, for Grade 316 steel, the effect of prior cyclic deformation on creep is to increase the time to rupture and reduce the strain rate, with the magnitude of the effect appearing to be relatively insensitive to applied stress [42], Fig. 11. For this steel, prior cyclic deformation is responsible for a significant increase in rupture ductility at all strain rates. The effects of prior cyclic

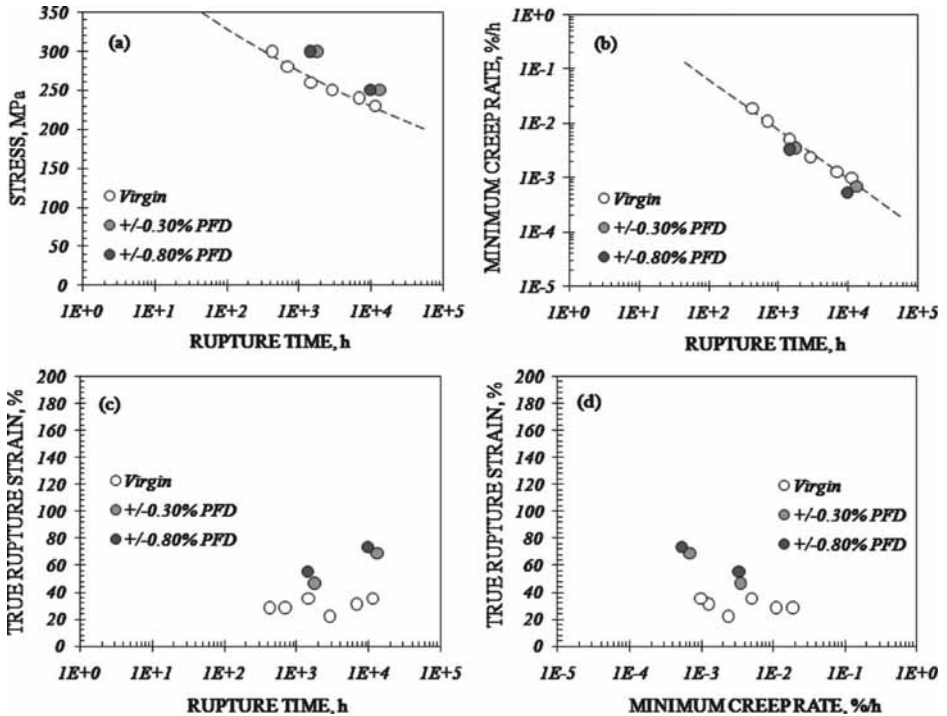


FIG. 11—Influence of prior fatigue deformation on creep rupture properties of Grade 316 steel 42.

deformation on creep rupture properties are dependent on material and temperature.

The use of fatigue endurance and creep rupture models involving the results from prior creep and prior fatigue tests can provide more effective estimates of D_F and D_C damage fractions, which can be formulated in an evolutionary way [5] but requires a significantly more sophisticated material property database than currently adopted in traditional approaches (e.g., Refs 1, 6–8). Nevertheless, such an approach provides a potential solution for the future.

Concluding Remarks

Typically, the defect-free assessment of high temperature components for which both creep and fatigue are the principal damage mechanisms involves (1) the determination of the stress/strain state at critical locations and (2) a numerical evaluation of the consequential damage condition. Creep-fatigue tests are a significant source of material input data for such assessments, and a new standard [32] provides the necessary guidance for their performance.

Cyclic/hold creep-fatigue tests are particularly informative in the provision of data to indicate the concurrent influences of cyclic loading on creep

deformation characteristics and creep deformation on cyclic plastic response, in addition to crack initiation endurance data to form the basis of creep-fatigue damage summation representations.

Sequential creep-fatigue testing can also be productively used to quantify the effects of prior creep deformation on fatigue endurance and prior cyclic deformation on creep rupture (for the determination of more physically realistic fatigue and creep damage fractions for certain materials).

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V. Kalyanasundaram,¹ A. Saxena,² S. Narasimhachary,¹
and B. Dogan³

ASTM Round-Robin on Creep-Fatigue and Creep Behavior of P91 Steel

ABSTRACT: The American Society for Testing and Materials (ASTM), through its Committee E08 on Fatigue and Fracture subcommittee E08.05 on Creep-Fatigue Crack Formation, has recently developed a new standard for creep-fatigue testing (ASTM E2714-09). This paper describes the plans and preliminary results from a round-robin being presently conducted in support and verification of this new standard. The choice of the test material (ASTM Grade P91), the design of the round-robin test matrix, and a machining plan for the specimens are described. The results of microstructural analysis, tensile testing, and creep deformation and rupture testing are also presented along with some preliminary results from creep-fatigue testing. A new analytical model for representing the creep deformation characteristics of this material is also presented and evaluated using the creep data generated as part of the round-robin program. The results of the round-robin creep-fatigue testing will be used to make appropriate modifications to the test standard.

KEYWORDS: creep-fatigue, crack formation, statistical data analysis, creep deformation and rupture

Introduction

ASTM E2714-09 [1] covers the determination of mechanical properties pertaining to creep-fatigue crack formation in nominally homogeneous materials by the use of test specimens subjected to uniaxial forces under isothermal

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¹Dept. of Mechanical Engineering, Univ. of Arkansas, Fayetteville, AR 72701.

²Dept. of Mechanical Engineering, Univ. of Arkansas, Fayetteville, AR 72701 (Corresponding author), e-mail: asaxena@uark.edu

³EPRI, Charlotte, NC 28262.

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conditions. It concerns creep-fatigue testing at strain rates and/or cycles involving sufficiently long hold times to induce creep deformation (and oxidation) during cyclic deformation where cycles to crack formation are affected by creep.

This test method is applicable to the determination of deformation and crack formation or nucleation properties as a consequence of either constant-amplitude strain-controlled tests or constant-amplitude force-controlled tests with hold times. It is primarily concerned with the testing of round bar specimens subjected to uniaxial loading in either force or strain control, wherein the latter is recommended. As a result of this round-robin, future improvements planned for this standard include, but are not limited to, a more definitive precision and bias statement.

A total of 16 participants from laboratories all over the world are conducting tests under a coordinated set of test conditions. The primary objective of the round-robin is to conduct creep-fatigue tests to characterize the number of cycles for crack formation while using the procedures specified in the standard to assess variability in the results. The list of participants, their affiliation, and the planned set of tests for each of them is provided in Table 1.

Test Material

The candidate materials considered for the round-robin were from a wide variety of materials that are used in high temperature applications and included gas turbine materials for aircraft and land-based engines, fossil power plants, and nuclear reactor materials. Other considerations included material availability in sufficient quantities, offers of sponsorship for material and machining costs, and the interest level in testing the material among the participating volunteers. The test material selected is ASTM Grade P91 steel (donated by the Electric Power Research Institute (EPRI), Charlotte) that has a creep rupture strength of 94 MPa at 600°C for a life of 10⁵ h. The nominal chemical composition of P91 steel in wt % is given in Table 2 [2]. There are two other round-robins with similar objectives that are under way. The Idaho National Laboratory is conducting a separate round-robin on IN617 (a nickel-based superalloy) and the Japan Society for Promotion of Science is conducting one on ASTM Grade P92 material. While these are parallel independent studies, an effort is planned to integrate the findings from all the studies at a later date.

P91 Steel—Brief Overview

Modified tempered martensitic heat resistant steels of type 9Cr-1Mo-Nb-0.2V (P91/T91 as per ASTM A335 [3]/ASTM A213 [4], respectively) developed in the United States in the early 1980s, are used in main steam pipes, superheaters, headers, boilers, and turbines in supercritical and ultra-supercritical fossil power generation plants [5]. As a replacement for low alloy ferritic and austenitic stainless steels, the high creep strength of P91 permits its use in relatively thinner-wall components. This class of steel also offers the advantages of

TABLE 1—List of participants and test matrix for the ongoing ASTM round-robin (Numbers indicate the total number of tests planned under those conditions).

Participants ↓	P91 Crack Formation						
	±0.25 %	±0.5 %	±0.75 %	±0.5 %/10 min Hold	±0.75 %/10 min Hold	±0.5 %/30 min Hold	
<i>Idaho National Laboratory: Laura Carroll</i>	2	2		2			
<i>EMPA Switzerland: Stuart Holdsworth</i>				4			4
<i>Tech U. Darmstadt: Alfred Scholz</i>	2	2	2				
<i>CRIEPI, Japan: Y. Takahashi</i>		2	2	2	2		2
<i>GE, Schenectady: David Knorr</i>		2	2	2			
<i>BAM, Berlin: Hellmuth Klingelhoefter</i>	2		2	2			
<i>NASA, Glenn: Brad Lerch</i>			2	2	2		
<i>ANSTO Australia: Warwick Payton</i>	2			2	2		
<i>Georgia Tech: Rick Neu</i>	2	2		2	2		
<i>Tohoku Univ.: A. T. Yokobori</i>	2	2		2	2		
<i>NRC, Canada: Jonathan Tsang</i>			2		2		2
<i>Univ. of Arkansas: Ashok Saxena</i>				Conduct creep deformation and rupture testing and metallography and TEM of the test material			
<i>MPA Stuttgart: Andreas Klenk/Karl Meile</i>							6
<i>British Energy: Mike Spindler</i>	2	2	2	2			
<i>Politecnico di Milano: S. Beretta</i>					2		2
<i>BiSS, India: R. Sunder</i>		2		2			4

TABLE 2—Nominal chemical composition of P91 steel (wt %).

Steel Grade	P91	C	Si	Mn	P	S	Ni	Cr	Mo	As	V	Nb	Al	Cu	N	Sb, Sn	Fe
		0.11	0.31	0.45	0.011	0.009	0.19	8.22	0.94	0.005	0.21	0.07	0.006	0.16	0.039	0.001	Bal.

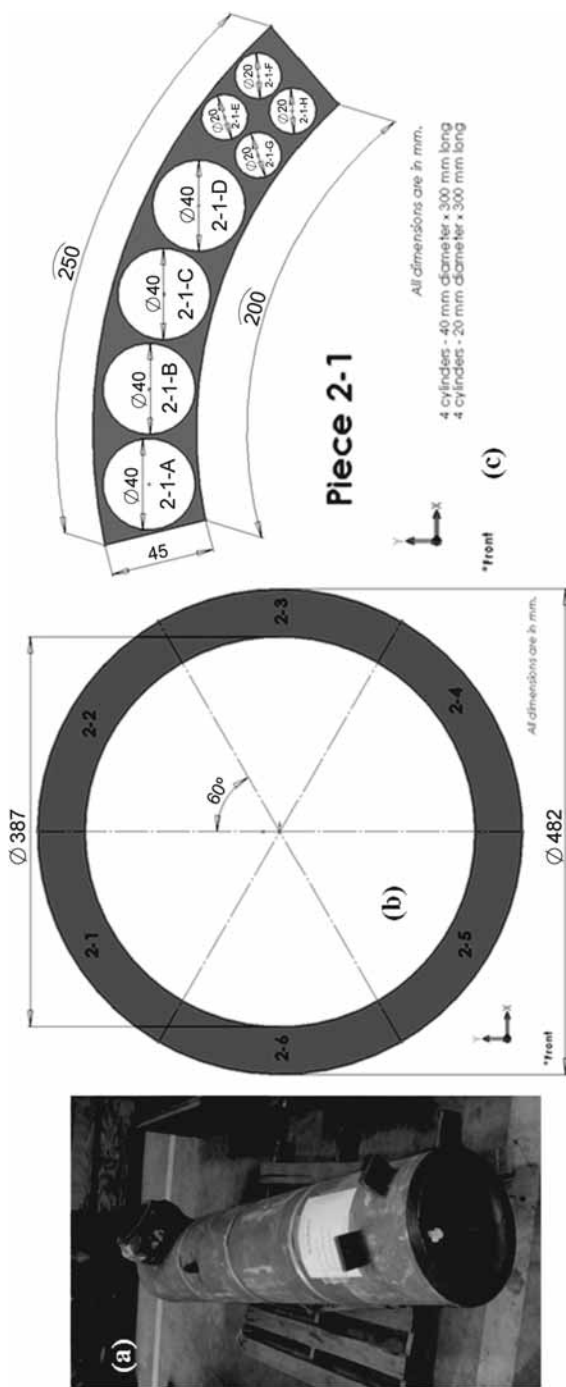


FIG. 1—(a) P91 pipe section used for the ASTM round-robin testing; (b) cross-sectional view of section 2 of the pipe section; (c) machining layout for subsection 2-1 (Material courtesy of EPRI).

low thermal expansion, high thermal conductivity, high corrosion cracking resistance, a low oxidation rate, and good inspectability and weldability [6]. These advantages allow this material to be a potential candidate to replace currently used P22 (2-1/4 Cr–Mo) steel and an alternative for future applications in power plants operating at service conditions greater than 600°C and a steam pressure of 300 bars.

Test Specimens and Machining Plan

Test specimens were machined from a pipe section 482 mm in outer diameter, ~1 m in length and 47.5 mm in wall thickness [see Fig. 1(a)]. The pipe was renormalized to ensure consistency with the original microstructure, then cut in to three sections along its length and labeled as pipe sections 1, 2, and 3, respectively. For this round-robin, the cut section 2 was again cut in to six equal subsections along the cross-section and labeled as shown in Fig. 1(b). Specimen blanks were machined from these subsections to accommodate various specimen sizes and geometries as provided by the participants. Figure 1(c) shows the machining layout for subsection 2-1 as an example. Specimen sizes and grip configurations were selected by the participants, but they were required to be within the allowable limits of the standard. Figures 2(a) and 2(b) show two typical specimen geometries and sizes requested by participants as part of the round-robin for performing creep-fatigue testing and for the supporting creep deformation and rupture testing, respectively.

Microstructural Characterization, Tensile and Creep Behavior

Figure 3(a) shows the microstructure of the renormalized P91 material wherein the martensite phase can be observed in the lath microstructure arranged in packets (30–40 μm in size) within prior austenite grains that were on average 100 μm in diameter. Optical micrographs at high magnifications revealed extensive carbide precipitates rich in chromium, namely, $M_{23}C_6$, where $M = \text{Cr}$ (sizes up to 100 nm), that can be located at grain and packet boundaries and fine carbonitride precipitation, namely, MX , where $M = \text{Nb}$ or V and $X = \text{C}$ and/or N , inside the packets similar to the schematic shown in Fig. 3(b). It has been found that this material has excellent creep resistance as its martensitic microstructure includes significant dislocation content in the lath structure introduced by heat treatment [7].

The University of Arkansas was responsible for conducting tensile (performed on-site at Bangalore Integrated System Solutions (BiSS), India) and creep rupture tests and the subsequent metallographic evaluation of the failed test coupons. Tensile tests were performed, at room temperature and at the high temperature (625°C) chosen for creep-fatigue testing, on dog-bone specimens with a gage length and diameter of 13 and 5 mm, respectively, at an initial strain rate of 0.002 s^{-1} . The tensile properties obtained from these tests are

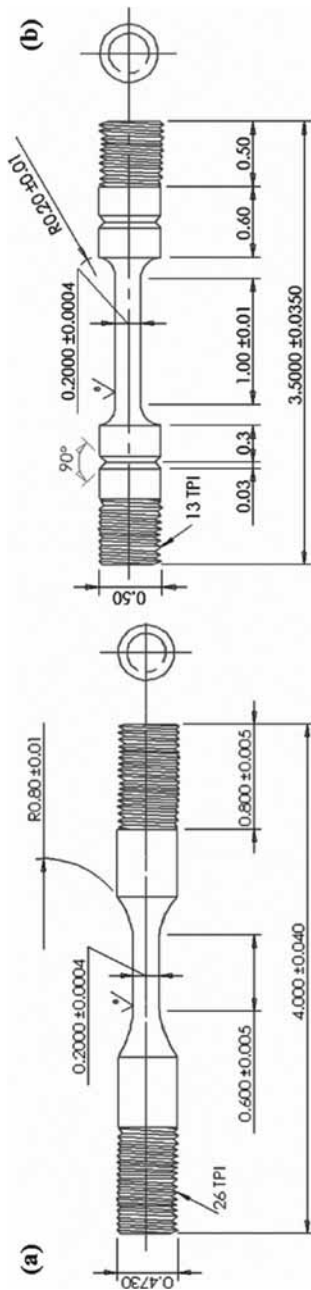


FIG. 2—(a) Creep-fatigue and (b) creep specimen layout used in the ASTM round-robin (all dimensions are in inches).

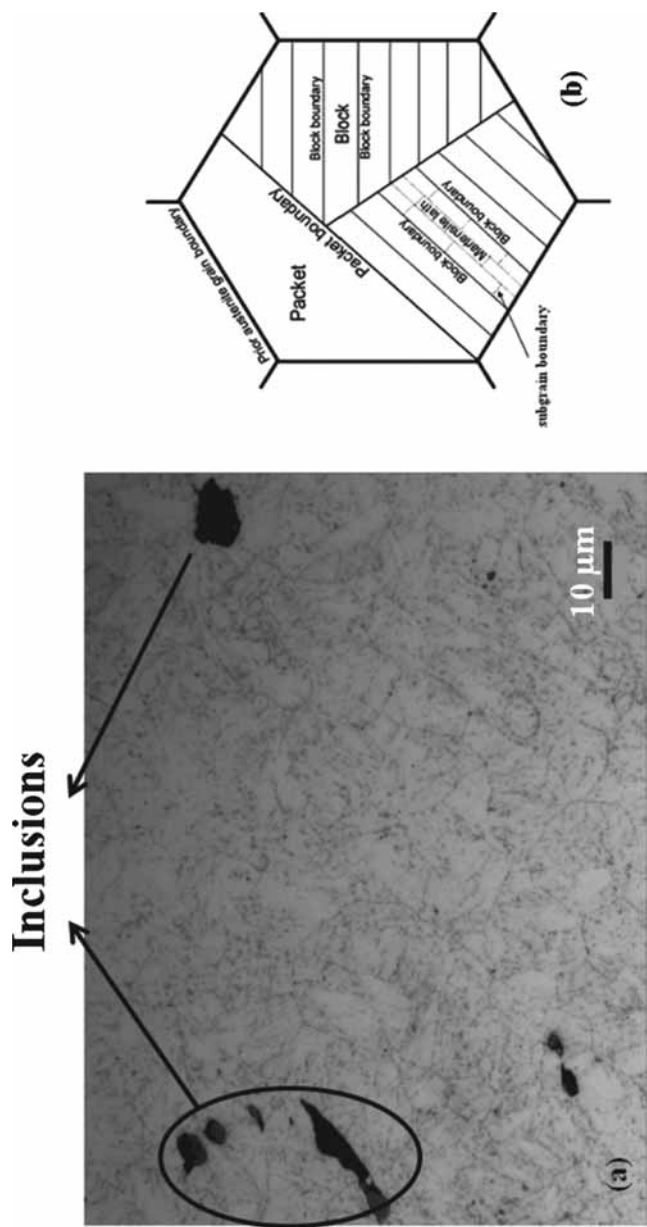


FIG. 3—(a) Optical micrograph of P91 steel microstructure with the inset showing inclusions and (b) shows a schematic of the microstructure 9.

TABLE 3—Tensile test results of P91 steel at room and high temperature (625°C).

Test Temperature (°C/°F)	0.2 % Yield Strength (MPa/ksi)	Ultimate Tensile Strength (MPa/ksi)	Elongation (%)
24/75	532.6/77.2	708.4/102.7	26
625/1157	325.1/47.1	343.7/49.8	33

listed in Table 3 and were found to be comparable to those published in the literature for P91 steels.

Creep tests were performed on smooth round specimens with a gage length and diameter of 25 and 5 mm, respectively, which were designed as per the ASTM E139 standard [8] [see Fig. 2(b)]. All tests were carried out at 625°C (898 K). Creep tests were carried out under uniaxial static (constant stress and temperature) loading conditions in a lab-controlled atmosphere (20±2°C and 50 % relative humidity). The external static load was applied using dead weights and a calibrated LVDT with a repeatability of 0.1 μm was employed to measure elongation during the tests. The test temperature was monitored continuously during the entire span of the tests using two K-type thermocouples wound mechanically at the top and bottom ends of the specimen gage length. The temperature difference between the top and bottom thermocouples during any test was continuously monitored and found to be within the allowable limits of ±2°C of the test temperature.

The Larson–Miller parameter (P_{LM}) is used as a predictive parameter to evaluate the stress level required for a given rupture time [10] and is denoted analytically as

$$P_{LM} = T_k * [C_{LM} + \log(t_r)]/1000 \tag{1}$$

where:

- T_k = test temperature (K),
- t_r = creep – rupture time (h), and
- C_{LM} = constant (30 for P91 steel).

The P_{LM} plotted against the applied engineering stress (external static load divided by the initial cross-sectional area of the specimen) σ in MPa is used to obtain the value of σ for a given t_r . For this round-robin, the test matrix was designed such that t_r ranges from 5000 to 150 h and the corresponding stress levels range from 102 to 152 MPa (see Fig. 4). The data from the tests conducted as part of this program are plotted with the data from the literature [11] in Fig. 4 and were found to lie perfectly within the trends of the literature data, thus once again confirming that the test material behavior is within the limits of typical P91 steels.

The creep deformation mechanism of P91 steels can be dominated depending on stress and temperature, by cross-slip and dislocation climb (> 70 MPa), and by grain boundary diffusion (< 70 MPa) [12]. It has been found by others that for martensitic steels, the metallurgical changes are of vital importance as they strongly affect creep resistance properties leading to loss in creep rupture strength [13]. From a microstructural perspective, it has been

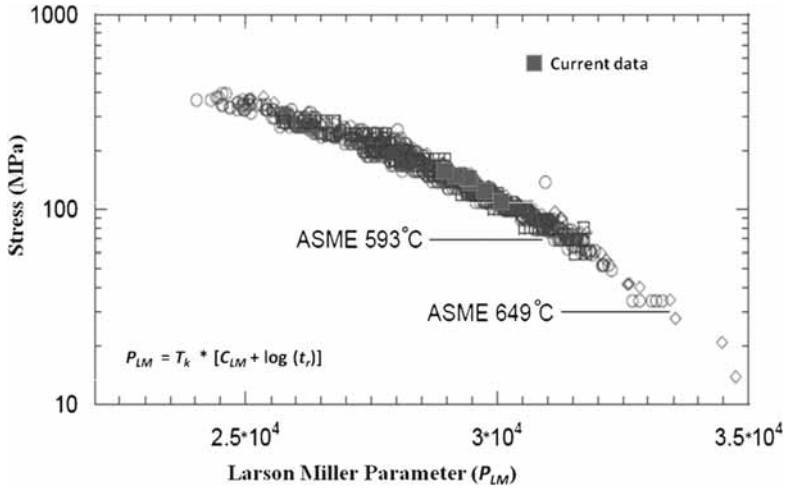


FIG. 4—Larson Miller parameter plot from Ref 11, overlaid with data from the current round-robin tests at 625°C.

observed that the $M_{23}C_6$ precipitates in P91 link up by rapid directional coarsening during the primary creep regime. The resistance to creep deformation by hindering of dislocation motion is increased by the evolution of this microstructure during the final stages of primary creep and during the entire secondary creep regime [14].

Creep strain versus time curves exhibited very short primary ($\sim 10\%$ of t_r) creep followed by secondary or steady-state and substantial tertiary creep stages under the tested conditions (see Fig. 5). Because of dynamic creep recovery due to subgrain growth and dislocation migration/annihilation in grain sub-boundaries, the primary creep strain rate starts to decrease until it reaches a plateau at the onset of secondary or steady-state creep. The secondary creep regime represents a dynamic equilibrium between work hardening and creep recovery processes, wherein a balance between generation of new dislocations and annihilation of existing dislocations is achieved [15]. The hindrance to dislocation motion due to the evolving precipitates mentioned above must also be part of the overall dynamic equilibrium during secondary creep stage. The minimum creep rate ($\dot{\epsilon}_{ss}$) as observed in the secondary creep regime is linearly fit using Norton's power law $\dot{\epsilon}_{ss} = A\sigma^n$ (where A and n are material constants) and the power law exponent (n) and constant (A) are found to be ~ 8 and 9.5×10^{-21} , respectively (see Table 4).

The onset of tertiary creep is characterized by accelerating creep rates as a result of a combination of the following reasons [16]:

- (1) increasing stress experienced in localized region(s) of the material due to necking phenomenon
- (2) increasing creep strain accumulation that leads to cavity formation and growth and consequent damage that develops in the form of creep cavities over time

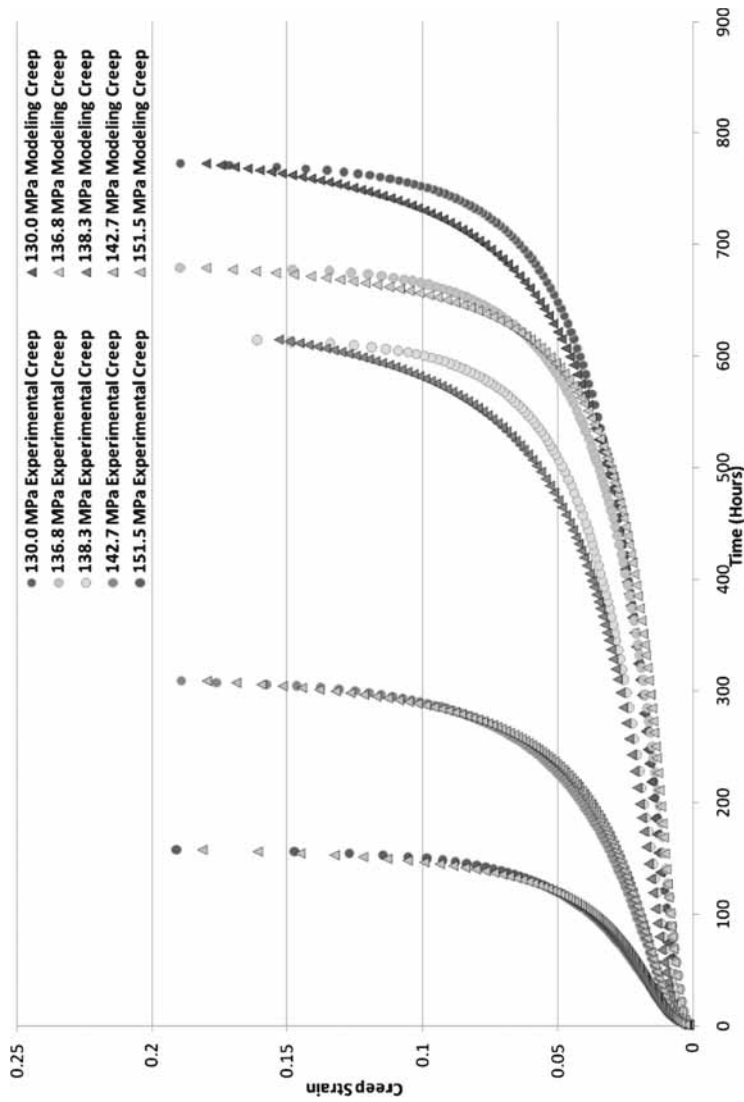


FIG. 5—Creep rupture (experimental versus modeled) curves for the completed ASTM round-robin tests of P91 steel at 625°C.

TABLE 4—Steady-state creep rate as a function of stress for P91 test material at 625°C.

Applied Stress (MPa)	Steady-State Creep Rate (h^{-1})	Power Law Constants
101.5	0.000 35	$n = 8.24$ $A = 9.53 \times 10^{-21}$
130.0	0.001 66	
136.8	0.003 81	
138.3	0.005 23	
142.7	0.007 79	
151.5	0.007 48	

(3) microstructural changes that are not associated with the accumulated damage

The creep rupture ductility for P91 steel is high with the final longitudinal elongation varying from 16–19 %. It can also be observed from Fig. 5 that the creep strains associated with the primary and the secondary creep regimes are relatively smaller than those associated with the tertiary creep regime, especially at the higher stress levels. Localized necking occurs at times past $\sim 95\%$ of the creep rupture time (i.e., t_r) when macro-cracking also appears and contributes further to rupture elongation. Similar such observations have been noted in other studies for this class of steels [9,12,13].

Optical microscopic analysis of ruptured creep specimens showed that the cracking mode during creep rupture is a predominantly transgranular (ductile) fracture with an array of microvoids growing through grains along crystallographic planes towards the final rupture location [see Fig. 6(a)]. It is possible that such a failure mode is due to plastic deformation at high stresses typically used in accelerated laboratory testing. It is further noted that these arrays were found most commonly near material inclusions [precipitates and/or secondary phases, see Fig. 6(b)].

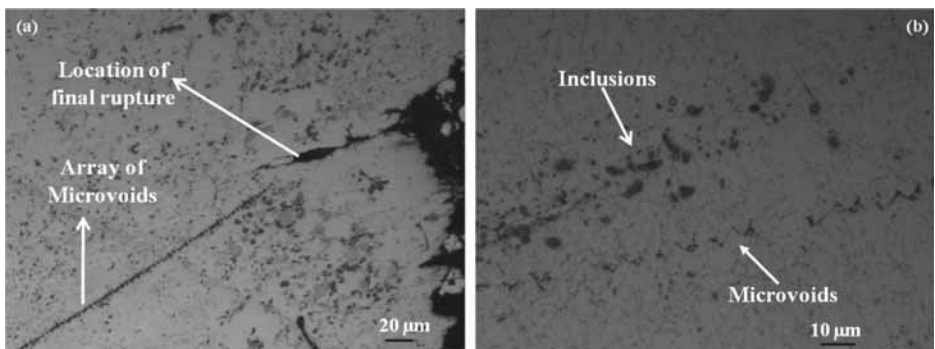


FIG. 6—Transgranular (ductile) fracture mode as observed in P91 steel (a) before and (b) after etching with Nital (3 % nitric acid in methanol) solution (Test condition: 151.5 MPa, 625°C).

Creep Deformation Model

In this section, the creep deformation and rupture data are represented using mathematical equations also known as extrapolation models. Such models are essential for data analysis and structural analysis. Since the creep data presented on P91 clearly exhibit primary, secondary, and tertiary creep behaviors, the focus of this exercise was on choosing a model that is capable of representing the creep behavior in all three regimes and one that also accounts for changes in deformation kinetics as a function of stress and temperature. Since all tests performed were at one temperature, the temperature capability of the chosen model cannot be verified from the current data set. A compilation of the existing models as applicable to any high temperature material and P91 steel in particular is provided in Ref [17]. After carefully evaluating all of these models, the authors found that the logarithmic creep strain prediction (LCSP) model to be most suitable for representing the data on P91 steel [18]. This model has the least number of fitting constants and its mathematical form has the natural shape of a creep curve at constant stress. However, in its current form, this model suffers from two shortcomings:

- (1) The model does not reduce to the correct form when the boundary conditions at the start and at the end of any creep test are applied (i.e., as one asymptotically approaches time $t = 0$ and $t = t_r$).
- (2) Further, the model does not take full advantage of parameters that can be measured from the actual tests to reduce the number of fitting constants.

Hence, a modified LCSP model is proposed here by addressing the above shortcomings of the original model (see Eq 2). As per the modified model, at any time t in a creep test beyond 1 h, engineering creep strain ϵ_t is given as

$$\log(\epsilon_t) = \left[f_s \left(\frac{\log(t_r) + \beta}{\log(t_\epsilon) + \beta} - 1 \right) + f_j \right]^{1/p} - C \quad (2)$$

where:

$$f_s = \frac{\beta}{\log(t_r)} [(\log(\epsilon_1) + C)^p - (\log(\epsilon_{ru}) + C)^p]; \quad f_j = (\log(\epsilon_{ru}) + C)^p$$

$p(\sigma, T) = p_0 + p_1 \log(\sigma) + [p_2 / (T + 273)]$ and β , p_0 , p_1 , p_2 , and C are fitting parameters,

T = test temperature ($^{\circ}\text{C}$),

σ = applied external stress (MPa),

ϵ_t = engineering creep strain at time t ,

t_r = time to failure by creep rupture (h),

t_ϵ = time to given engineering creep strain (h), and

$\epsilon_1, \epsilon_{ru} = \epsilon_t$ at time $t = 1$ h and uniform strain at rupture, respectively.

As noted earlier, typically from 95 % of t_r and beyond, macro-cracking starts to appear because of localized necking and results in an unstable specimen

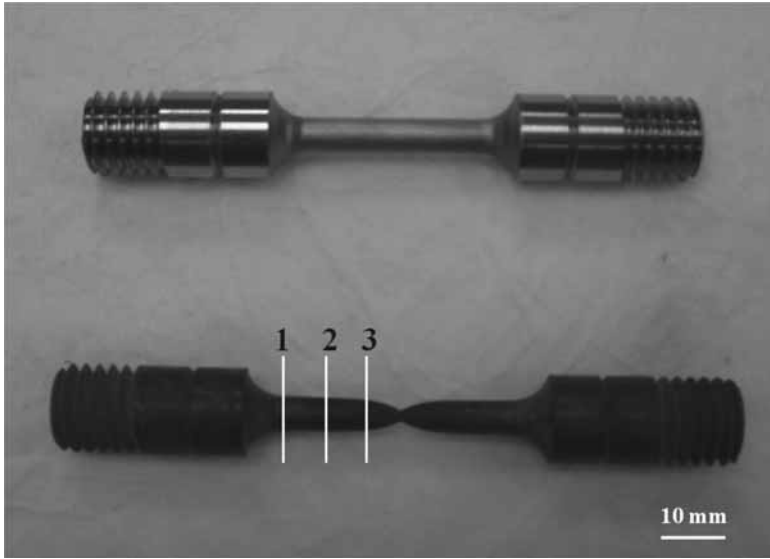


FIG. 7—Pictorial representation to empirically measure the uniform strain at rupture, ϵ_{ru} . Vertical lines numbered 1, 2, and 3 indicate locations where diameter is measured to compute d_{unif} . Untested specimen in the top is kept as a reference for the creep ruptured specimen below.

response thereafter. We also know that the reduction in area varies significantly based on the necking characteristics. Hence, the engineering creep strain at rupture (ϵ_t at time $t = t_r$) is taken to be the measured uniform strain at rupture, ϵ_{ru} (i.e., strain generated in the specimen's gage length before the onset of necking in a localized region). This strain can be empirically obtained by evaluating the reduction in area (A_{red}) in the specimen gage length in regions away from the neck. Rather than using the conventional ductility equation for A_{red} , an analytical form as shown in Eq 3 is employed in this work. As shown in Fig. 7, the diameter of the specimen at locations 1, 2, and 3 is measured and averaged to obtain d_{unif} and this value is used to compute A_{unif} .

$$A_{red} = \frac{A_{orig} - A_{unif}}{A_{orig}} \quad (3)$$

where:

A_{orig} = original cross-sectional area in the specimen gage length and

A_{unif} = uniform cross-sectional area obtained by using d_{unif} (the average of diameters in regions away from the neck).

Minimizing human measurement errors and the effect of complex strain profiles across the specimen gage length were the primary reasons for using the average value of d_{unif} rather than using a single diameter value in the “unnecked” region. The choice of position of locations to compute d_{unif} does not

TABLE 5—Modified LCSP model fitting constants for P91 steel at 625°C.

Parameter	Value
β	0.27
p_0	-1.65
p_1	-9.15
p_2	2553.74
C	6.8

seem to play a significant role, as long as they are chosen to be approximately equidistant (to capture the diameter variation across the specimen gage length of the creep ruptured specimen) away from the neck to the edge of the gage length markers. Also, it was observed in this work that a choice of three such locations is optimal and a larger number of locations does not improve the consistency in the average value of d_{unif} .

To perform non-linear regression (NLR) of the available creep rupture data, the SOLVER[®] software routine, available as an additional plug-in in Microsoft Excel, is employed in this work. A few of the tested creep rupture data were randomly chosen and few commercial NLR softwares including SOLVER[®] were used to non-linearly fit this data by minimizing the squares of errors (difference between the actual and predicted strain values). It was found that SOLVER[®] did equally as well, if not better, in all of these cases. Thus, the performance of this package as compared with the commercially available NLR softwares was carefully analyzed and evaluated before its eventual selection. This routine searches for the set of values of the fitting factor constants or parameters that minimize the sum of squares of residuals, i.e., of differences between each observation and the corresponding prediction. From the available test data for P91 steel, the modified LCSP model provided the fitting constants as shown in Table 5. As can be seen, the model has relatively fewer fitting constants as compared to the other models and is consequently more robust [18]. The fitted strain versus time histories are compared to actual measurements in Fig. 5. The model successfully describes the data at various stress levels. Figure 8 demonstrates the fitting efficiency of the modified LCSP model for P91 steel by comparing predicted and measured creep rates. Once again, the prediction seems to correlate well with the experimental data.

Creep-Fatigue Tests

The creep-fatigue round-robin test matrix has been designed to address precision in the results obtained from tests when utilizing the procedure outlined in the new standard E2714-09. More specifically, the objective is to quantify intra- and inter-laboratory variability in the creep-fatigue data. It is anticipated that the standard will be revised as warranted by the results of the round-robin. For example, questions such as what percentage drop in the force provides the most consistent measure of the number of fatigue cycles for crack formation are

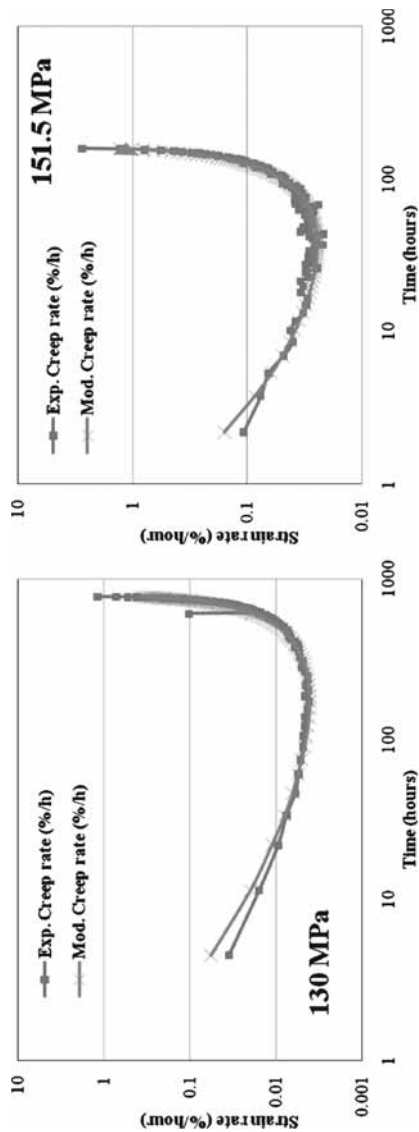


FIG. 8—Predicted creep strain rate by modified LCSP model for P91 steel as compared with that of experimental values for two different test conditions at 625°C.

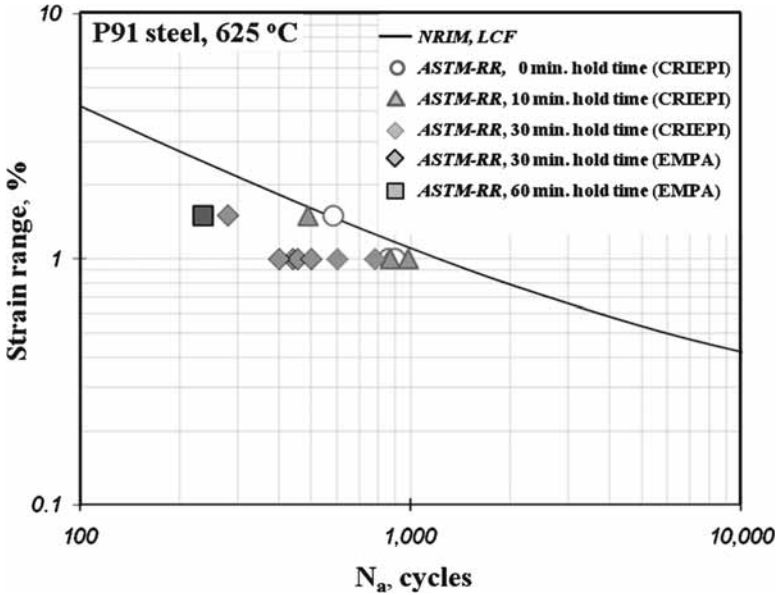


FIG. 9—Pilot tests' results from creep-fatigue testing of P91 steel, where N_a is the number of cycles to failure. Data courtesy of Dr. Stuart Holdsworth of EMPA, Switzerland and Dr. Yukio Takahashi of CRIEPI, Japan.

expected to be addressed. As noted earlier, 16 laboratories worldwide are involved in conducting the creep-fatigue testing as per the test matrix given in Table 1. The creep-fatigue tests are being conducted at 625°C at strain amplitudes of ± 0.25 , ± 0.5 , and ± 0.75 %. The hold times for the tests are 0, 10, and 30 min at the peak tensile strain. The strain rate chosen for loading and unloading in a triangular wave-form will be 0.025 % per second. The participants are asked to report the number of cycles at various level drops in force starting with a 2 % drop. To determine the creep-fatigue round-robin test conditions, pilot tests were conducted by Dr. Stuart Holdsworth of Swiss Federal Laboratories for Materials Science and Technology (EMPA), Switzerland and Dr. Yukio Takahashi of Central Research Institute of Electric Power Industry (CRIEPI), Japan and the results of these pilot tests are shown in Fig. 9.

These results demonstrate that the selected test conditions reported in Table 1 meet the criteria for round-robin testing. The test temperatures and hold times chosen are sufficiently high for creep-fatigue interactions to be present during the testing.

Summary

In order to further refine the precision and bias statements for the recently developed ASTM standard E2714-09 for creep-fatigue testing, a round-robin program has been initiated on P91 steel with the participation of 16 laboratories

worldwide. The specimen machining, material characterization, and pilot test studies to finalize test parameters for this round-robin have been completed. As part of this round-robin, this paper also reports the results of microstructural analysis, tensile and creep deformation, and rupture tests performed *thus far* on P91 steel. Preliminary microscopic analysis of ruptured creep specimens shows that material undergoes predominantly transgranular (ductile) fracture with an array of microvoids found most commonly near precipitates and/or secondary phases. A modified LCSP model has been successfully developed to accurately represent the creep deformation characteristics of this material with the available data. Some preliminary results from the creep-fatigue tests are also presented.

Acknowledgments

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A. Mehmanparast,¹ C. M. Davies,¹ and K. M. Nikbin¹

Evaluation of the Testing and Analysis Methods in ASTM E2760-10 Creep-Fatigue Crack Growth Testing Standard for a Range of Steels

ABSTRACT: The development of the ASTM E2760-10, “Creep/Fatigue Crack Growth Testing” standard has initiated a phase of testing, analysis, and round robin initiatives associated with high temperature cyclic loading crack growth tests. Creep and fatigue are two complex independent mechanisms which may assist each other to drive the crack. A simplification of the macro creep/fatigue crack growth response of structures is therefore required. Using a linear cumulative damage method, as proposed in ASTM E2760, and other codes of practice, data for a range of steels are analysed. Limited crack growth data, using test methods similar to ASTM E2760, from previous collaborative projects for a range of steels are presented and analysed to identify the difficulties in the process. Creep-fatigue crack growth tests at frequencies ranging from 10 to 0.001 Hz are examined to consider the testing methodologies and in order to assess the effect of data scatter, dwell times, and the interaction region between creep crack growth (CCG) and fatigue crack growth mechanisms. Given the limited data available and the level of scatter, it can still be shown that the linear cumulative summation of static and high frequency data may be sufficient to predict creep/fatigue interaction both in terms of C^* and ΔK , and the level of creep ductility will dictate appropriateness of the correlating parameter employed. Finally, in order to show that the level of data scatter can be dealt with in a predictive manner, the crack initiation and growth rates are also analysed using the NSW CCG model.

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¹Dept. of Mechanical Engineering, Imperial College London, South Kensington Campus, London SW7 2AZ, United Kingdom.

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KEYWORDS: Creep-Fatigue interaction, P91, P22, 1CrMoV, 316L(N), ASTM E2760-10, Crack Growth

Introduction

High temperature components are generally subjected to various stress and thermal cycles resulting from, e.g., start-ups and shut downs in power plants. This load cycling at high temperatures may cause creep-fatigue interaction effects that can lead to component failure through crack initiation and growth. The interaction of these two failure mechanisms are complex and difficult to quantify and predict in any one cycle. A phenomenological methodology is therefore required to analyse crack growth results derived from standard test pieces in order that the data can be used as a “material property” to predict failure in plant components. High frequency fatigue crack growth tests and static creep crack growth testing are more likely to be performed in order to understand the influence of stress cycling on high temperature cracking. However, it is essential that in some cases combined creep-fatigue tests are performed over a range of load-cycling frequencies to identify interaction effects if they exist. This will enable enhanced and more reliable component lifetime predictions to be achieved. The aim of ASTM E2670 is therefore to standardise and unify the testing and analysis methods to perform these complex tests.

Static creep crack growth tests and constant temperature cyclic load tests have previously been conducted [1] on compact tension, *C(T)*, specimens from a range of ferritic steels 1CrMoV (cast), P22 and P91 and austenitic type 316L(N) steel. The testing temperature employed for each material is identified in Table 1. The cyclic load tests have been performed at a range of frequencies (around 0.0008–10 Hz).

Data from these tests have been collected and analysed in order to identify the trends in the different material behaviours, and highlight the issues

TABLE 1—*Creep and fatigue crack growth properties for different steels.*

Material	Condition	Temperature (°C)	CT size	<i>D</i>	ϕ	<i>f</i> (Hz)	λ	<i>p</i>
316L(N)	Base	650	CT25	1.0E+00	0.66	0.1	1.3E−08	3.34
						10	4.7E−09	3.34
316L(N)	Base	650	CT50	1.8E+00	0.69	...	...	...
1CrMoV	Cast	530	CT25	7.0E+00	0.72	0.01	3.5E−06	1.48
						0.1	1.5E−06	1.48
1CrMoV	Cast	530	CT50	2.9E+01	0.93	...	...	...
P22	Base	565	CT25	2.3E+00	0.64	0.1	4.5E−09	3.26
						1	2.6E−09	3.26
P22	Base	565	CT50	4.5E+00	0.69	...	...	...
P91	Base	625	CT25	7.6E+00	0.80	0.001	1.5E−08	3.57
						0.01−1	2.0E−09	3.57
P91	Base	625	CT50	4.4E+00	0.69	...	...	...

associated with interactive creep-fatigue crack growth testing and data analysis. First, an outline of the testing and data analysis procedure is presented and then the results from each material are described.

Testing Procedure

The ASTM test procedures E1457 [2] and E647 [3] provide guidance on static creep crack growth and fatigue crack growth testing, respectively. Standardised test procedures under combined creep-fatigue cyclic loading conditions at high temperatures have not been widely established. ASTM E2670 aims to fill this gap and unify the testing methods under creep/fatigue conditions. An outline of the procedure employed for the tests analysed in this work is given below.

Specimen Geometry

The compact tension, $C(T)$, specimen geometry was employed as recommended in fatigue and creep crack growth testing standards [2,3], thus the thickness to width ratio, B/W , was 0.5 and the initial crack length, a_0 , was half the specimen width ($a_0/W=0.5$). In order to examine any size effects, specimens of width $W=25$ and 50 mm were examined.

Cycle Shape

A schematic illustration of loading level versus time is shown in Fig. 1. A constant loading R -ratio of 0.1 was used throughout. In the low frequency creep-fatigue tests ($f < 0.1$ Hz), the load ramping rates were such that the load increase/decrease may be considered to occur almost instantaneously hence giving a square cycle loading, as schematically illustrated in Fig. 1(a). The hold time during the creep dwell (*i.e.* time at max load) and time on the minimum load were equal. The time for the ram to lower and rise was approximately 5 and 1 s, respectively. For the relatively high frequency fatigue tests ($f \geq 0.1$ Hz) a sinusoidal cycle shape was exhibited, as illustrated in Fig. 1(b).

Crack Growth Measurements

The potential drop (PD) technique has been used to measure the extent of crack growth in each cycle. Both AC and DC PD techniques were employed in different cases. The load line displacement, force, the PD voltage and test temperature were continuously monitored during testing. The tests are normally stopped prior to full fracture, when rapid increase in PD and displacement is observed. The specimen is then broken open using a cyclic load at room temperature. The termination of the test prior to fracture helps to distinguish the extent of crack growth due to fatigue from the fast fracture and facilitates the metallurgical analysis subsequent to test completion [4]. The initial and final crack lengths can be measured using a 9 point average measurement across the specimen thickness [2] or more recently by measuring the crack area and dividing it by the thickness of the fracture surface (see, e.g., Ref. [5]). The former technique was used in this case.

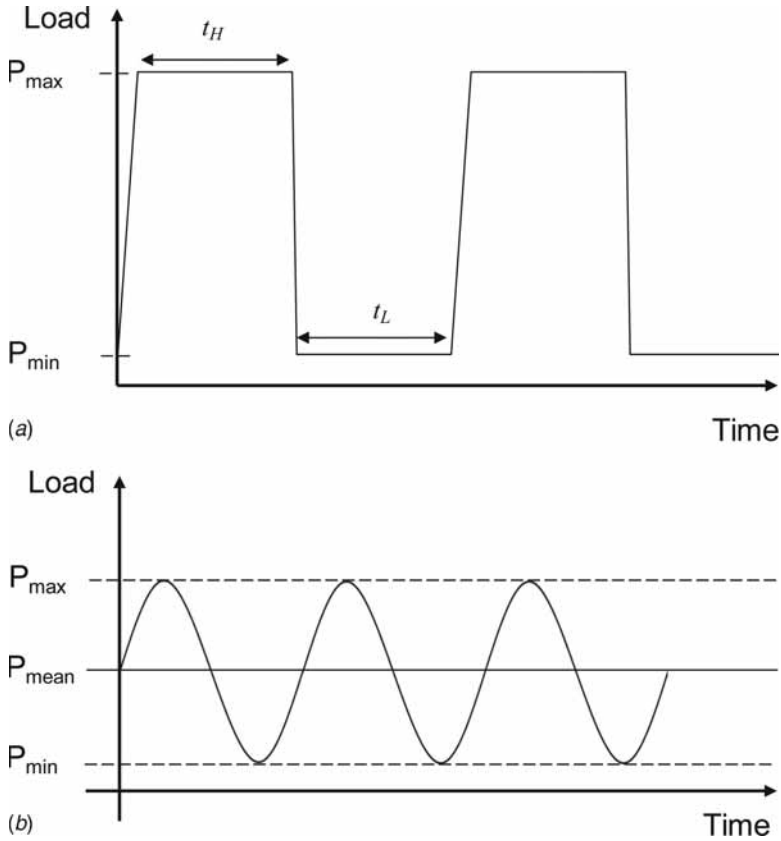


FIG. 1—Schematic illustration of (a) square loading cycles for the low frequency creep-fatigue tests ($f < 0.1$ Hz) and (b) sinusoidal loading cycles for the high frequency tests ($f \geq 0.1$ Hz).

Having measured the initial and final crack length, the instantaneous crack extension can be calculated from

$$a = a_0 + (a_f - a_0) \times \frac{V - V_0}{V_f - V_0} \quad (1)$$

where a_0 , a_f , and a are the initial, final and instantaneous crack lengths, respectively, and V_0 , V_f , and V are the corresponding initial, final, and instantaneous value of the PD voltage.

Data Analysis Procedure

The fracture mechanics parameters employed to correlate the data are ΔK and C^* . Where creep is dominant then C^* at maximum load correlates the data and

where fatigue is dominant ΔK is used. The extent of creep ductility will also dictate the correct choice of parameter. In high temperature cyclic tests, creep is generally considered to be the dominant failure mechanism when the frequency is sufficiently low ($f < 0.01$) and thus the crack behaviour is expected to be similar to static CCG, whilst at higher frequencies ($f > 0.1$ Hz) fatigue damage dominates. At intermediate frequencies, an interaction between creep and fatigue failure mechanisms may lead to failure. In order to understand and analyse combined creep-fatigue mechanism, both static creep and high cycle fatigue data analysis procedures are next described.

Creep Crack Growth Rates

At long times, where a steady state of creep deformation and damage has developed at a crack tip, the CCG rate, $\dot{a}$, may be described by the crack tip parameter C^* according to the power law relationship

$$\dot{a} = DC^{*\phi} \quad (2)$$

where D and ϕ are material constants, which may be temperature and stress state dependent [6], and $\dot{a}$ versus C^* data appear as a straight line on log-log axes [6]. Prior to steady state conditions being achieved, data points will appear as a "tail" on the $\dot{a}$ versus C^* plot.

The C^* parameter may be determined experimentally in a CCG test from the load line displacement rate measurements, $\dot{\Delta}$, using the relation [7]

$$C^* = \frac{P \dot{\Delta}}{B_n(W-a)} H \eta \quad (3)$$

where $\dot{\Delta}$ is the load line displacement rate, P is the applied load, B_n is the net thickness between the side-grooves, W is the specimen width, and H and η are geometry dependent constants. For a $C(T)$ specimen $H = n/(n+1)$ and $\eta = 2.2$ where n is the creep power law stress exponent [7]. Note that in the case of a cyclic load crack growth test, $\dot{\Delta}$ in Eq (3) is interpreted as the load line displacement rate at max load.

For situations where the crack tip deformation is predominantly elastic (e.g., short times and/or high crack velocity) it is expected that crack growth will be controlled by the linear elastic stress intensity factor K

$$\dot{a} = D'K^m \quad (4)$$

where D' and m are material constants, which may be temperature and stress state dependent.

Creep Crack Initiation Time

During a static creep crack growth test a significant period of time may elapse prior to the commencement of crack extension from a pre-existing defect due to

creep strain/damage accumulation at the crack tip. This period of time is generally referred to as the creep crack initiation (or incubation) time. Creep crack initiation (CCI) is often defined as the time for a small measurable crack extension, Δa , of 0.2 or 0.5 mm [2]. Provided that CCI occurs under predominantly C^* controlled steady state creep conditions, the initiation time, t_i , may also have a power-law relation with the C^* parameter. The initiation time may be estimated by assuming that CCG is a continuous process that occurs on initial loading under steady state creep conditions (i.e., steady state distribution of damage and C^* crack tip field control [6]). Thus the initiation time for a crack extension, Δa , may be predicted using

$$t_i = \frac{\Delta a}{DC^*\phi} \quad (5)$$

where the constants D and ϕ , are as defined in Eq (2).

An initiation time may not exist for fatigue crack growth conditions. For lifetime prediction purposes it is conservative to assume that there is no CCI period; however this may lead to overly conservative results.

Creep Crack Initiation and Growth Models

It is important to present models that can be used to predict cracking in order to validate them with actual data. The NSW crack growth models [6,8] can predict creep crack growth under static load conditions, where creep dominates, and consider the effect of constraint due to geometry, creep properties, and loading. Therefore it could be appropriate to use the same models for the low frequency cyclic loading conditions with a hold time in which crack growth may occur due to a combination of low order fatigue effects and a dominant creep damaging mechanism. Furthermore, any deviation of the cyclic load crack growth data from the NSW model could indicate a significant fatigue contribution to crack growth and possibly fatigue dominant conditions which may occur at higher frequencies. The NSW models are described below and will therefore be used to compare the cyclic load crack growth behaviour to static load creep crack growth predictions in order to give an insight into the competing damage mechanisms as well as to predict the overall crack initiation and growth rates in the steels presented.

NSW-MOD Crack Growth Model—The NSW-MOD model [8] has been derived to predict the CCG rate under steady state creep conditions and considers the dependence of creep strain on both the crack tip angle, θ , and the creep stress exponent, n , in addition to the stress state. Crack growth is considered to occur when a critical quantity of damage, measured by a creep ductility exhaustion approach, is attained at a characteristic distance, r_c , ahead of the crack tip. The creep crack growth rate prediction from this model is given by

$$\dot{a}_{\text{NSW-MOD}} = (n+1)(Ar_c)^{1/n+1} \left[\frac{C^*}{I_n} \right]^{n/n+1} \frac{\tilde{\epsilon}_{\text{eq}}(\theta, n)}{\epsilon_f^*(\theta, n)} \Big|_{\text{max}} \quad (6)$$

Crack growth is predicted using the maximum value of the angular function $\tilde{\epsilon}_{eq}/\epsilon_f^*$, where $\tilde{\epsilon}_{eq}$ is a non-dimensional function of θ and n , values of which are tabulated in Ref. [9]. The value of the multiaxial ductility, ϵ_f^* , for a range of stress states may be estimated from the uniaxial failure strain, ϵ_f , using an appropriate model, such as the Cocks and Ashby void growth and coalescence model [10] which has been applied in this study. The characteristic distance r_c is generally taken to be the grain size of the material. In Eq (6) I_n is a dimensionless integration constant that depends on n .

The CCI time, defined as the time for a small measurable crack extension, Δa , may be estimated by assuming that the crack grows from the point of initial loading at a constant rate. Lower and upper bound predictions of the CCI time may be obtained using either the steady state CCG rate, $\dot{a}_s$, or the initial CCG rate (CCG rate at initial crack length, a_0), $\dot{a}_0$, respectively. The initial CCG rate, $\dot{a}_0$, for a given value of C^* may be estimated to be $(n+1)$ times less than the steady state value [6], hence the CCI time is predicted from

$$\frac{\Delta a}{\dot{a}_s} \leq t_i \leq \frac{\Delta a}{\dot{a}_s} (n+1) \quad (7)$$

Using the NSW-MOD CCG rate prediction, the CCI time may be predicted using

$$\frac{\Delta a}{(n+1)(Ar_c)^{1/n+1}} \left(\frac{I_n}{C^*} \right)^{n/n+1} \frac{\epsilon_f^*}{\tilde{\epsilon}_{eq}} \Big|_{\max} \leq t_i \leq \frac{\Delta a}{(Ar_c)^{1/n+1}} \left(\frac{I_n}{C^*} \right)^{n/n+1} \frac{\epsilon_f^*}{\tilde{\epsilon}_{eq}} \Big|_{\max} \quad (8)$$

Approximate NSW Crack Growth Model (NSWA)—It has been found that for a range of steels the CCG rate is most sensitive to the multiaxial creep ductility ϵ_f^* and that the steady state CCG rate may be reasonably approximated by [11]

$$\dot{a}_s^{\text{NSWA}} = \frac{3C^{*0.85}}{\epsilon_f^*} \quad (9)$$

where $\dot{a}$ and C^* have the units mm/h and MPam/h, respectively, and ϵ_f^* is taken as the uniaxial failure strain, ϵ_f , for plane stress conditions and $\epsilon_f/30$ for plane strain [12].

Following Eq (7), the CCI time may also be estimated from the NSWA model using

$$\frac{\Delta a \epsilon_f^*}{3C^{*0.85}} \leq t_i \leq \frac{\Delta a (n+1) \epsilon_f^*}{3C^{*0.85}} \quad (10)$$

Fatigue Crack Growth Correlation

Fatigue crack growth is a simpler mechanism which is largely independent of constraint effects. Under predominately fatigue controlled conditions, the crack growth per cycle N , may be described by the Paris law [13]

$$\frac{da}{dN} = \lambda \Delta K^p \quad (11)$$

where ΔK is the stress intensity factor range, λ is a material parameter which depends on the R ratio, and p is a material constant.

Creep-Fatigue Crack Growth Rate

The crack growth rate can be calculated by plotting the crack length versus time or cycle. The crack growth rate, da/dt , can be converted to the amount of crack growth per cycle, da/dN , using the following relationship

$$da/dN = \frac{da/dt}{3600f} \quad (12)$$

where f is the frequency. The total crack growth per loading cycle will be attributed to both creep and fatigue deformation. The total combined crack growth rate per cycle can be conceded as a linear summation of the creep and fatigue contribution, i.e.

$$\left(\frac{da}{dN}\right)_{\text{total}} = \frac{(da/dt)_{\text{creep}}}{3600f} + \left(\frac{da}{dN}\right)_{\text{fatigue}} \quad (13)$$

where f is the frequency of the load cycle in Hz. The creep crack growth rate for a given value of C^* is given by Eq (2), and the fatigue contribution for the corresponding value of ΔK obtained from Eq (11). Hence, Eq (13) may be rewritten as

$$\left(\frac{da}{dN}\right)_{\text{total}} = \frac{DC^{*\phi}}{3600f} + \lambda \Delta K^m \quad (14)$$

Results

In the proceeding sections, the data from four steels are analysed and compared. The data are analysed in terms of crack growth initiation and growth rate and the NSW models described above are used to predict the bounds for the data presented.

Comparison of Crack Growth Behaviour

The crack extension versus time data for all tests are compared for each material in Figs. 2–5. In these figures the time is normalized by the test duration, t_f . Note that the final value is subject to the point where the test was stopped prior to full specimen fracture. First, the results for 316L(N) at 650°C are presented in Fig. 2. In these figures static tests are represented by black and white open symbols and crosses, whereas the cyclic tests are shown in colour. It is clearly seen in Fig. 2 that the crack growth rates in cyclic creep-fatigue tests accelerates with

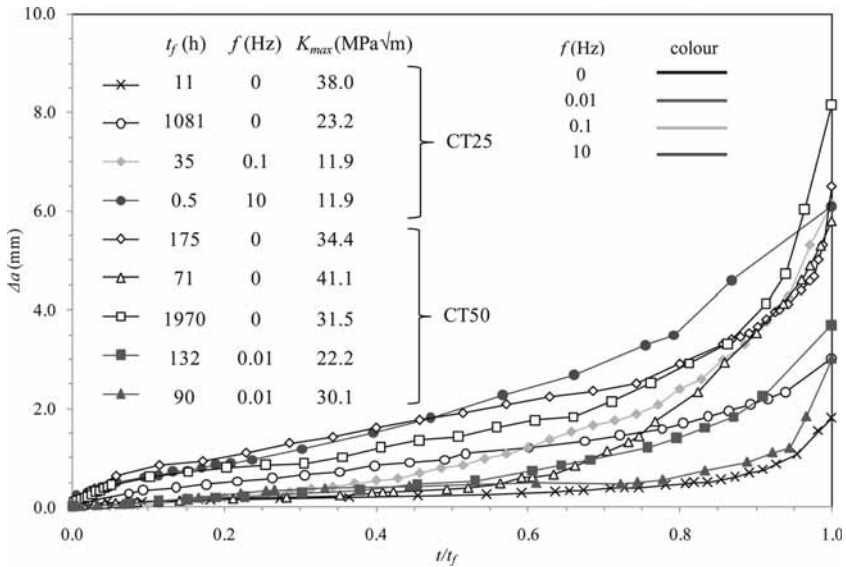


FIG. 2—Comparison of the crack extension versus normalized time data for static and cyclic crack growth tests on 316L(N) parent material C(T) specimens at 650°C.

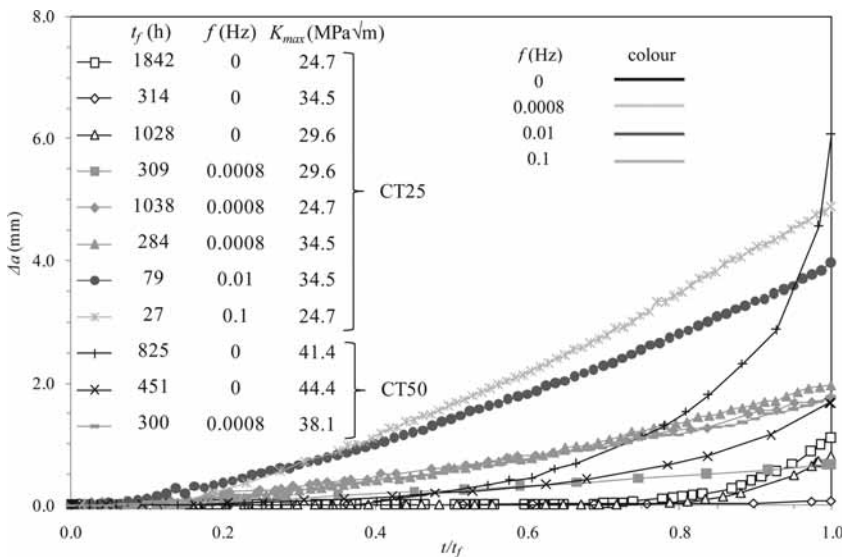


FIG. 3—Comparison of the crack extension versus normalized time data for static and cyclic crack growth tests on 1CrMoV (cast) material C(T) specimens at 530°C.

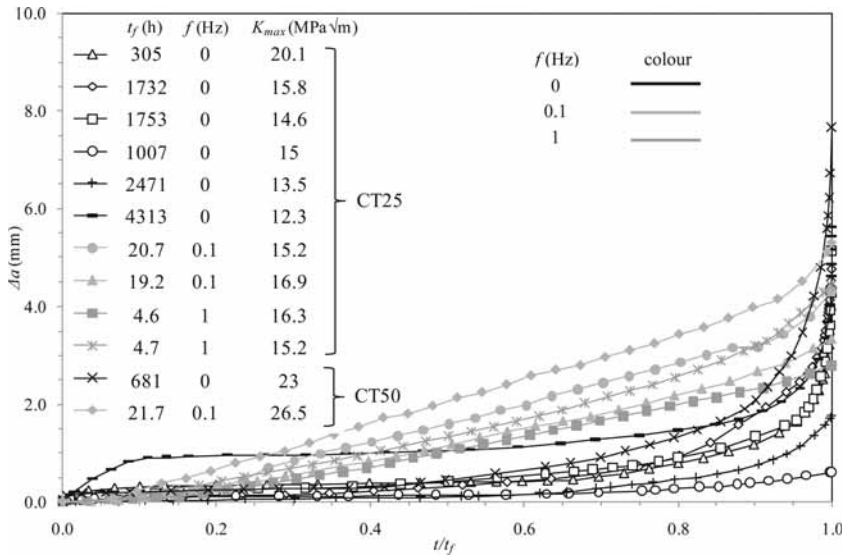


FIG. 4—Comparison of the crack extension versus normalized time data for static and cyclic crack growth tests on P22 parent material circumferential direction C(T) specimens at 565°C.

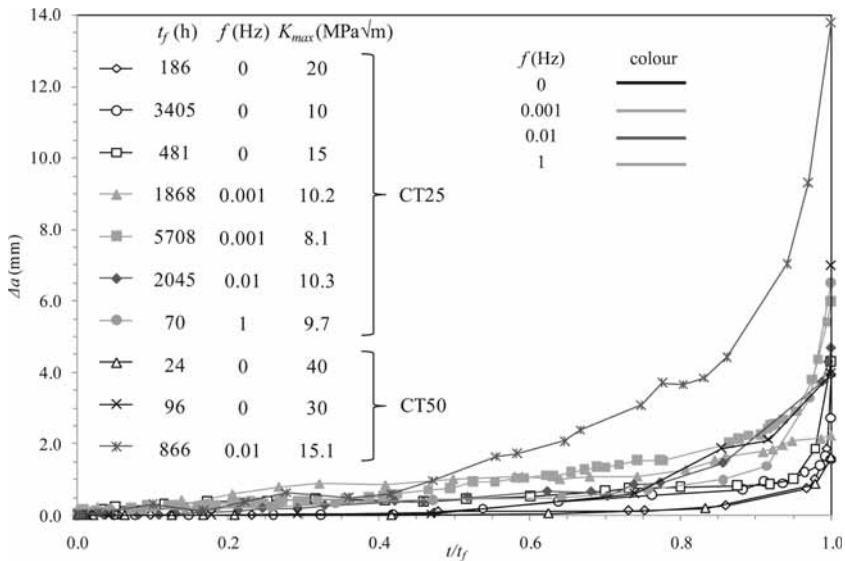


FIG. 5—Comparison of the crack extension versus normalized time data for static and cyclic crack growth tests on P91 parent material circumferential direction C(T) specimens at 625°C.

an increase in frequency and subsequently the time to failure drops significantly. A relatively long period of time is observed for the low frequency tests (shown in red) before any significant crack extension occurs. It may also be inferred that during the respective test durations, larger crack extensions are exhibited in the CT50 specimens (of width $W = 50$ mm) compared to that of the CT25 specimens (of width $W = 25$ mm).

Similar trends are generally seen in Figs. 3, 4, and 5 for 1CrMoV (cast), P22, and P91, respectively, within the inherent data scatter. For the majority of the static CCG tests, as seen particularly for the ferritic steels in Figs. 3–5, little crack extension is observed for the first 80% of the tests duration, corroborating the concept of the creep crack initiation/incubation period. It is clear from these figures that the failure time in creep-fatigue tests is a function of initial applied load and frequency, such that the specimen fails faster with an increase in any of these parameters.

Crack Growth Rate Correlations with ΔK

The crack growth per cycle, da/dN , has been plotted against the stress intensity factor (SIF) range, ΔK , for the cyclic load creep-fatigue tests in Fig. 6–9. For clarity some extensive tails have been removed from these figures. A regression fit has been made to the high cycle data ($f \geq 0.1$ Hz) and the Paris law fatigue crack growth (FCG) constants determined (see Eq (11)) and are given in Table 1. In addition, a regression fit has been made to intermediate frequency data, assuming the same slope as the high frequency data, and the influence of frequency on the crack growth rate per cycle determined. In Figs. 6–9, the high

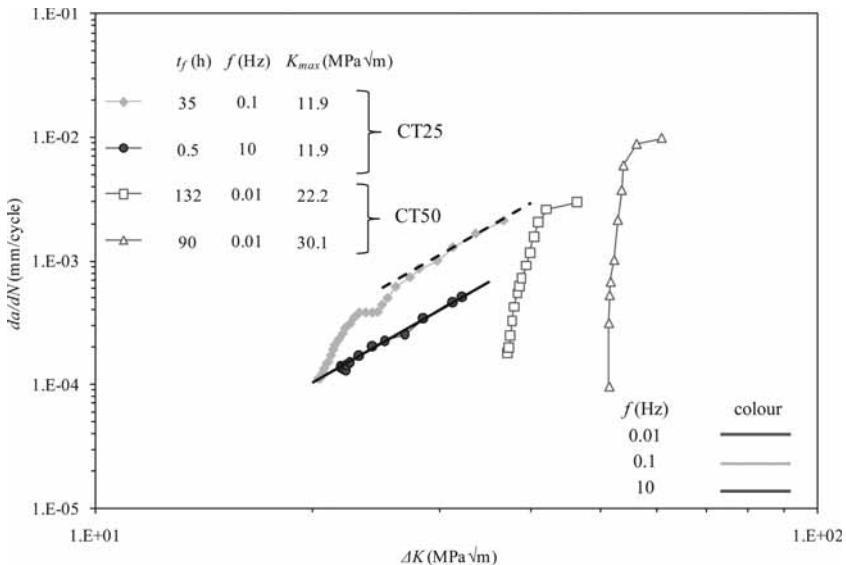


FIG. 6—Comparison of crack growth per cycle versus ΔK data for cyclic crack growth tests on 316L(N) C(T) specimens at 650°C.

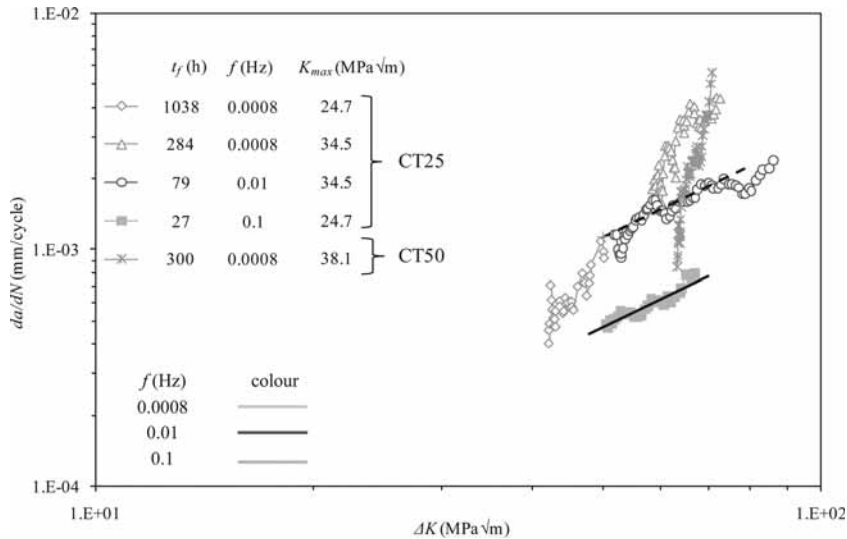


FIG. 7—Comparison of crack growth per cycle versus ΔK data for cyclic crack growth tests on 1CrMoV C(T) specimens at 530°C.

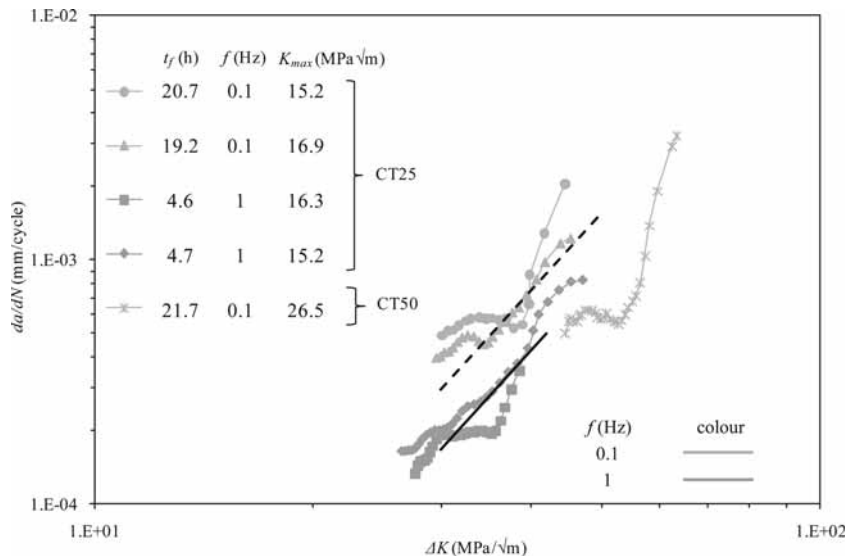


FIG. 8—Comparison of crack growth per cycle versus ΔK data for cyclic crack growth tests on P22 C(T) specimens at 565°C.

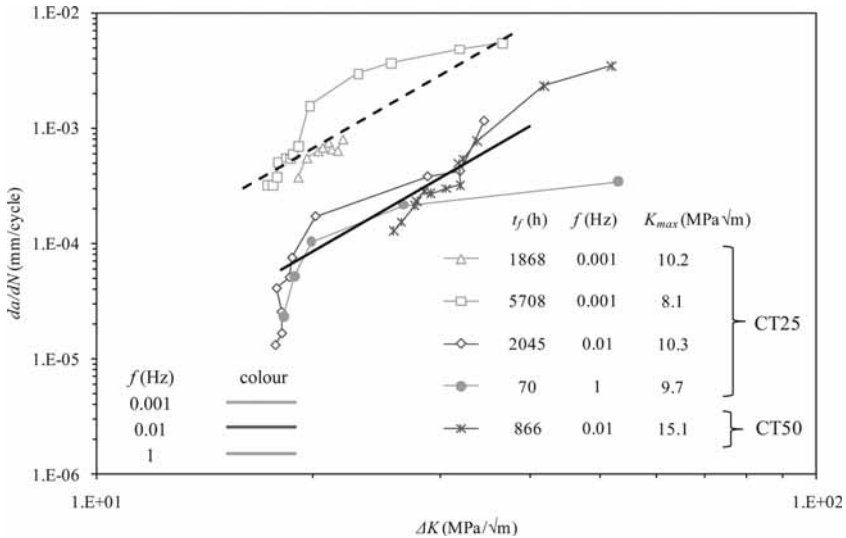


FIG. 9—Comparison of crack growth per cycle versus ΔK data for cyclic crack growth tests on P91 C(T) specimens at 625°C.

frequency ($f \geq 0.1$ Hz) cyclic tests on the CT25 geometry, which have been considered to be fatigue dominant specimens, have been shown in solid symbols. Due to insufficient number of cyclic tests on CT50, the Paris law constants have not been determined for the larger geometry and hence no values for CT50 are provided in Table 1.

The FCG analysis on 316L(N) is shown in Fig. 6. A linear relationship between da/dN and ΔK can be clearly seen for high frequency creep-fatigue test ($f = 10$ Hz). A similar relation is also seen for part of the $f = 0.1$ Hz data; however, the FCG rate for a given ΔK is approximately 2.8 times higher than that for $f = 10$ Hz, indicating that the contribution of creep to cracking is significant. At lower frequencies however no similar correlation between da/dN and ΔK are seen, indicating that fatigue is not the predominant mechanism for crack growth in these tests.

Similar results are generally seen in Fig. 7–9 for the ferritic steels. Due to an inadequate number of tests on the CT50 geometry, no conclusions can be made on the influence of specimen size on the da/dN versus ΔK crack behaviour. Note that there is no data at $f = 10$ Hz for the 1CrMoV material in Fig. 7, however in this case a similar trend is seen between the data for $f = 0.1$ Hz and $f = 0.01$ Hz, and a power law fit has been made to these data sets, assuming a constant slope (given by a regression fit to the $f = 0.1$ Hz data). For a given ΔK , da/dN is approximately 2.4 times higher for $f = 0.01$ Hz than that for $f = 0.1$ Hz.

For the P22 steel, shown in Fig. 8, data for only 1 Hz and 0.1 Hz are available. On average, the 0.1 Hz data for CT25 is a factor of 1.7 higher than that for 1 Hz. For P91 however, it can be seen that the data for 1 and 0.01 Hz fall about the same trend, and the trend line for the 0.001 Hz also follows a similar slope.

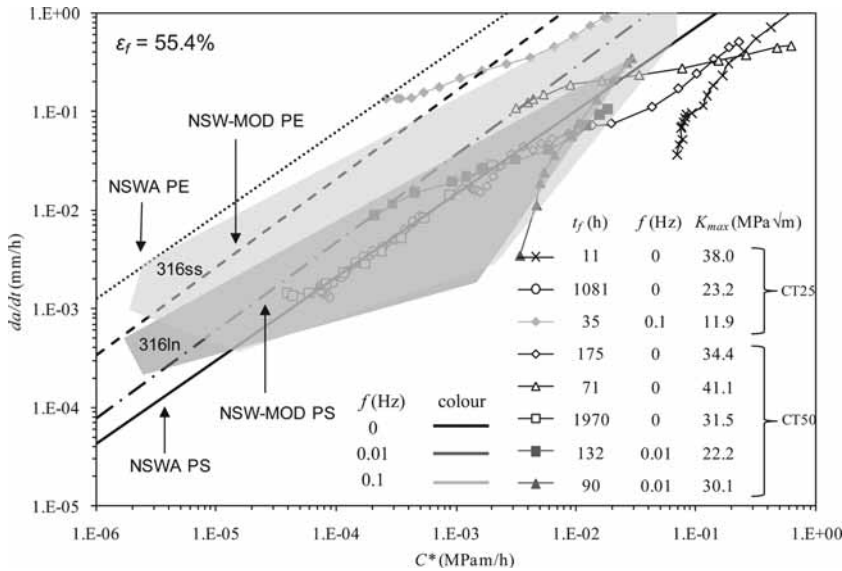


FIG. 10—Comparison of CCG rate versus C^* data of cyclic crack growth tests with static CCG tests on 316L(N) parent material $C(T)$ specimens at 650°C. Also included are data bands for 316ss at 550–800°C and 316L(N) for 550–650°C [14].

For a given ΔK , the data is around 8 times higher for $f=0.001$ Hz than that for $f=1$ and 0.01 Hz. Overall, the rate of increase in crack growth rate appears to increase as the frequency decreases.

*Crack Growth Rate Correlations with C^**

The crack growth rate, $\dot{a}$, in static CCG and low frequency ($f \leq 0.1$ Hz) creep-fatigue tests have been analysed in terms of the creep fracture mechanics parameter, C^* and shown in Figs. 10–13. The results have been compared to NSW-MOD (Eq (6)) and NSWA (Eq (9)) CCG predictive models and to the experimental CCG data bands available within a range of temperatures. In the NSW models calculations, the creep process zone has been taken to be 0.05 mm for all the steels studied in this paper. In addition, the reduction in area measurements (RA) (see Table 2) have been employed for the uniaxial creep ductility, ϵ_f , which is generally considered to give a more accurate measure of the failure properties local to the rupture region, and the minimum creep strain rate properties, given in Table 2, have been used. The properties required for the NSW-MOD model are given in Table 3.

The static CCG data for $\Delta a < 0.2$ mm have been removed from these plots since they are considered to comprise the transient creep region where damage is building up to a steady state at the crack tip, and are hence considered invalid for a correlation with the C^* parameter [2]. A power-law fit, as described in Eq 2, has been made to the data sets for a given geometry, and the constants derived are given in Table 1.

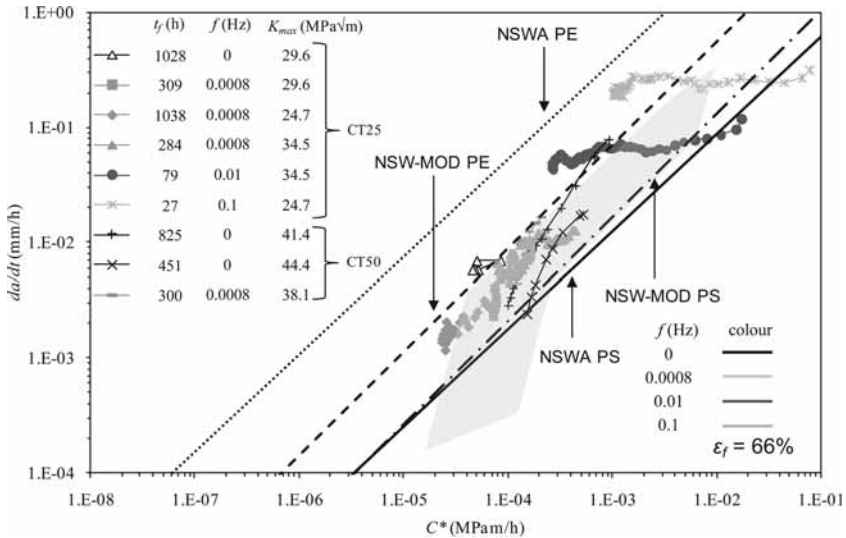


FIG. 11—Comparison of CCG rate versus C^* data of cyclic crack growth tests with static CCG tests on 1CrMoV (Cast) C(T) specimens at 530°C. Also included is a data band for 1CrMoV at 550°C [14].

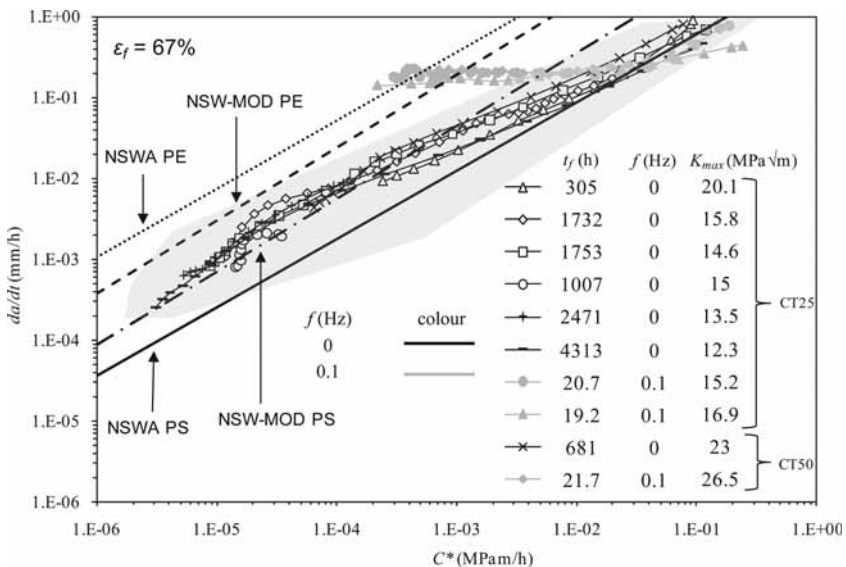


FIG. 12—Comparison of CCG rate versus C^* data of cyclic crack growth tests with static CCG tests on P22 C(T) specimens at 565°C. Also included is a data band for P22 at 538–570°C [14].

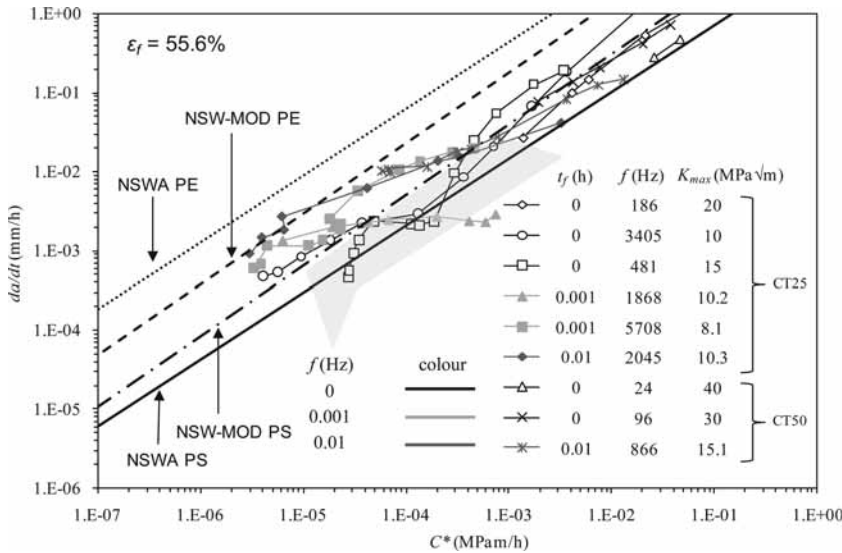


FIG. 13—Comparison of CCG rate versus C^* data of cyclic crack tests with static CCG tests on P91 C(T) specimens at 625°C. Also included is a data band for Parent and Weld material of P91 steel at 538–593°C [14].

Note that in these figures the crack growth rate and LLD rates used for C^* calculations are based on total time. It may be considered that crack growth due to creep occurs during the period of max load, which for this loading cycle, where the time at max load and min load are equal, effectively means that the time for CCG is half the total time. Thus, the crack growth rate and LLD rate would effectively be doubled. However, this does not change the overall data trends since both are increased in proportion. Thus, the data would simply be displaced along its current trend line.

Figure 10 shows the crack growth rate data against the C^* parameter for 316L(N) at 650°C. The static CCG data generally fall upon the plane stress NSW prediction, and the NSW-MOD line provides a conservative prediction to the data. The low frequency (0.01 Hz) data also fall around the plane stress prediction lines. Accelerated cracking, by approximately a factor of 10 for a given value of C^* , is however shown for $f=0.1$ Hz, signifying that C^* alone does not characterise the crack tip conditions in this case. Also included in this figure are the experimental data bands for 316L(N) for a temperature range of 550°C to 650°C, and 316 stainless steel (ss), taken from [14] for a temperature range of 550 to 800°C. All the low frequency (<0.1 Hz) data fall within the 316L(N) data band as expected.

For 1CrMoV material at 530°C, the very low frequency tests (0.0008 Hz) behave as static CCG tests and fall between the plane stress and plane strain NSW-MOD predictions. In addition the majority of the data fall within the experimental data band for 1CrMoV material at 550°C. Almost all of the data are bounded by the NSW prediction lines; however data for $f \geq 0.1$ Hz has a clearly

TABLE 2—Uniaxial creep properties for different steels, $\dot{\epsilon} = A\sigma^n$; $\dot{\epsilon} = [h^{-1}]$, $\sigma = [MPa]$, and $t_r = B_r\sigma^{-10}$; $t_r = [h]$.

Material	Type	Temperature	$A_{\min}$	$\sigma_{0\min}$	$n_{\min}$	A_{ave}	$\sigma_{0\text{ave}}$	n_{ave}	B_r	v_r	ϵ_f	RA
316L(N)	Base	650	9E-24	436.3	8.7	2E-22	383.0	8.4	2.75E+20	8.40	0.59	0.55
1CrMoV	Cast	530	9E-27	910.9	9.2	6E-31	440.6	11.4	4.64E+22	9.43	0.12	0.66
P22	Base	565	1E-25	318.1	10.0	2E-24	280.7	9.7	1.44E+19	8.40	0.37	0.67
P91	Base	625	5E-23	391.9	8.6	1E-29	247.0	12.1	1.78E+18	7.75	0.12	0.56

TABLE 3—NSW-MOD model values for a range of steels.

Material	$\tilde{\epsilon}_{eq}(\theta, n)$		$\epsilon_f^*(\theta, n)/\epsilon_f$		$\epsilon_f(\tilde{\epsilon}_{eq}/\epsilon_f^*)_{max}$		I_n	
	Plane stress	Plane strain	Plane stress	Plane strain	Plane stress	Plane strain	Plane stress	Plane strain
316L(N)	1	0.964	1.485	0.129	1.934	11.889	3.155	4.567
1CrMoV	1	0.962	1.486	0.128	1.931	11.420	3.128	4.532
P22	1	0.959	1.555	0.124	1.927	11.686	3.088	4.480
P91	1	0.964	1.485	0.129	1.934	12.034	3.163	4.577

different trend. For a given C^* the crack growth rate for $f=0.1$ Hz is a factor of around 5 higher than that for $f=0.01$ Hz.

For P22 steel at 565°C (Fig. 12), the static data at low C^* , fall upon the plane stress NSW-MOD prediction and tend to the plane stress NSWA prediction at high C^* and generally falls within the data band. No correlation is, however, shown between the crack growth rate with respect to time and C^* for $f=0.1$ Hz, indicating that fatigue processes dominate for this frequency, material and temperature. Apart from the P22 data for $f=0.1$ Hz in Fig. 12, the crack growth rate is conservatively predicted by the plane strain NSWA crack growth model.

The static and cyclic data shown in Fig. 13 for P91 at 625°C, generally fall between the plane stress and plane strain NSW-MOD predictions and the data band is in agreement with the static CCG data. The cyclic crack growth data for both $f=0.01$ Hz and 0.001 Hz fall close to each other; thus it is apparent that these tests are significantly influenced by the creep processes.

It can be seen in these graphs that the data for $f=0.01$ Hz behaves differently for various materials at the temperatures considered, i.e., for P91 and 316L(N) the $f=0.01$ Hz data generally shows a correlation with the C^* parameter, whereas for 1CrMoV, no correlation is shown between the crack growth rate and the C^* parameter. For $f<0.01$ Hz, a correlation with the C^* parameter has been observed for the cases considered.

A comparison of the creep crack growth data for all the materials tested with $f\leq 0.01$ Hz is given in Fig. 14, for the temperatures corresponding to their operating conditions. For a given value of C^* , similar crack growth rates are shown for 316L(N), P91 and P22 steel, though the test temperature of P22, 565°C is significantly lower than that of P91 and 316L(N) at 625 and 650°C, respectively. The crack growth rate for 1CrMoV steel, at the lower operating temperature of 530°C is on average an order of magnitude less than that of the other steels considered.

*Crack Initiation Time Correlations with C^**

In a creep-fatigue test, the crack growth can be caused by creep, fatigue or the interaction between two mechanisms. Assuming that the crack growth is due to creep and that the transition time is exceeded [15], the initiation (incubation)

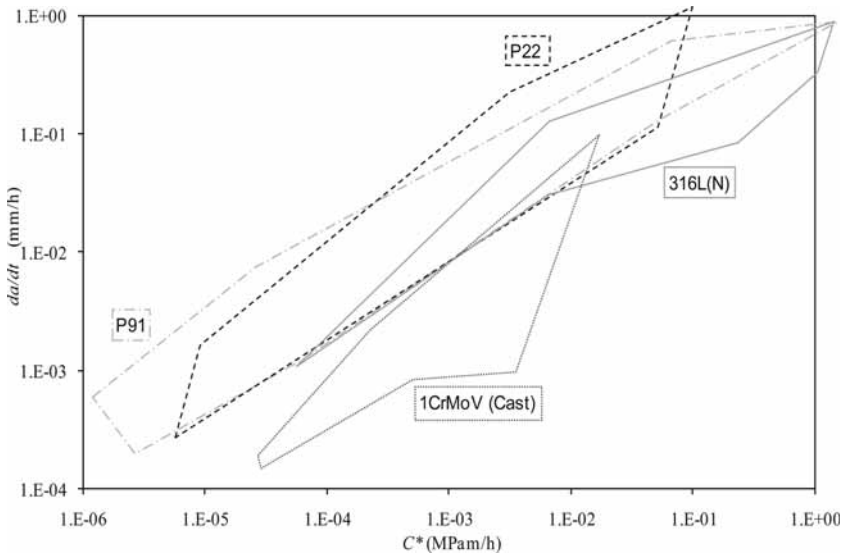


FIG. 14—Comparison of CCG rate versus C^* data bands in cyclic (creep-fatigue) crack growth tests for different steels ($f \leq 0.01$).

time in static CCG tests on $C(T) = 50$ mm and $C(T) = 25$ mm has been analysed using the creep fracture mechanics parameter, C^* .

In practice the time for a small crack extension of 0.2 mm is generally referred to as the creep crack initiation (CCI) time. The creep crack initiation time, t_i , has been correlated with the C^* parameter for all the static CCG tests and compared to that of the cyclic interactive creep-fatigue tests in Figs. 15–18. Also included in these figures are the appropriate predictions from the NSW-MOD and NSW models.

As shown in Eq (7), the upper bound predictions based on the steady state CCG rate is a factor of $(n + 1)$ times higher than the lower bound predictions in NSW initiation models, where n is the creep power law exponent. The two extreme predictions of initiation time are then provided by the lower bound model (based on the steady state CCG rate, $\dot{a}_s$) under plane strain conditions and the upper bound model (based on the initial CCG rate $\dot{a}_0$) under plane stress conditions. The former prediction is the more conservative (i.e., shorter initiation times predicted).

It can be seen in Figs. 15–18 that the static CCG data generally fall between the upper and lower bound plane stress predictions based on the NSW-MOD model. The static CCG data for 1CrMoV at low C^* fall close to the plane strain NSW-MOD prediction line calculated based on $\dot{a}_s$, whereas the data for higher C^* lie closer to the plane stress prediction line. Generally the cyclic data follows the same trend as the NSW models' prediction lines; however, the initiation times for the cyclic tests are generally around 10 times less than that for the static CCG data, for a given value of C^* .

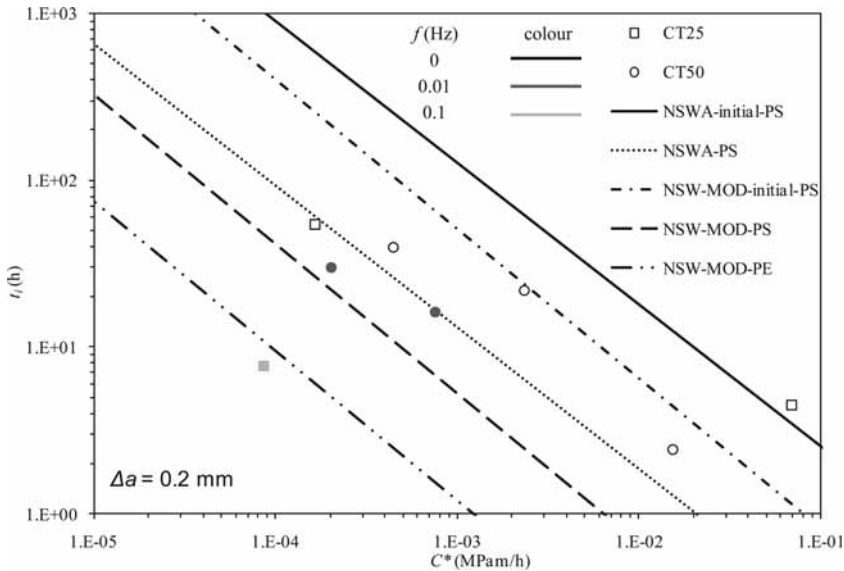


FIG. 15—Comparison between creep crack initiation time for static and cyclic crack tests on 316L(N) C(T) specimens of $W=50$ and 25 mm to NSW models at 650°C.

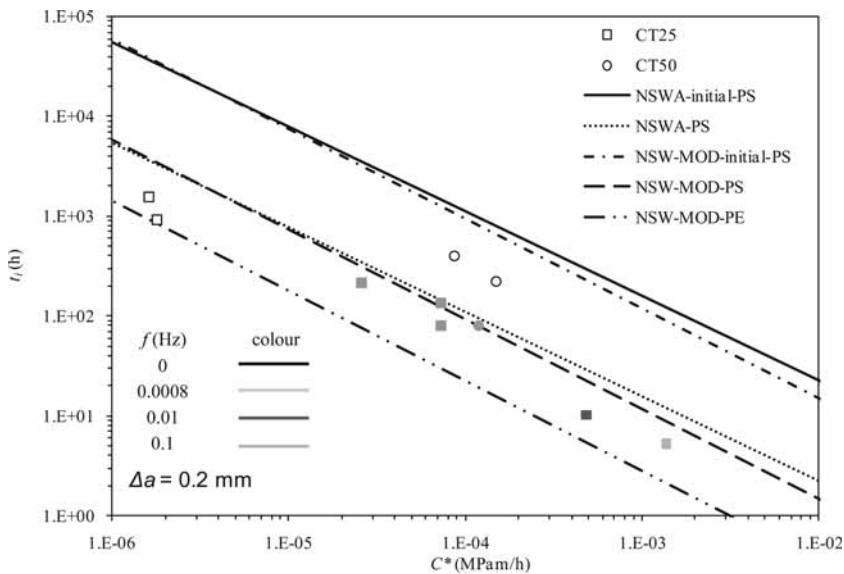


FIG. 16—Comparison between creep crack initiation for static and cyclic crack growth tests on 1CrMoV (Cast) C(T) specimens of $W=50$ and 25 mm to NSW models at 530°C.

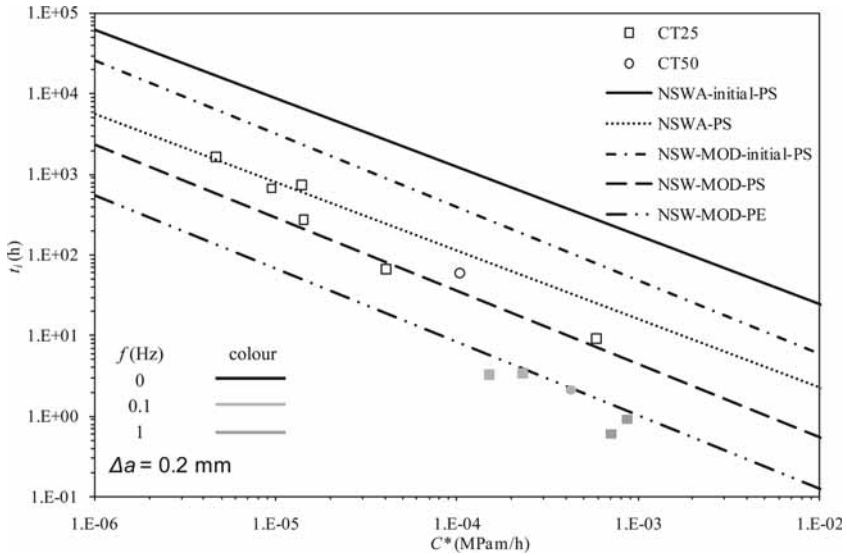


FIG. 17—Comparison between creep crack initiation time for static and cyclic crack growth tests on P22 C(T) specimens of $W = 50$ and 25 mm to NSW models at 565°C .

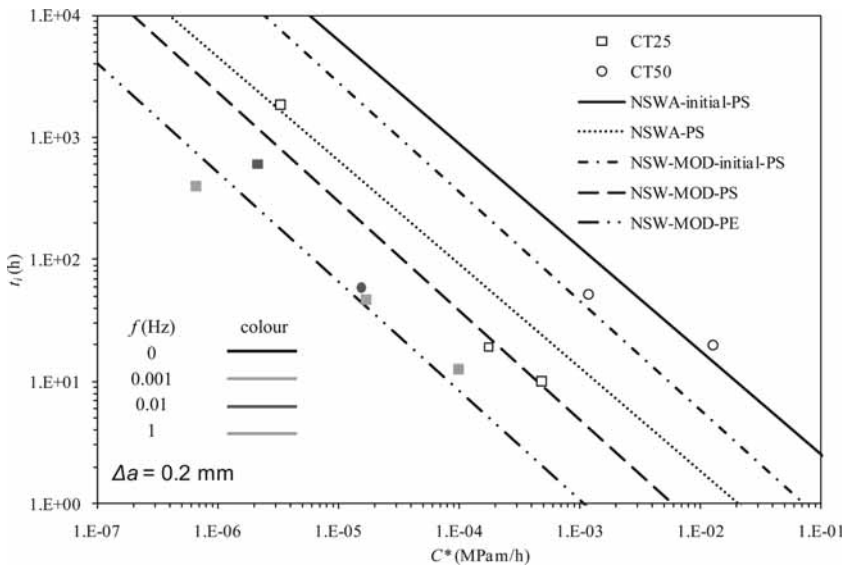


FIG. 18—Comparison between creep crack initiation time for static and cyclic crack growth tests on P91 C(T) specimens of $W = 50$ and 25 mm to NSW models at 625°C .

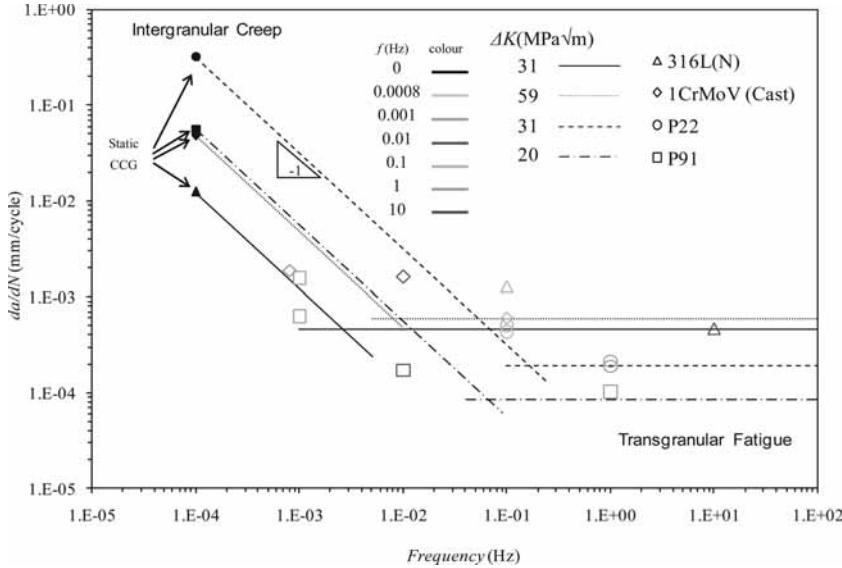


FIG. 19—Sensitivity of crack growth per cycle to frequency in different steels at constant temperature, $R=0.1$, and for given ΔK values.

Influence of Frequency on Crack Growth

The influence of frequency on creep-fatigue interaction has been examined, by plotting the crack growth rate per cycle versus frequency, using Eq (14), as shown in Fig. 19. At high frequencies, where fatigue mechanism dominates, the first term in Eq (14) tends to zero and da/dN_{total} tends to the Paris law. The Paris law for each material appears as a horizontal line and denoted the “transgranular fatigue line” in Fig. 19. At low frequencies, time dependent creep failure mechanism dominates, and the second term in Eq (14) is negligible compared to the first “creep” term. Hence, as frequency tends to zero, da/dN in Eq (14) becomes inversely proportional to frequency and a slope of -1 is seen when da/dN is plotted against frequency on log-log scale, as shown by the “intergranular creep line” in Fig. 19.

In Fig. 19, the crack growth per cycle in cyclic tests at a constant temperature has been plotted against a range of frequencies ($0.0008 \leq f \leq 10$) for each material. In all creep-fatigue tests examined in this work, the R load ratio is 0.1 and thus it has been assumed that the crack growth per cycle is a function of frequency alone. Thus, the sensitivity of crack growth to frequency in different steels can be investigated at given values of ΔK . The results for each material are compared, as shown in Fig. 19, for a given stress intensity factor.

Note that since the tensile and creep properties of each material at their respective test temperatures differ, the loads applied to specimen of each material are also different. Hence the stress intensity factor range considered in Fig. 19 is significantly higher for 1CrMoV (59MPa√m) compared to that of P22 and 316L(N) (31MPa√m) and that of P91 steel (20MPa√m). It must also be noted

that a comparison has only been made between $C(T) = 25$ mm specimens since the fatigue law constants are not available for $C(T) = 50$ mm due to insufficient number of tests on this geometry size samples.

It is expected as frequency tends to zero, the data should tend to the static CCG data limit. Assuming that the static creep data can be considered as a very low frequency cyclic test, the static creep crack growth data has been included in Fig. 19. These points have been estimated by inserting the creep crack growth rate measured in the CCG tests, for a given K corresponding to the ΔK used in the cyclic tests, into the first term on the right hand side of Eq (13) and taking the frequency to 1×10^{-4} Hz. The intergranular creep lines included in these figures intersect the static CCG data with a slope defined by Eq (13). Note that considerable scatter exists in the $\dot{a}$ versus K data which suggests difficulties in its interpretations. Hence upper and lower bound fits were made to the $\dot{a}$ versus K data, by offsetting the mean line by $\pm$ standard deviations.

The static CCG data shown in Fig. 19 are also based on the mean values from the regression fits to the data. Note that due to the scatter in the static CCG data, a significant error bar is associated with the crack growth per cycle calculated for these points which has been estimated to be a factor of up to 8 times higher/lower for the 316L(N) material and similarly by factors of 3.5, 1.2, and 2.4 for the 1CrMoV, P22, and P91 steels, respectively.

At intermediate frequencies, where both creep and fatigue processes are significant, the failure mechanism changes from creep to fatigue mode as frequency increases. As explained in Ref. [6], both modes of cracking are likely to progress intermittently in this region, through or around individual grains. Thus at intermediate frequencies when one mechanism becomes arrested locally, the other may take over to allow cracking to progress at a rate equal to the sum of the individual rates [6].

As can be seen in Fig. 19, the intermediate frequency data for P22 falls between the transgranular and intergranular lines shown. The frequency range where a transition between creep and fatigue dominance occurs differs for the various materials at their respective test temperatures and defined values of ΔK . In the transition regime, the data for P91 falls below that of the intergranular creep line; however, should the lower bound intergranular creep line be used, then the correct trend will be seen. Similarly, due to the inherent data scatter, the data point for 1CrMoV at $f = 0.0008$ Hz falls below the intergranular line and the crack growth per cycle for this point is only marginally higher than that for $f = 0.01$ Hz, indicating there is some error associated with these points. The limited data available for 316L(N) appear to fall within the expected region of the transgranular and intergranular lines shown.

In order to compare the response of various materials at the appropriate operating temperatures for a given loading condition, the crack growth per cycle in cyclic tests at a constant temperature has been plotted against a range of frequencies ($0.001 \leq f \leq 10$) for each material at a fixed, common value of ΔK . A linear regression fit has been made to the crack growth rate (with respect to time) data against ΔK for all tests, the equation for which is of the same form as Eq (4) but replacing K by ΔK . The regression fit constants derived are given in Table 4. The crack growth rate for ΔK of 30 MPa $\sqrt{\text{m}}$ has been obtained for each

TABLE 4—The power law constants in da/dt versus ΔK graph for cyclic tests at different frequencies. $da/dt = A\Delta K^B$: $da/dt = [mm/h]$, $\Delta K = [MPa\sqrt{m}]$.

Material	Condition	Temperature	CT size	f (Hz)	A	B
316L(N)	Base	650	CT25	0.1	6.0E – 06	3.24
				10	2.0E – 04	3.24
1CrMoV	Cast	530	CT25	0.0008	2.0E – 04	0.73
				0.01	3.0E – 03	0.73
				0.1	1.0E – 02	0.73
P22	Base	565	CT25	0.1	7.0E – 07	3.49
				1	4.0E – 06	3.49
P91	Base	625	CT25	0.001	5.0E – 12	6.58
				0.01	1.0E – 11	6.58
				1	5.0E – 10	6.58

material. Note that in some instances the trend line of a given test had to be extrapolated to $\Delta K = 30 \text{ MPa}\sqrt{m}$. The crack growth rate (with respect to time) obtained has then been employed in Eq 13, and the fatigue crack growth per cycle for $\Delta K = 30 \text{ MPa}\sqrt{m}$ predicted using the appropriate Paris law constants from Table 1. The results are shown in Fig. 20.

As can be seen in Fig. 20, for the testing temperatures considered, similar mean creep crack growth rates are predicted for 1CrMoV and 316L(N) steel for low frequencies whereas the cracking rates predicted for P22 and P91 at a given

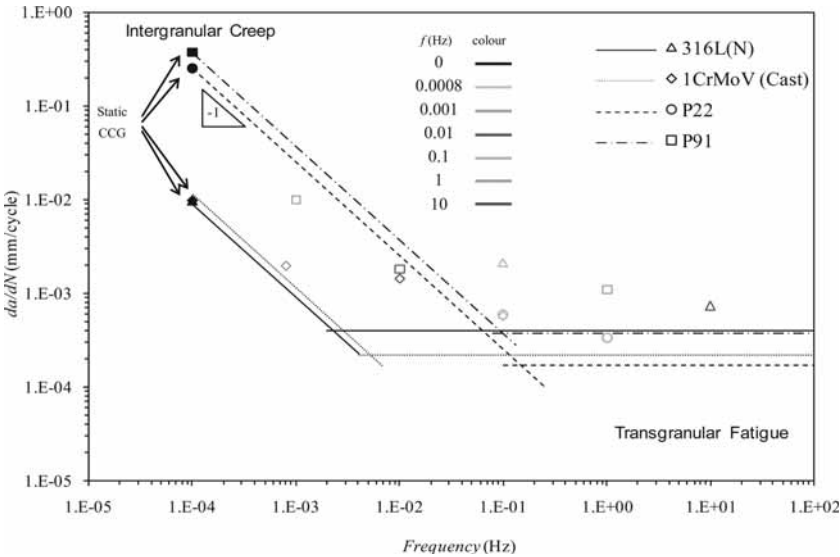


FIG. 20—Sensitivity of crack growth per cycle to frequency for a ΔK value of $30 \text{ MPa}\sqrt{m}$.

frequency are approximately 10 times higher. It must be noted that these factors are subjective to the data band used. At high frequencies, however, the crack growth rates for P91 and 316L(N) are almost identical and around twice that of P22 and 1CrMoV steels. Again, as in Fig. 19 the data for P91 falls below the intergranular creep line; however, should the lower bound intergranular creep line be used then the right trend will be seen. The correct trends are indicated by the other materials. However, within the extend of data scatter for the limited data available, no distinct difference can be observed between the data trends of the various materials within the transition regime.

Conclusions

The creep and creep/fatigue crack growth results for four steels taken from the HIDA collaborative project have been analysed and the interactive creep/fatigue behaviour of these steels identified. The new ASTM standard for creep-fatigue testing has been used as guideline to analyse the results. The present study also highlights the difficulty of deriving such information from tests that inherently produce substantial scatter. Furthermore, the results indicate that further round robin testing and analysis is necessary to identify clearly the additional recommendations needed in the standards in order to improve test methods and their subsequent analysis.

The crack growth rates separate into two distinct categories. Generally below 0.1 Hz the cracking rate is time dependent and above 0.1 Hz it is cycle dependent. The shift in this dependency can be due to the extent of creep ductility and constraint levels. The crack growth per cycle can vary by up to an order of magnitude with a small change in frequency. The change in failure mechanism was determined by relating the departure of cyclic test results from the $\dot{a}$ versus C^* correlation, demonstrated by the static load tests dominated by creep, towards the da/dN versus ΔK correlation shown by the high frequency (continuous fatigue) tests. Considering the data scatter and subsequent error associated with the data shown, the correct trends are shown on the da/dN versus frequency plot for all materials, provided an appropriate data bound on the creep crack growth data is used.

Overall it is evident that considerable scatter exists in the data leading to the upper/lower bounds of data to be considered to aid data interpretation. To improve the confidence in the data trends seen, it is clear that more data are required. This suggests that recommendations should be given in the standard for a minimum number of tests per frequency required (for a given temperature and loading ratio). Furthermore the aims and objectives of the new standard E2760-10 needs to be further clarified and coordinated to compliment the previously available ASTM E1457 and E647 for creep and fatigue testing. Otherwise users may misinterpret the finding when comparing their results in the different standards.

Acknowledgment

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R. Sugiura,¹ A. T. Yokobori, Jr.,¹ T. Nakagawa,¹ T. Adachi,²
I. Nonaka,¹ M. Tabuchi,³ Y. Hasegawa,⁴ and T. Matsuzaki¹

Characterizations of Creep-fatigue Crack Initiation and Growth Life for P92 using Circular Notched Round Bar Specimen

ABSTRACT: A testing method under creep-fatigue interaction conditions has been proposed as the ASTM standard E2714-09. In accordance with the ASTM standard, round bar smooth specimens have been recommended as a testing specimen. Additionally, creep-fatigue crack initiation and growth life are usually defined as the attainment of a specific rate of decrease in the maximum tensile stress or the modulus of elasticity ratios. However, these criterions of crack initiation and growth life on ASTM standard are not characterized in terms of physical crack size. In this study, the creep-fatigue crack initiation and growth tests for P92 were conducted using circular notched round bar specimens. An attempt is made to measure the crack length during the tests by the direct current potential drop method, which results in the establishment of a high accuracy crack length measurement. Additionally, the life of creep-fatigue crack growth was characterized by linear cumulative damage law combined with the Q^* concept which has been proposed as fracture parameter to describe the life of creep crack growth.

KEYWORDS: P92, creep-fatigue, crack initiation life, crack growth life, circular notched round bar specimen, linear cumulative damage law, Q^* concept

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¹ Graduate School of Engineering, Tohoku Univ., Aoba 6-6-01, Aramaki, Aoba-ku, Sendai-City 980-8579, Japan.

² Faculty of Science and Engineering, Ishinomaki Senshu Univ., Shinmito, Minamisakai, Ishinomaki-City 986-8580, Japan.

³ National Institute for Materials Science, Sengen 1-2-1, Tsukuba 305-0047, Japan.

⁴ Steel Research Laboratories, Nippon Steel Corporation, Shintomi 20-1, Futtsu-City 293-8511, Japan.

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Introduction

W-strengthened 9 % Cr ferritic heat-resistant steel, ASME code case P92, has been developed for the application of a boiler material for the ultra super critical (USC) and advanced ultra super critical (A-USC) power plants [1,2]. The developed P92 has been used under conditions of high temperature creep-fatigue interaction. Concerning the maintenance of the operational safety and minimizing operational costs, it is necessary to establish unified laws of predicting the crack initiation and growth life under the conditions of creep-fatigue interaction.

At present, a testing method under creep-fatigue interaction conditions has been proposed as the ASTM standard [3]. In accordance with the ASTM standard, a round bar smooth specimen have been recommended as a testing specimen. Additionally, creep-fatigue crack initiation and growth life are usually defined as the attainment of a specific ratio of decrease in the maximum tensile stress or the modulus of elasticity ratios [3]. However, these criteria of crack initiation and growth life on the ASTM standard are not characterized in terms of physical crack size.

In this study, the creep-fatigue crack initiation and growth tests using circular notched round bar specimens for P92 under displacement-controlled conditions were conducted. Additionally, an attempt is made to measure the crack length during the tests on the basis of the direct current potential drop (DCPD) method, which results in the establishment of a high accuracy crack length measurement. As a result, the criterion of creep-fatigue crack initiation and growth were proposed in terms of physical crack size. On the basis of proposed criterions, the life of creep-fatigue crack growth was characterized by the linear cumulative damage law combined with the Q^* concept which has been proposed as a fracture parameter to describe the life of creep crack growth.

Material and Specimen

The material used for tests was the W-strengthened 9 % Cr ferritic heat-resistant steel, ASME code case P92, which is used for a boiler material. The chemical composition and mechanical properties of this material are shown in Tables 1 and 2 respectively. The specimen used was manufactured into a round bar specimen with a circular notch to measure the crack initiation and growth length under the creep-fatigue interaction conditions, as shown in Fig. 1. The tip angle and depth of the circular notch are 60° and 1.5 mm. The coefficient value of stress concentration of an initial notch tip is 3.95 [4]. This circular notched

TABLE 1—Chemical composition of P92 (mass %).

C	Si	Mn	P	S	Cr
0.085	0.26	0.49	0.006	0.0012	8.85
Mo	W	Nb	V	N	B
0.50	1.80	0.06	0.20	0.0510	0.0030

TABLE 2—Mechanical properties of P92.

Temperature (°C)	0.2% Proof Stress (MPa)	Tensile Stress (MPa)	Elongation (%)
20	583	731	27.6

round bar specimen produces the local multi-axial stress field due to the plastic constraint although the applied load is in a uniaxial tensile condition [5].

Test Method and Testing Condition

Crack initiation and growth tests were performed on the basis of ASTM E2714-09 [3]. The machine used was a Servo-controlled tension-compression fatigue machine with electric resistance furnace manufactured by TOSHIN KOGYO (Tokyo, Japan). A schematic illustration of the measurement method for deformation in the test specimen and crack length using the DCPD method is shown in Fig. 2. To control the deformation in the test specimen, the extensometer is used and the dynamic longitudinal extension was measured. The extensometer tips contact the parallel gage section of the test specimen and the distance between them is 20 mm, as shown in Fig. 2. The crack length can be measured by an electrical potential method [6,7]. An electrical dc of 4 A is applied to the

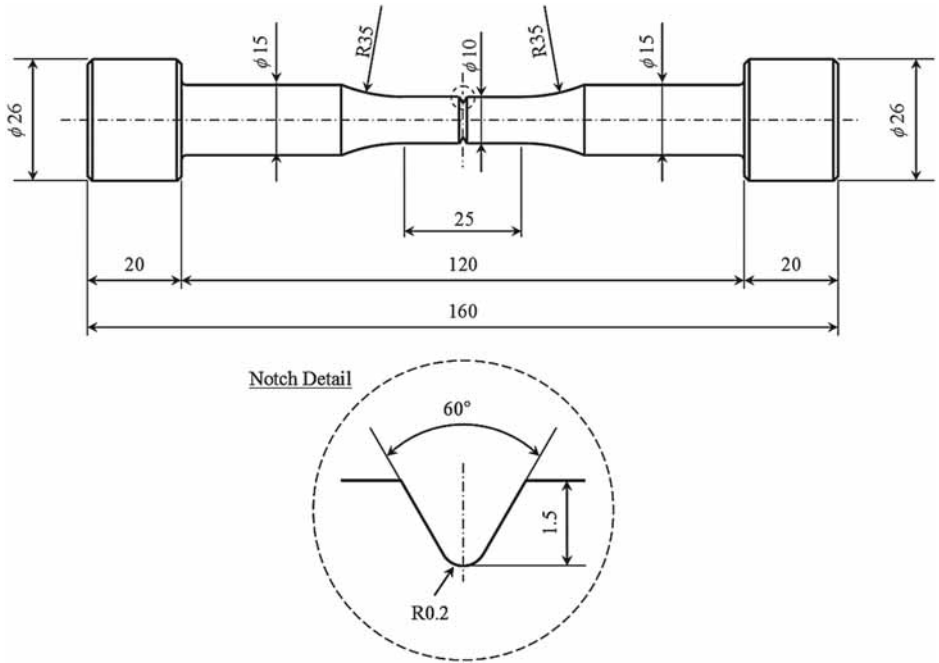


FIG. 1—Geometry and size of a circular notched round bar specimen.

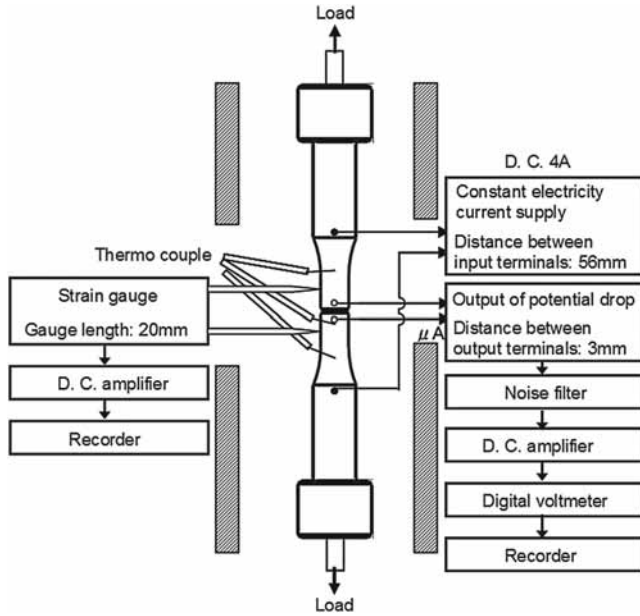


FIG. 2—A schematic illustration of measurement method for the deformation in test specimen and crack length using DCPD method.

specimen and the value of the electric potential drop is measured, as shown in Fig. 2. The distance between input terminals is 56 mm and the distance between output terminals, which are located to the edges of initial notch, is 3 mm.

The crack length is usually calculated using Johnson's equation for the C(T) specimen given by [6,7]

$$a = \frac{2W}{\pi} \cos^{-1} \left[\frac{\cosh(\pi y/2W)}{\cosh\{(u/u_0)\cosh^{-1}[\cosh(\pi y/2W)/\cos(\pi a_0/2W)]\}} \right] \quad (1)$$

where a is the length of the crack (mm), a_0 is the initial crack length (mm), y is half the distance between the output terminals (mm), W is the width of the specimen (mm), u_0 is the initial voltage (μV), u is the actual value of the potential (μV). However, Johnson's formula [6] is not applied to the circular notched round bar specimen. Previously, a modified equation of Eq 1 was derived to measure a crack length from a circular notch given by Eq 2 [8]

$$a = 0.667a_0 \exp \left[0.434 \frac{D}{\pi a_0} \cos^{-1} \left\langle \frac{\cosh(\pi y/D)}{\cosh\{U/U_0 \cosh^{-1}[\cosh(\pi y/D)/\cos(\pi a_0/D)]\}} \right\rangle \right] \quad (2)$$

where D is the diameter of specimen (mm).

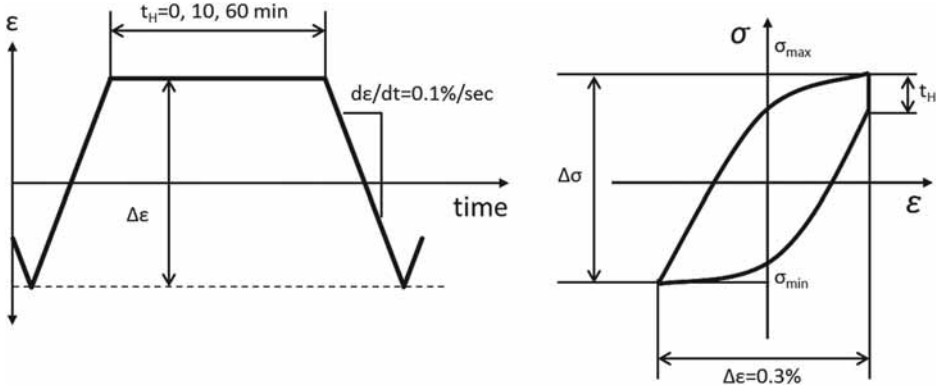


FIG. 3—Examples of creep-fatigue cycle shape and stress-nominal strain hysteresis diagrams with holding time.

The accuracy of predicting the crack length using DCPD method and proposed Eq 2 is discussed in the next section.

The test temperatures are 600 °C and 630 °C, respectively, which were kept at a specified temperature with the precision of ± 2 °C. Strains applied to the parallel gauge section of test specimen were measured in the longitudinal direction as shown in Fig. 2, which are denoted as nominal strains. Nominal strain rate and strain range are 0.1 %/s and 0.3 %, respectively. Crack initiation and growth tests were conducted at three holding times, i.e., 0, 10, and 60 min. Examples of creep-fatigue cycle shape and stress-nominal strain hysteresis diagrams with holding time are shown in Fig. 3.

Frequency under creep-fatigue displacement-controlled conditions is calculated by the following equation

$$f = \frac{1}{(t_H + t_{LOAD} + t_{UNLOAD})} \quad (3)$$

where t_{LOAD} and t_{UNLOAD} are loading and unloading times, respectively, which are given by

$$t_{LOAD} = t_{UNLOAD} = \frac{\Delta \epsilon}{\dot{\epsilon}} \quad (4)$$

In the present paper, with increase in t_H , frequency f will decrease due to the constant of strain range and rate.

Experimental Results

Accuracy of Predicting the Crack Length using the DCPD Method

The crack front shows a circle of which the center was almost identical to the center of the original round bar specimen with a circular notch, as shown in

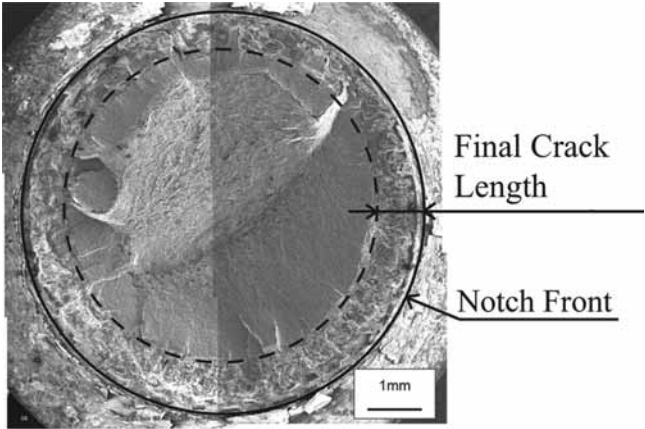


FIG. 4—The fracture surface of a circular notched round bar specimen for the P92 ($T=600^{\circ}\text{C}$, $t_H=60\text{ min}$).

Fig. 4. Figure 5 shows the comparison of actual total crack length including initial notch depth with calculated total crack length including initial notch depth by the DCPD method. Calculated total crack length was found to predict the actual total crack length with a precision of 7.3 % accuracy. This has resulted in the validity of high accuracy of crack length measurement.

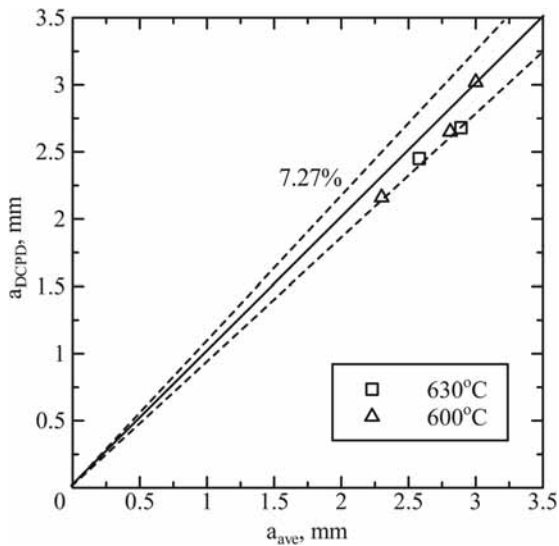


FIG. 5—Comparison of actual total crack length including initial notch depth with calculated total crack length including initial notch depth from DCPD method.

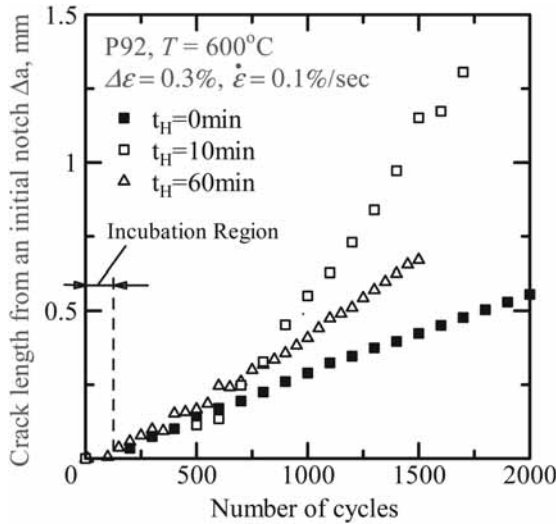


FIG. 6—Relationship between crack length from an initial notch tip and number of cycles at a temperature of 600°C .

Creep-fatigue Crack Growth Behavior

The relationships between crack length from an initial notch tip obtained by DCPD method and number of cycles at the temperatures of 600°C and 630°C are shown in Figs. 6 and 7, respectively. From Figs. 6 and 7, crack growth

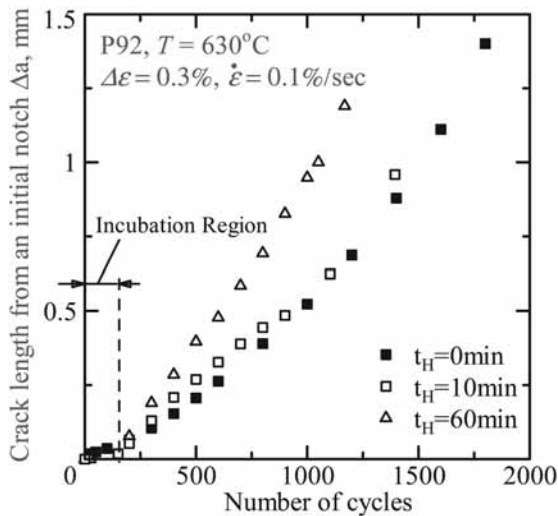


FIG. 7—Relationship between crack length from an initial notch tip and number of cycles at a temperature of 630°C .

behaviors under the creep-fatigue interaction condition were found to show the incubation region of crack growth at the early stage. After that, crack growth from an initial notch tip shows the linear relationship against the number of cycles.

Cycle Sequential Characteristics of Tension and Compression Peak Stresses

Cycle sequential characteristics of tension and compression peak nominal stresses at temperatures of 600 °C and 630 °C are shown in Figs. 8 and 9. These stresses were defined as the net section stresses σ_{net} without considering the growing crack length from a notch tip during the tests. From Figs. 8 and 9, in the case of $t_H=0$, cyclic softening behaviors in maximum and minimum net section stresses was found out at the early stage. On the other hand, under the creep-fatigue condition, at the early stage, maximum tensile net section stress was found to rapidly decrease and maximum compression net section stress was found to rapidly increase, which result in the occurrence of kinematic hardening behavior which is different from the experimental result of $t_H=0$. This behavior becomes eminent with an increase in strain hold time.

Cycle Sequential Characteristic Characteristics of Stress-Strain Hysteresis Diagrams

Cycle sequential characteristics of net section stress-nominal strain hysteresis diagrams at a temperatures of 600 °C are shown in Figs. 10–12. The experimental data in each figure show diagrams for the early, middle and final stages of tests.

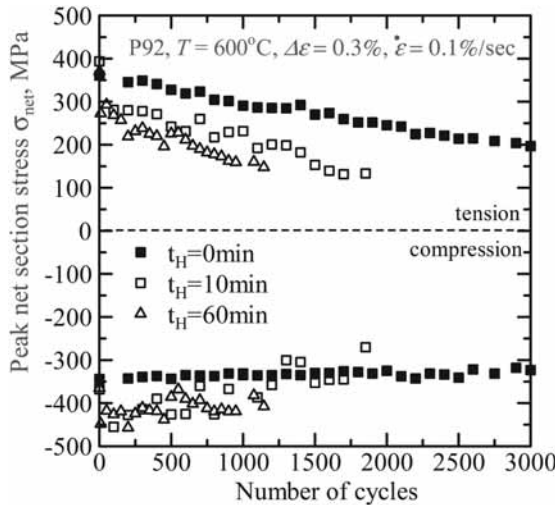


FIG. 8—Change characteristics of tension and compression peak net section stresses at a temperature of 600 °C.

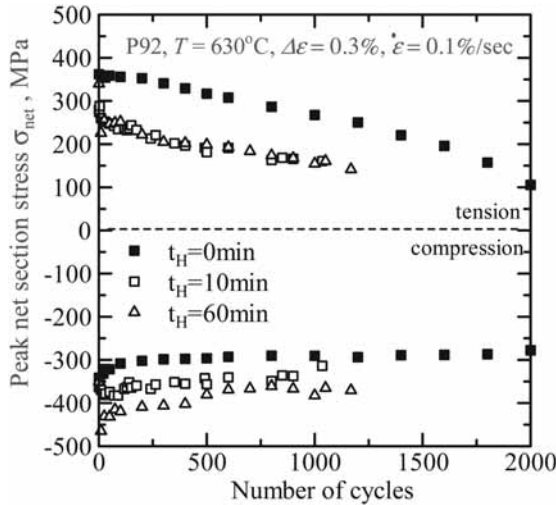


FIG. 9—Change characteristics of tension and compression peak net section stresses at a temperature of 630 °C.

From the experimental result of $T = 600\text{ }^{\circ}\text{C}$, $t_H = 0\text{ min}$, as shown in Fig. 10, the peak net section stress in tension was found to decrease with increase in the number of cycles; however, the origin of hysteresis loop does not shift to the compression side. On the other hand, from the experimental results of $T = 600\text{ }^{\circ}\text{C}$, $t_H = 10$ and 60 min as shown in Figs. 11 and 12, the origin of hysteresis loops

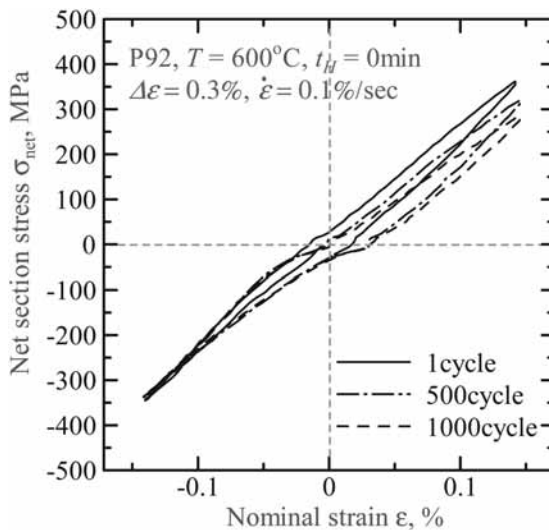


FIG. 10—Change characteristics of net section stress - nominal strain hysteresis diagrams at the temperature of 600 °C without holding time.

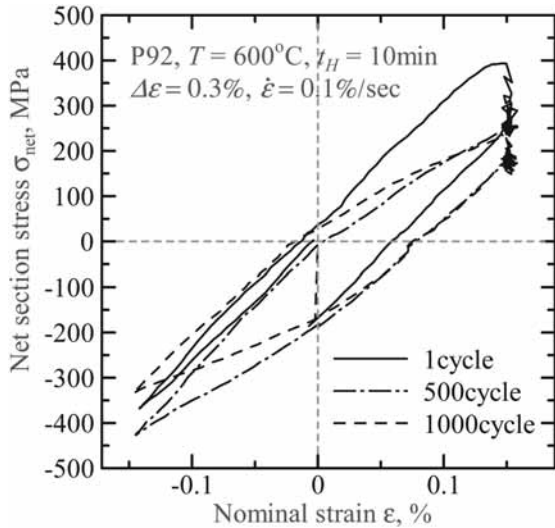


FIG. 11—Change characteristics of net section stress-nominal strain hysteresis diagrams at the temperature of 600°C with $t_H = 10\text{ min}$.

under the creep-fatigue interaction conditions were found to shift to the compression side, which correspond with the experimental results in Fig. 9, due to the holding time.

The experimental characteristics of $T = 630^\circ\text{C}$ are similar to those of $T = 600^\circ\text{C}$.

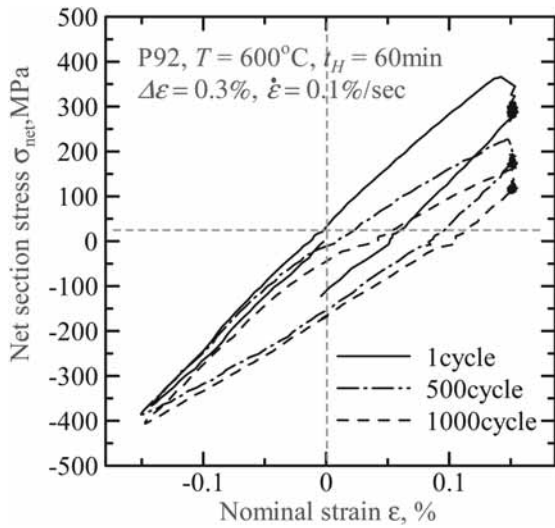


FIG. 12—Change characteristics of net section stress-nominal strain hysteresis diagrams at the temperature of 600°C with $t_H = 60\text{ min}$.

In accordance with the ASTM standard, round bar smooth specimen have been recommended as a testing specimen [3]. Additionally, creep-fatigue crack initiation and growth life are usually defined as the attainment of a specific rate of decrease in the maximum tensile stress or the modulus of elasticity ratios [3]. However, on the basis of the ASTM standard mentioned above, it is considered to be the difficulty of characterizing the creep-fatigue crack initiation/growth behaviors and its life for P92 which show the different characteristics of peak net section stresses in tension and compression sides. Therefore, in the next section, cyclic deformation behavior is estimated by using the parameter of net section stress range $\Delta\sigma_{\text{net}}$.

Cyclic Sequential Characteristics of Net Section Stress Range $\Delta\sigma_{\text{net}}$

Cyclic sequential characteristics of the stress range $\Delta\sigma_{\text{net}}$ of stress-strain hysteresis diagrams at the temperatures of 600 °C and 630 °C is shown in Fig. 13.

Figure 13 shows that the values of the stress range $\Delta\sigma_{\text{net}}$ under all conditions decrease monotonously with increase in the number of cycle. There is no rapidly decrease in the stress range $\Delta\sigma_{\text{net}}$ at the early stage as shown by Figs. 8 and 9. Therefore, it is considered that the stress range $\Delta\sigma_{\text{net}}$ shows regular characteristics which will gives reasonable criterions of crack initiation and growth life.

Criteria of Crack Initiation and Growth Life

In accordance with the ASTM standard E2714-09 [3], round bar smooth specimen have been recommended as a testing specimen [3]. Additionally, creep-fatigue crack initiation and growth life are usually defined as the attainment of a specific ratio of decrease in the maximum tensile stress or the modulus of

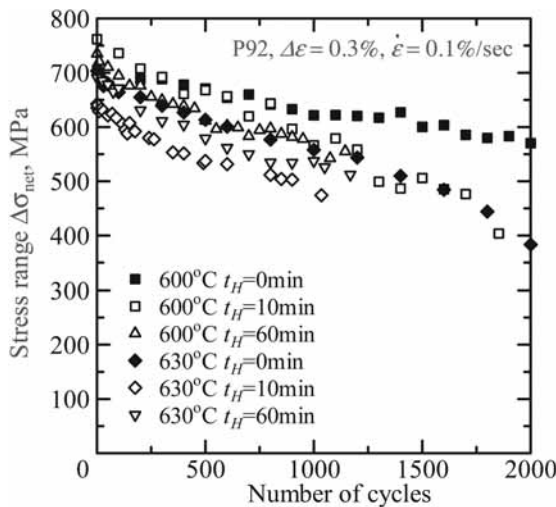


FIG. 13—Cyclic deformation behavior estimated by the parameter of stress range $\Delta\sigma_{\text{net}}$.

elasticity ratios [3]. However, on the basis of ASTM standard mentioned above, it is considered to be a difficult to characterize the behaviors of creep-fatigue crack initiation and growth, and to predict its life for P92. Therefore, in the section Cyclic Sequential Characteristics of Net Section Stress Range $\Delta\sigma_{\text{net}}$, the parameter of the stress range $\Delta\sigma_{\text{net}}$ was proposed to predict the life of creep crack initiation and failure. In this section, the definition of crack initiation and growth life was discussed relating to the parameter of the stress range $\Delta\sigma_{\text{net}}$.

The relationship between the crack length from an initial notch obtained by DCPD method and the reduction ratio of stress range $\Delta\sigma_{\text{net}}$ at the temperatures of 600 °C and 630 °C are shown in Figs. 14 and 15. From Figs. 14 and 15, the crack length from an initial notch tip was found to show the linear relationship with the reduction ratio of stress range where $(\Delta\sigma_0)_{\text{net}}$ and $(\Delta\sigma)_{\text{net}}$ are initial and current stress range. Additionally, at the stage of reduction ratio of stress range $\Delta\sigma_{\text{net}}$ with about 25 %, an adequately uniform crack growth was found out under all conditions. On the other hand, at the stage of reduction ratio of stress range $\Delta\sigma_{\text{net}}$ with about 10 %, it was found to correspond with the period of incubation region of crack growth as shown in Figs. 6 and 7. These results show that criterions of crack initiation from an initial notch and growth life can be defined as the attainment of 10 and 25 % reduction of stress range $\Delta\sigma_{\text{net}}$, respectively.

Separate Estimation of the Cycle-Dependent Mechanism from the Time-Dependent Mechanism on Crack Initiation and Growth Life

The separate estimation of the effect of cycle-dependent (CD) mechanism on crack initiation and growth life from that of time-dependent (TD) mechanism is

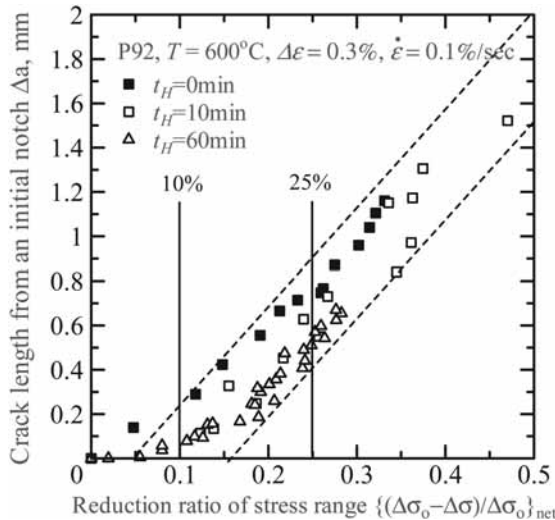


FIG. 14—Relationship between the crack length from an initial notch tip and reduction ratio of stress range $\Delta\sigma_{\text{net}}$ at a temperature of 600 °C.

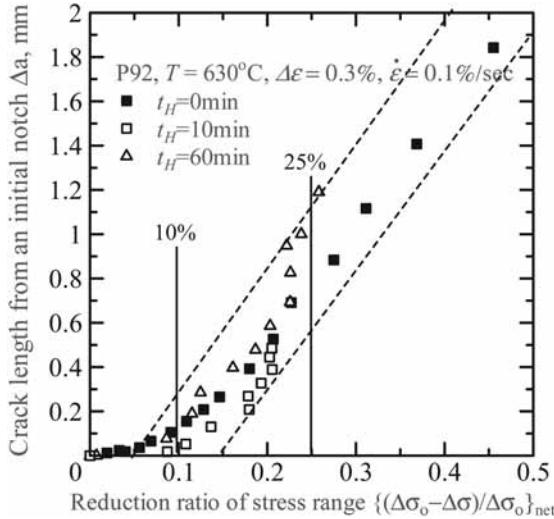


FIG. 15—Relationship between the crack length from an initial notch tip and reduction ratio of stress range $\Delta\sigma_{net}$ at a temperature of 630 °C.

important from the view points of the estimation of mechanical performance of materials and the construction of predicting methods of crack initiation and growth life.

The schematic illustration of the method for separate estimation of the cycle-dependent mechanism from the time-dependent mechanism is shown in Fig. 16 [9,10]. When the crack initiation life t_i and crack growth life t_f are dominated by the cycle-dependent mechanism, the relationship between $\log 1/t_i$ or $1/t_f$

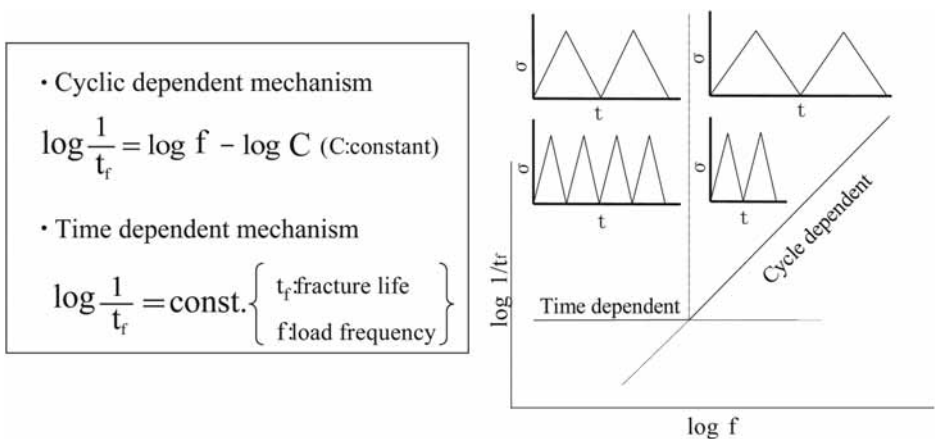


FIG. 16—Separated estimation method of the cycle-dependent mechanism from the time-dependent mechanism.

t_f and f are linear with a gradient of 45° , respectively. On the other hand, when the crack initiation life t_i and crack growth life t_f are dominated by the time-dependent mechanism, the relationship between life $1/t_i$ or $1/t_f$ and f are linear and parallel to the $\log f$ axis. Therefore, from the experimental relationship between $1/t_i$ or $1/t_f$ and f , the determination of whether the crack initiation and growth life are dominated by the cycle-dependent, time-dependent, or a competitive mechanism can be assessed [9,10].

The relationship between the inverse value of crack growth life t_f and frequency f is shown in Fig. 17. In Fig. 17, the experimental results show that the inverse value of crack growth life was found to decrease approximately parallel to the life for the cycle-dependent mechanism. Therefore, the crack growth life on the basis of proposed criterion will be dominated by the cycle-dependent mechanism.

The relationship between the inverse value of crack initiation life t_i and frequency f is shown in Fig.18. In Fig. 18, the experimental results also show that the inverse value of crack growth life decreases parallel to the life for the cycle-dependent mechanism. Therefore, the crack initiation life on the basis of proposed criterion will be also dominated by the cycle-dependent mechanism.

Prediction of Crack Growth Life under Creep-Fatigue Conditions

To predict the life of crack growth under high temperature creep-fatigue conditions, linear [11–13] and non-linear [14] cumulative damage laws for smooth specimens have been proposed. For example, on the basis of ASME Boiler and Pressure Vessel Code N-47 [12], linear accumulative damage law was given by Eq 5

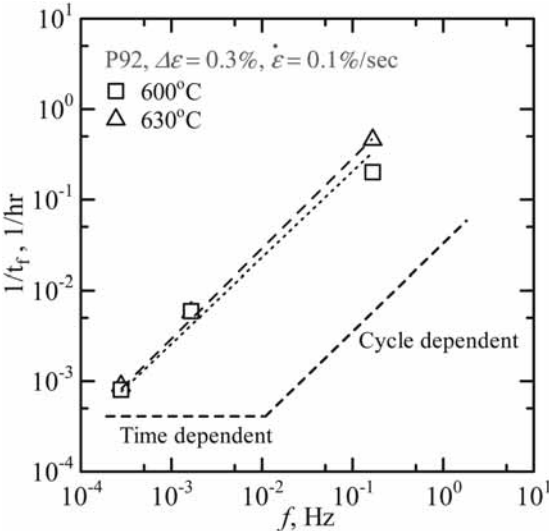


FIG. 17—Relationship between the inverse value of crack growth life and f .

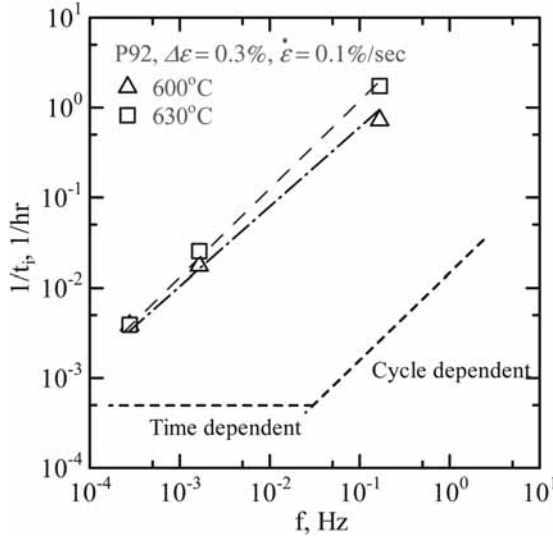


FIG. 18—Relationship between the inverse value crack initiation life and f .

$$\phi_f + \phi_c \leq D \quad (5)$$

$$\phi_f = \sum_{j=1}^{j=u} \left(\frac{N_j}{N_{dj}} \right) \quad (6)$$

$$\phi_c = \sum_{k=1}^{k=v} \left(\frac{t_k}{t_{f,k}} \right) \quad (7)$$

where ϕ_f and ϕ_c are accumulated fatigue and creep damages, N_j is the number of cycles in the strain range j , N_{dj} is the number of cycles to failure under a sufficiently high strain rate in the strain range j , t_k is total sum of stress holding time under the condition of stress k , $t_{f,k}$ is creep rupture time under the condition of stress k .

In this study, the strain range was controlled steadily during the tests; therefore, the fatigue accumulated damage given by Eq 6 is written by Eq 7

$$\phi_f = \frac{N_f}{N_{f-f}} \quad (8)$$

On the other hand, creep accumulated damage given by Eq 7 is written by Eq 9

$$\phi_c = \sum_{N=1}^{N=N_f} \left\{ \int \left(\frac{dt}{t_{f,k}} \right)_N \right\} \quad (9)$$

If the creep accumulated damage during compression process is negligibly small as compared with that of tensile process, creep accumulated damage of Eq 9 is given by Eq 10

$$\phi_c = \sum_{N=1}^{N=N_f} \left\{ \int_0^{t_H} \left(\frac{dt}{t_{f,k}} \right)_N \right\} \quad (10)$$

Equation 10 represents the accumulation of creep damage on N th cycle during the tensile process. If the creep damage is not changed by strain cycle, Eqs 9 and 10 are written as follows:

$$\phi_c = N_f \left\{ \oint \left(\frac{dt}{t_{f,k}} \right)_N \right\} \quad (11)$$

$$\phi_c = N_f \left\{ \int_0^{t_H} \left(\frac{dt}{t_{f,k}} \right)_N \right\} \quad (12)$$

In this study, the creep damage during the compression process assumed to be negligibly small as compared with that of tensile process, Eqs 5, 8 and 12 are used to predict the life of crack initiation and growth under creep-fatigue conditions.

Linear Cumulative Damage Law Combined with the Q^* Concept

To calculate the creep cumulative damage given by Eq 12, the prediction of fracture life, $t_{f,k}$, under statically creep condition is necessary. In this section, an attempt is made to predict the creep fracture life, $t_{f,k}$, in Eq 12 on the basis of Q^* concept [15,16] and the newly creep-fatigue cumulative damage law is proposed.

The Q^* parameter [15,16] has been derived to correlate creep crack growth rate based on local stress dependent thermally activated process theory. The main purpose of Q^* parameter [15,16] is to predict high temperature mechanical failure of materials and creep crack growth life.

For creep-brittle materials, the Equation of creep crack growth rate is defined by Eq 13 on the basis of the Q^* concept [16]

$$\frac{da}{dt} = A \exp(Q^*) = A K_{in}^n \exp\left(-\frac{\Delta H_g}{RT}\right) \quad (13)$$

where A is a constant, Q^* is the Q^* parameter, K_{in} is the initial stress intensity factor ($\text{MPa m}^{1/2}$), n is the exponent of K_{in} , ΔH_g is an activation energy (kJ/mol), R is the gas constant ($= 8.3145 \text{ J/K mol}$), T is the absolute temperature (K).

The life of creep crack growth t_f is obtained by integrating Eq 13 from a_i to a_f , as given by Eq 14

$$t_f = \int_0^{t_f} dt = \int_{a_i}^{a_f} \frac{da}{A \exp(Q^*)} = \frac{1}{A} \exp(-Q^*) (a_f - a_i) = \frac{1}{A^*} \exp(-Q^*) \quad (14)$$

where t_f is the life of creep crack growth (h), a_i is the initial crack length (mm), a_f is the final crack length (mm), A^* is a const[$= A/(a_f - a_i)$].

From Eq 14, Eq 15 was obtained as follows:

$$\frac{1}{t_f} = A \exp(Q^*) = A^* K_{in}^n \exp\left(-\frac{\Delta H_g}{RT}\right) \quad (15)$$

Based on Eq 15, the Q^* parameter for P92 under the creep condition has been derived [17]. This is givenby

$$Q^* = 6.8 \ln(K_{in}(\sigma_k)) - \frac{499 \times 10^3}{RT} \quad (16)$$

From Eqs 15 and 16, the creep rupture time $t_{f,k}$ in the equation of the creep damage, as shown in Eq 12, is given by Eq 17

$$t_{f,k} = \frac{1}{1.27 \times 10^{19}} (K_{in}(\sigma_k))^{-6.8} \exp\left(\frac{499 \times 10^3}{RT}\right) \quad (17)$$

When Eq 17 is substituted in Eq 12, we can calculate the creep damage. In next section, creep accumulated damage was estimated by this method and it was compared with the fatigue accumulated damage.

Evaluation of Creep and Fatigue Accumulated Damages

The accumulated creep damages obtained by the method mentioned previous section are plotted with the accumulated fatigue damages, as shown in Fig. 19. The results for SUS304, 316, and P91 steels are also shown in Fig. 19 [18].

Figure 19 shows that crack growth life for P92 and P91 are dominated by fatigue, in contrast to the results for SUS304 and 316. This will be due to that fact that the W-added 9Cr steels have a strengthening microstructure such as martensitic lath structure, which causes an increase in the resistance against the creep crack growth.

This has resulted in the evidence of the occurrence of cycle-dependent mechanism on crack initiation and growth life in the previous section 6.

Comparison of Experimental Life with Predicted Life on the Basis of Linear Cumulative Law Combined with Q^* Concept

To derive the predictive law of crack initiation and growth life on the basis of linear cumulative law combined with Q^* concept, critical damage value of D in Eq 5 must be decided. Therefore, using the non-linear least square method, the calculated data in Fig. 19 were approximately given by Eq 18

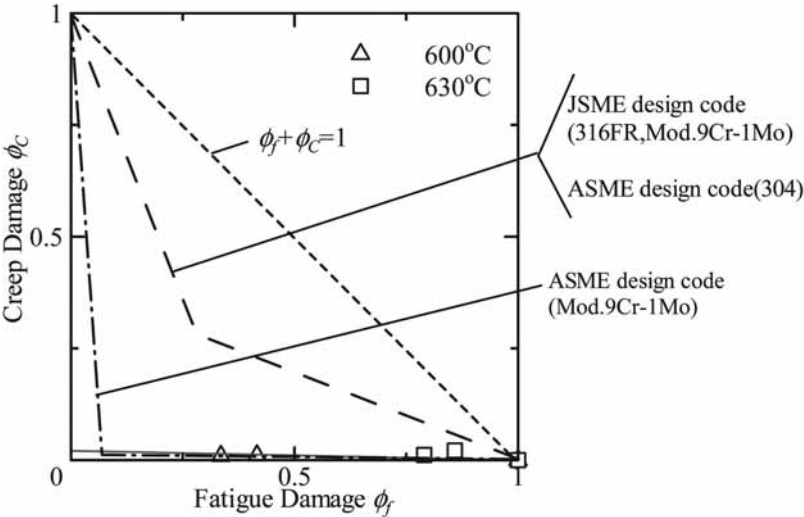


FIG. 19—Calculated results for creep and fatigue accumulated damages.

$$D = 0.021 + 0.979\phi_f \tag{18}$$

Substituting Eqs 8, 12 and 18 in Eq 5, the number of cycles to failure can be estimated by the following equation:

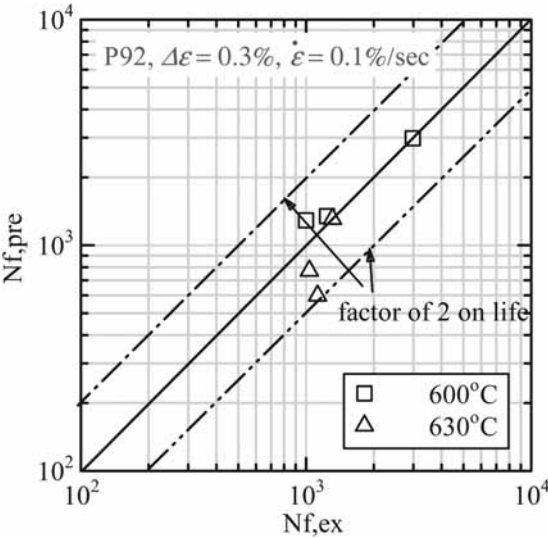


FIG. 20—Comparison of experimental life with predicted life on the basis of linear cumulative law combined with Q^* concept.

$$N_f = \frac{0.021N_{f-f}}{0.021 + N_{f-f}\phi_{c-1}} \quad (19)$$

where ϕ_{c-1} is creep accumulated damage on $N_f/2$ cycle.

Figure 20 shows the comparison of experimental life with predicted life obtained by Eq 19. As shown in Fig. 20, predicted life for various condition are in agreement with the experimental life within a factor of 2. This result will be reasonable to predict the crack growth life for P92 on the basis of linear cumulative damage law combined with Q^* concept.

Conclusions

Creep-fatigue crack initiation and growth tests using circular notched round bar specimen were conducted for P92. The following results were obtained.

- (1) On the basis of DCPD method and proposed equation, crack length during the tests was measured, which results in the establishment of high accuracy crack length measurement.
- (2) As a result of result 1, criterions of crack initiation and growth were defined as the attainment of 10 % and 25 % decrease in the stress range $\Delta\sigma_{\text{net}}$.
- (3) In the range of frequency on this experiment, crack initiation time and crack growth life were found to be dominated by cyclic-dependent mechanism.
- (4) Life prediction method on the basis of linear accumulative law with the Q^* concept was established.

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Jonathan Parker¹

Creep Fatigue Behavior of Creep Strength Enhanced Ferritic Steels

ABSTRACT: Creep Fatigue knowledge is necessary for meaningful reliability assessment of existing power generating plant because cyclic operation of an aging power plant fleet requires better understanding of factors affecting damage initiation and propagation, as well as technologies and methods to predict damage accumulation for individual components and overall safe/economic life. Utilities typically adopt operating practices which involve more severe cycles, either in terms of number or magnitude or both, so that there is, and will continue to be, a requirement to assess the creep fatigue performance of traditional boiler and turbine components. The recent installation of large numbers of combustion turbines and the associated heat recovery steam generators has significantly increased the number of materials and components which are operating under cyclic conditions at high temperatures. Indeed, the materials and component types susceptible to creep-fatigue is increasing since much of the new high efficiency coal plant will be operating with steam/metal temperatures much higher than those of traditional plant. Recent expert review has reemphasized that with most practical scenarios regarding future generation mix, fossil plant will continue to be important. Thus, there is the expectation that creep-fatigue damage will increase in occurrence, and affect an increased range of components. Different challenges will be faced since the new components will involve a greater range of alloys, methods of manufacture, and types of operation than conventional fossil generating stations. It is therefore apparent that operation under cyclic conditions at temperatures where creep can occur the potential for creep/fatigue damage is, and will continue to be, a matter of significant concern. The present paper considers specific challenges associated with creep/fatigue in creep strength enhanced ferritic steels such as Grade 91 and Grade 92.

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¹EPRI, Formally at Structural Integrity Associates, Inc., 1300 West WT Harris Blvd., Charlotte, NC 28262.

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KEYWORDS: creep, fatigue, steel, damage, analysis

Introduction

Operation under cyclic conditions at temperatures at which creep can occur is a matter of significant concern. In components that operate at high temperatures, changes in conditions especially at the beginning and end of the operation, result in transient temperature gradients [1]. If these transients are repeated, the differential thermal expansion during each transient results in a thermally induced cyclic stress. The extent of the resulting fatigue damage depends on the nature and frequency of the transient, the thermal gradient in the component, and the material properties.

Components that are subject to thermally induced stresses often operate within the creep range so that damage due to fatigue and creep must be taken into account. In steam-turbine rotors and casings, the large section size of the component results in large temperature gradients even under conditions of long start-up and pre-warming of the component. Under fast starting conditions from relatively low temperatures, severe thermal stresses can result. This problem has been exacerbated in recent years by placing base load units into cyclic operation for economic reasons. Also, surface craze cracking can occur in a variety of components operating at elevated temperatures due to condensation of steam on relatively cold metal surfaces during certain modes of operation. In severe situations, this condensate impingement can lead to through wall steam leaks or even fast fracture.

The complexity associated with understanding damage development under creep-fatigue conditions is further increased because there are many different types of operating cycles (for a single component location, the types of cycles could vary with time), and, in some circumstances, environmental factors can influence crack initiation and growth.

Some examples of the different types of cycle that affect in-service performance include the following [1]:

- Temperature increase or decrease which is in-phase or out-of-phase with pressure or load changes.
- Hold periods at different positions in the cycle.
- Stress cycles that involve the interaction of both primary and thermal stresses.
- Residual stress effects.

In developing damage rules, the most common testing involves strain-controlled fatigue tests with hold periods at constant stress or strain. Sometimes interspersed creep and fatigue tests have been performed so that the damage is introduced sequentially. Strain-controlled fatigue tests under athermal conditions are also performed, although these are generally known as *thermo-mechanical* fatigue tests. The endurance for elevated temperature cycles is reduced compared to behavior at room temperature. Further reduction in cycles to failure is observed when hold times are introduced into the cycle.

Creep-fatigue has also increased in importance because as units age and are subject to increased cyclic operation to meet energy demands, the risk of failure becomes more significant. Indeed, both the number and severity of cycles are

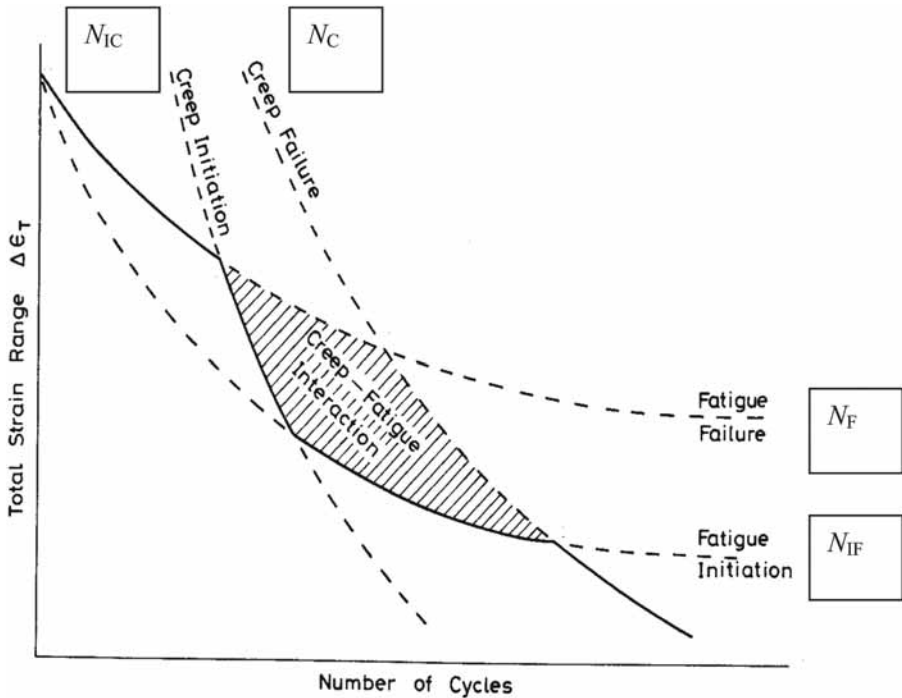


FIG. 1—A simplified representation of the Hale's creep-fatigue diagram. The shaded region shows the conditions where creep-fatigue interaction will occur [4].

increasing since dispatch requirements can lead to severe ramp rates both for unit starts and shut downs. Greater understanding of the fundamental mechanisms that comprise creep-fatigue is therefore necessary to effectively manage plant component life and several review documents have been produced to provide further insight on these issues [1,2,3].

Damage Morphology Issues

Creep-fatigue interaction and its dependence on specimen size is shown schematically in Fig. 1 [4]. Fully reversed strain-controlled low cycle fatigue tests were conducted with tension dwells up to 1000 mins duration on an austenitic steel at 600°C [4]. Many tests were interrupted prior to failure to allow a metallographic assessment of grain boundary cavities. Importantly, this assessment involved full sectioning so damage development within the body of the sample was recorded. The first signs of damage were detected after a 1-minute dwell. Under these conditions fatigue cracks growing from the surface encountered cavitated boundaries. There was a clear demonstration of increased cavitation with increased dwell time, so that at the maximum hold level, the specimen failed by bulk intergranular failure. When no dwells were imposed, failure occurred by transgranular crack propagation, and under these conditions

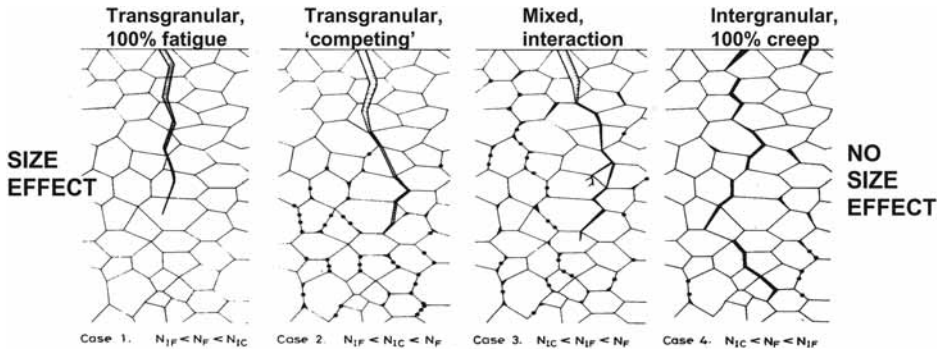


FIG. 2—Schematic diagram illustrating different types of damage which can develop in high temperature components, Ref [5].

striations were visible on the fracture surface. With increasing dwell period; therefore, the growing crack interacted more and more with cavitated areas due to creep, and it was considered that a true creep-fatigue interaction had taken place.

In the majority of cases laboratory evaluations concentrate on generating data of cycles to failure rather than performing detailed characterization of the morphology of damage. In many alloys tested under cyclic loading conditions specific 'maps' of damage appearance when creep fatigue interaction takes place are not established. This then raises issues for identification of early stages of damage in service components. Additional complexities associated with assessment of thick section components such as headers and rotors will be present since under plant operating conditions there will frequently be the need to take into account gradients in temperature and/or stress.

The generally accepted view of how fatigue and creep damage may be characterized is shown schematically in Fig. 2 [2]. This figure indicates that fatigue is typically fully transgranular and creep is typically fully intergranular. It should be noted that with traditional alloys, exhibiting equiaxed microstructure, creep voids generally develop on grain boundaries. These voids then increase in number and size leading to microcrack formation. It should be emphasized that with advanced steels creep damage can also take the form of precipitate coarsening and microstructural changes. These changes may in some circumstances be described as strain softening and will clearly significantly change material strength.

When operating conditions are such that both creep and fatigue damage can occur there are at least two different options for damage development. One is where the two forms of damage both occur but act independently of each other. An example of this situation is shown in a sample removed from a low alloy steel weld. In this weld, inside surface connected fatigue cracking and mid-wall creep damage have developed apparently independently, Fig. 3. The other condition is where damage is mixed and there is an interaction [3]. A typical photograph showing damage in a laboratory sample is shown in Fig. 4 [4].

A different type of interaction where cycles have apparently accelerated the formation of creep cracking is shown in the photomicrograph from an

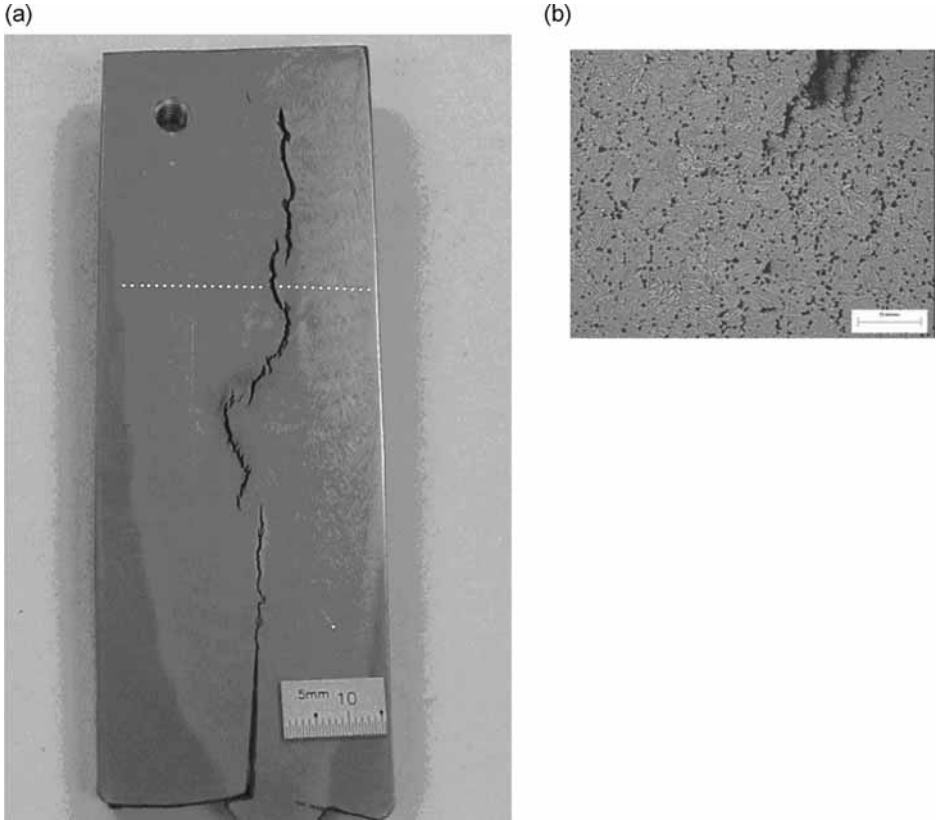


FIG. 3—Example of cracking from an ex-service weld where independent creep and fatigue related defects have been formed (a). The macro defect initiated at the inside surface appears to be related to fatigue. The midwall damage in transgranular and is typical of creep, see detailed micrograph (b).

ex-service component, Fig. 5 [5]. This photograph illustrates the fact that damage in service components maybe significantly more difficult to identify as creep fatigue than is the case when laboratory tests are performed under controlled conditions. Analysis of the operating conditions associated with the cracking shown in Fig. 5 indicated that the cyclic stresses were relaxed by creep which occurred in the Type IV region of the weld heat affected zone (HAZ). Thus, the cyclic loading resulted in locally higher stress conditions than would have been experienced from pressure alone and it was the local damage developed from the stress relaxation (i.e., relaxation occurs by strain accumulation) which resulted in cracking. It is worth noting that while this example is a special case, in service damage frequently occurs at welds. This occurs in part because welds typically exhibit heterogeneous microstructure and properties (the so called metallurgical notch) and in part because welds often have local geometric stress concentrations. It is well understood that relaxation of high stress may cause

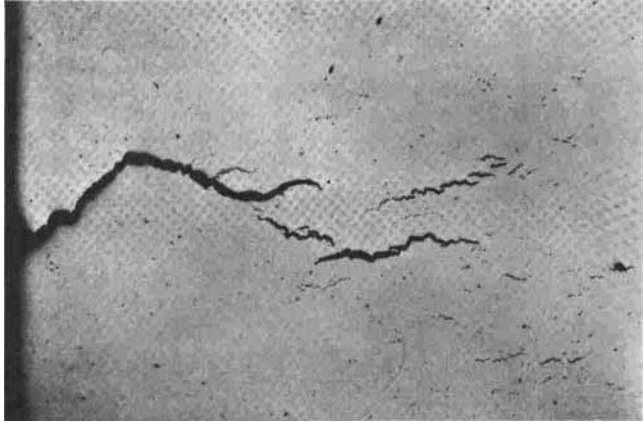


FIG. 4—Example of creep fatigue damage developed in a laboratory test specimen [6].

cracking in brittle regions, e.g., as is the case with stress relief cracking. However, systems or cyclic loads can be concentrated in the weaker regions of a weld leading to accelerated creep damage and crack formation, e.g., as with Type IV cracking which occurs in the HAZ of several alloy steels.

Laboratory Testing Methodologies

Creep-fatigue tests are typically carried out to generate material property input data for use in models for assessment of component behaviour. In some cases these tests are performed under idealised cycles since, in

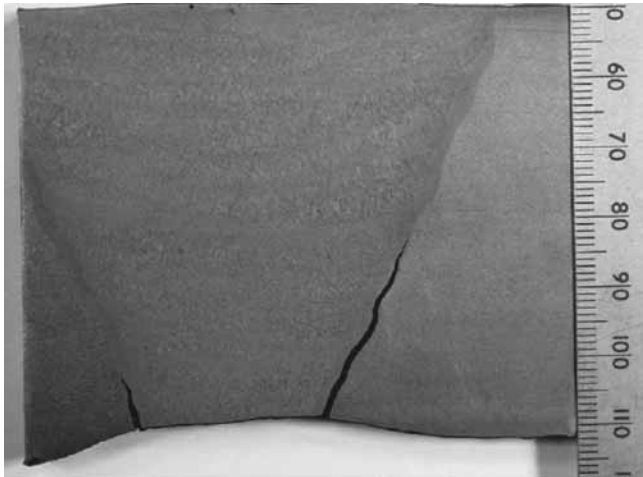


FIG. 5—Example of damage developed in a low alloy steel weldment after elevated temperature operation with both steady and cyclic stresses present [7].

general, relatively simple cycles facilitate subsequent analysis. These tests lead to the development of models such as cyclic/hold deformation and endurance data models. Alternatively, testing may be performed under conditions believed to directly represent service. Thus, by application of a repeated typical, representative service - related cyclic/hold, it may be possible to assess life empirically. Furthermore, these tests provide results which can be used to verify the effectiveness of creep-fatigue assessment procedures in close-to-service conditions.

It is important to show that the methods used to correlate laboratory data with plant behavior are relevant. In that regard, component targeted feature-specimen tests have often been considered essential to validate models designed to link small scale tests to more complex behavior. Typically these feature tests also allow benchmarking of design methods and evaluation of remaining life predictions made using available creep-fatigue assessment procedures. One opportunity which feature tests provide is that it is normally the case that instrumentation can be included to monitor the development of damage. The ability to identify the factors affecting crack initiation and the details of how crack link up and propagation take place are particularly important when planning inspection and maintenance campaigns for plant components.

Analytical Methodologies to Estimate Crack Initiation

Methodologies for performing analytical evaluation of crack initiation include:

- Time fraction.
- Strain fraction/Ductility exhaustion.
- Simplified methods for total strain.
- Estimate creep strain with time for applicable stress/temperature combinations.
- Strain to Failure, elongation/Reduction of Area or other method for different conditions.

The behavior of different steels has been reported in major studies as well as in individual papers. It should be emphasized that full analysis requires knowledge of a range of materials information. In many cases, full data has not been developed for an alloy under consideration the analysis is based on certain assumptions. In addition, in most cases laboratory assessments have concentrated on parent properties because less data exists for weld metals, HAZ or overall weldment performance.

Review of recent developments associated with application of the British Energy R5 code with regard to estimation of crack initiation demonstrates that:

- A 'stress modified' ductility exhaustion approach has been derived from consideration of cavity nucleation and growth. This approach gives less pessimistic predictions of failure for cycles where the creep dwell is positioned at a stress below the peak in the cycle and when the strain range is low.
- Models have been developed to include making assessments which include the effects of multiaxial states of stress.

- The majority of creep-fatigue problems are associated with welds. The current advice for creep-fatigue initiation for welds is complicated and can be excessively conservative. Thus, changes have been proposed to R5 Vol. 2/3 App. A4. These changes involve replacement of fatigue strength reduction factors (FSRF) by:
 - Weld strain enhancement factor (WSEF), which takes account of material mismatch and geometry and weldment endurance reduction (WER), which takes account of the reduced endurance of a weldment.

Although in combination these give a similar result to a FSRF so the fatigue damage is unchanged when only the WSEF is used as in the determination of the start of dwell stress, the calculated creep damage fraction is reduced.

In addition, there are several proposed modifications to account for the effects of compressive dwells. These can be summarized as:

- Fatigue strain range is enhanced for the presence of a creep dwell.
- The fatigue damage should be calculated by removing the nucleation phase of fatigue endurance. This makes allowance for the physical processes that reduce fatigue endurance in the presence of a compressive dwell.
- The creep damage per cycle is zero.

Future considerations for developments of the R5 methodology include:

- Engineering or True Strain at Failure?
- Further validation of 'stress modified' approach for type 316H parent (testing is advanced) and for type 316H HAZ (creep-fatigue testing to start, creep tests in progress). Evaluation is also planned for 1CrMoV rotor steel (using existing data).
- Improved prediction of stress relaxation.
- Improved prediction of thermo-mechanical cycle and start of dwell stress:
 - Modified elastic modulus methods, which provide an approximate computational means of estimating more accurate values of start-of-dwell stress, elastic follow-up factor and strain range than are obtained using a basic shakedown analysis.
 - Inelastic analysis including the development of constitutive models for austenitic parent and weld metals.
- Further work on the influence of high residual stresses in weldments.
- Derivation of the WSEFs for ferritic materials.

Materials Information

An evaluation of available creep fatigue information has highlighted concerns regarding a lack of metallurgical data, in particular, concerning:

- Long term creep behaviour.
- Mechanical behaviour.
- Behaviour of welds.
- Fabricability/manufacturing.

It should also be noted that in many cases the results of a test program are limited to creep rupture lives and endurance data either for different ramp rates or for different hold times. It is apparent that material creep ductility will exhibit a critical effect on the type of damage developed and how that damage will influence creep fatigue interaction. The data presented in Fig. 6 provide an illustration of this effect. Three turbine low alloy steels were tested and all exhibited similar creep rupture strength. However, the creep ductility of the three alloys was different and this difference was a factor in the level of life reduction noted in cyclic tests with hold time. Thus, the alloy with the highest ductility exhibited very little life reduction when hold times were included in the cycle. Conversely, creep brittle material revealed a significant reduction in cycles to failure for cycles with hold time [8].

Creep Fatigue in Creep Strengthened Ferritic Steels

The behavior of creep strengthened ferritic steels such as Grade 91 and Grade 92 has been evaluated by a number of investigators. Typically, this type of evaluation involves performing low cycle fatigue tests at different ramp rates or performing cyclic tests which include hold time. In both cases data are normally compared to the results of tests performed with rapid cyclic ramps. Typical results for Grade 91 steel are illustrated in Fig. 7. Here both the tests with slow ramps and hold times show reduced lives. Interestingly tests where the hold was in tension or compression or both show similar lives.

This observation differs in detail to results from an alternative study considering creep-fatigue-oxidation interactions in Grade 91 steel. Since this steel typically enters service with a tempered martensitic structure it is susceptible to microstructural changes with exposure time at high temperature. Detailed microstructural studies have shown that cyclic work softening was accelerated by hold times at peak temperature. Thus, marked differences in microstructure were found for creep fatigue exposure compared to similar samples cycled without hold; for example Fig. 8. In this study the most extreme life reductions for creep fatigue exposure were noted in samples which had experienced compressive holds. This behavior was consistent with the greater incidence of surface cracking observed under these conditions, Fig. 9. The cracking observed was consistent with scale disruption and internal oxidation which promoted defect initiation in the steel substrate.

The additional complications involved with assessment of creep fatigue behavior of strengthened Ferritic steels arise in part because:

- The initial microstructure is complicated and can be varied by the fabrication process used, these variations may or may not be identified by optical microscopy.
- The creep strength and ductility depend on both the detailed chemical composition and processing details. These differences in strength cannot simply be identified using optical metallography.
- Surface effects in general and scale formation in particular can also be involved in crack initiation. The details of how the scales form, whether

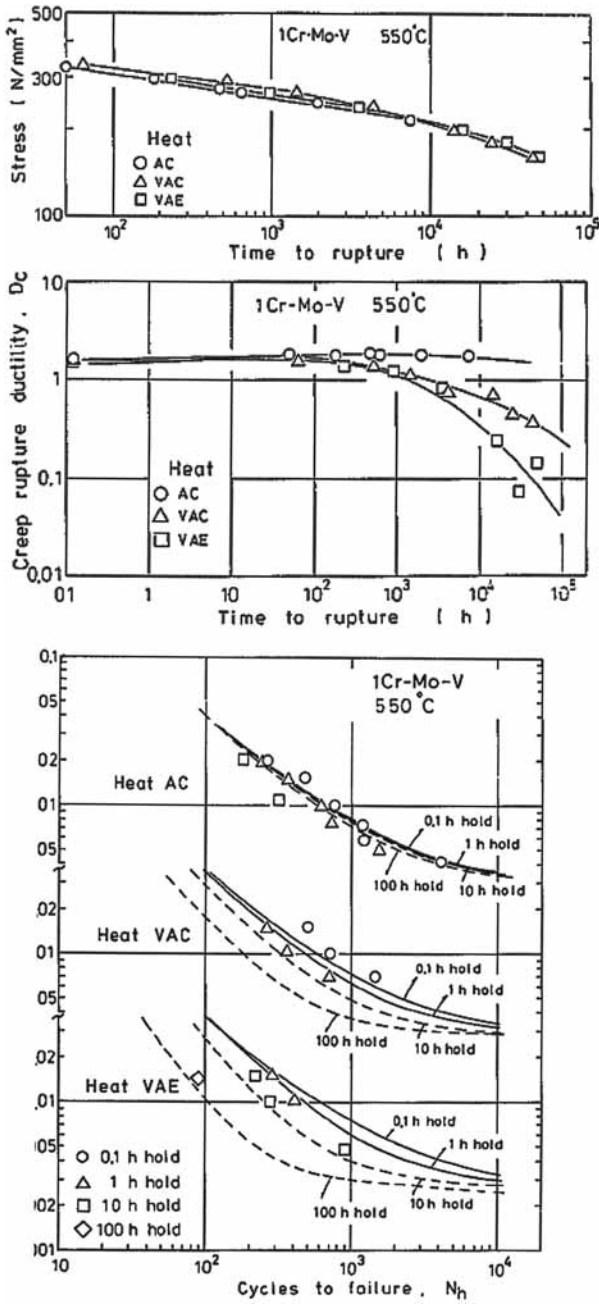


FIG. 6—Creep rupture strength and ductility for three 1CrMoV steels (a) and cyclic endurance curves for these steels showing the different effect of hold time (b) [8]. Although all three steels have apparently similar creep strength the differences in creep ductility influence the creep fatigue interaction behavior.

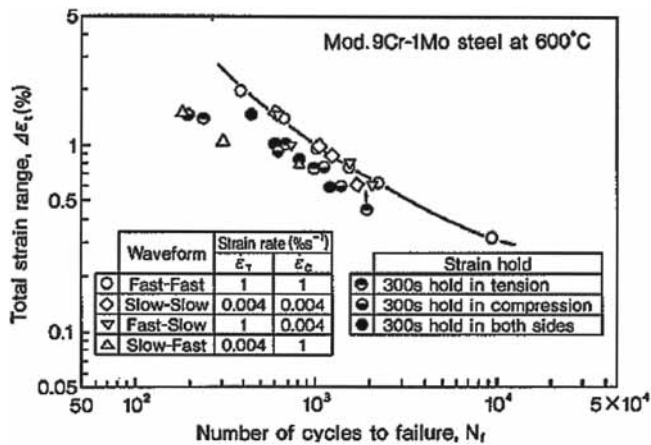


FIG. 7—The effect of different ramp rates and hold times on the number of cycles to failure of Grade 91 steel tested at 600°C, [9].

they crack and if so whether these cracks initiate defects in the underlying substrate requires a level of understanding not presently established.

In view of these and other complications in understanding and analysis of creep fatigue behavior it is not surprising that when data are analyzed using the traditional approaches the results can appear very pessimistic. For example, assessment of data in the ASME section NH applied a very conservative interpretation to the expected creep behavior for Grade 91 steel, Fig. 10. As shown

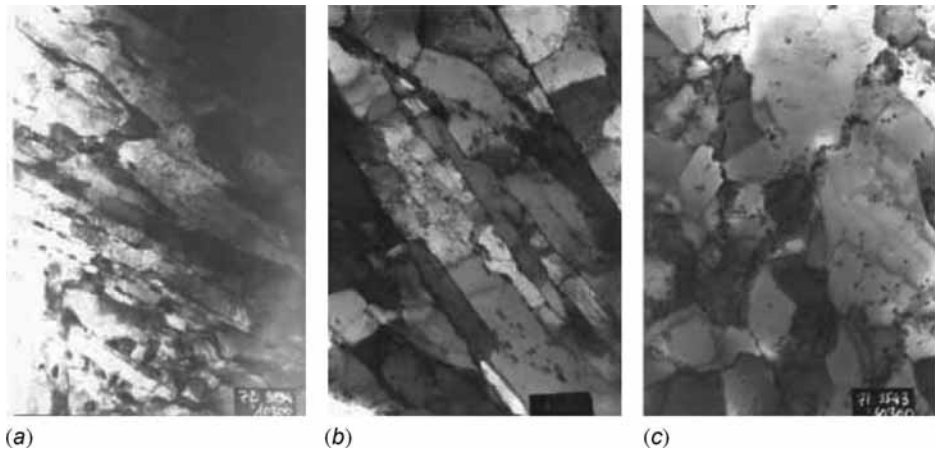


FIG. 8—Microstructures observed using transmission electron microscopy of Grade 91 steel. The micrographs show the pretest microstructure (a), the microstructure after cycling at elevated temperature without hold (b), and the microstructure after cycling at elevated temperature with hold times (c) [10].

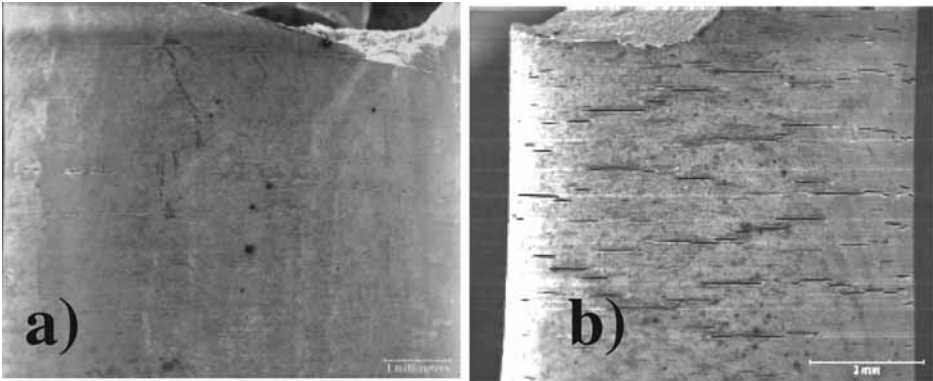


FIG. 9—Comparison of the specimen surface of Grade 91 samples tested under creep fatigue conditions. The sample after holding on the tensile stress side, shown in (a), exhibited limited surface cracking while the sample after compressive holds (b) revealed many surface cracks and microcracks, Ref [11].

this conservative approach implied that the creep fatigue endurance for Grade 91 steel was markedly worse than that for conventional boiler steels. This conservatism is likely a function of the fact that the analysis approaches did not take into account the complications associated with variable strength, microstructural changes, strain softening and the influence of surface scale. Indeed, recent analysis which accounts for the strain softening behavior provides a reasonable approach to creep-fatigue of Grade 91 steel, Fig. 11.

Work performed by CRIEPI in Japan has summarized analysis of many creep-fatigue tests carried out on the austenitic alloy 316FR and the 9 and

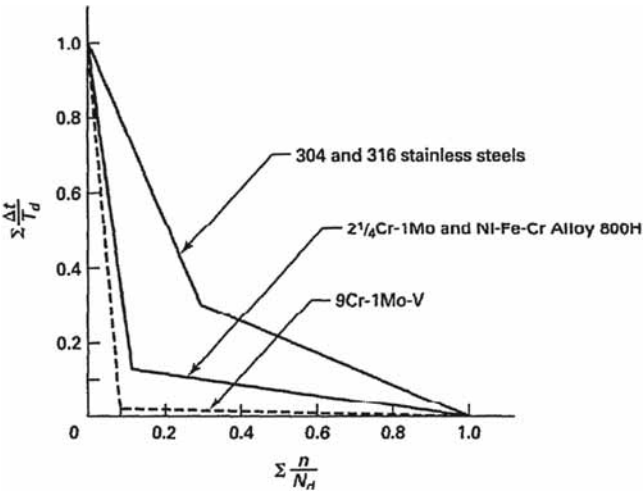


FIG. 10—Creep fatigue diagram suggested by the ASME section NH [12].

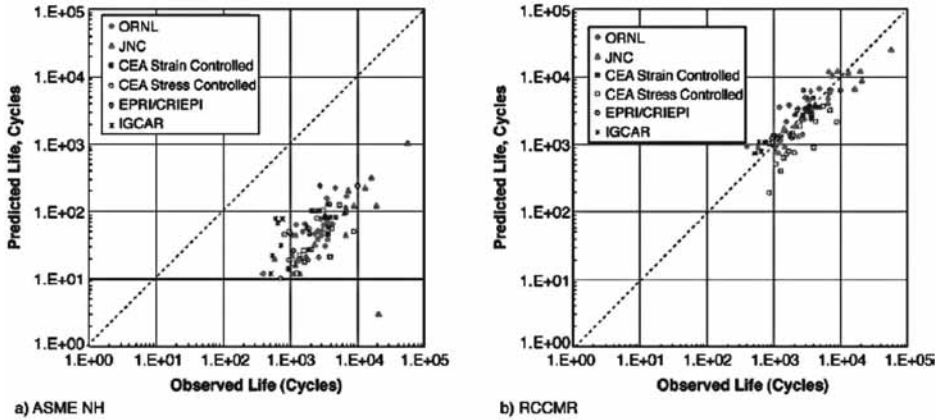


FIG. 11—Comparison of life fraction estimates for Grade 91 steel using the procedures in (a) ASME NH and (b) RCCMR [13].

12 % Cr steels Grade 91, Grade 122, and TMK1. These test results included data from several long term tests. The main findings of this work were summarized as:

- Time fraction type stress based creep damage is insufficient for predicting life reduction due to holding. This is especially the case at the small strain ranges of practical interest. This cannot be covered by any interaction diagram.
- Calculation of ductility exhaustion type strain based creep damage tends to overestimate creep fraction used per cycle when all inelastic strain during hold period is counted.
- By decomposition of inelastic strain or extracting ductility exhausting effect using a rate dependent overall ductility, life reduction can be estimated with reasonable accuracy (within factor of two).
- Low cycle fatigue endurance can be described using the Coffin Manson method but creep fatigue damage analysis should recognize that when there is interactive damage there is an acceleration of fatigue damage because of ductility reduction as a consequence of the accumulated creep damage.

Review of data generated in Japan demonstrated that the new steels P23, P92 and P122 have superior creep strength, and comparable low-cycle fatigue and creep-fatigue strength compared with conventional steel P22 and P91. However, plotting creep fatigue data on the conventions life fraction interaction diagram indicates very short life expectancies. It appears that the excessive life reduction in creep-fatigue testing of laboratory specimens can be mainly attributed to the strain concentration in the specimen due to the deformation, such as bulging and necking induced during creep-fatigue test. The detailed measurements performed in this evaluation are of a type rarely undertaken. The fact that post test laboratory evaluation provides key data to assist interpretation of results was noted and offers further weight to the earlier recommendation

regarding the need for laboratory test programs to include post test examination.

It was also noted that the creep-fatigue endurance life of P23, P92 and P122 can be evaluated by using an inelastic strain range including creep strain and Coffin-Manson relationship obtained by low-cycle fatigue test without a creep effect. However, it should be noted that Yamaguchi showed that good relationships required that the creep strain was normalized by the appropriate value of creep ductility.

Especially for boiler materials uncertainty exists about extrapolation from short term tests to the long-term creep range so that there is reduced confidence that the expected performance will have the normal factors of safety. Since in service experience is not available regarding damage mechanisms (e.g., corrosion, creep damage) for these materials and operating conditions it will be necessary to adopt a series of measures to evaluate performance. These include:

- Additional instrumentation to permit measurement of key factors (e.g., temperature, pressure, creep deformation, displacement) at critical component locations.
- Storage and retrieval of information in component data bases to allow more accurate life time consumption calculation.
- Generation of additional laboratory materials data, in parallel to service operation so that the accuracy of extrapolation methods can be reviewed. In addition, by conducting tests on sections used to manufacture actual components it will be possible to develop heat specific data which allows the determination of the location within the scatter band.

Concluding Remarks

Operators of power generating plant are adopting operating practices that lead to increasingly severe cycles, both in terms of number or magnitude or both. This cyclic behavior may involve the traditional conditions associated with starts and stops. However, very many other operating factors can introduce transient loading. These include thermal effects from use of attemperators, quenching from condensate and even stress transients from operational induced pressure changes. The increased complexity of these transients results in an on-going requirement to assess the creep-fatigue performance and reliability of many boiler and turbine components.

Recent installations of combustion turbines and heat recovery steam generators have increased the number of materials and components that operate at high temperature under cyclic loading conditions. New high-efficiency coal plants will operate with even higher steam and metal temperatures than those of traditional plants. New components will use different alloys and manufacturing methods than those of conventional fossil plants. As a result, creep-fatigue damage will occur more frequently and in more components. Understanding the ways in which cyclic operation modify high temperature performance and being able to monitor damage development in components with confidence are increasingly key aspects for ensuring safe and reliable operation of power generating plant.

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Mike W Spindler¹ and Warwick M Payten²

Advanced Ductility Exhaustion Methods for the Calculation of Creep Damage During Creep-Fatigue Cycling

ABSTRACT: The thermal mechanical cycling of high temperature components can result in creep-fatigue cycles in which the creep dwell has a wide variety of positions within the cycle. During in-phase cycling the creep dwell is placed at the maximum strain in the cycle. Out of phase cycling gives compressive dwells and phase cycling between 0 and 180 gives a cycle in which the dwell is placed before the maximum strain is reached, i.e., at an intermediate position within the cycle. Furthermore, components often experience cycles with a variety of different strain amplitudes, which can result in cycle sequencing effects. The effects of these different types of creep dwells have been investigated as part of the development programme for the R5 High Temperature Life Assessment Procedure. The materials included in the R5 programme were two low alloy ferritic steels; a cast 1CrMoV and a cast 1/2 CrMoV and three austenitic stainless steels; a type 316H steel, a cast type 304L, and a type 347 weld metal. The analysis of these tests has resulted in the proposal of a new method to calculate creep damage. This has been shown to give better predictions for the creep damage at failure in laboratory tests compared with both the current R5 ductility exhaustion approach and the time fraction approach, which is used in typical design codes such as ASME III and RCC-MR. In particular, the new method gives significantly improved predictions of creep damage at failure for creep-fatigue cycles with intermediate dwells and for cycles with low strain ranges, which are of particular relevance to the service cycles in real plant. This paper reviews the

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¹ EDF-Energy, Assessment Technology Group, Barnett Way, Barnwood, Gloucester, GL4 3RS, United Kingdom, e-mail mike.spindler@edf-energy.com

² Australian Nuclear Science and Technology Organisation, BMP 1, Menai 2234 NSW, Australia.

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findings of the work on the new method, the effects of multiaxial states of stress and the effects of compressive dwells.

KEYWORDS: creep-fatigue, creep damage, austenitic stainless, low alloy ferritic

Nomenclature

- a_0 = crack depth used to define crack initiation in specimen or structure
- a_i = crack depth of 0.02 mm which is used to define crack nucleation in specimen or structure
- A_1, P_1, m_1, n_1 = parameters in stress modified ductility exhaustion model, Eqs 12, 15, and 16
- b, B'' = parameters in stress relaxation Eq 3
- B, n_1 = parameters in current R5 creep ductility model, Eq 7
- c, f = location of interaction locus of bilinear creep-fatigue damage interaction diagram, (c, f) , where c is coordinate of creep damage and f is coordinate of fatigue damage
- d_c = creep damage per cycle
- d_f = fatigue damage per cycle
- D_c = total creep damage per test
- D_f = total fatigue damage per test
- $N_{5\%}$ = number of cycles to 5 % decrease in maximum stress
- $N_{25\%}$ = number of cycles to 25 % decrease in maximum stress
- N_0 = continuous cycling fatigue endurance to create crack of depth a_0
- N_f = number of cycles failure as defined by complete separation
- N_g = number of cycles to grow fatigue crack from nucleation size of 0.02 mm to a_0
- N_i = number of cycles to nucleate fatigue crack of size of 0.02 mm
- N_{pred} = predicted number of cycles to initiation of creep-fatigue crack
- N_{obs} = observed number of cycles to the chosen failure criterion
- p, q = Parameters in current R5 model for effect of multiaxial states of stress on creep ductility, Eq 8
- t = time
- t_f = time at failure
- t_h = dwell time
- T = temperature
- Z = elastic follow-up factor
- $\Delta\epsilon_T$ = total strain range
- $\Delta\sigma$ = stress drop during creep dwell, $\sigma_0 - \sigma$
- $\dot{\epsilon}_c$ = instantaneous creep strain rate
- $\bar{\epsilon}$ = von Mises creep strain rate
- ϵ_f = uniaxial creep ductility
- $\bar{\epsilon}_f$ = von Mises creep ductility
- ϵ_L = lower shelf creep strain at failure

- ε_U = upper shelf total inelastic strain at failure
- σ = instantaneous stress
- $\bar{\sigma}$ = von Mises stress
- σ_0 = initial stress during creep dwell
- $\sigma_{1,2,3}$ = principal stresses where σ_1 is maximum principal stress and $\sigma_2 \geq \sigma_3$
- σ_H = hydrostatic stress
- σ_{rup} = multiaxial rupture stress

Introduction

The thermal mechanical cycling of high temperature components can result in creep-fatigue cycles in which the creep dwell has a wide variety of positions within the cycle. During in-phase cycling the creep dwell is placed at the maximum strain in the cycle (Fig. 1(a)). Out of phase cycling gives compressive dwells (Fig. 1(c)) and phase cycling between 0° and 180° gives a cycle in which the dwell, which is at the maximum temperature, is placed before the maximum strain is reached (Fig. 1(b)), i.e., at an intermediate position within the cycle (or intermediate dwell). Furthermore, power plant components often experience cycles with different strain amplitudes (Fig. 1(d)), which lead to cycle sequencing effects and the creep dwells do not necessarily experience pure strain control but may contain elastic follow-up (Fig. 1(a)).

The effects of these different types of creep dwells have been investigated as part of the development programme for the R5 High Temperature Life Assessment Procedure [1]. Creep and fatigue damages have been calculated following the approaches that are used in high temperature design codes, such as ASME III [2] and RCC-MR [3] and life assessment procedures such as CRIEPI [4,5] and R5 [1]. In the ASME III and RCC-MR design procedures, calculations of creep damage are carried out using the time fraction approach. The CRIEPI and R5 life assessment procedures use ductility exhaustion approaches to calculate creep damage. The materials included in this study were three austenitic stainless steels; a type 316H steel [6], a cast type 304L [7], and a type 347 weld metal [8,9] and two low alloy ferritic steels, a cast 1CrMoV [10] and a cast 1/2CrMoV [11]. For convenience isothermal cycling of uniaxial specimens has been used for all of these tests. Nevertheless, a wide range of different types of dwells have been studied. For each of these materials both the time fraction and ductility exhaustion approaches were used to calculate the creep damage and all other calculations remained the same. Notably, the work programme has resulted in the proposal of an improved method to calculate creep damage. This new method is referred to as the "stress modified ductility exhaustion" approach (SMDE).

It was shown in the research papers [6,7,9–11] that the new SMDE approach gave significantly improved predictions of creep damage at failure for cycles with intermediate dwells and for cycles with low strain ranges and peak dwells. These types of cycle and dwell are typical of most practical plant cycles and thereby the new SMDE approach would have a significant benefit under the most common cycles for which engineering assessments are conducted. In order to implement the SMDE approach in practical applications it is also

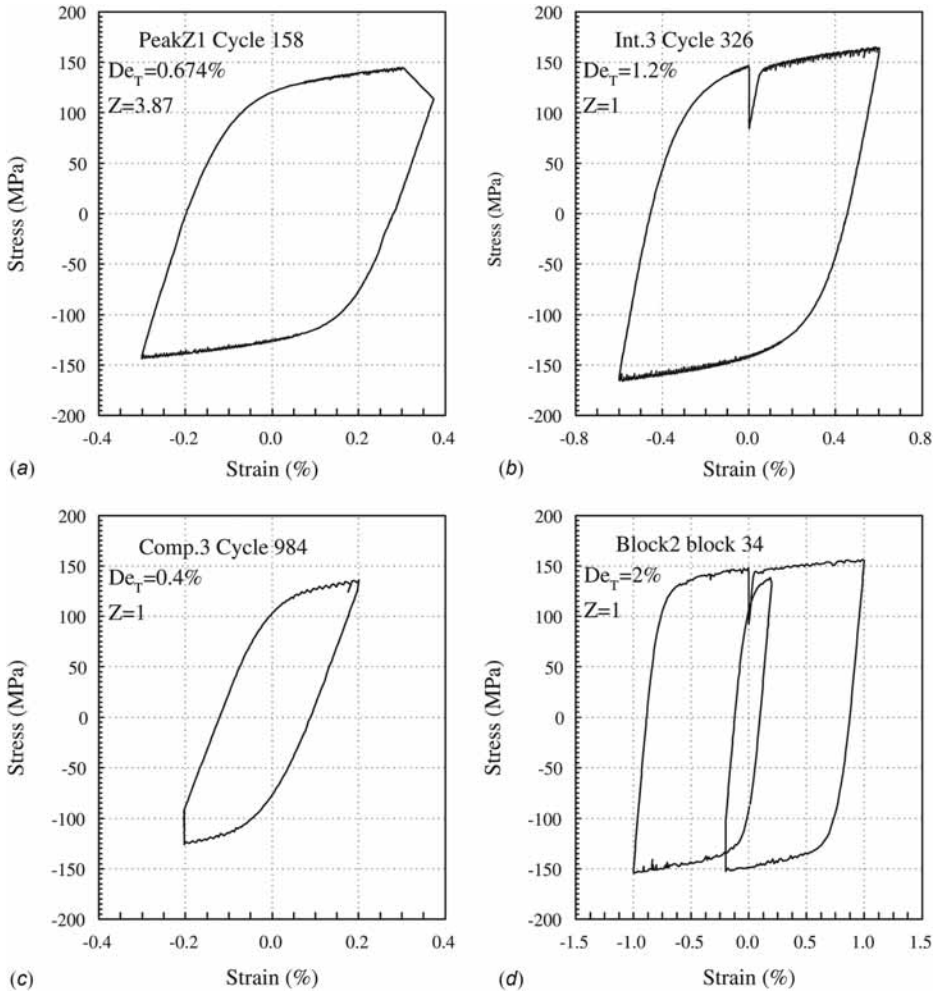


FIG. 1—Cycle types for creep fatigue tests on cast type 304L [7], (a) peak dwell with elastic follow-up, (b) intermediate dwell, (c) compressive dwell, (d) block loading with intermediate tensile dwell in the major cycle and a compressive dwell in the minor cycle.

necessary to have a complementary model for the effects of compressive dwell and for dwells with multiaxial states of stress [12], which are also described in this paper.

This paper reviews the findings of the previous research papers and in particular on the implications for practical engineering assessments of real plant components. The purpose of this paper is to provide background to the implementation of the SMDE approach into the R5 high temperature life assessment procedure [1].

Uniaxial Creep-Fatigue Tests

The creep-fatigue tests on type 316H Cast BQ at 570°C [6] were reported previously by Wareing et al. [13]. These tests were conducted with peak dwells. This is a large dataset of 26 tests with strain ranges from 0.35 to 1.85% and dwell times from 0.5 to 47.5 h.

The cast type 304L material [7] was tested at 650°C. Four types of creep-fatigue tests were conducted; (i) tests with peak dwells (Fig. 1(a)), (ii) tests with intermediate dwells at 0% strain (Fig. 1(b)), (iii) tests with compressive dwell (Fig. 1(c)), and (iv) two cycle sequencing tests, which involved 15 or 16 minor cycles followed by a major cycle with an intermediate dwell at 0% strain. The minor cycle of one sequencing test experienced a compressive dwell (Fig. 1(d)). The dwells in ten of the tests were conducted under strain control and in two of the tests the dwells were controlled under combined load/strain control to an elastic follow-up factor of approximately 3.8 (Fig. 1(a)). The dwell times were up to one hour.

The data on type 347 weld metal [8,9] included tests conducted by Bretherton et al. [14]. All of the tests were conducted on welds made using the same welding wire. Some tests [9] were conducted with the specimens manufactured in the axial orientation. Whereas the specimens tested by Bretherton et al. [14] were manufactured in the transverse orientation. All specimens were tested in the as-welded condition at a temperature of 650°C. The creep-fatigue tests on axial specimens were at strain ranges of 0.6 and 1.2%, with creep dwells of 1 h duration, which were placed either at the peak of the fatigue cycle or at an intermediate position. The dwells in eight of the tests, on axial specimens, were conducted under strain control and in two of the tests the dwells were controlled under combined load/strain control to an elastic follow-up factor of approximately 2. The creep-fatigue tests on transverse specimens [14] were all conducted with peak dwells, strain ranges of between 0.37 to 1.51% and with dwell times of 1, 5, 10, and 23.3 h.

The cast 1CrMoV steel was tested at 550°C as part of the Brite-Euram Project No. BE 5245 "C-FAT." Creep-fatigue tests were conducted with either peak dwells or with intermediate dwells at 0% strain. The strain ranges were between 0.6 and 2%. The effect of elastic follow-up was investigated by comparing tests with elastic follow-up factors of 1 (strain control), 1.5, and 2.5. The majority of the tests had dwell times of 1/2 an hour, although two of the tests, with peak dwells, had 16 h dwells.

The data for cast 1/2CrMoV steel at 550°C was taken from the work of Batte et al. [15]. Both push-pull and reversed bend creep fatigue tests were conducted. The push-pull tests had peak dwells and the reversed bend tests had either peak dwells or intermediate dwells at 0 displacement. The reverse bend tests were conducted under displacement control and examination of the hysteresis loops showed that during the dwell elastic follow-up occurred with an average value of 2.5. The majority of the tests had dwell times of 16 h, although three of the tests, with peak dwells, had 1/2 hour dwells and one had a 2 h dwell. Of particular note was an unfailed test which experienced 3547 cycles with a 16 h dwell and thereby lasted 57k hours.

For each of the materials investigated and in the case of the weld metal for both orientations there are continuous cycling fatigue, tensile, and creep data at the same temperature as the creep-fatigue tests.

Current Methods to Calculate Creep and Fatigue Damage

The models to predict creep-fatigue failure that have been investigated in this paper calculate total creep damage, D_c , and total fatigue damage, D_f , separately. Failure is conceded when the sum of the fatigue and creep damages reaches some failure criterion; for example, in R5 [1] a linear damage summation is used and failure is conceded when $D_c + D_f \geq 1$. It is usual for the failure criterion to vary depending on the method used to calculate the creep damage and the interaction diagram used in ASME III [2], RCC-MR [3], for austenitic steels, is bi-linear with a locus at (0.3,0.3), whereas for low alloy ferritic steels ASME III uses a locus at (0.15,0.15).

For the interpretation of uniaxial creep-fatigue test data in the research papers [6,7,9–11] mean fatigue endurance and mean creep failure data for the same heat of steel or for the same weld metal were used. Thus, best estimates of creep and fatigue damages were calculated. For plant assessments lower bound fatigue endurance and creep failure data, which have been derived from multi-heat datasets, would be used to give conservative upper bound estimates of the damages. These upper bound estimates of creep and fatigue damage are presented for the first time in this paper. The methods for calculating fatigue and creep damage are described in the following sections.

Fatigue Damage

The approach for the calculation of fatigue damage per cycle, d_f , in R5 [1] is simply

$$d_f = 1/N_0(\Delta\epsilon_T, T) \quad (1)$$

where N_0 is the continuous cycling fatigue endurance to create a crack of depth a_0 at the appropriate total strain range, $\Delta\epsilon_T$, and temperature, T . The total fatigue damage, D_f , is then summed for the total number of cycles and taking account of any cycles with different strain ranges that result in different damages per cycle. In the R5 procedures the fatigue strain range is enhanced for the presence of a creep dwell by adding the relaxation strain. For plant analyses of thin-section components a_0 is often set to 10% of the cross-section so that an un-cracked body stress analysis is appropriate. However, for multi-heat datasets only the number of cycles to the complete separation of a test specimen, N_f , is commonly reported, which can be larger than N_0 . Thus for plant analyses, an empirical model for fatigue crack nucleation followed by a short fatigue crack growth law is used to calculate N_0 from lower bound values of N_f . This will be discussed later as it is not required when analysing the results of tests.

For the analysis of the uniaxial creep-fatigue tests it is important that the same failure criterion is used for both the continuous cycling fatigue and creep-

fatigue tests, as this ensures that the crack sizes are approximately the same. However, because of differences between the available data on each of the five materials different fatigue failure criteria have been used in the research papers. For the cyclic hardening materials a 5 % drop in the peak load was used for the cast type 304L and a 25 % drop in the peak load was used for the type 316H. For the cyclic softening materials failure was defined by either the number of cycles to a percentage decrease in stress from the trend of decrease in peak tensile stress, which was 2 % for the cast 1CrMoV and 5 % for the type 347 weld metal, or by complete separation for the cast 1/2 CrMoV steel.

Time Fraction Approach

The time fraction rule for the estimation of creep damage is widely used in design codes such as ASME III [2] and RCC-MR [3]. The time fraction rule for the creep damage per cycle is

$$d_c^{\text{TF}} = \int_0^{t_h} \frac{dt}{t_f(\sigma, T)} \quad (2)$$

where t_f is the creep rupture time and is calculated as a function of the stress and temperature, which prevail for the increment of time, dt , during stress relaxation and t_h is the dwell time. For the analysis of the uniaxial creep-fatigue tests the stress history during relaxation was described by fitting the following equation to the experimental stress drop, $\Delta\sigma$

$$\Delta\sigma = \sigma_0 - \sigma = \sigma_0 B'' \ln(bt/Z + 1) \quad (3)$$

where σ and σ_0 are the instantaneous and initial stresses, respectively, t is the time, and B'' and b are material constants and Z is the elastic follow-up factor.

For the time fraction approach the creep damage under compressive and multiaxial states of stresses is calculated using a multiaxial rupture stress, σ_{rup} . For type 304 and type 316 stainless steels ASME III [2] uses

$$\sigma_{\text{rup}} = \bar{\sigma} \exp \left[0.24 \left(\frac{3\sigma_H}{\sqrt{\sigma_1^2 + \sigma_2^2 + \sigma_3^2}} - 1 \right) \right] \quad (4)$$

where σ_1 , σ_2 , and σ_3 are the principal stresses, $\bar{\sigma}$ is the von Mises stress, and σ_H is the hydrostatic stress. RCC-MR [3] uses

$$\sigma_{\text{rup}} = 0.133 \times 3\sigma_H + 0.867 \bar{\sigma} \quad (5)$$

R5 Ductility Exhaustion Approach

The ductility exhaustion approach of R5 to calculate the creep damage per cycle, d_c^{R5} , is given by

$$d_c^{R5} = \int_0^{t_h} \frac{\dot{\epsilon}_c}{\epsilon_f(\dot{\epsilon}_c, T)} dt \quad (6)$$

where $\dot{\epsilon}_c$ is the instantaneous creep strain rate and $\epsilon_f(\dot{\epsilon}_c, T)$ is the corresponding creep strain at failure at the appropriate temperature, T , as a function of the creep strain rate.

The current approach in R5 is to use the creep strain at failure at an appropriate temperature in conjunction with the average creep strain rate, ϵ_f/t_f , to define the relationship between ductility and strain rate. The creep strain at failure is calculated from the elongation at failure minus the plastic loading strain. The relationship between creep strain at failure and average creep strain rate usually takes the form of a lower shelf at low strain rates, ϵ_L , and an upper shelf at high strain rates, ϵ_U , which are connected by a transition region in which the ductility is a function of the creep strain rate. This relationship can be represented by

$$\epsilon_f = \min[\epsilon_U, \max(\epsilon_L, B\dot{\epsilon}_c^{n_I})] \quad (7)$$

where B and n_I are material constants

In R5 the effect of multiaxial states of stress on creep damage are considered by using the von Mises creep strain at failure and the von Mises creep strain rate. The von Mises creep strain at failure is calculated by multiplying the uniaxial creep strain at failure by a multiaxial ductility factor, MDF, which is given by

$$\bar{\epsilon}_f(\dot{\epsilon}_c, T) = \epsilon_f(\dot{\epsilon}_c, T) \text{MDF} = \epsilon_f(\dot{\epsilon}_c, T) \exp \left[p \left(1 - \frac{\sigma_1}{\bar{\sigma}} \right) + q \left(\frac{1}{2} - \frac{3\sigma_H}{2\bar{\sigma}} \right) \right] \quad (8)$$

where p and q are constants. The multiaxial creep strain at failure of the type 304 and 316 data have been analysed [16] and shown that for type 304 at 593°C $p=2.38$ and $q=1.04$ and for type 316 at 593 to 600°C $p=0.15$ and $q=1.25$. For compressive dwells the R5 approach calculates creep damage using the upper shelf ductility. Thus, Eq 6 simplifies to $d_c = Z\Delta\sigma/(E\epsilon_U)$ where E is Young's Modulus. For all of the tests the instantaneous creep strain rate was obtained by differentiating Eq 3 and replacing $\dot{\sigma}$ by $-\dot{\epsilon}_c E/Z$, so that

$$\dot{\epsilon}_c = ZB''\sigma_0 b/[E(bt + Z)] \quad (9)$$

Results of Best Estimate Creep and Fatigue Damage Calculations for Uniaxial Creep-Fatigue Tests

For each material the same fatigue data and the stress relaxation data were used. This means that the fatigue damages were always the same and that the only differences between the different cases considered were the method for calculating creep damage and the creep failure data (creep rupture or creep strain at failure). The fatigue and creep damages per cycle (d_f and d_c ,

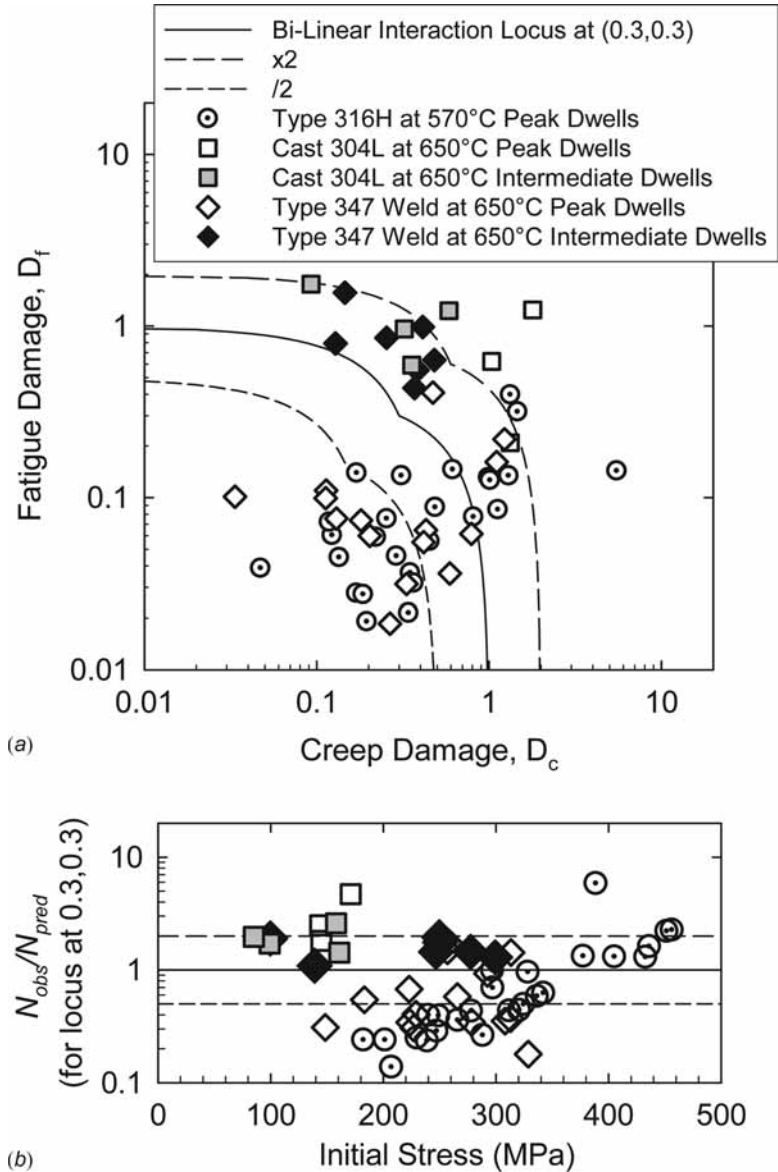


FIG. 2—Best estimate damages for austenitic stainless steels using the time fraction approach; (a) creep-fatigue interaction diagram, (b) N_{obs}/N_{pred} versus initial stress for a dwell at half life.

respectively) are calculated using the methods outlined above. The total fatigue and total creep damages (D_f and D_c , respectively), which are given by the per cycle damage multiplied by the observed number of cycles to the chosen failure criterion, N_{obs} , are then plotted on the creep-fatigue interaction

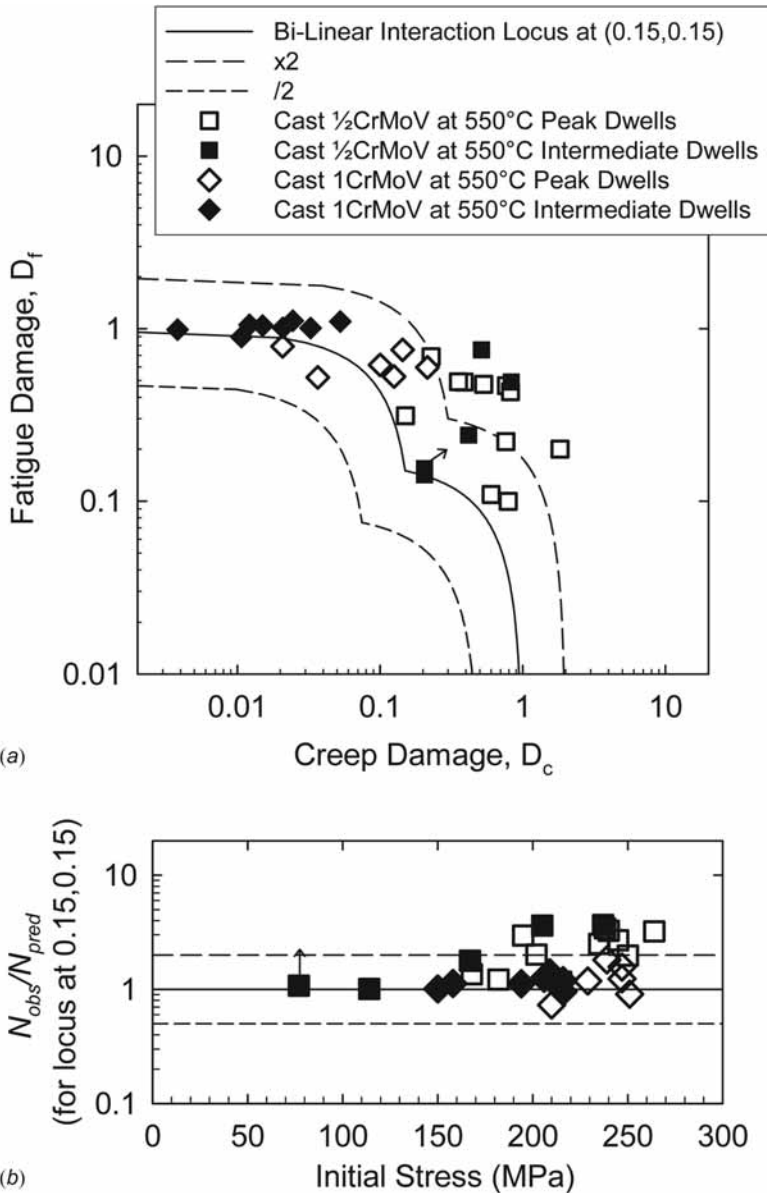


FIG. 3—Best estimate damages for low alloy ferritic steels using the time fraction approach; (a) creep-fatigue interaction diagram, (b) N_{obs}/N_{pred} versus initial stress for a dwell at half life.

diagrams, Figs. 2(a), 3(a), 4(a), and 5(a). Also graphs of the observed number of cycles to the chosen failure criterion, N_{obs} , divided by the predicted number of cycles to the initiation of a creep-fatigue crack, N_{pred} , are plotted versus the initial stress of a mid life dwell in Figs. 2(b), 3(b), 4(b), and

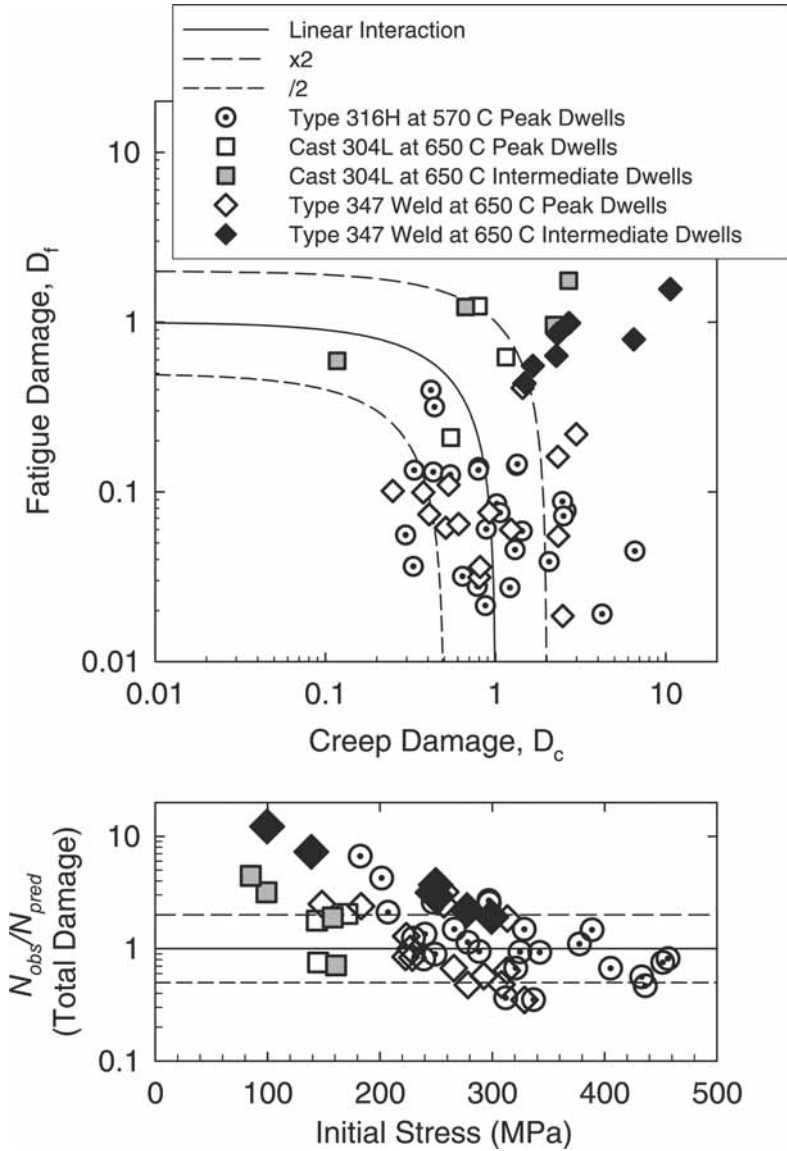


FIG. 4—Best estimate damages for austenitic stainless steels using the R5 ductility exhaustion approach; (a) creep-fatigue interaction diagram, (b) N_{obs}/N_{pred} (or total damage) versus initial stress for a dwell at half life.

5(b). It is useful to note that for the linear interaction the predicted number of cycles to the initiation of a creep-fatigue crack, N_{pred} , equals $1/(d_f + d_c)$ and thereby N_{obs}/N_{pred} equals the total damage ($D_f + D_c$). Thus, these graphs simultaneously show how well both the endurance and also the total damage

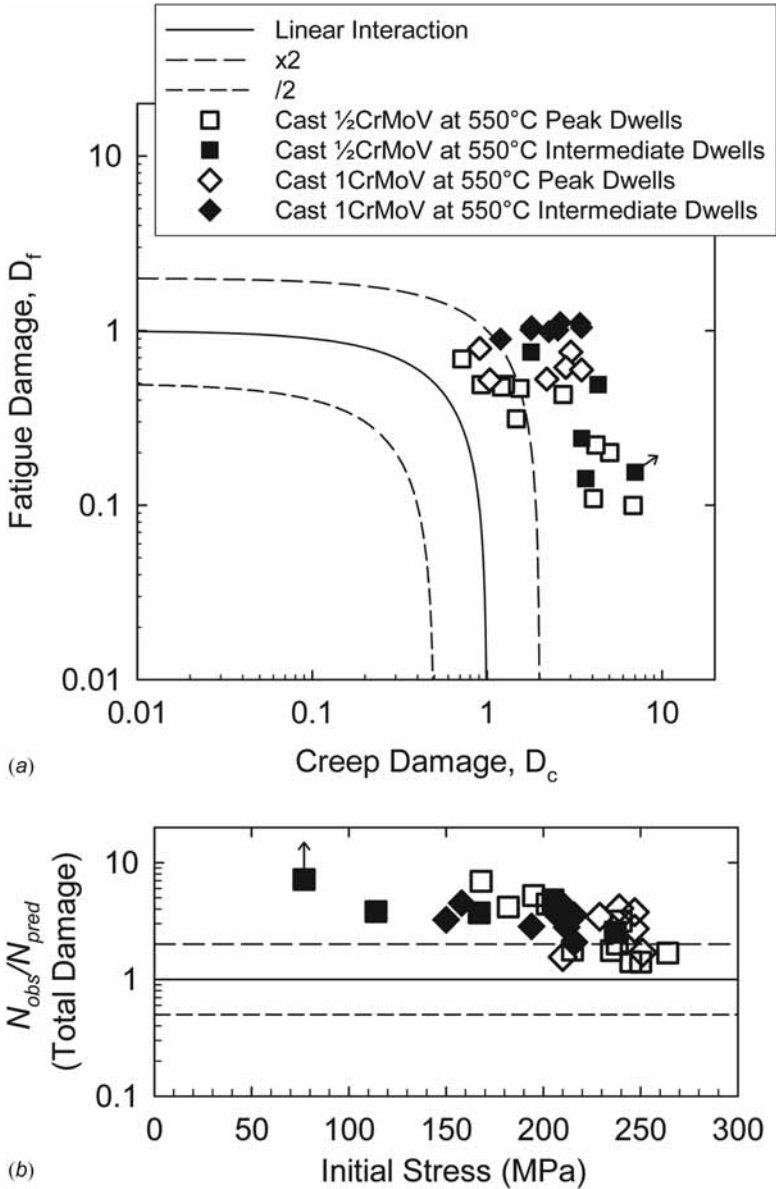


FIG. 5—Best estimate damages for low alloy ferritic steels using the R5 ductility exhaustion approach; (a) creep-fatigue interaction diagram, (b) N_{obs}/N_{pred} (or total damage) versus initial stress for a dwell at half life.

are predicted and in particular because the fatigue damage is always the same these graphs show how well the creep damage is predicted as a function of the initial stress. With regards to the graphs for the time fraction the calculation of N_{pred} is a little more complicated as the bi-linear interaction needs to be taken into account. For an interaction locus at (c, f) , where c is the coordinate of the creep damage and f is the coordinate of the fatigue damage; then for constant amplitude testing this leads to

$$\text{if } d_f/d_c \geq f/c \text{ then } N_{\text{pred}} = \left(d_f + [(1-f)/c]d_c \right)^{-1} \quad (10a)$$

$$\text{if } d_f/d_c \leq f/c \text{ then } N_{\text{pred}} = \left([(1-c)/f]d_f + d_c \right)^{-1} \quad (10b)$$

Similar graphs of $N_{\text{obs}}/N_{\text{pred}}$ versus both the total strain range and N_{obs} have also been prepared but for brevity are not reproduced here. In general, the trends are consistent with those shown by the initial stress.

Time Fraction Approach

The damage interaction diagrams for the austenitic and ferritic materials are shown in Figs. 2(a) and 3(a). For austenitic steels the data in Fig. 2(a) shows a wide degree of scatter and the bi-linear interaction with a locus at (0.3,0.3) does not appear to follow any of the trends in the data. The tests with intermediate dwells are conservatively predicted by the bi-linear interaction whereas the tests with peak dwells are particularly widely scattered. In particular, a large number of tests are non-conservatively predicted by the bi-linear interaction. The graph of $N_{\text{obs}}/N_{\text{pred}}$ versus the initial stress Fig. 2(b) shows that while the type 316H data show a clear trend with stress the data for the type 347 weld metal and cast type 304L show no particular trend. This simply means that it is not obvious how the time fraction approach can be improved for all materials using a consistent modification. Of particular concern for practical engineering applications is the apparent trend in the type 316H data toward non-conservatism at low initial stresses, see Fig. 2(b). This is because real plant applications usually involve relatively low initial stresses.

For the ferritic steels Fig. 3(a) the time fraction approach, with a locus at (0.15,0.15) shows reasonable predictions, albeit time fraction tends to be conservative for the creep damage dominated tests, which is the opposite of that observed for the austenitic steels. In particular, the cast 1/2 CrMoV tests tend towards increasing conservatism at high start of dwell stresses, see Fig. 3(b). Nevertheless, at low start of dwell stresses relevant to real plant applications the predictions are very good, see Fig. 3(b).

R5 Ductility Exhaustion Approach

The damage interaction diagrams for the R5 ductility exhaustion approach are shown in Figs. 4(a) and 5(a). For the austenitic steels, it is clear from Fig. 4(a)

that the scatter for the peak dwells is less than that exhibited by the time fraction approach. This supports the use of the R5 ductility exhaustion approach in preference to the time fraction approach for austenitic steels. However, for the tests with intermediate dwells the predictions are highly conservative. The graph of $N_{\text{obs}}/N_{\text{pred}}$ versus the initial stress Fig. 4(b) shows that the reason why the predictions for the tests with intermediate dwells are conservative is because these tests have low start of dwell stresses. These observations suggest an effect of stress on creep damage, which is not currently taken into account by ductility exhaustion approach. This is of particular relevance to real plant assessments which usually involve relatively low initial stresses and means that the R5 ductility exhaustion approach will tend to give overly conservative predictions under real plant conditions.

A similar observation regarding the effect of stress can be drawn from the data on ferritic steels shown in Figs. 5(a) and 5(b). Nevertheless, the predictions for ferritic steels are overly conservative for all tests.

The Stress Modified Ductility Exhaustion Approach

The ductility exhaustion approach of R5 can be simply modified to include the effect of stress on creep damage by treating the ductility as a function of both strain rate and stress. The resulting expression for the creep damage per cycle calculated from a ductility exhaustion approach, which includes the effect of stress, d_c^{SM} , is given by

$$d_c^{\text{SM}} = \int_0^{t_h} \frac{\dot{\epsilon}_c}{\epsilon_f(\dot{\epsilon}_c, \sigma, T)} dt \quad (11)$$

where t_h is the dwell time, $\epsilon_f(\dot{\epsilon}_c, \sigma, T)$ is the inelastic strain at failure at the appropriate temperature as a function of both the creep strain rate and the stress. It should be noted that another difference between the R5 ductility exhaustion approach and the SMDE approach is that it is intended to use the true inelastic strain at failure rather than engineering creep strain at failure. The true inelastic strain at failure can be defined by either (i) the true total inelastic strain at failure, calculated from $\ln(1 + \text{elongation at failure})$ or (ii) the true local inelastic strain at failure as calculated from the reduction of area using $\ln[1/(1 - R \text{ of } A, \text{ in absolute units})]$. Using Eq 11 to calculate creep damage there are two values that need to be known: the strain rate during the dwell and the inelastic strain at failure.

It has been shown experimentally [6–8,10–12] that the uniaxial creep ductility of austenitic and ferritic steels can be represented by

$$\epsilon_f = \min \left[\epsilon_U, A_1 \exp \left(\frac{P_1}{T} \right) \dot{\epsilon}_c n_1 \sigma^{-m_1} \right] \quad (12)$$

where A_1 , n_1 , m_1 , and P_1 are material constants. It should be noted that the upper shelf ductility, ϵ_U , is necessary for high ductility constant rate and creep

tests that fail by necking followed by ductile transgranular failure, either by shear at 45° or by “cup and cone.”

Consideration of the effects of multiaxial states of stress is given by using the von Mises strain and strain rate and the maximum principal stress in Eq 12 and by multiplying by a cavity growth factor (CGF)

$$\bar{\epsilon}_f(\dot{\epsilon}_c, \sigma_1, T) = \epsilon_f(\dot{\epsilon}_c, \sigma_1, T) \text{CGF} \quad (13)$$

It has been shown by an analysis of biaxial creep rupture data on type 304 [12] that the appropriate CGFs for use with Eq 12 are for the high ductilities, i.e.,

$$\epsilon_U \text{CGF}_{\text{RiceTracey}} = \exp\left(\frac{1}{2} - \frac{3\sigma_H}{2\bar{\sigma}}\right) \quad (14a)$$

and for the creep failures

$$\text{CGF}_{\text{combined}} = \frac{\bar{\sigma}}{\sigma_1} \exp\left(\frac{1}{2} - \frac{3\sigma_H}{2\bar{\sigma}}\right) \quad (14b)$$

Thus, combining Eqs 12, 14a and 14b gives

$$\bar{\epsilon}_f = \min\left[\epsilon_U \exp\left(\frac{1}{2} - \frac{3\sigma_H}{2\bar{\sigma}}\right), A_1 \exp\left(\frac{P_1}{T}\right) \dot{\epsilon}_c^{n_1} \sigma_1^{-m_1} \frac{\bar{\sigma}}{\sigma_1} \exp\left(\frac{1}{2} - \frac{3\sigma_H}{2\bar{\sigma}}\right)\right] \quad (15)$$

Indeed it has been shown [12] that Eq 15 gives good fit to the biaxial creep ductility data on type 304 at 593°C and uniaxial creep ductility data at 482 to 816°C

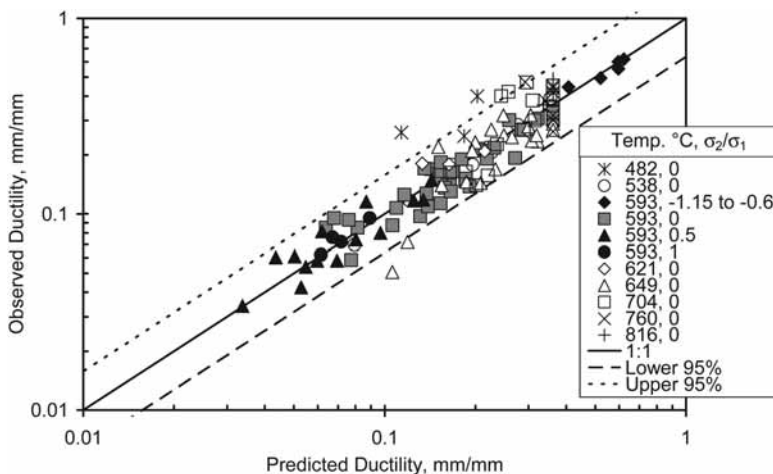


FIG. 6—The observed uniaxial (482–816°C) and biaxial (593°C) creep ductility of type 304 versus the predicted creep ductility [12].

(Fig. 6), and is also better than current R5 approach for multiaxial creep ductility, which uses Eqs 78.

Results of Stress Modified Creep Damage Calculations

In each of the research papers [6,7,9–11] the SMDE approach was compared with the time fraction and R5 ductility exhaustion approaches. For each material the same fatigue data and the stress relaxation data were used as before.

As well as introducing an effect of stress into the ductility exhaustion approach two further changes to the approach were found to be necessary [6,7,9]. The first of these improvements was with regard to the upper shelf ductility. In the work on type 316H [6] and cast type 304L [7] it was found that removing the upper shelf from the creep damage model gave an even better prediction of creep damage. The upper shelf ductility in constant rate and creep tests corresponds to necking followed by ductile shear failures, typically at 45° to the applied stress in austenitic steels and cup and cone in ferritic steels. This type of failure cannot physically occur during a strain controlled creep-fatigue cycle and it is inclusion in the creep damage model results in overly conservative estimates of creep damage. Thereby, although Eq 12 was used to analyse the creep ductility data (which can fail by shear) when the model was used to predict creep damage during the creep-fatigue tests this was simplified to

$$\bar{\varepsilon}_f = A_1 \exp\left(\frac{P_1}{T}\right) \dot{\varepsilon}_c^{n_1} \sigma_1^{-m_1} \frac{\bar{\sigma}}{\sigma_1} \exp\left(\frac{1}{2} - \frac{3\sigma_H}{2\bar{\sigma}}\right) \quad (16a)$$

which for uniaxial loading further simplifies to

$$\varepsilon_f = A_1 \exp\left(\frac{P_1}{T}\right) \dot{\varepsilon}_c^{n_1} \sigma^{-m_1} \quad (16b)$$

The effects of excluding the upper shelf was not considered in the paper on type 347 weld metal [9]. Nevertheless this has been considered in this paper and the creep damages given here for the SMDE approach are lower than those reported previously in Ref 9.

The second of these improvements was to the methods for analysing creep failure data. The analysis of creep ductility data as a function of the average creep strain rate assumes that the explanatory variables stress, strain rate and temperature are constant during the tests. However, in the case of a creep test the strain rate can vary by orders of magnitude and in constant strain rate tests the stress varies. In particular, the strain rate during initial loading can be many orders of magnitude greater than the average creep strain rate. In addition, there has always been an issue regarding the use of the creep strain at failure as both the response variable and as an input to calculating the explanatory variable, the average creep strain rate. Clearly, this would produce an unreasonably high correlation between the two variables. These problems have been addressed in Refs 6, 7 and 9 by adopting a new method to optimise the material parameters as a damage model rather than as a creep ductility model. This

creep damage modelling approach uses the instantaneous true inelastic strain rate and the true stress as calculated from the available creep and constant strain rate test data in conjunction with an appropriate ductility model, Eq 12, to calculate creep damage using the ductility exhaustion model, Eq 11. The creep damage model is then solved, using non-linear regression, to minimise the sum of squares of the residual in the natural logarithm of the damage (e.g., $(\ln(d_c))$ [2]). The method optimises the coefficients in the creep damage model so that it gives an accurate prediction of creep damage at failure in constant load creep tests and can include constant strain rate tests as well. Further details of this approach are given in Refs 6, 7 and 9.

With regard to the ferritic steels [10,11] the added complexity of the method to optimise the material parameters as a damage model rather than as a creep ductility model was not found to be necessary. Thereby, for the ferritic steels the creep ductility model, Eq 12, was fitted using conventional approaches as described in Ref ¹². Nevertheless, in the same way as for the austenitic steels, while it was necessary to include an upper shelf when fitting the creep data this was then excluded from the damage model when used to predict failure in the creep-fatigue tests (i.e., Eq 16b). Furthermore, an important change to the definition of creep ductility has been made for the ferritic steels. The SMDE approach uses the true inelastic strain at failure and for the ferritic steels this was defined by the true local inelastic strain at failure as calculated from the R of A using $\ln \{1/(1-R \text{ of A, in absolute units})\}$. For high stress (high strain rate) creep tests on ferritic steels the true local inelastic strain at failure is much larger than the engineering creep strain at failure, which is used in R5, and thereby this leads to less pessimistic values for the calculated creep damage.

The damage interaction diagrams that have been calculated using the SMDE approach are shown in Figs. 7(a) and 8(a), from which it can be seen that the SMDE approach gives good predictions of the damages at failure for all tests on both austenitic and ferritic steels. Comparison with Figs. 2(a), 3(a), 4(a) and 5(a) shows that the predictions of the SMDE approach are better than both the current R5 ductility exhaustion and the time fraction approaches. An examination of Figs. 7(b) and 8(b) shows that as intended the SMDE approach has removed the effect of the initial stress on the calculated creep damage and that this and the use of the true local inelastic strain at failure for the ferritic steels are the reasons for the improved predictions of creep damage.

The Effects of Compressive Dwells

For states of stress where one of the principal stresses is tensile, Eq 16a will predict finite values for creep ductility. However, when all of the principal stresses are zero or negative then Eq 16a predicts infinite ductility, which results in the prediction of zero creep damage. Thus, the SMDE approach predicts that compressive dwells cause no creep damage. This is different to the R5 ductility exhaustion approach and is more consistent with the physical understanding of creep cavitation mechanisms, which suggest that creep cavities can neither nucleate nor grow under compressive stresses. Nevertheless, the experimental

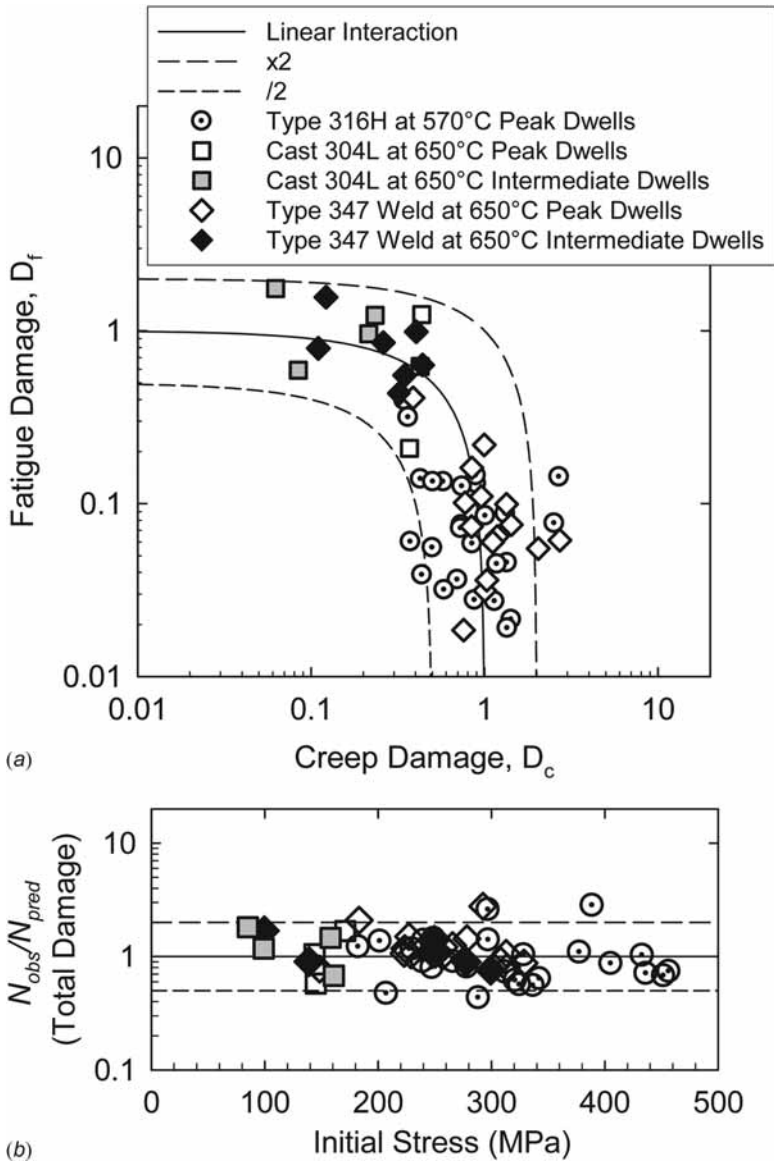


FIG. 7—Best estimate damages for austenitic stainless steels using the stress modified ductility exhaustion approach; (a) creep-fatigue interaction diagram, (b) N_{obs}/N_{pred} (or total damage) versus initial stress for a dwell at half life.

evidence [1,7] shows that when only compressive dwells are present the effect on the endurance can be significant. Thus is it necessary to increase the fatigue damage to take account of the observation that compressive dwells lead to reduced endurance.

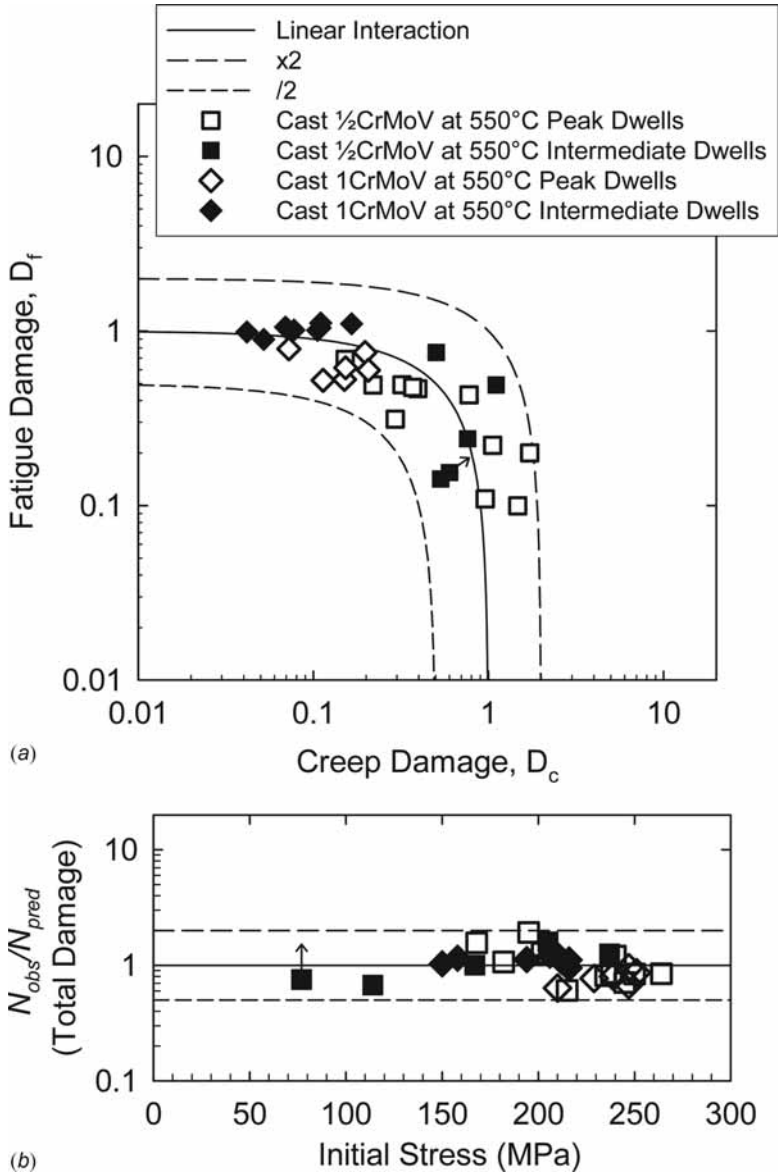


FIG. 8—Best estimate damages for low alloy ferritic steels using the stress modified exhaustion approach; (a) creep-fatigue interaction diagram, (b) N_{obs}/N_{pred} (or total damage) versus initial stress for a dwell at half life.

It is proposed that the fatigue damage should be increased by removing the nucleation phase of fatigue endurance. The continuous cycling fatigue endurance, N_0 in Eq 1, is replaced by N_g , the number of cycles to grow a fatigue crack from a_i , the nucleation size of 0.02 mm to a_0 . For the analysis of the tests N_g is

calculated from $N_5\% - N_i$, where N_i the number of cycles to nucleate a crack of size a_i is calculated from the empirical equation

$$\ln(N_i) = \ln(N_f) - 8.06N_f^{-0.28} \quad (17)$$

where N_f is the number of cycles to failure, as defined by complete separation. Equation 17 is a simplification of a model due to Pineau [17] which was based on data for type 304 and 316 austenitic steels; nevertheless, it has also been successfully applied to predict nucleation in ferritic steels as reported in the validation appendix of R5 [1]. The fatigue damage for a cycle with a compressive dwell is then calculated from

$$d_f = \frac{1}{N_g} \quad (18)$$

Thus, compressive dwells are assumed to remove the nucleation phase of fatigue failure and a crack of size a_i (0.02 mm) is conceded immediately. As in the current R5 procedures, the creep relaxation strain should be included when calculating the total strain range.

With regard to validating this approach for the effect of compressive dwell the test data described earlier include only a relatively few tests (on the cast type 304L [7]) and thereby a collation of published test data were compared to the predictions of Eq 18. These published data had previously been used for the validation of R5 [1]. The predictions of cyclic endurance are plotted in Fig. 9 which compares the observed number of cycles to failure with that predicted using the above methods. It can be seen from Fig. 9 that the proposed method gives a better prediction of the number of cycles to failure than the current method that is used in R5, which tends to be overly conservative.

Discussion

Standard methods to calculate creep damage have been applied to 85 uniaxial creep-fatigue data on three austenitic steels and two ferritic steels. It has been shown that both the time fraction and the R5 ductility exhaustion approaches can result in unrealistic estimates of creep damage (Figs. 2–5). Whereas, a new stress modified ductility exhaustion approach gives realistic predictions of damage at failure for all tests (Figs. 7 and 8). Indeed, it can be seen from Figs. 7(b) and 8(b) that the SMDE approach removes the dependence of the total damage at failure on the initial stress. This means that the SMDE approach gives markedly better predictions of creep damage for a wider range of start of dwell stresses than both the time fraction and the current R5 approaches. In particular, the predictions are better for low stress creep dwells, which are found in the intermediate dwells and low strain range cycles that are typical of plant cycles.

However, Figs. 2–5 and 7,8 used the mean fatigue endurance and mean creep failure data for the same heat of steel or for the same weld metal to

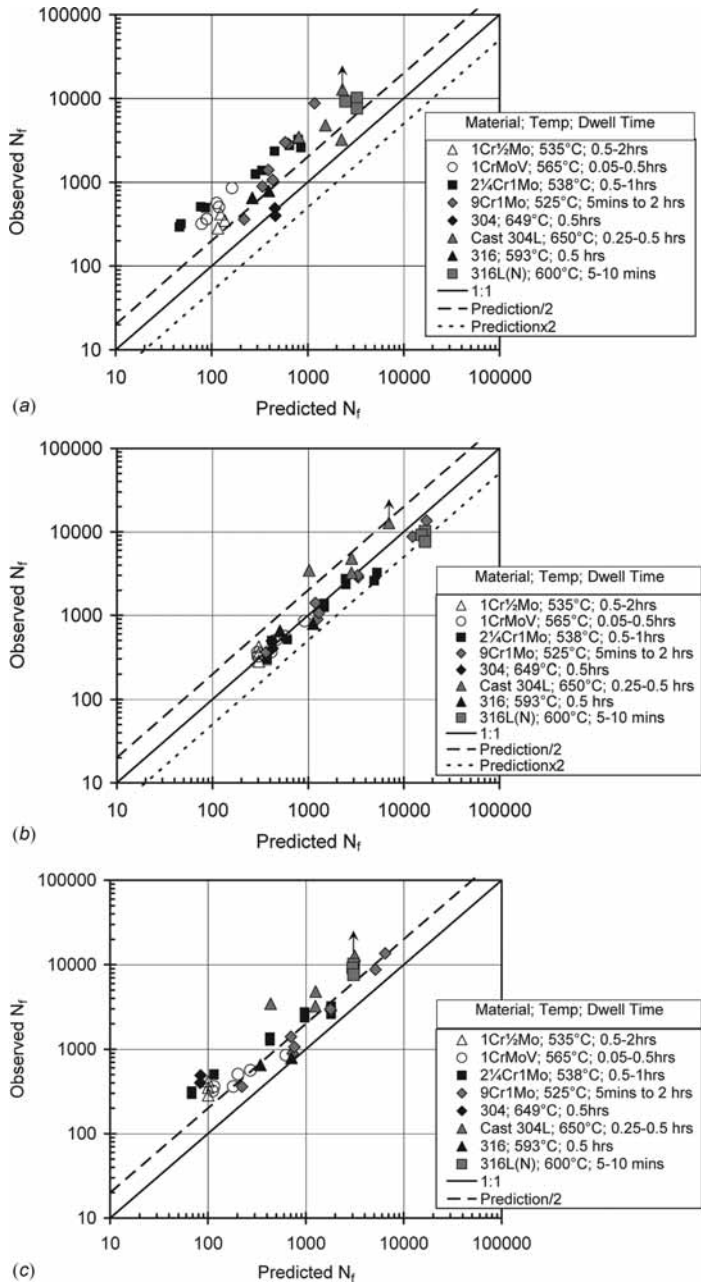


FIG. 9—The observed versus predicted number of cycles to failure for fatigue tests with compressive dwell; (a) mean endurance predicted using the current R5 approach, (b) mean endurance predicted using the proposed approach, (c) lower bound endurance predicted using the proposed approach.

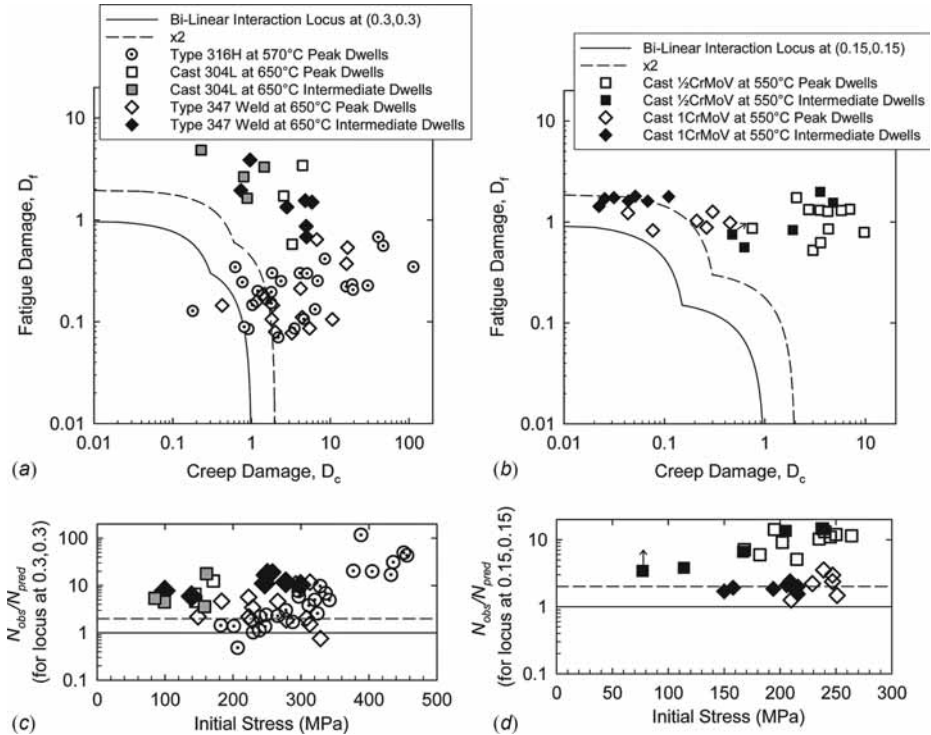


FIG. 10—Upper bound damages for using the time fraction approach; (a) austenitic steels creep-fatigue interaction diagram, (b) N_{obs}/N_{pred} versus initial stress for a dwell at half life; (c) ferritic steels creep-fatigue interaction diagram, (d) N_{obs}/N_{pred} versus initial stress for a dwell at half life.

calculate damage. Thus, best estimates of creep and fatigue damages were calculated.

For plant assessments lower bound fatigue endurance and creep failure data, which have been derived from multi-heat datasets, would be used to give conservative upper bound estimates of the damages. Upper bound estimates of; fatigue damage and creep damage have been calculated for the R5 ductility exhaustion, time fraction and SMDE approaches and are presented in Figs. 10–12. As for the best estimate calculations of damage the SMDE approach (Fig. 12) gives much more realistic predictions of damage at failure for all tests than either the time fraction (Fig. 10) or R5 ductility exhaustion (Fig. 11). Indeed, the values of N_{obs}/N_{pred} for the SMDE approach range from 0.96 to 7.9 (two tests being non-conservatively predicted), whereas for the R5 ductility exhaustion approach N_{obs}/N_{pred} ranges from 0.74 to 27 (four tests being non-conservatively predicted) and for the time fraction approach N_{obs}/N_{pred} ranges from 0.48 to 115 (two tests being non-conservatively predicted). What this means for engineering assessments of real plant is that the time fraction approach would on average be relatively conservative, this is shown by the

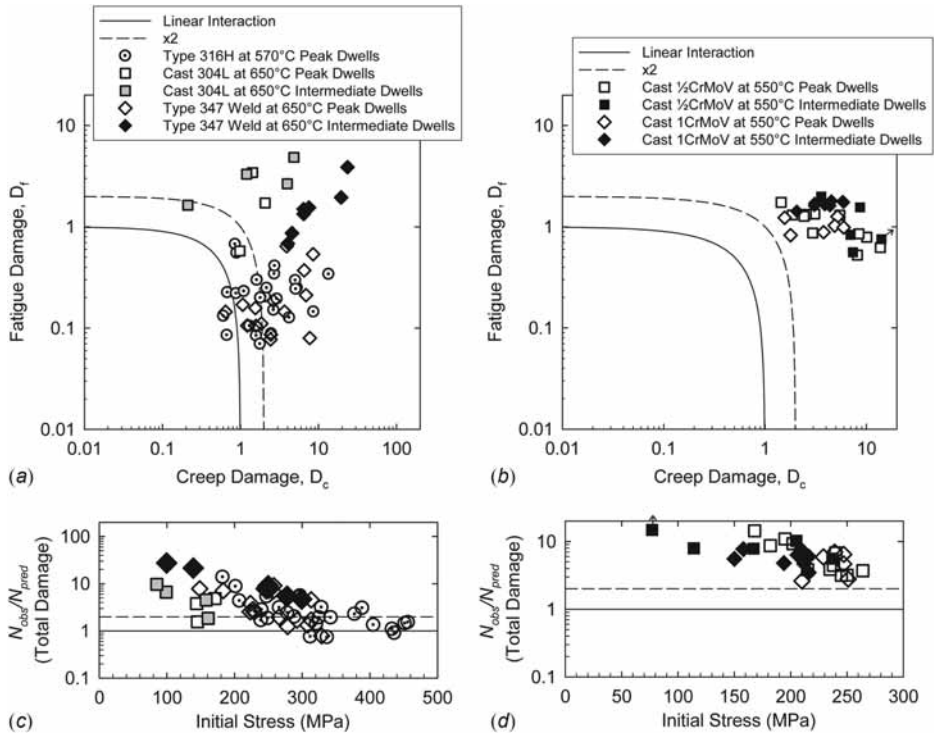


FIG. 11—Upper bound damages for using the R5 ductility exhaustion approach; (a) austenitic steels creep-fatigue interaction diagram, (b) N_{obs}/N_{pred} versus initial stress for a dwell at half life; (c) ferritic steels creep-fatigue interaction diagram, (d) N_{obs}/N_{pred} versus initial stress for a dwell at half life.

average of N_{obs}/N_{pred} (assuming a log normal distribution), which is 5 when lower bound failure data are used (i.e., upper bound estimates of N_{obs}/N_{pred}). However, for the time fraction approach the scatter is quite large, indeed the scatter at a $\pm 95\%$ level is given by factors of 7.9 and 1/7.9, which being greater than the average of 5 and means that the time fraction approach can be both non-conservative and overly conservative in different assessments. Unfortunately, there is no clear predictor for which assessments would give non-conservative estimates of failure. Nevertheless, Fig. 10(b) does indicate that assessments with high initial stresses tend to give overly conservative estimates of failure. With regard to the R5 ductility exhaustion approach, the average of N_{obs}/N_{pred} (upper bound estimates) is 3.9, and the scatter at a $\pm 95\%$ level is given by factors of 4.7 and 1/4.7, which are better than the values for the time fraction approach. This supports the use of the R5 ductility exhaustion approach in preference to the time fraction approach. However, Fig. 11(b) clearly shows that effect of initial stress on the calculated values of N_{obs}/N_{pred} means that the R5 approach becomes increasingly over conservative for assessments with low start of dwell stresses which are common in the intermediate

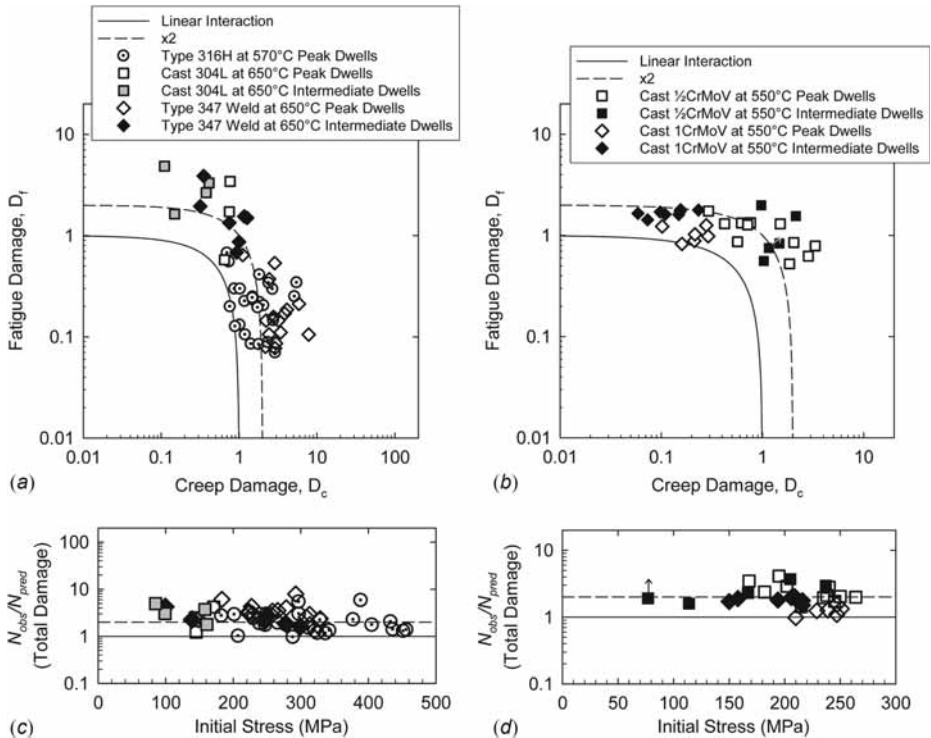


FIG. 12—Upper bound damages for using the stress modified ductility exhaustion approach; (a) austenitic steels creep-fatigue interaction diagram, (b) N_{obs}/N_{pred} versus initial stress for a dwell at half life; (c) ferritic steels creep-fatigue interaction diagram, (d) N_{obs}/N_{pred} versus initial stress for a dwell at half life.

dwells and low strain range cycles that are typical of plant cycles. Indeed, at start of dwell stresses of around 75–125 MPa the values of N_{obs}/N_{pred} range from 6.6 to 27 and are clearly overly conservative. Nevertheless, for the SMDE approach the average of N_{obs}/N_{pred} (upper bound estimates) is 2.22, and the scatter at a $\pm 95\%$ level is given by factors of 2.44 and $1/2.44$, which are significantly better than the values for the time fraction and R5 approaches. As can be seen from Fig. 12(b) there is no effect of the initial stress on the calculated values of N_{obs}/N_{pred} and the scatter is much smaller than that for the time fraction and R5 approaches. Indeed, at low start of dwell stresses, which typical of real plant cycles, of around 75–125 MPa the upper bound estimates of N_{obs}/N_{pred} range from 1.6 to 5.0. These values are realistically conservative without undue pessimism. It is thereby anticipated that the SMDE approach will be particularly beneficial for real plant cycles, which typically have low start of dwell stresses. Thus, it is proposed that the SMDE approach should be included into the R5 assessment procedure to provide an alternative to the current R5 ductility exhaustion approach.

Similarly, it has been shown that the revised approach for compressive dwell gives a better prediction of the number of cycles to failure than the current method that is used in R5, which tends to be overly conservative, see Fig. 9(a). The over conservatism of the R5 approach is because it calculates a creep damage for a compressive dwell, which is not physically correct since creep cavities can neither nucleate nor grow under compressive stresses. However, it should be remembered that the revised approach does not explicitly take into account the physical process that occur during a compressive dwell, such as, (i) an increase in the mean stress, (ii) increased oxidation, (iii) increased hysteresis energy, and (iv) thermal softening. Nevertheless, it is anticipated that these processes and in particular oxidation would have a greater effect on the fatigue crack nucleation phase, which constitute the greatest proportion of the fatigue life at large endurances (as predicted by Eq 17), where compressive dwells have the greatest effect. Modifying the number of cycles to grow a crack, N_g , would have had the contrary effect and have a greater affect at low endurances, which conflicts with the observations on compressive dwells. Thus, it is proposed that the new approach for compressive dwells, which removes the nucleation phase of fatigue endurance, should replace the approach that is currently used in R5.

Conclusions

- (1) The time fraction approach, which is used to calculate creep damage in design procedures can predict both unrealistically high values and non-conservatively low values of creep damage and is therefore not used in the R5 assessment procedure for cases involving significant cyclic loading.
- (2) The stress modified ductility exhaustion approach give less pessimistic predictions of creep damage than the current R5 ductility exhaustion approach, for cycles with low stress creep dwells. This includes cycles where the creep dwell is positioned at a stress below the peak in the cycle and cycles with low strain ranges which are typical of the types of service cycles experienced by plant. Thus, it is proposed that the stress modified ductility exhaustion approach should be included into the R5 assessment procedure to provide an alternative to the current R5 ductility exhaustion approach.
- (3) A revised approach for the prediction of fatigue endurance for cycles containing a compressive dwell has been proposed. It has been shown that this approach gives a better prediction of the number of cycles to failure than the current method that is used in R5, which tends to be overly conservative. Thus, it is proposed that the new approach for cycles with compressive dwells should replace the approach that is currently used in R5.

Acknowledgments

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Meimei Li,¹ S. Majumdar,¹ and K. Natesan¹

Modeling Creep-Fatigue Behavior of Mod.9Cr-1Mo Steel

ABSTRACT: This paper presents three models that were developed explicitly for predicting the creep-fatigue behavior of Mod.9Cr-1Mo steel. The Cyclic Softening Model incorporated tensile strain hardening and creep deformation, and described the cyclic stress variation as a function of cycle number under hold-time creep-fatigue loading. The Stress Relaxation Model predicted the stress relaxation curve during the hold time of cyclic loading, based on the creep properties of the material. The Interactive Damage Rate Equation predicted the hold-time creep-fatigue life using a methodology, in which the key materials parameters were determined by the creep properties and continuous fatigue data. The model predictions captured the trends of creep-fatigue interaction in Mod.9Cr-1Mo steel, and were in good agreement with experimental data in the literature.

KEYWORDS: creep-fatigue, Mod.9Cr-1Mo, ferritic steel, cyclic softening, stress relaxation, damage rate equation

Introduction

Ferritic-martensitic steels are the lead structural materials for various types of advanced fission and fusion energy systems [1]. These alloys have excellent resistance to irradiation-induced swelling, good thermal properties, and low cost. Mod.9Cr-1Mo ferritic-martensitic steel is a prime candidate for high temperature structural applications in advanced sodium-cooled fast reactors [2]. Mod.9Cr-1Mo is a Fe-9%Cr-1%Mo steel modified with V and Nb. The high temperature strength of Mod.9Cr-1Mo relies on tempered martensite stabilized by $M_{23}C_6$ carbides and a fine distribution of vanadium/niobium carbon-nitride

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¹ Argonne National Laboratory, 9700 South Cass Ave., Argonne, IL 60439 (Corresponding author), e-mail: mli@anl.gov

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(MX) precipitates. The improved high temperature strength of Mod.9Cr-1Mo allows construction of thinner-wall components, resulting in lower thermal stresses during transients and lower material costs of advanced reactors.

The current reference design of an advance sodium-cooled fast reactor envisions an outlet temperature of 550°C [2]. As the operating temperature of sodium-cooled fast reactors are significantly higher than that in current light water reactors (LWRs), the design of elevated temperature components must comply with American Society of Mechanical Engineers (ASME) Boiler and Pressure Vessel Code Section III (Rules for Construction of Nuclear Facility Components Component), Subsection NH (Class I Components in Elevated Temperature Service). Time-dependent effects on mechanical properties such as creep, creep-fatigue, and creep ratcheting must be considered.

Creep-fatigue interaction is a complex dynamic process involving combined effects of creep and fatigue (and environmental effects) on the accumulated damage. Fatigue life can be reduced significantly, if time-dependent creep damage becomes significant. The process depends on a number of mechanical and metallurgical factors including test temperature, strain rate, hold time, environment, thermo-mechanical treatment, microstructure, etc. The mechanisms of creep-fatigue damage are not well-understood, and consequently, predictive capabilities are limited. The current ASME design rule of bilinear damage summation puts severe limits of fatigue and creep loads for Mod.9Cr-1Mo (Grade 91) steel [3]. Improved creep-fatigue design rules are needed for adequate prediction of life for high-strength ferritic/martensitic steels and for nuclear reactor component designs.

A large number of creep-fatigue data is available in the literature [4]. Creep-fatigue experiments reveal that Mod.9Cr-1Mo steel exhibits unique deformation characteristics that are significantly different from those of austenitic stainless steels and low alloy steels. Mod.9Cr-1Mo is subjected to considerable cyclic softening during strain-controlled cyclic loading [5–7]. The stress amplitude decreases as the number of cycle increases. This is in contrast with austenitic stainless steels, Types 304 and 316, which cyclically harden [8]. The cyclic behavior of Mod.9Cr-1Mo steel is also different from that of annealed 2.25Cr-1Mo, in that annealed 2.25Cr-1Mo experiences cyclic hardening, cyclic softening, or a mixture of cyclic hardening and softening, depending on temperatures, strain rate, and hold times [9]. Cyclic softening in Mod.9Cr-1Mo steel occurs at all temperatures between 25°C and 600°C in both continuous fatigue and creep-fatigue tests [5]. When hold times are introduced in fatigue cycles, cyclic softening is accelerated in comparison with pure fatigue. The initial high strength advantage of the alloy is lost even after a small number of loading cycles. Depending on the strain range and strain rate, this stress may drop to nearly one-half of its starting value [10]. Considerable efforts have been made to model the cyclic softening behavior in Mod.9Cr-1Mo steel [11–13].

To evaluate the creep damage under creep-fatigue loading conditions, it is necessary to have a good knowledge of stress variation during the hold time. When a hold time is applied at peak strains either in tension or compression or a combination of tension and compression, stress relaxes with time due to conversion of the elastic strain into the inelastic strain when the total strain is kept constant. Stress relaxation data during the hold time is critical in evaluating

materials inelastic behavior and deformation characteristics and estimating creep damage under creep-fatigue conditions. Various approaches have been adopted for evaluating stress relaxation behavior during hold times. In ASME NH, stress relaxation is determined based on the isochronous stress-strain curves. Both RCC-MR and DDS codes use creep strain curves and the strain hardening law to determine the stress relaxation response, and generally provide a better description of stress relaxation behavior. Overall, all three procedures tend to predict higher stresses than experimental observations [11].

A number of life prediction models have been developed to compute the creep-fatigue life of structural components, e.g., linear damage rule [15,16], frequency separation equation [17], strain-range partitioning equation [18], damage rate equation [19], simplified model test approach [20], continuum damage mechanics model [21], etc. Asayama [22] reviewed the applicability of several creep-fatigue damage evaluation methods to Mod.9Cr1-Mo steel, including modified ductility exhaustion model, strain range separation model, time fraction and ductility exhaustion hybrid model, etc., and concluded that these models predicted experimental data accurately but predictions among these models differ considerably in long-term regions. Several mechanistic models have also been developed based on the physical mechanisms of creep-fatigue damage, e.g., cavitation damage model [23], crack initiation and propagation model [24], coupled deformation damage model [25–27], etc. These mechanistic models are not amenable to design life prediction at the current stage, but they provide important guidelines as to how the existing database could be extrapolated to design loading conditions. There is an apparent need that an improved creep-fatigue predictive model is developed for high-strength ferritic/martensitic steels.

This paper presents the modeling results of three models explicitly established for Mod.9Cr-1Mo steel, i.e., the Cyclic Softening Model, the Stress Relaxation Model and the Interactive Damage Rate Model. Modeling results are compared with literature data of creep-fatigue tests for Mod.9Cr-1Mo steel. The limitations of the models are discussed.

Cyclic Softening Model for Mod.9Cr-1Mo Steel

Previous work has shown that cyclic softening in Mod.9Cr-1Mo steel is attributed to elimination of initial fine lath and sub-grain boundaries and coarsening of sub-grain microstructure, reduction of dislocation densities and coarsening of precipitates [11,12,28]. Based on the observation of the microstructural evolution under cyclic loading, several micromechanical models have been proposed to link the microstructural recovery with the macroscopic cyclic softening in Mod.9Cr-1Mo steel [11–13]. These models were able to reproduce the cyclic softening effect, and yet the accurate, quantitative predictions of strain and hold time effects have not been achieved. Rather than considering the recovery process only, the macroscopic cyclic softening behavior of Mod.9Cr-1Mo steel was modeled in this study by including both tensile strain hardening and creep recovery terms. The competition between these two processes leads to an overall effect of cyclic softening. The instantaneous stress level, $\sigma(N)$, at the number of cycle, N , is given by Eq. 1

$$\sigma(N) = \sigma_0(T, \Delta\epsilon_t) \cdot (\epsilon_p)^k \cdot \left(\frac{1}{N}\right)^m \quad (1)$$

where:

σ_0 = stress constant for a given temperature, T and a total strain range, $\Delta\epsilon_t$,
 ϵ_p = plastic strain in a cycle,
 k, m = material constants.

The value of the stress constant, σ_0 , was obtained from the tensile stress-strain curve of Mod.9Cr-1Mo steel for a given temperature and total strain range. The power exponent, k was assumed the same as the power exponent of strain hardening of the tensile stress-strain curve. The power exponent, m , was determined by the power-law creep for Mod.9Cr-1Mo steel, i.e.

$$\begin{aligned} \dot{\epsilon} &= A\sigma^n \\ m &= 1/n \end{aligned} \quad (2)$$

where:

$\dot{\epsilon}$ = steady-state creep rate,
 σ = applied stress,
 n = power-law exponent.

The tensile and creep properties of Mod.9Cr-1Mo steel were used in Eqs. 1 and 2 to predict its cyclic softening behavior. Figure 1 shows the calculated

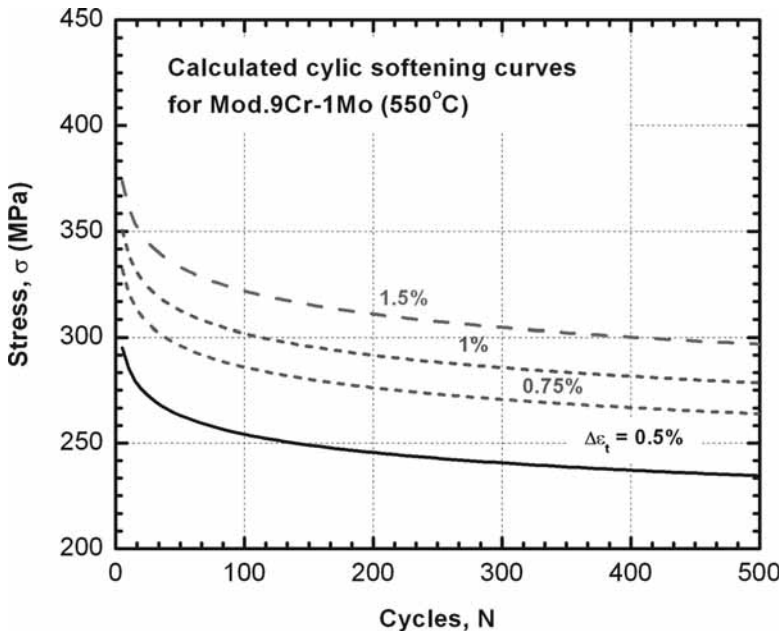


FIG. 1—Calculated cyclic softening curves for Mod.9Cr-1Mo steel for total strain ranges of 0.5–1.5% at 550°C.

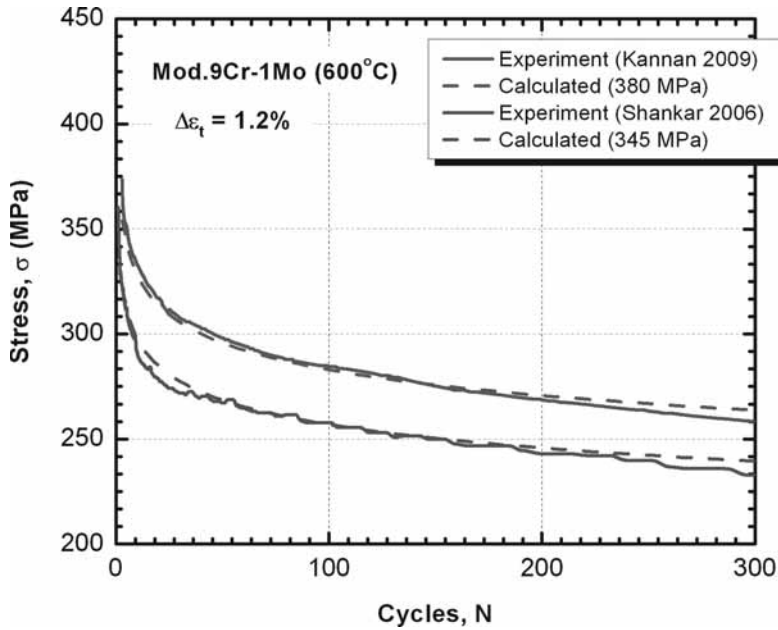


FIG. 2—Comparison of the calculated cyclic softening curves with the experimental data taken in the literature [6,7] for Mod.9Cr-1Mo steel tested at the total strain ranges of 1.2% at 600°C.

cyclic softening curves for different total strain ranges, 0.5%, 0.75%, 1.0%, and 1.5% at 550°C. The calculated values well captured the trends of cyclic softening behavior of Mod.9Cr-1Mo steel, in that the peak stress decreases with increasing number of cycles, and cyclic softening is stronger at a higher total strain range.

The calculated cyclic softening curves were compared with the experimental data reported in the literature [6,7]. Note that the data taken from Ref. [6] is from the test carried out in air. The comparisons are shown in Fig. 2 for Mod.9Cr-1Mo steel, tested at the total strain range of 1.2% at 600°C. The difference in the cyclic softening curves obtained under the same testing condition in two different studies is likely due to heat variations as well as the differences in material strength resulting from heat treatments of Mod.9Cr-1Mo steel. A tensile strength of 380 MPa was used to calculate the cyclic softening curve for the Mod.9Cr-1Mo steel studied by Kannan et al. [6], and a tensile strength of 345 MPa was used for the calculation of cyclic softening curve for the Mod.9Cr-1Mo steel investigated by Shankar et al. [7]. A good agreement was achieved in both cases between the experiments and predictions.

Stress Relaxation Model for Mod.9Cr-1Mo Steel

The stress relaxation behavior during hold time of creep-fatigue loading for Mod.9Cr-1Mo steel was modeled using the modified Feltham stress relaxation equation [29]. The relaxed stress $\sigma(t)$ at time, t , during hold is expressed as

$$\sigma(t) = \sigma(0) - \frac{E\dot{\epsilon}_0}{\beta} \ln(\beta t + 1) \quad (3)$$

$$\dot{\epsilon} = A\sigma(0)^{n'}$$

where:

$\sigma(0)$ = stress at time $t = 0$,

E = Young's modulus,

β and A = constants,

n' = power exponent.

The constants, A and n' were determined from the power-law creep.

The inelastic strain rate during hold, $\dot{\epsilon}_{in}(t)$ was calculated from the rate of change of the stress, given by

$$\dot{\epsilon}_{in}(t) = -\frac{1}{E} \frac{d\sigma(t)}{dt} \quad (4)$$

Figure 3 shows the calculated stress relaxation curve during hold time of the creep-fatigue loading for Mod.9Cr-1Mo steel for the total strain range of 1% and 550°C. The constant, A and the power exponent, n' in Eq. 3 were determined by the dislocation creep equation of Mod.9Cr-1Mo steel. The calculated curve is compared with the literature experimental data [30] and shown in Fig. 3. The

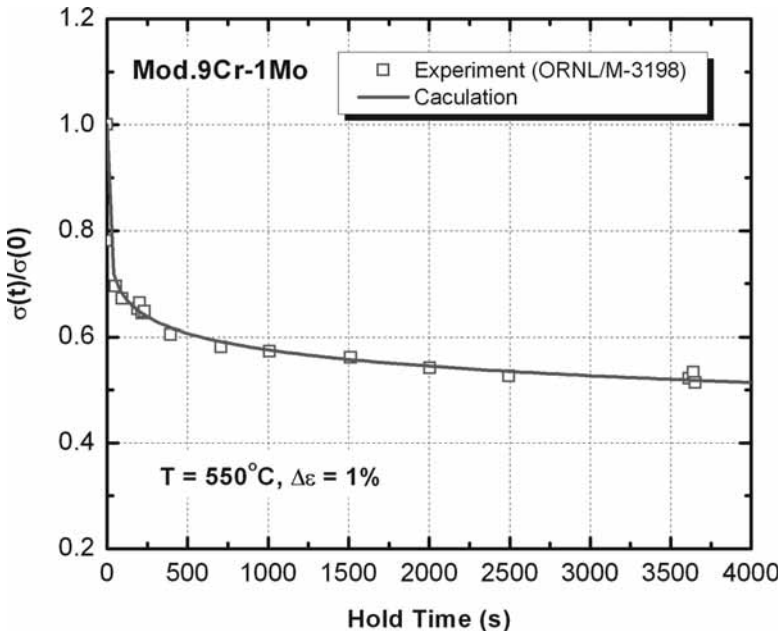


FIG. 3—The calculated stress relaxation curve during hold time of the creep-fatigue loading and its comparison with the experimental data taken in the literature [30] for Mod.9Cr-1Mo steel tested at the total strain range of 1% at 550°C.

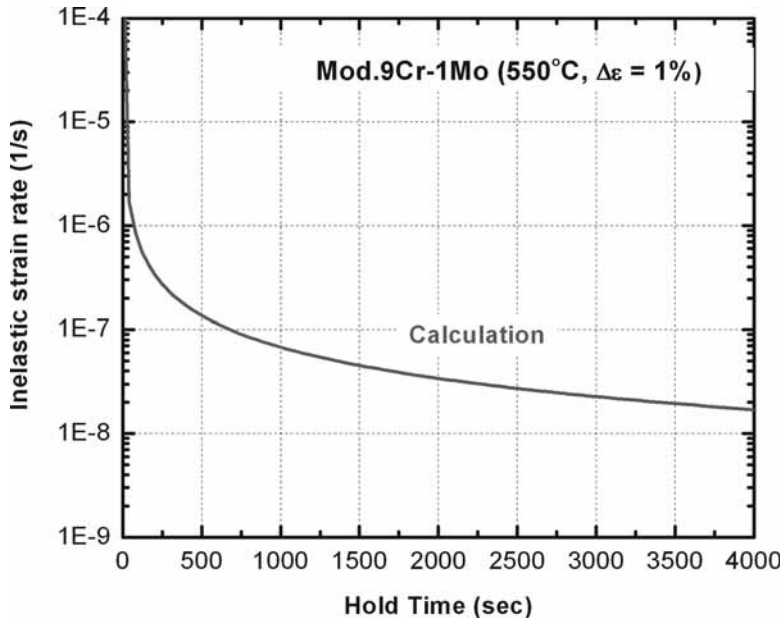


FIG. 4—The calculated inelastic strain rate during the hold time of creep-fatigue loading for Mod.9Cr-1Mo steel, tested at the total strain range of 1% at 550°C.

calculation correctly predicts the stress-relaxation behavior of Mod.9Cr-1Mo steel and is in good agreement with the experimental observation. The inelastic strain rate during the hold time calculated using Eq. 4 is presented in Fig. 4. A two-stage behavior is evident in Figs. 3 and 4. At the first stage the stress is relaxed drastically, while in the second stage stress relaxed gradually. The strain rate is significantly higher during the first stage than during the second stage.

Interactive Damage Rate Model for Mod.9Cr-1Mo Steel

The Damage Rate Model was proposed by Majumdar and Maiya in 1976 [19]. It has been successfully applied in several materials, e.g., Type 04 austenitic stainless steel and 2.25Cr-1Mo ferritic steel, to predict the creep-fatigue data under various loading conditions [31–33]. The applicability of this model to Mod.9Cr-1Mo steel has not been evaluated and is explored in this study.

A single rate equation was initially proposed to account for fatigue crack growth in the Damage Rate Model. The model was later modified to take into account the grain boundary cavitation under hold time during cycling. The model was further modified to take into account the interactive process of fatigue-induced crack growth and creep-induced grain boundary cavitation, so-called “Interactive Damage Rate Equation.” The Interactive Damage Rate Equation is given as

$$\frac{1}{a} \frac{da}{dt} = \left\{ \frac{T}{C} \right\} (1 + \alpha \ln c/c_o) |\dot{\epsilon}_{in}|^m |\dot{\epsilon}_i n|^k \quad (5)$$

where:

a = current crack length,

$\varepsilon_{in}, \dot{\varepsilon}_{in}$ = current total inelastic strain as measured from last reversal and inelastic strain rate, respectively,

c_o and c = initial and current cavity sizes, respectively,

T, C, α, m, k = material parameters.

"Failure" is assumed to result either by the growth of a crack or by the coalescence of cavities, and failure occurs when the crack length reaches a value of a_f or cavity size reaches a value of c_f , whichever occurs earlier.

The fatigue life can be determined by the integration of Eq. 5. For symmetric continuous cycling at zero mean stress and a given strain rate, the fatigue life, N_f is given as

$$N_f = \frac{m+1}{2A} (\Delta\varepsilon_{in})^{-(m+1)} \dot{\varepsilon}_{in} 1 - k \quad (6)$$

and

$$A = \frac{T+C}{2} \left/ \ln \frac{a_f}{a_o} \right. \quad (7)$$

where:

$\Delta\varepsilon_{in}$ = total inelastic strain range,

a_o = initial length of a crack,

a_f = final crack size.

For the case of slow-fast loading (slow tensile and fast compressive) with an average slow tensile inelastic strain rate $\dot{\varepsilon}_s$, and an average fast compressive inelastic strain rate $\dot{\varepsilon}_f$, the fatigue life is given as

$$D_T = \frac{1}{m+1} (\Delta\varepsilon_{in})^{m+1} [C_T \dot{\varepsilon}_s k - 1 + C_C \dot{\varepsilon}_f^{k-1}] \quad (8)$$

$$D_C = \frac{A_g}{m+1} (\Delta\varepsilon_{in})^{m+1} [\dot{\varepsilon}_s k_c - 1 - \dot{\varepsilon}_f, k_c - 1] \quad (9)$$

and

$$N_f = \left[-1 + \left\{ 1 + \frac{2D_C}{D_T} \right\}^{1/2} \right] / D_C \quad (10)$$

For the case of fast-slow loading with an average fast tensile inelastic strain rate $\dot{\varepsilon}_f$ and an average slow compressive inelastic strain rate $\dot{\varepsilon}_s$, $D_C = 0$

$$D_T = \frac{1}{m+1} (\Delta\varepsilon_{in})^{m+1} [C_T \dot{\varepsilon}_f, k - 1 + C_C \dot{\varepsilon}_s, k - 1] \quad (11)$$

For hold-time cycles with the inelastic strain range, $\Delta\varepsilon_{in}$, the hold time t_T at the peak tensile limit, the hold time t_C at the peak compressive limit, the following

equations may be written, if the incremental plastic strain during hold is considered negligible in comparison with the plastic strain range during the ramp

$$D_T = \frac{C_T + C_C}{m+1} (\Delta \varepsilon_{in})^{m+1} \dot{\varepsilon}_{in}^{k-1} + C_T (\Delta \varepsilon_{in})^m \int_0^{t_T} |\dot{\varepsilon}_{in}(t)|^k dt + C_C (\Delta \varepsilon_{in})^m \int_0^{t_C} |\dot{\varepsilon}_{in}(t)|^k dt \quad (12)$$

and

$$D_C = A_g (\Delta \varepsilon_{in})^m \left[\int_0^{t_T} |\dot{\varepsilon}_{in}(t)|^{k_c} dt - \int_0^{t_C} |\dot{\varepsilon}_{in}(t)|^{k_c} dt \right] \quad (13)$$

where $\dot{\varepsilon}_{in}(t)$ represents the inelastic strain rate during hold.

Accurate predictions of the fatigue life by the Interactive Damage Rate Model require optimization of a number of materials parameters. To establish the Interactive Damage Rate Model for Mod.9Cr-1Mo steel, we propose a methodology that can be used to predict the life under hold-time creep-fatigue loading conditions using known material properties. The concept of the methodology is illustrated in Fig. 5. The concept is based on numerous observations that the creep-fatigue damage under hold-time cyclic loading are similar to the damage caused by the effects of waveshapes and strain rates [34]. Continuous fatigue with various waveshapes and strain rates may be used as an accelerated approach to simulate the damage under hold-time creep-fatigue loading. Key material parameters in the model can be determined by creep properties and fatigue data obtained under symmetric continuous cycling (fast-fast and slow-slow) and asymmetric cycling (slow-fast and fast-slow). The established model, based on creep and fatigue properties, can then be used to predict the life under creep-fatigue loading

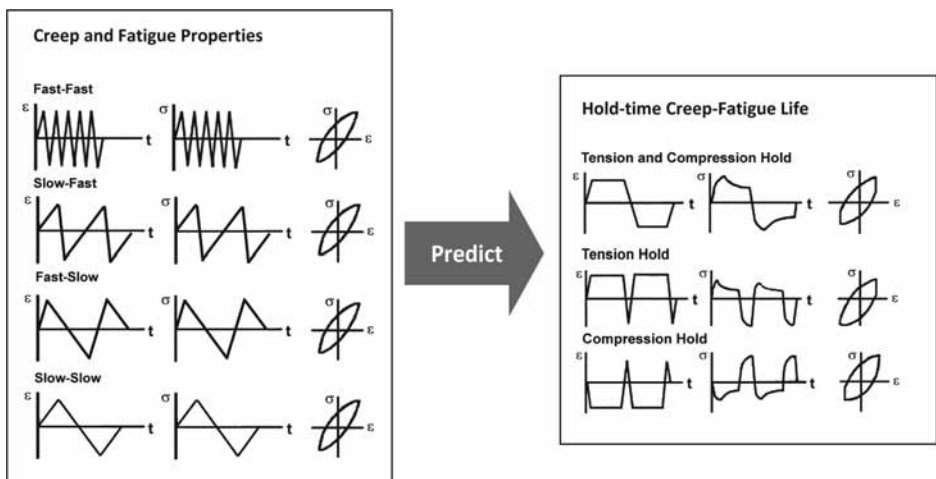


FIG. 5—Methodology of using creep and fatigue properties to predict the life under hold-time creep-fatigue loading.

TABLE 1—Material parameters in the Damage Rate Model for Mod.9Cr-1Mo.

	k	m	A	k_c	C_T	C_C	A_g
550°C	1	0.475	9.02×10^{-4}	0.927	0.0012	0.0006	0.004
600°C	0.971	0.428	5.86×10^{-4}	0.935	0.00078	0.00039	0.012

with hold-time in either tension or compression or combined tension and compression, and validated by the hold-time creep-fatigue data.

A set of creep and fatigue data under different temperatures, strain rates, and waveforms from the literature [35,36] was used to determine the material parameters in the Interactive Damage Rate Model, and their values are given in Table 1. The material parameter, k , m , A were determined from a least-squares fit of the continuous fatigue data to the crack-growth equation. The material parameter, C_T , C_C were determined under the assumption that $T=2C$. The material parameter, k_c was estimated using the creep rupture data at 550°C and 600°C for Mod.9Cr-1Mo steel. The parameter, A_g was obtained from the best fit of the fatigue data under slow-fast cycling.

The life under hold-time creep-fatigue loading was predicted for Mod.9Cr-1Mo using the established Interactive Damage Rate Model. Modeling results are compared with the literature experimental data for the steel tested under constant strain amplitude at a strain rate of 0.005 s^{-1} with a tension hold time of 0.1 h or 1 h at 550°C and 600°C in air [14]. The comparison is shown in Fig. 6,

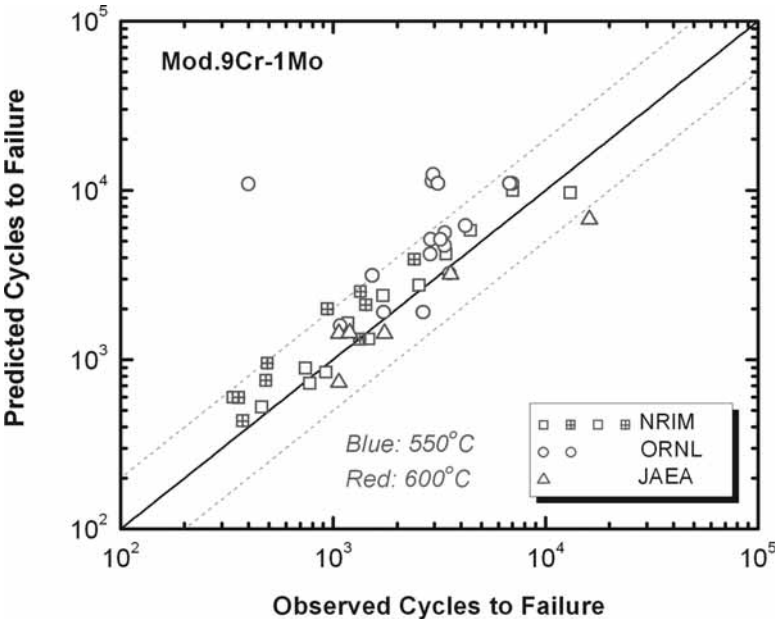


FIG. 6—Comparison of predicted fatigue lives with literature experimental data [14] for Mod.9Cr-1Mo tested at 550 and 600°C with a tensile hold time of 0.1 or 1 h.

where the predicted lives are plotted as a function of the observed lives. The solid line represents the best fit, and the dash lines represent a deviation by a factor of 2 from the perfect agreement. The predicted lives are all within a factor of 2 of the observed lives with one exception for a 550°C test.

Discussion

The creep-fatigue behavior of Mod.9Cr-1Mo steel was predicted by using Cyclic Softening, Stress Relaxation, and Interactive Damage Rate models. Key material parameters in the models were determined using known tensile, creep, and fatigue properties. The Cyclic Softening Model takes into account tensile strain hardening and creep recovery (dislocation creep), and predicts well the cyclic softening behavior of Mod.9Cr-1Mo steel. The temperature and strain range dependence of the cyclic softening curves was well captured. The difference in the material strength due to heat-to-heat variations and/or different heat treatment conditions, and their effect on cyclic softening can be described by the tensile strength. The model is applicable before the onset of a macro-crack, where a significant load drop takes place. Simplification was made in the model by assuming a constant plastic strain throughout the cyclic process, which can be improved. Further model improvement can also be made by incorporating microstructural parameters in the strain hardening and recovery terms. The Cyclic Softening Model does not account for any hold-time effects. It is unclear how the location and length of hold time affects cyclic softening of Mod.9Cr-1Mo steel.

The stress relaxation behavior of Mod.9Cr-1Mo steel, during the hold time of cyclic loading, was described by the Stress Relaxation Model. The model was derived from the primary and secondary creep of the material. It captures the two-stage behavior of stress relaxation, in which the stress relaxes rapidly at the beginning and then decreases gradually over the relaxation time. The model predicts well the stress relaxation curves at the half-fatigue life for Mod.9Cr-1Mo steel. Refinements of the model are required to accurately describe the dependence of stress relaxation responses on test temperature, strain range, length of hold time, hold type, and cyclic deformation history.

A methodology was developed to predict the life under hold-time creep-fatigue loading by the Interactive Damage Rate Model using creep and continuous fatigue property data for Mod.9Cr-1Mo steel. The Interactive Damage Rate Model appears to predict the lives well for Mod.9Cr-1Mo. It is important to realize the assumptions and limitations of the Interactive Damage Rate Model. First, the nucleation times for a micro or submicro-size crack and cavity are assumed to be negligible. Further, the Interactive Damage Rate Model assumes that a certain number of cavities are nucleated under tensile stress early in life and grow, but it ignores the fact that under tensile stress, cavities are nucleated continuously so that their number and spacing may change with time. Also, the effects of a microcrack on the growth of cavities near its tip are ignored in the analysis. It was noted that creep damage is very small in comparison with fatigue damage under all the conditions in this study. Tests that can produce

significant creep damage, such as slow-fast cyclic tests, cyclic creep tests (i.e., tensile hold under constant load), sequential loading tests, i.e., monotonic creep with significant creep damage followed by fatigue cycling to failure, etc., are needed to further validate the model.

Conclusions

Three models, explicitly developed to describe the creep-fatigue behavior of Mod.9Cr-1Mo steel, are presented in the paper, including the Cyclic Softening Model, the Stress Relaxation Model, and the Interactive Damage Rate Model. The Cyclic Softening Model incorporated tensile strain hardening and creep recovery, and well captures the cyclic softening response of Mod.9Cr-1Mo steel under creep-fatigue loading. The Stress Relaxation Model correctly simulates the two-stage behavior of stress relaxation during the hold time of cyclic loading. The Interactive Damage Rate Model for life prediction for Mod.9Cr-1Mo steel was established using known creep properties and fatigue data obtained under symmetric and asymmetric continuous fatigue conditions. The hold-time creep-fatigue life predicted by the Interactive Damage Rate Equation showed good agreement with literature experimental data for Mod.9Cr-1Mo steel. Integration of these three models will be considered in future work.

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R. P. Skelton¹

Models for Small Crack Growth under Creep-Fatigue in Austenitic Steels

ABSTRACT: Low cycle fatigue (LCF) endurance data have a valuable part to play in the lifetime assessment of components and structures. These data comprise the initiation and growth stages, but the growth relations themselves and their practical use are not as familiar as those employed for deeper cracks. Early work modelled continuous-cycling fatigue crack growth by assuming a succession of miniature LCF specimens at the crack tip, the field then being extended by investigators examining behaviour at high temperatures. Models were developed allowing for the concomitant contribution of creep damage for comparison with continuous-cycling properties where striation spacings recorded cyclic crack progress. Alongside such modelling, empirical laws were deduced describing the progress of short cracks. Expressions may be derived linking LCF with linear-elastic fracture-mechanics (LEFM) crack growth, using the parameter ΔJ (equivalent stress-intensity parameter). However, the purpose of this review is to compare and contrast those models which employ an easily measured surface parameter (such as total or plastic strain range) as the governing variable. Crack growth normally adopts an exponential form so that the rate of growth per cycle accelerates as the crack deepens. The distinguishing feature is that the process zone at the crack tip is itself surrounded by cyclically yielding material, in contrast with LEFM. Energy methods may also be employed, where the process zone at the crack tip fails when the accumulated energy density reaches a critical value. An upper bound relation is provided, accounting for the deleterious effects of creep-fatigue-oxidation interaction, if empirical data are not to hand. A conservative assessment may thus be made of cyclic crack growth

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¹ Consultant, 4 Irwin Road, Guildford, Surrey GU2 7PP, UK, e-mail: peter@pskelton.wanadoo.co.uk

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rate at a specified depth. This review examines the capability of each model to allow for such creep-fatigue effects.

KEYWORDS: striation spacings, austenitic steels, cyclic plastic zone, process zone, damage factors, grain-boundary cavitation, inter/transgranular cracking, interactive crack growth

Introduction

This review concerns modelling of the growth of short cracks under low cycle fatigue (LCF) conditions. It follows on from previous reviews of the topic [1–7], incorporating work that was not there included and concentrates particularly on the interaction of creep and fatigue. Crack propagation properties and growth relations at elevated temperatures in the LCF range are not as familiar as those in the linear-elastic fracture-mechanics (LEFM) range, although results in fatigue and creep-fatigue ranging from low-alloy ferritic steels [8–11], advanced ferritic steels [12], austenitic steels [13–27] through to single crystals [28,29] have been steadily reported since the mid 1960s. The laws governing crack growth appear as “Method II” in Volume 4/5 of the R5 assessment procedure [30].

It is remarkable that there appear to be very little published data on the Ni-based superalloys and related materials. This is possibly a reflection that no crack growth whatever can be tolerated in associated components for power generation or aerospace practice. This review concentrates mainly on the austenitic steels and frequently makes use of reported striation spacings as the method for determining baseline crack growth rates.

Two points must be made clear at the outset. Firstly, interactive growth for deep cracks in the more familiar LEFM regime will not be covered. It has been argued [31] that the ratio of (crack growth) papers published in the LEFM/LCF regimes is about two orders of magnitude. Secondly, it is assumed that the damaging effect of creep during an imposed dwell period does not cause an increment of crack growth to occur during the dwell itself [32]. Any enhancement of cyclic fatigue crack growth is assumed to occur during the tension-going ramp after a previous reversal into compression.

It would appear that most published papers in the field of elevated-temperature fatigue either concern deep crack growth in the LEFM range, as noted above, or on total endurance in the LCF range where attendant crack growth rates are not reported. Indeed, the degree of crack advance must be deduced, somewhat insensitively, from the accompanying tensile load drop during LCF testing [33,34]. Much more accurate determinations may be had by careful examination of the post-fracture surface or by employing electrical potential-drop techniques [35]. It is of interest to note that as early as 1936 in a paper from MIT Cambridge [36], the standard method of fatigue of plain polished specimens (recording only the number of cycles to failure) was described as “how to discover a minimum of information with a maximum expenditure of time, labour, and expense”. This, then, may be regarded as the birth of the separate notions of crack initiation and propagation. Studies of LCF crack growth

appear to have flowed out of the main stream of traditional testing, forming a useful tributary enabling the prediction of early crack growth in the surface regions of components undergoing cyclic duty. Expressions may indeed be derived linking LCF with LEFM crack growth, using the parameter ΔJ , (equivalent stress-intensity parameter) [2,5,37,38]. However, the purpose of this review is to compare and contrast those models which employ an easily measured parameter (such as surface total or plastic strain range) as the governing variable. To prevent undue length, the case for notched testpieces will not be considered.

Early Work

Crack Tip Parameters

Some notable contributions on modelling, pre-dating experiments on creep-fatigue interaction are briefly summarised in chronological order as follows. McClintock [39,40] introduced two useful concepts, applying at the crack tip: firstly, a damage criterion summing to unity for incremental advance involving the shear yield strain γ_y and the fracture yield strain γ_f , the latter approximated by the monotonic ductility. Secondly, the paper is probably the first of the class “ ρ/N ” where ρ is the assumed size of a process zone at the crack tip and N is the cycles required to traverse that zone. These considerations led to an expression of the form

$$\frac{1}{\rho} \frac{da}{dN} = 8 \frac{\gamma_y}{\gamma_f} \left(\frac{r_p}{\rho} \ln \frac{r_p}{\rho} \right) \quad (1)$$

where $r_p > \rho$ is the plastic zone size ahead of the crack tip and a is crack depth.

By considering the change in shape of an elliptical crack between maximum tension and compression, then applying the Neuber relation [41] to relate crack tip and remote conditions, Grosskreutz [42] arrived at the following relation for crack growth

$$\frac{da}{dN} = r_o \left\{ \left[\frac{2\Delta\epsilon_p}{1+\beta} \left(4 \frac{a}{r_o} \right)^{1/(1+\beta)} + 1 \right]^{(1+\beta)/2} - 1 \right\} \quad (2)$$

where:

$\Delta\epsilon_p$ = the plastic strain range at the surface,

r_o = the crack tip radius at maximum compression, and

β = the cyclic strain hardening exponent (discussed later).

The dependence of cyclic growth rate with depth was found to be sensitive to the manner in which the root strain, r_o , was defined. The author then went on to integrate Eq (2) between suitable limits to find the number of cycles spent in “stage II” growth, finding good agreement with LCF endurance data available at the time.

The idea of successive re-initiation at the crack tip was also adopted by Weiss [43,44] where the Neuber relation [41] was used together with the assumption that crack tip strain and the strain at some distance ahead of the tip

were influenced by the term β . Again, crack growth was considered to occur over this distance when strain there exceeded the unidirectional fracture strain. Some of these early models have been summarised by Coffin [45]. The present author also derived an alternative expression involving an effective crack tip radius [46], although in common with many other “re-initiation” models which require a large number of arbitrary constants (reviewed in Ref [47]), such relations can be difficult to apply in engineering assessments [31]. However, using the ρ/N re-initiation model, a solution was derived later [47] in terms of the strain-energy density generated ahead of the fatigue crack, the process zone failing when the accumulated energy density reaches a critical value, W_c . Values for this parameter are generally easily obtained from bulk failure in LCF tests to some given damage criterion (e.g., specified percentage load drop). The analysis was extended to damage accrual in other process zones in an array ahead of the crack tip, i.e., before the arrival of the crack tip. The basic relation is

$$\frac{da}{dN} = \frac{1}{1 + \beta} \left(\frac{\Delta J}{I_\beta W_c} \right) f(\beta) \quad (3)$$

where the terms I_β and $f(\beta)$ are easily accessible tabulated constants, see for example Ref [48]. A very similar relation was deduced in terms of ΔK for the elastic case, where ΔK is the range of stress intensity. It may be noted that the value of ρ does not appear explicitly in Eq (3) but is incorporated in the term W_c characterizing the number of cycles to produce a crack of specified depth. Furthermore since the term W_c is known to decrease as strain rates decrease or tensile dwell periods increase [6], Eq (3) predicts an increase in crack growth rate under these conditions (in comparison with continuous cycling at typical reversal rates of 10^{-3} to 10^{-4} /s).

Remote Parameters

Equation (3) is an example of the case where a crack-tip parameter (ΔJ) is employed which is in turn derived from remote conditions. The first steps to define an LCF cyclic strain-intensity parameter ΔK_e defined as $\Delta \varepsilon \sqrt{\pi a}$ (where $\Delta \varepsilon$ is a representative strain range) were taken by Boettner et al. [49] who took $\Delta \varepsilon = \Delta \varepsilon_p$, the remote plastic strain range. Solomon [50] later amended this to be the remote total strain range, $\Delta \varepsilon_t$. A still further adjustment [51] showed that the strain range for which the crack was open gave the best agreement with experiment: this was in turn demonstrated to be identical to the parameter ΔJ [38]. This latter parameter is thus capable of unifying crack growth rates between the LCF and LEFM regimes [52]. It can be partitioned into fatigue and creep components [5,53], but the parameter does not appear to have been widely used for service assessments. The remainder of this review will thus concentrate on empirical relations and models which employ an easily measured surface strain parameter.

Empirical Relations

The propagation rate of short cracks can either be described in terms of plastic strain range or total strain range at the surface [17]. Total strain range is more

useful when conducting engineering assessments, and this parameter will generally be used throughout the review. The relation between them is given by [54]

$$\Delta \varepsilon_t = \Delta \varepsilon_e + \Delta \varepsilon_p = \frac{\Delta \sigma}{E} + \left(\frac{\Delta \sigma}{A} \right)^{1/\beta} \quad (4)$$

where:

A = the cyclic strength coefficient and

β = the cyclic hardening exponent.

In Eq (4), values of A and β are taken to define this full range behaviour and thus describe the shape of the hysteresis loop. Approximations using a locus description (involving the tips of separate hysteresis loops) have been discussed previously [55].

The growth of cracks is described by the relation

$$\frac{da}{dN} = Ba^Q \quad (5)$$

for growth in mm/cycle, where the value of B depends on strain range and Q is a constant for a given alloy, temperature, and sometimes environment. Values of Q have been summarised elsewhere [1,4,6]. Most commonly, materials adopt a value $Q=1$ (exponential growth). An example is provided in Fig 1(a), the data being taken from Ref [17]. The value of B is identified as the crack growth rate pertaining at unit crack depth, independent of the value of slope Q . Taking values of B at the intercepts in Fig 1(a), these may then be plotted against total strain range. The results will be incorporated later.

The data in Fig 1(a) show, at a given crack depth, that the crack growth rate increases with increasing (total) strain range. For any strain range and crack depth, the growth rate generally increases as the dwell time is extended or the strain rate reduced, see for example Refs. [17] and [32]. An example for a ferritic steel is shown in Fig 1(b) [11].

Crack growth cycles, N_g , from an initial defect size a_o to a final defect size a_f are obtained by integration giving

$$N_g = \frac{1}{B(Q-1)} (a_o^{1-Q} - a_f^{1-Q}) \quad (6a)$$

or when $Q=1$

$$N_g = \frac{1}{B} \ln \left(\frac{a_f}{a_o} \right) \quad (6b)$$

Such relations are useful in verifying cycles between “beach marks” on the fracture surface or in striation counting where the spacing between striations increases with increasing crack depth [20,21,56].

“Modelling” in this review concerns the behaviour of cracks under tensile (or compression) dwell conditions or at very low strain rates. Values of the term B against corresponding total strain range deduced from medium- to long-term

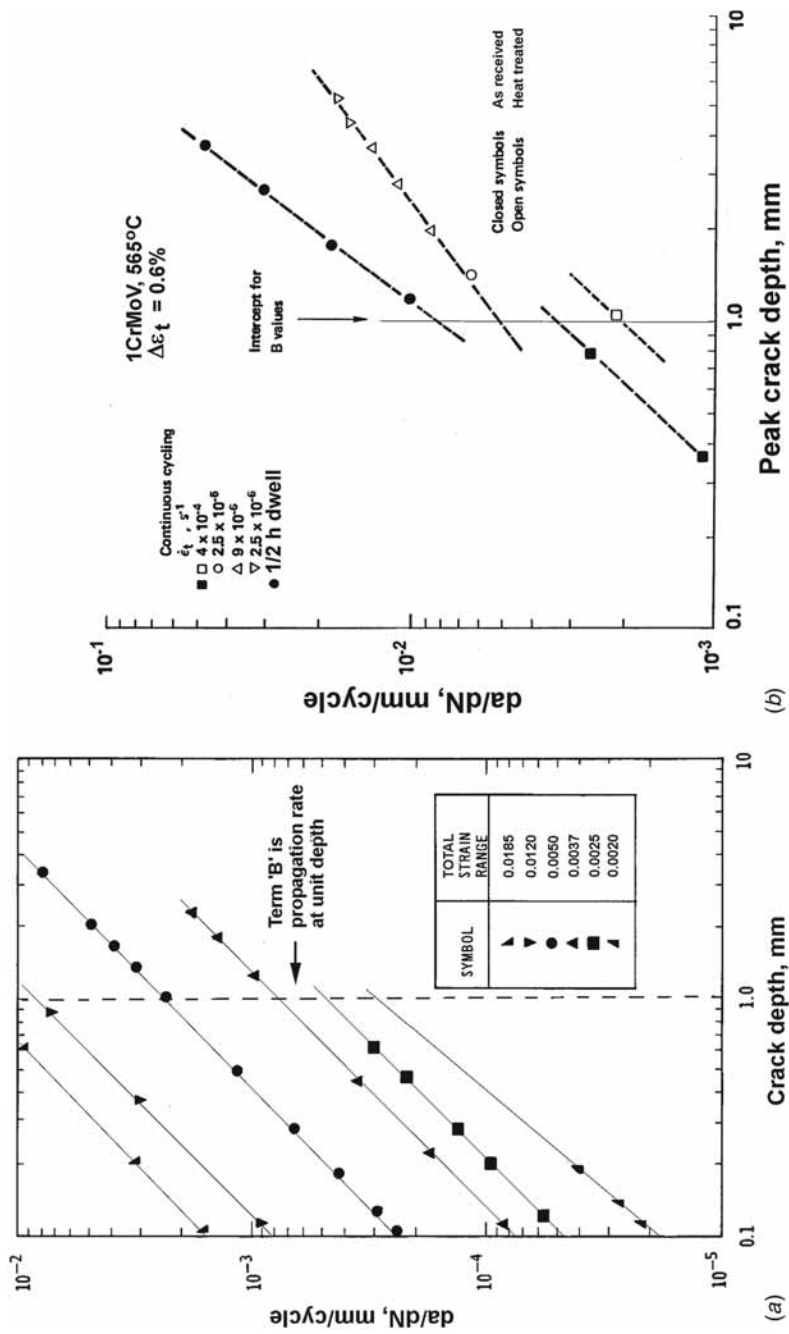


FIG. 1—(a) Experimental of crack growth rates for 20Cr25NiNb steel at 750°C. After Refs [1] and [17]. (b) Example of crack growth rates with dwell [11]. Reproduced with permission, Science Reviews 2000 Ltd.

tests in the literature are plotted in Fig 2 for 316 steel at 570 and 600°C, respectively [3,27], 321 steel at 700°C [22], and 1/2Cr1MoV, 1CrMoV steels, respectively, at 550°C [32]. The 24 h dwell test on 316 steel for example lasted 9000 h [27]. It is notable that the data lie almost exactly on the “upper bound” line [4] used in the R5 assessment procedure [30]

$$B = 2.61 \times 10^4 \Delta \epsilon_t^{2.85} \quad (7)$$

A quick check shows that the values of B deduced from Fig 1(a) for example all lie below the values predicted from Eq (7). From all reviewed data involving a dwell or high temperatures [1,2,4,6], Eq (7) has proved to be a robust upper bound for assessment. It will be used throughout this review as a reference line.

Continuous-Cycling Data

Elevated temperature crack growth data in the form of Fig 2 have been summarised previously [1,4,6,11] and need not be repeated here. Opportunity is taken however to present reference data where authors have determined the term $S = da/dN$ by careful striation counting, where the term S represents the striation spacing.

In some early work on 316 austenitic steel at 600°C, Dawson et al. [14] presented tabulated data and one example of a plot of $\log S$ (in centimeter) against linear a (in centimeter) of slope k , there being little difference from a “log-log” plot at short crack depths. It can be shown in this case that, expressed in millimeter, $B = S_0 10^{k'}$ where S_0 is the first detectable striation spacing and $k' = k/10$. Calculated values are plotted in Fig 3 where it is seen that the data lie below the upper bound of Eq (7).

Again at 600°C Nilsson [57] tabulated values of B deduced from striation spacings for several austenitic steels. Corresponding values of total strain range may be obtained from a companion paper [58]. The data lie considerably below Eq (7), see Fig 4.

Wareing and Vaughan [21] tested smooth cylindrical (of varying surface finish) and square specimens on two casts of 316 steel tested in LCF at 400°C. Minimum and maximum values of striation spacing of 0.2 and 5 μm were typically observed. Parameters were plotted against plastic strain range, but these may be converted to total strain range by means of Eq (4) taking $A = 2577 \text{ MPa}$ and $\beta = 0.241$ [21] and taking the variation of Young’s modulus for this steel as

$$E = U - VT \quad (8)$$

where $U = 2 \times 10^5 \text{ MPa}$ and $V = 75 \text{ MPa}/^\circ\text{C}$ for any temperature T . The results are plotted in Fig 5. For comparison, measurements of crack growth on the surface of square specimens by an optical telescope are included. These differences appear to overshadow cast-to-cast variations in growth rate but the data at 400°C generally lie well below the upper bound of Eq (7).

A similar comparison (at 1 cpm) for LCF tests on 316 steel was made at 625°C, both for smooth specimens (6.35 mm diameter) and notched (0.05 mm

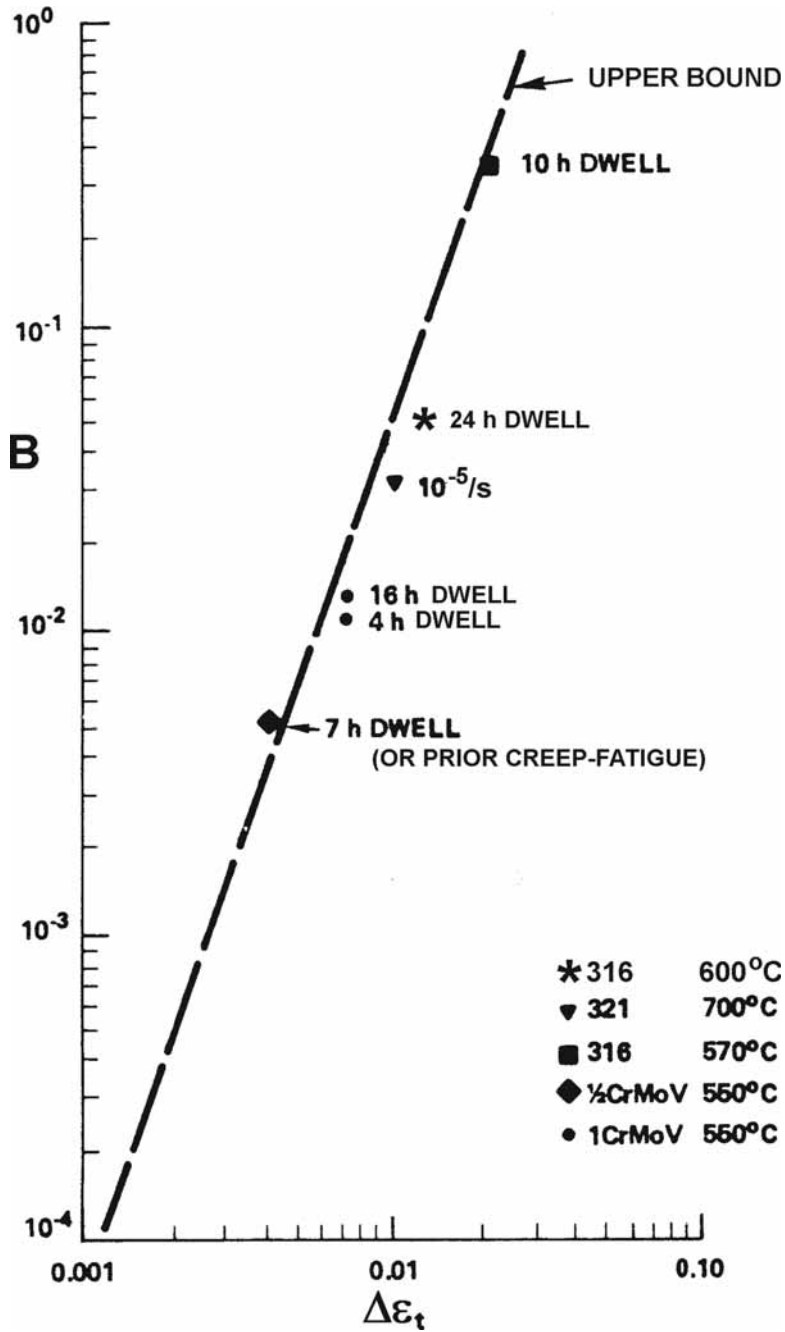


FIG. 2—Upper bound growth relation Eq (7) for several alloys under tensile dwell conditions or low strain rates, see text for data sources. After Ref [6].

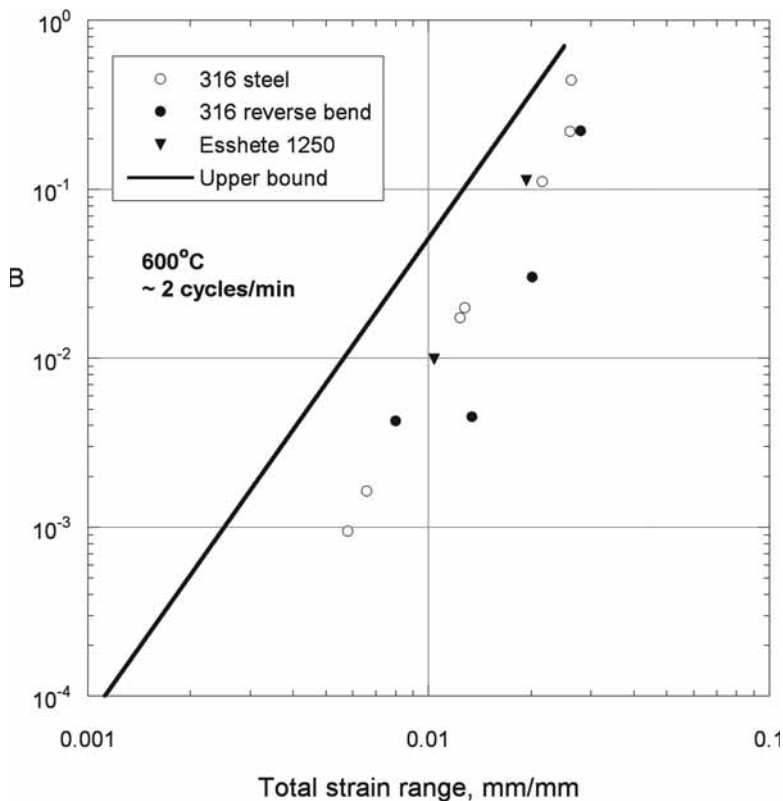


FIG. 3—Continuous-cycling crack growth data deduced from striation data in Ref [14] for comparison with upper bound Eq (7).

deep) specimens of 5 mm square section [20]. (Lower-strain specimens had a notch depth of 0.3 mm.) Stage I crack growth consisted of heavily oxidised intergranular cracking, giving way to stage II cracking (striations). Cracking was predominantly transgranular, though some intergranular stages were present; at greater depths, large striations sometimes contained smaller ones. In contrast, at low ($< 0.2\%$) plastic strain-range levels, several cycles could be associated with a single striation. The minimum spacing was observed to be $0.5 \mu\text{m}$, increasing to a maximum of $9 \mu\text{m}$. The results are plotted in Fig 6, using Eq (8) and the values $A = 1715 \text{ MPa}$ and $\beta = 0.240$ in Eq (4) [17] to convert to total strain range. Neglecting differences in technique, it is seen that the straight-fronted cracks in square specimens (optical method) travel approximately twice as fast as a semi-elliptical crack in cylindrical specimens. This is in contrast to the 400°C data of Fig 5.

Again using Eq (8) and the values $A = 2038 \text{ MPa}$ and $\beta = 0.199$ in Eq (4) [59], striation-based results are plotted in Fig 6 for 316 steel at 570°C , undertaken at a strain rate of $4 \times 10^{-4}/\text{s}$. As expected, the data lie further away from the reference Eq (7). Further striation data for 316 steel at 600°C in air and

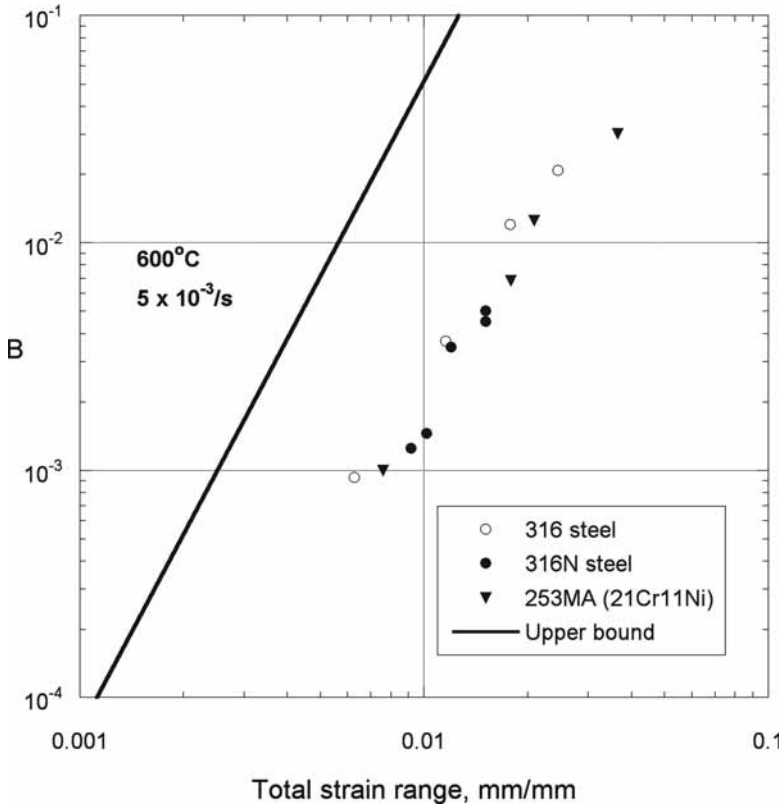


FIG. 4—Continuous-cycling crack growth data deduced from striation data in Refs [57] and [58] for comparison with upper bound Eq (7). Compare with Fig 3.

vacuum deduced from the experiments of Levaillant et al. [60] are also plotted in Fig 6. At the strain range investigated, vacuum growth rates are about half those in air.

Striation spacings reflect the magnitude of crack opening displacement (COD) and (neglecting minor striations within a major striation) offer a useful measure of the maximum cyclic crack growth rate under given conditions. This aspect will be returned to later for dwell conditions. At crack depths below 0.25 mm, striations have in fact been observed with dwell for 316 steel at 625°C [3,59], see Fig 7. This is referred to as “phase I”. However, the longer the dwell, the more restricted is this crack depth. It would thus be incorrect to extrapolate to unit crack depth in Fig 7 for example, in an attempt to deduce a value for the term B .

Notwithstanding, the benefits of Eqs (5) and (7) and the data for example in Figs 2–6 can now be seen. If the cyclic strain range is known at the surface of a component or structure, an immediate assessment may be made of cyclic crack growth rate at a specified depth. In practice this depth must not exceed $1/4$ of the cross sectional width of the component, a similar ratio applying to a specimen in the laboratory [4].

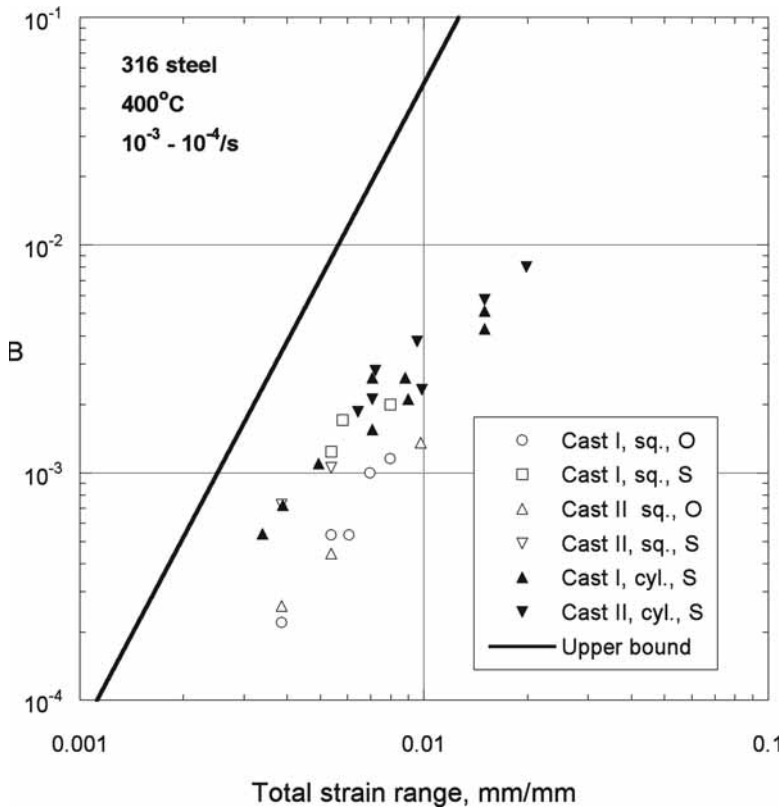


FIG. 5—Comparison of techniques in 316 steel at 400°C in square or cylindrical cross section specimens (O, optical method; S, striation method). Data taken from Ref [21]. Compare also with upper bound Eq (7).

Enhancement of Growth Law—Damage Factors

Taking Eq (7) plotted in Figs 2–6 for example, there is no mechanism for this empirical line, although since it is known that tensile dwells or slow cycling produce an increase in cyclic growth rates [6]; it is expected that crack growth conforming to this line would be intergranular. This may be shown by examining Solomon and Coffin's [15,16] data on alloy A286 at 593°C, which are re-plotted in Fig 8. The total strain range has been calculated making use of Eq (4) and cyclic stress-strain data (strictly, "frequency-modified" data) supplied in a companion paper [61]. Cyclic crack growth rates in both air and vacuum increased as the testing frequency (i.e., strain rate) was reduced. The effect was greater in air than under vacuum, such that intergranular cracking occurred at 10^{-3} Hz ($\sim 3 \times 10^{-5}$ /s) in air but was delayed until 10^{-4} ($\sim 3 \times 10^{-6}$ /s) Hz in vacuum. Superimposing the upper bound Eq (7) in Fig 8, it is seen that at frequencies $< 10^{-4}$ Hz ($\sim 3 \times 10^{-6}$ /s) the upper bound is transcended in these experiments, cracking lying entirely in the intergranular region.

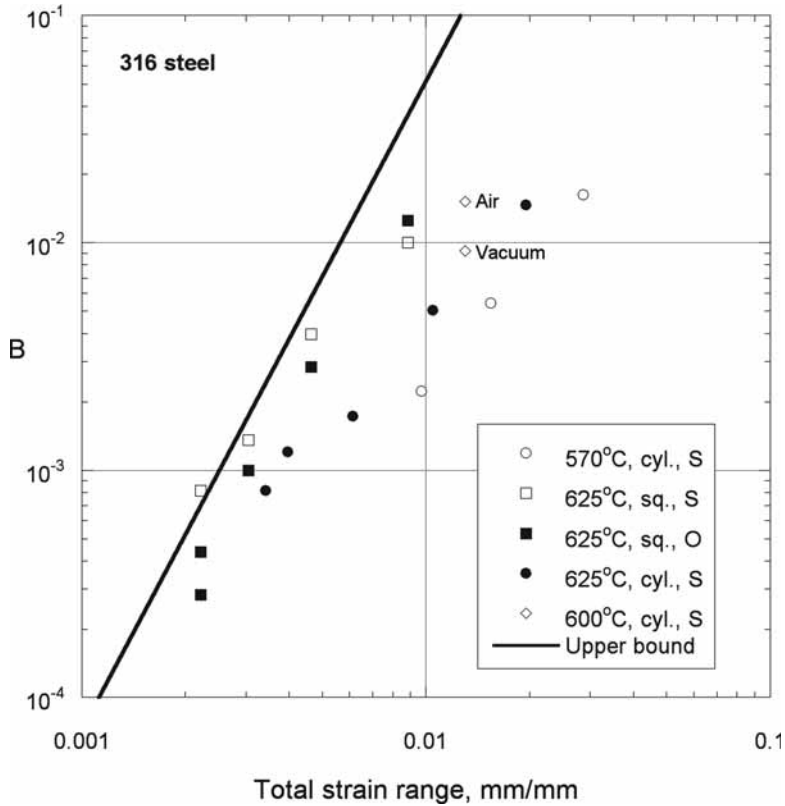


FIG. 6—Results for 316 steel at in square or cylindrical cross section specimens (O, optical method; S, striation method). Data at 570, 600 and 625°C are from Refs [59], [60], and [20], respectively.

Mechanistically Based Models

As discussed by Rémy [62], many creep-fatigue models are of the additive type

$$\left(\frac{da}{dN}\right)^* = \left(\frac{da}{dN}\right)_f + \left(\frac{da}{dN}\right)_c \quad (9)$$

where the subscripts 'f' and 'c' stand for fatigue (continuous cycling) and creep, respectively, while the asterisk denotes the combined effect. However this type of relation appears to have been applied mostly in the LEFM (or ΔJ) region. In the LCF regime, a strain-enhanced oxidation term is often added [62,63]. In the following models, creep damage generally appears rather as a multiple of the continuous-cycling crack growth rate.

Reuchet and Rémy [64] found that for MAR M509 alloy at 900°C, Eq (6a) with $Q = 1.8$ and $a_o = 0.06$ mm gave a predicted life of 3900 cycles compared with an observed plain specimen endurance of 1700 cycles. They suggested that as cracking proceeds into cumulatively damaged regions the relation becomes

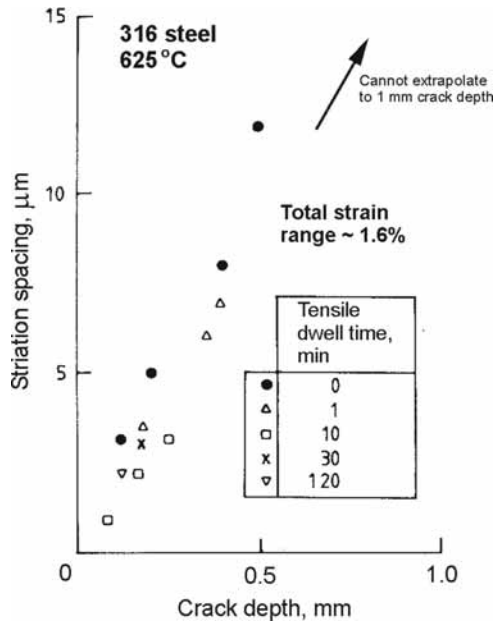


FIG. 7—Striation spacings observed in austenitic steel over limited crack depth. After Refs [3] and [59].

$$\frac{da}{dN} = Ba(1 + qf) \quad (10)$$

where q is a constant and f is the volume fraction of cavities, which in their experiments developed as $(\text{time})^{4/3}$, that is, as $N^{4/3}$. Thus the value of B in Eq (10) effectively increases with time. Upon integrating Eq (10), a propagation life of only 1350 cycles was predicted [64]. It is notable that by incorporating a term of negative sign into Eq (10), the “healing” effect of crack re-welding in the tip regions may in addition be accounted for [1].

Majumdar and Maiya [65,66] introduced a model where cyclic crack growth rate could be accelerated or decelerated according to the presence of cavities ahead of the crack. Such cavities grow in the presence of a tensile dwell or sinter under the influence of a compressive dwell. This was supported by experimental evidence on type 304 steel at 593°C and a set of coupled equations (“damage rate equations”), both containing the applied plastic strain rate, were derived describing the rate of crack advance and cavity growth and their interaction. The derivation is complex, but Pineau [25] has shown that these authors’ considerations reduce to

$$\left(\frac{da}{dN}\right)^* = \left(\frac{da}{dN}\right)(1 + \gamma GIN) \quad (11)$$

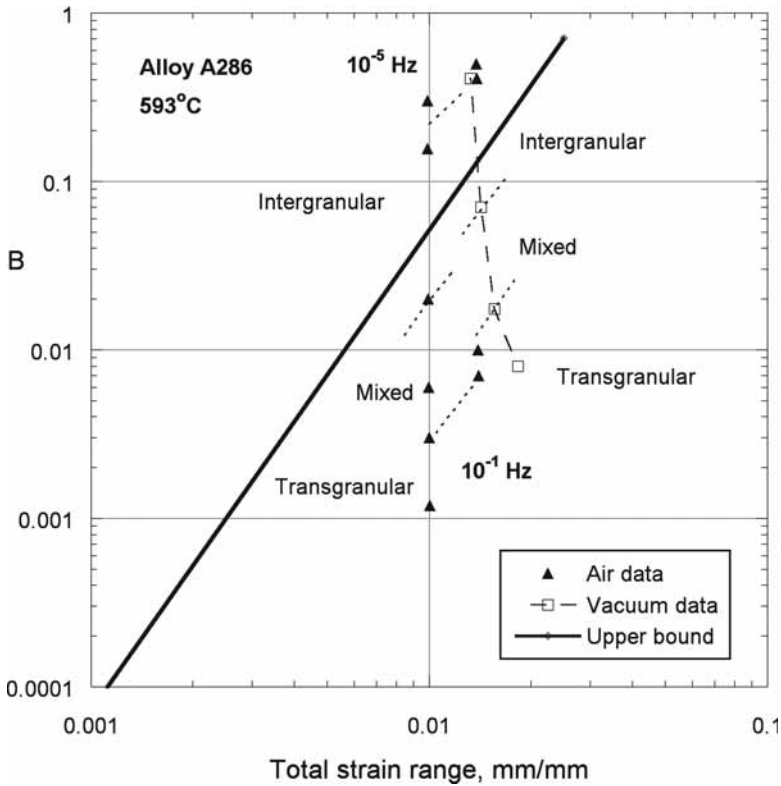


FIG. 8—Mechanism map on Alloy 286 for comparison with upper bound Eq (7). Compiled from data in Refs [15], [16], and [61].

where:

γ = a parameter related to the G coefficient (positive or negative sign) used in the presence of tensile or compressive stress,

I = an integral involving the plastic strain range and strain rate over a single cycle, and

N = the number of cycles.

Equation (11) is effectively a continuous-cycling growth rate given by Eq (5) multiplied by a damage factor. The subscript 'f' henceforth is dropped for convenience.

Priest and Ellison [67] and Miller et al. [68] have quantitatively compared the Majumdar and Maiya [65,66] approach (in its integrated form) with other models for creep-fatigue endurance prediction. They concluded that the method was powerful, especially in its ability to account for the healing of damage under compressive dwell conditions, but noted that a large number of constants were required which in turn depended on complex and time consuming dwell tests.

Levaillant and Pineau [24,27] based their damage model on the metallographically deduced fraction of cracked grain boundaries per cycle, D_c , modified by a further coefficient μ and exponent p . Their studies on a 316 steel at 600°C led to the expression

$$\left(\frac{da}{dN}\right)^* = \left(\frac{da}{dN}\right) \frac{1}{(1 - \mu N^* D_c^p)^2} \quad (12)$$

In the later work [27] the factor μ was connected with the stress drop during relaxation, during which creep damage accumulates.

Beere and Miller's [69] model considers, during "slow-fast" deformation (that is, a slow tension-going deformation rate followed by a fast compression-going rate, the latter at a constant value of $10^{-3}/s$), grain-boundary cavities growing within the plastic zone and ahead of it in the "elastic" zone. As such, the model is not strictly of LCF type, since yield is not general but it gives a useful mechanism map separating the possible damage mechanisms. An equation was set up that balanced the applied load with the load-bearing capacity of the two zones. Cavities in the zone increased in size on approaching the crack tip. The resulting expression is

$$\left(\frac{da}{dN}\right)^* = \Delta \epsilon_p \frac{(\sigma_m / \sigma_y)}{1 - (p^2 / \lambda^2) - (\sigma_m / \sigma_y)} a \quad (13)$$

where:

σ_m = the maximum tensile stress,

σ_y = the cyclic yield stress, and

p and λ are the average values of void size and spacing in the plastic zone.

When cavitation is small, the term in p , λ is negligible and the equation approaches that for continuous cycling. When cavitation is extensive the p , λ terms dominate, causing accelerated growth. The resulting mechanism map for 20Cr25Ni/Nb steel over a temperature range is shown in Fig 9.

In such models, the cavity growth rate is calculated from the knowledge of grain-boundary diffusion coefficients, atomic volume, Boltzmann's constant, etc. A similar approach has been taken by Rie et al. [70], except that these authors employ a more realistic crack growth rate, which by comparing Eqs (4), (5), and (7) requires there to be an exponent to the term $\Delta \epsilon_p$. A value of this exponent is typically 1.5 for austenitic steels [1-3,25].

Empirical Models

In the linear damage approach, the incremental fatigue damage (for continuous cycling) is the reciprocal of the cycles involved, i.e., $1/N$. To this is added the creep damage, d_c , so the total damage per cycle is given by [30,32]

$$\frac{1}{N^*} = \frac{1}{N} + d_c \quad (14)$$

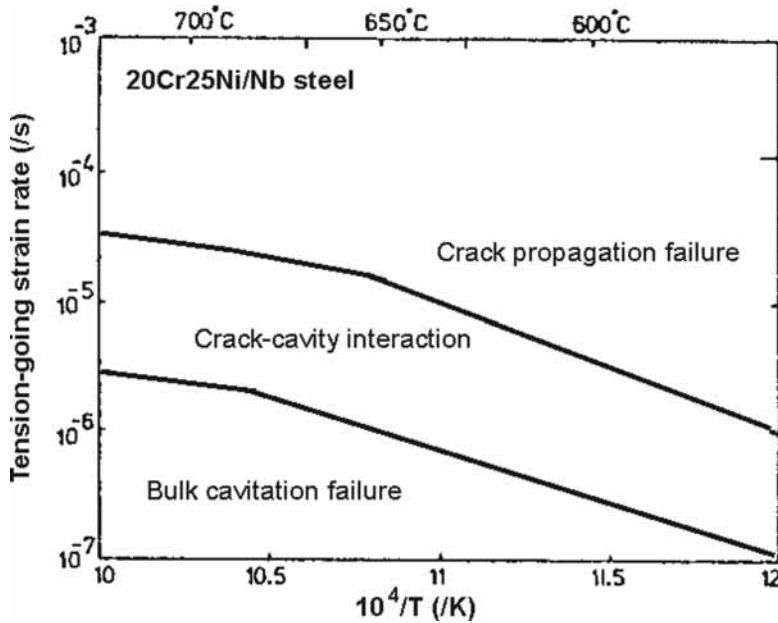


FIG. 9—Mechanism map for austenitic steel as a function of temperature and tensile strain rate. After Ref [69].

Thus if N denotes the number of cycles involved in a continuous-cycling step and N^* the reduced number of cycles in that same step due to creep-fatigue interaction, then it is possible to show [32] by differentiation of Eq (14)

$$\left(\frac{da}{dN}\right)^* = \left(\frac{da}{dN}\right)(1 + Nd_c)^2 = \left(\frac{da}{dN}\right) \frac{1}{(1 - N^*d_c)^2} \quad (15)$$

The second relation shows a classic damage function ω increasing towards unity as the number of creep-fatigue cycles accumulates, i.e., the term d_c now explicitly accounts for the creep damage per cycle during the crack growth phase. Under these conditions therefore the basic LCF continuous-cycling law given by Eq (5) becomes enhanced owing to the creep damage accumulated during the dwell. Substituting Eqs (5) and (6a) into Eq (15) the result is

$$\left(\frac{da}{dN}\right)^* = Ba^Q \left\{ 1 + \frac{d_c}{(Q-1)B} (a_o^{1-Q} - a^{1-Q}) \right\}^2 \quad (16a)$$

A very similar relation applies [32] when $Q = 1$ and is

$$\left(\frac{da}{dN}\right)^* = Ba \left\{ 1 + \frac{d_c}{B} \ln \left(\frac{a}{a_o} \right) \right\}^2 \quad (16b)$$

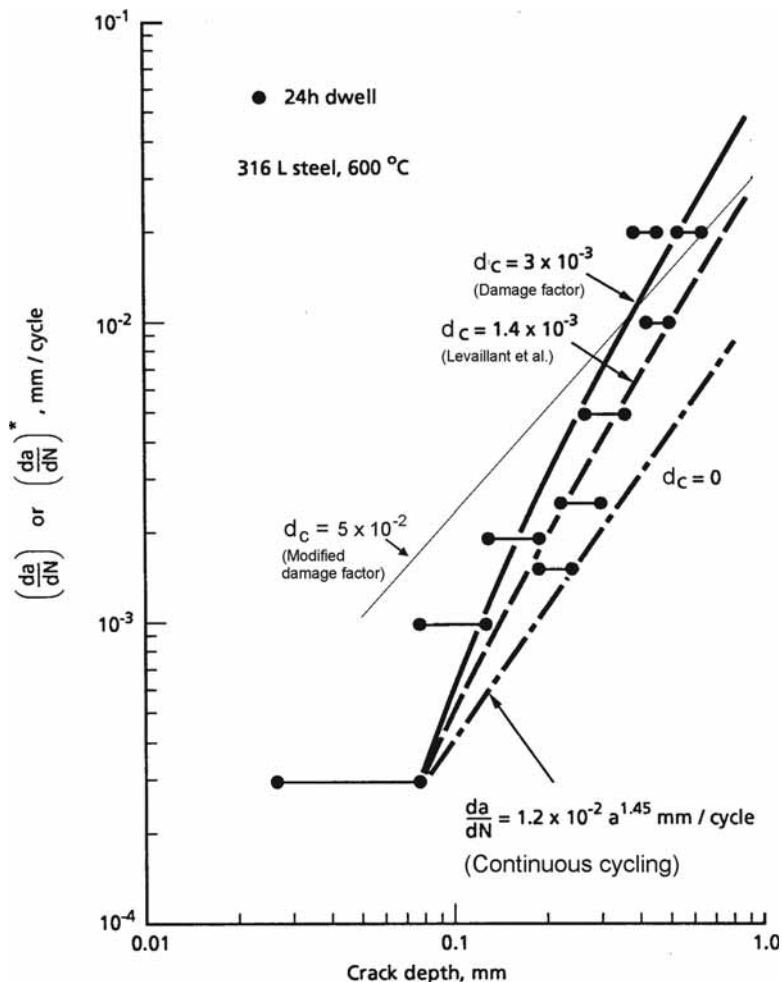


FIG. 10—Comparison of damage factors on original theory (full line), modified theory (fine full line) and that based on metallographic observation (broken line). Original data points from Levaillant et al. [27] represent markings/secondary cracking. After Refs [32] and [52].

Values of d_c have been determined empirically for creep-fatigue growth in a variety of alloys [32]. Equation (16a) is shown plotted in Fig 10 (full line) assuming $a_0 = 0.075$ mm and $B = 1.2 \times 10^{-2}$ from the 24 h dwell-time results of 316L steel at 600°C [27]. A good fit to the data is obtained by assuming $d_c = 3 \times 10^{-3}$ in Eq (16a).

It can be shown [52] that for the 24 h dwell data in Fig 10 (for which $D_c = 2.78 \times 10^{-4}$, $p = 2.76$, and $\mu = 2.76$ [24,27]) that the equivalent damage factor, d_c in Eq (16a) is 1.4×10^{-3} . This is also shown plotted in Fig 10 as the

broken line and may be regarded as more physically founded than the 100% empirical fit from Eq (16a) accorded to the full line in this figure.

Takahashi et al. [71,72] have derived (from a different standpoint) a series of damage curves for the creep-fatigue interaction. It may be shown [52] that their approach effectively allows the square exponent 2 in Eq (15) to become a variable term, α , as follows

$$\left(\frac{da}{dN}\right)^* = \left(\frac{da}{dN}\right) \frac{1}{(1 - \phi_c)^\alpha} \quad (17)$$

This relation will be returned to in the Discussion section.

Modification to Damage Factors for LCF

Strictly, as pointed out by Tomkins [73,74], the (reversed-plastic) zone, r_p , which is a region of concentrated strain encountered in LCF, is itself surrounded by a matrix which undergoes reversed “general” yielding (as opposed to reversed elasticity occurring in LEFM), indicated in Fig 11. This Figure also indicates a schematic size for the process zone. This may be defined as an imaginary damage-accumulating region at the crack tip whose size is determined by the crack-growth model adopted. It is contained within the actual cyclic plastic zone which is much larger; moreover the latter is a real entity because its size can often be identified by metallographic means [52].

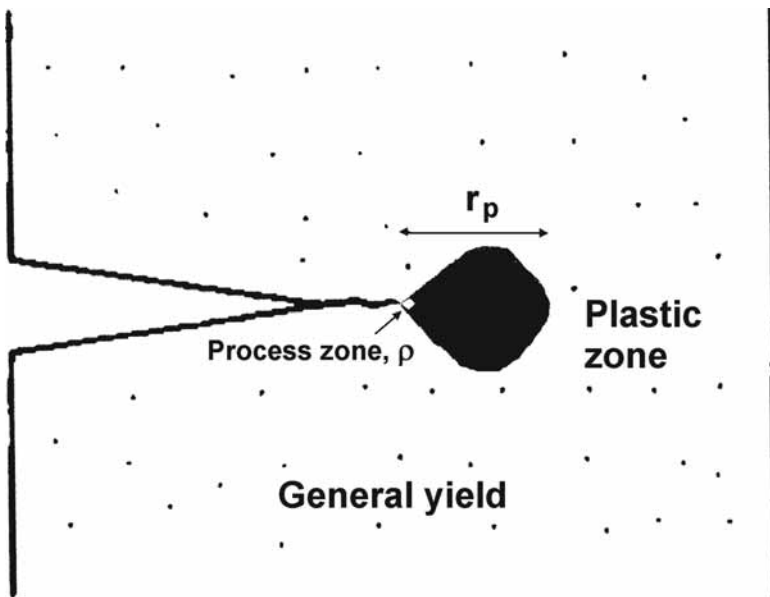


FIG. 11—In LCF, the process zone is embedded in the cyclic plastic zone, in turn embedded in material undergoing reversed yielding.

Lloyd and Wareing, who worked in terms of the cyclic plastic zone [75], in contrast argued that there is no real difference at the crack tip—only the boundary conditions are changed. The size of the zone is given [73,74] by

$$r_p = \left[\sec \left(\frac{\pi \Delta \sigma}{4 T^-} \right) - 1 \right] a \quad (18)$$

where:

$\Delta \sigma$ = the stress range and

T^- = the UTS of “heavily cycled material.”

The plastic zone size thus increases in proportion to the crack depth. It may then be shown [52] (by allowing creep damage to accrue only in the plastic zone and not generally in the ligament, assuming that the creep-fatigue cycles to traverse the zone is given by $N^* = r_p / (da/dN)^*$) that

$$\left(\frac{da}{dN} \right)^* = \frac{Ba^Q}{2} [1 + 2M + (1 + 4M)^{1/2}] \quad (19)$$

where in turn $M = d_c r_p / Ba^Q$. Using appropriate values for material constants [52] (for example, $A = 2050$ MPa, $\beta = 0.256$ for the 316 steel [27]), the value of d_c was re-assessed using Eq (19) for the experimental data shown in Fig 10 (fine full line). It is seen that it is a factor 15 greater than before. The method does not capture the slope conveyed by the experimental data. For creep-fatigue LCF it is concluded that creep damage is best allowed to accumulate in the remaining ligament rather than in the intense zone at the crack tip.

Tomkins/Wareing Approach

These authors deserve separate consideration since their work during the 1970s and 1980s has been, and continues to be, widely quoted. We now re-examine the basic relations for application in terms of total strain range.

The crack-tip-shear-decohesion mechanism for stage II crack growth [3,76] which incorporates Eq (18) predicts the following relation for propagation

$$\frac{da}{dN} = \Delta \varepsilon_p \left[\sec \left(\frac{\pi \Delta \sigma}{4 T^-} \right) - 1 \right] a \quad (20)$$

where (to emphasize) $\Delta \varepsilon_p$ is the plastic strain range prevailing at the surface. The model predicts exponential growth, i.e., $Q = 1$ in Eq (5). Thus, with knowledge of appropriate cyclic stress-strain properties via Eq (4), predictions on crack growth may be made according to material state. For example, it has been shown [55] that while the terms β and A may vary during evolutionary cyclic hardening or softening, the parameter β/A nevertheless remains approximately constant such that in a typical case

$$\frac{\beta}{A} = 1 \times 10^{-4} \quad (21)$$

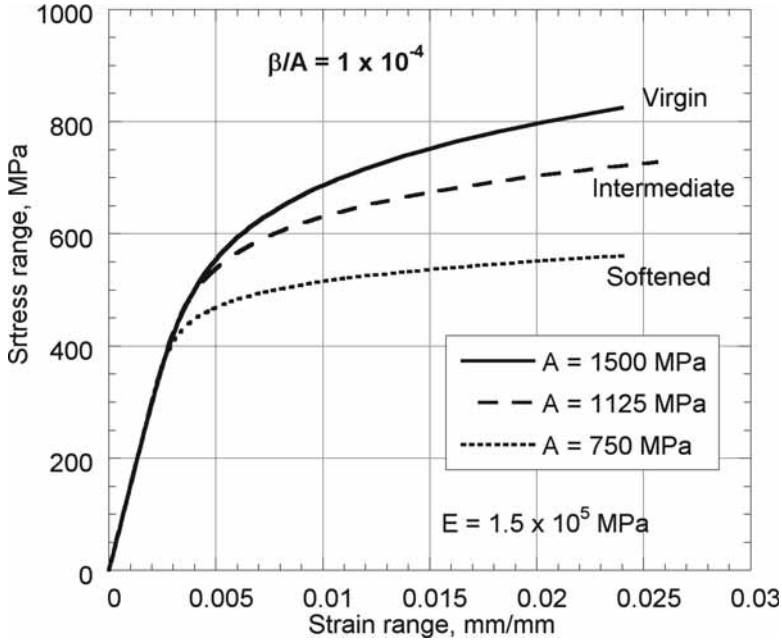


FIG. 12—Model of evolutionary cyclic softening behaviour [52]. Reproduced with permission, Science Reviews 2000 Ltd.

at any time. The model example in Fig 12 shows a virgin curve ($A = 1500$ MPa, $\beta = 0.150$, $E = 150$ GPa) and curves with factored (i.e., from Eq (21)) values of $A = 1125$ MPa; $\beta = 0.113$ (“intermediate softening”) and $A = 750$ MPa; $\beta = 0.075$ (“fully softened”)

The calculation may be performed in reverse for a cyclically hardening material.

The term preceding “ a ” of Eq (20) is thus identified with the parameter B of Eq (5). Taking now values of A and β used for the “virgin,” “intermediate,” and “softened” states used in Fig 12 and substituting into Eq (20), values of B are plotted against total strain range in Fig 13 for the three modelled states. Comparison is also made with the upper bound Eq (7), and it is seen that for a given total strain range, a higher growth rate is predicted for fully softened material. Again, the calculation may be done in reverse, predicting a slower crack growth rate for the hardened state.

This model also predicts its own “upper bound,” given as half the crack tip opening displacement (CTOD/2) [3,76]. The corresponding equation for growth in this case is

$$\frac{da}{dN} = \Delta \epsilon_p \left[\frac{\pi \Delta \sigma^2}{8 T^- E} + \frac{\pi \Delta \sigma \Delta \epsilon_p}{2 T^- (1 + \beta)} \right] a \quad (22)$$

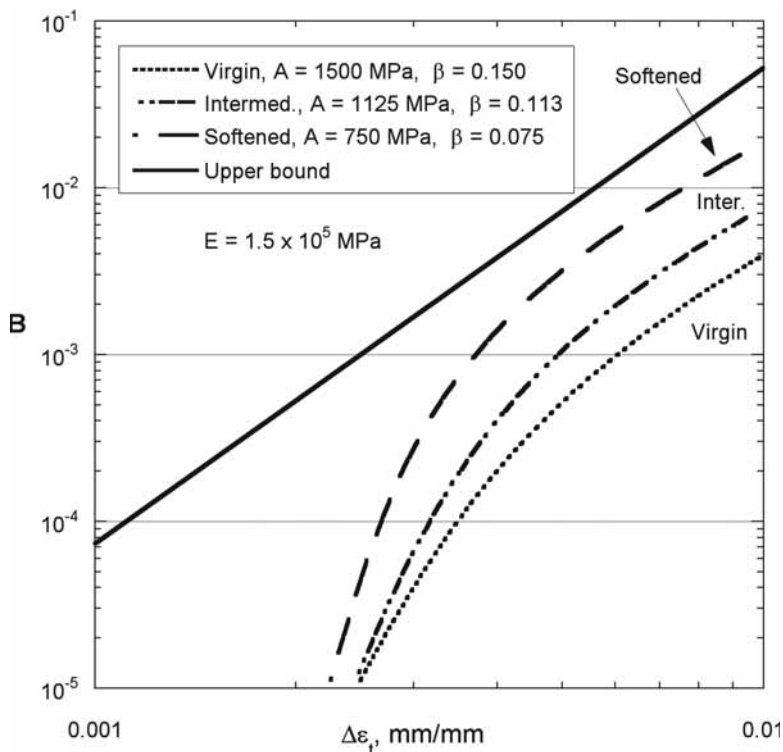


FIG. 13—Comparison of upper bound Eq (7) with crack-tip shear decohesion model [52]. Reproduced with permission, Science Reviews 2000 Ltd.

The “ B ” term in Eq (22) is shown plotted in Fig 14 for the three “strength” states. In this case the difference between the material states is not so marked, although the same order is preserved. It is notable that the data straddle the upper bound Eq (7).

Block Loading/Sequential Tests and “Zone Size”

In lieu of models based on the process (or plastic zone, r_p) zone size, authors sometimes undertake block loading tests in order to ascertain a zone size left by a previous step. Similar tests have also been conducted in the LEFM range [77–79], reviewed in Ref 52. Cases for LCF are now considered.

During tests on type 321 steel at 700°C [22], strain rates on the same specimen were alternated between $6.7 \times 10^{-3}/s$ and $6.7 \times 10^{-5}/s$. Corresponding cracking on the fracture surface was transgranular (showing striations) and intergranular (no striations). The demarcation lines were clearly indicated, enabling the intergranular rates to be determined.

In some “fast-fast” tests ($\sim 10^{-3}/s$) on type 304 steel at 600°C, temporary excursions to “slow-fast” ($\sim 10^{-5}$ then $10^{-4}/s$) cycling increased crack growth

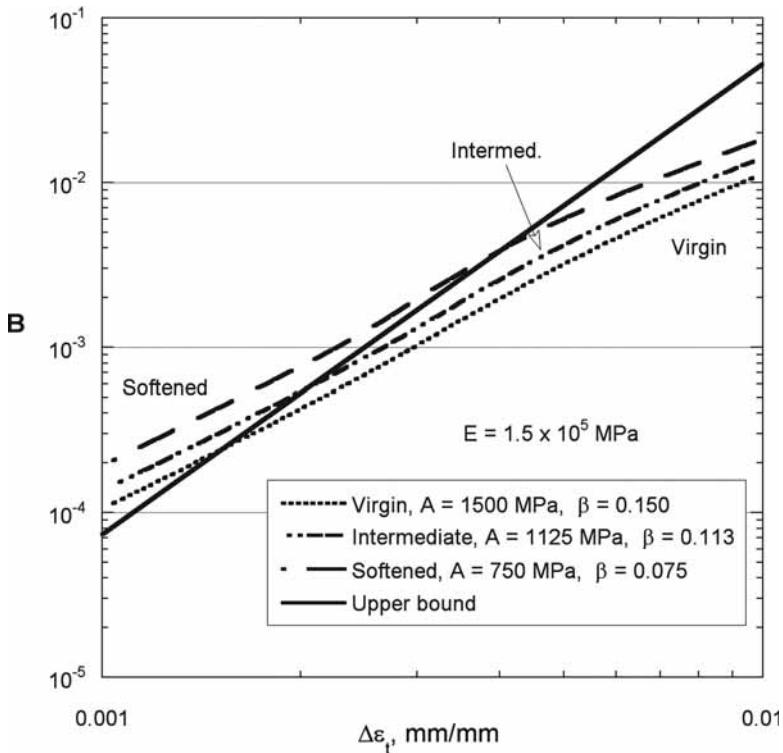


FIG. 14—Comparison of upper bound Eq (7) with crack-opening-displacement model. Reproduced with permission, Science Reviews 2000 Ltd.

rates by a factor 3, the original rate being restored on reverting to fast-fast cycles [80]. Thus there was little history effect after the 25 excursion cycles had taken place. Appropriate values of B at 0.8% total strain range for comparison with Fig 2 are 5×10^{-3} and 2×10^{-2} , respectively, at tensile strain rates of $2 \times 10^{-4}/s$ and $2 \times 10^{-5}/s$, respectively [81].

In some tests on 304 steel at 593°C at 1% total strain range Majumdar and Maiya [65] propagated a crack to a depth of 1 mm under a 10 min tensile dwell condition ($B = 5.1 \times 10^{-3}$ in Eq (5)). At this point cycling continued with a 4 min compression-only dwell. The crack growth rate was observed to slow down ($B = 3.5 \times 10^{-3}$ in Eq (5)) towards the continuous-cycling growth rate condition ($B = 3.2 \times 10^{-3}$ in Eq (5)) observed in a separate test. The effect was attributed to a shrinkage of cavities under the compressive hold. It may be noted that respective values of B remain below the upper bound of Eq (7). These and similar experiments have formed the basis of the authors' interactive model of crack growth, discussed previously.

Aged type 316 specimens were tested at 625°C [82]. The standard waveform was a 5 min compressive dwell, followed by a 60 min tensile dwell (The short compression dwell was believed to accelerate cavity formation during the

tensile dwell.) Comparison was made with continuous-cycling data on a separate specimen. In the dwell test, cracking was transgranular up to about 1 mm crack depth. There was then a gradual transition to intergranular cracking and a marked acceleration in growth rate. A compressive dwell of 6 h was then applied. On resuming the standard dwell cycle the growth rate was reduced to the reference continuous-cycling rate and the path was again transgranular. Further similar sequential tests led the authors to believe that the entire (creep cavity) damage was initiated in the first dwell cycle. Lloyd and Wareing [75] questioned this particular conclusion.

Discussion

It has been demonstrated that the relations deduced for interactive growth in creep-fatigue have a distinct similarity. Throughout, comparisons have been made against the reference Eq (7). The slope of Eq (7) was chosen as an empirical conservative prediction for ferritic data available at the time [4], and was based on cyclic oxidation effects at the specimen surface/crack tip [83] rather than, say, the CTOD model of Tomkins et al.

Interaction Coefficient

It has been pointed out [52,84] that Eq (14) is not strictly an interaction relation, merely a linear addition to unity of two damage criteria. If a term I_{cf} describes a measure of the interaction upon pure fatigue crack growth for accumulated creep damage φ_c , then it can be demonstrated [52] that

$$\left(\frac{da}{dN}\right)^* = \left(\frac{da}{dN}\right) \frac{1 - I_{cf}\phi_c^2}{[1 + I_{cf}\phi_c^2 - (1 + I_{cf})\phi_c]^2} \quad (23)$$

Since $\varphi_c = N^*d_c$ (see also Eq (14)), it is noted that Eq (23) reduces to Eq (15) when $I_{cf} = 0$.

Equation (23) is shown plotted in Fig 15, normalised to the continuous-cycling growth rate (da/dN), indicating the degree of enhancement for various values of I_{cf} . It is compared with the simpler Eq (15) which is based on simple linear addition of creep and fatigue damage. A very similar relation may be obtained by manipulating Eq (17) [52] which is based on the work of Takahashi et al. [71,72].

Minimum Crack Depth

It has been demonstrated [4,25,27] that Eq (5) breaks down below some value a_{min} , so that after crack initiation, and up to a depth $a_{min} \sim 0.20$ mm the crack growth rate is constant. Kitamura et al. [85] have shown that in type 304 steel at 650°C in vacuum, coalescing cracks appear to undergo “fits and starts” of propagation, some arresting at grain boundaries, so that on average, propagation rate is constant in this region. We may thus write

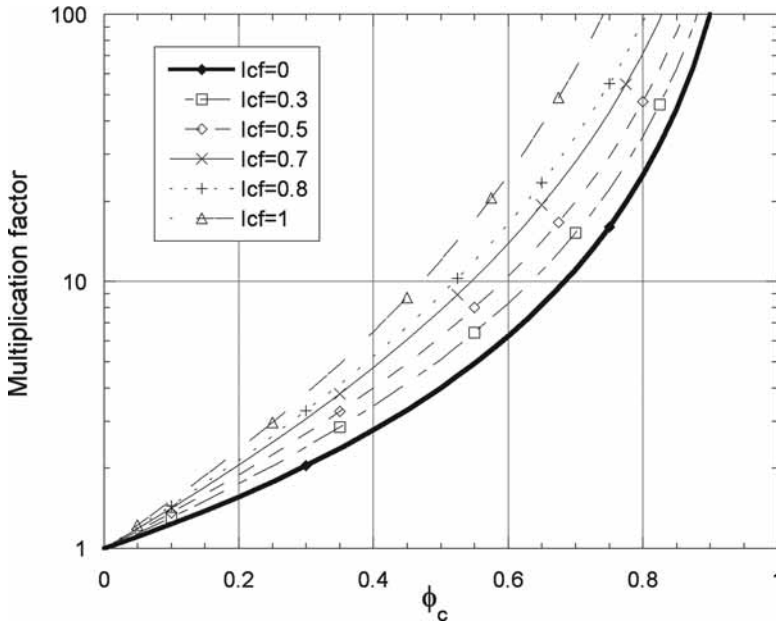


FIG. 15—Enhanced crack growth rate due to creep-fatigue interaction, Eq (23). Case for $I_{cf}=0$ corresponds to Eq (15). After Ref [52].

$$\frac{da}{dN} = Ba_{\min} = \text{constant} \quad (24)$$

Hence if it requires $N_{\min}$ cycles to propagate up to $a_{\min}$, beyond which power-law growth occurs

$$N_{\min} = \frac{a_{\min}}{Ba_{\min}^Q} = \frac{1}{B} a_{\min}^{1-Q} \quad (25a)$$

whereas if exponential growth applies

$$N_{\min} = \frac{a_{\min}}{da/dN} = \frac{1}{B} \quad (25b)$$

Thus for exponential growth, the term $1/B$ may alternatively be interpreted as a measure of cycles up to $a_{\min}$. In the R5 Code [30], $a_{\min}$ is assumed to be 0.20 mm, but there may be a case for reducing this value generally to 0.14 mm or thereabouts. It is emphasized that Eqs (25a) and (25b) may not apply under dwell conditions where bulk creep damage can cause early growth to be widespread, discussed later.

Propagation Depth with Damage

Equations (16a) and (16b) are most useful in their integrated forms for block loading over N_i cycles. In continuous cycling these take the respective forms

$$a_f = \left(a_o^{1-Q} - (Q-1) \sum_{i=1} B_i N_i \right)^{\frac{1}{1-Q}} \quad (26a)$$

$$a_f = a_o \exp \left(\sum_{i=1} B_i N_i \right) \quad (26b)$$

where the damage function has been set to zero for power-law or exponential growth.

The relation may also be used in a “non cumulative damage” mode

$$a_f = \left\{ a_o^{1-Q} - (Q-1) \sum_{i=1} \frac{B_i}{d_{ci}} \left(\frac{1}{1 - N_i d_{ci}} - 1 \right) \right\}^{\frac{1}{1-Q}} \quad (27a)$$

$$a_f = a_o \exp \left\{ \sum_{i=1} \frac{B_i}{d_{ci}} \left(\frac{1}{1 - N_i d_{ci}} - 1 \right) \right\} \quad (27b)$$

This form allows creep damage to accumulate over each block of identical cycles only, resetting the value of d_c for each block. Damage in previous cycles is thus ignored, and this could occur physically when a period of intergranular cracking is followed by transgranular cracking, as in the tests of Majumdar and Maiya [65,66] discussed above, or those of Raj and Min [82] where creep damage is healed by a compressive dwell.

For true cumulative damage, all damage that has occurred previously in the specimen or component must be accounted for. It can be shown that the respective forms are

$$a_f = \left\{ a_o^{1-Q} - (Q-1) \sum_{i=1} \frac{B_i N_i}{(1 - \omega_o)(1 - \omega_t)} \right\}^{\frac{1}{1-Q}} \quad (28a)$$

$$a_f = a_o \exp \left\{ \sum_{i=1} \frac{B_i N_i}{(1 - \omega_o)(1 - \omega_t)} \right\} \quad (28b)$$

where ω_o is all previous cumulative damage up to the current i th block and ω_t is this damage plus that accumulated in the i th block itself.

The exponential forms have been presented previously [6,32] but the power-law versions are provided for a sensitivity analysis. An initial defect size of 0.2

TABLE 1—*Postulated loading regime for component.*

	Cycles/block	$\Delta\epsilon_t$, mm/mm	d_c	B in Eq (7)
1st application	50	0.006	3×10^{-4}	1.21×10^{-2}
2nd application	200	0.005	2×10^{-4}	7.22×10^{-3}
3rd application	300	0.004	1×10^{-4}	3.82×10^{-3}

mm is now assumed. A typical block loading regime is summarised in Table 1. The final depths of crack penetration for the cases of Eqs (26) are, respectively, presented in Table 2. In the case of power-law growth with $Q = 1.75$, penetration is less for the chosen value of B because, as may be seen by comparison with Fig 1, overall growth rates are less in the early stages for a steeper slope. Clearly, for each case, penetration depths will converge as $Q \rightarrow 1$.

Cavitation and Damage

In comparing the various crack-growth models, the original schematic of Hales [86] for creep-fatigue crack initiation may with benefit be modified for growth, as shown in Fig 16. A fatigue crack is allowed to start at one face of a specimen or component while creep damage builds up elsewhere. Thus in Fig 16:

- Case A shows the fatigue crack propagating further along a transgranular path, whilst creep damage accumulates throughout the specimen/component quite independently. This is “competitive” crack and cavity growth, each mechanism obeying its own “failure criterion.”
- Case B shows (as an example) approximately 70% fatigue failure and 30% creep failure. Although they initially act independently, failure has been acknowledged since their sum equals unity. This is the “additive” mode and represents a formative type of interaction.
- Case C emphasizes true creep-fatigue interaction, i.e., a transgranular crack is perturbed by accumulating creep damage, and becomes intergranular. This is the basis of a gradual increase in the value of B in Eq (5), as adopted by many of the models.

Further discussion on metallographic aspects of creep-fatigue damage accumulation may be found in Ref [84].

Suppose that B^* is the maximum value of B observed that is consistent with true creep-fatigue interaction, that is where the path of a transgranular crack is diverted and accelerated by developing grain-boundary cavitation, see for

TABLE 2—*Predicted crack growth from separate damage models. Power law assumes $Q = 1.75$.*

Type of damage	Depth, mm (power)	Depth, mm (exponential)
None (continuous cycling)	1.08	4.88
Non-cumulative	1.21	5.41
Cumulative	1.53	6.56

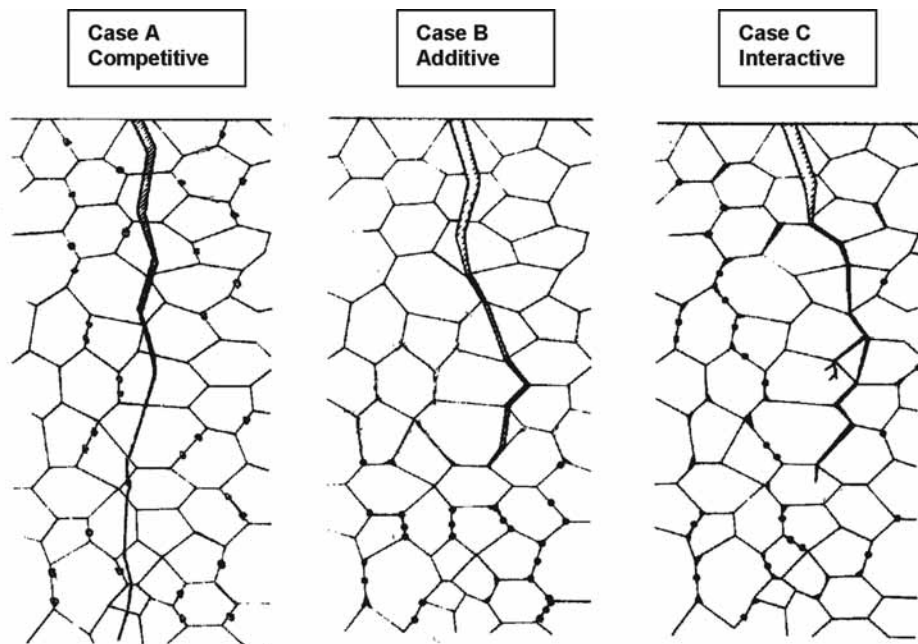


FIG. 16—Three types of crack growth, based on the concept of Hales [86]. Case B illustrates linear damage summation, see text. After Ref [84].

example Refs [32,84]. As discussed previously, this is very well represented by Eq (7). The magnification factor, F , in terms of the equivalent continuous-cycling test at the same strain range may thus be expressed

$$F = \frac{B^*}{B} \quad (29)$$

In Eqs (5), (16(a)), and (16(b)), the term B is numerically equal to the cyclic crack growth rate at unit crack depth (1 mm). Thus $B^* = (da/dN)^*$ at that depth. Re-arranging Eqs (16(a)) and (16(b)) and putting $a = 1$ we find

$$d_c = \frac{B^* (Q - 1) (F^{1/2} - 1)}{F a_o^{1-Q}} \quad (30a)$$

$$d_c = \frac{B^* (F^{1/2} - 1)}{F \ln(1/a_o)} \quad (30b)$$

for power-law growth and exponential growth respectively, so that cyclic creep damage is conveniently expressed in terms of B^* which itself is a function of strain range via Eq (7).

Over a wide range of experimental data [4,6] it is found that, for all strain ranges, there is an order of magnitude increase in creep-fatigue crack growth/

cycle compared with continuous cycling so that $F \sim 10$ in Eq (29). Taking a total strain range of 0.5% (0.005 mm/mm) in Eq (7) and substituting in Eq (30b) with $a_o = 0.14$ mm, the optimum value for d_c becomes 7.9×10^{-4} per cycle. If the strain range is increased to 1% (0.01 mm/mm) this value increases to 5.7×10^{-3} per cycle. If d_c is increased above these respective values at some critical level d_{crit} , it may be argued that creep damage is widespread internally and (grain boundary) failure is beginning to occur independently of crack advance.

Energy Method

For exponential crack growth it has been shown [52] that Eq (3) may be expressed in terms of total strain range as

$$\frac{da}{dN} = \frac{\pi E f(\beta) \Delta \epsilon_t^2}{(1 + \beta) I_\beta W_c} a \quad (31)$$

where the terms $f(\beta)$ and I_β are exactly as defined in Eq (3), so that the parameter B for short crack growth in Eq (5) is now identified from energy considerations. Taking typically for the case of plane stress $E = 1.5 \times 10^5$ MPa, $\beta = 0.2$ and a failure energy (5% load drop stage) $W_c = 1$ J/mm³ ($= 10^3$ MPa) at 550°C [47], comparison of Eqs (31) and (5) gives

$$B = 113 \Delta \epsilon_t^2 \quad (32)$$

This result gives a propagation parameter generally below the upper bound equation Eq (7) as shown plotted in Fig 17.

The value of W_c is a variable; it has been demonstrated that much lower values of W_c are associated with creep damage [6]. This defines the element of “process zone” of width ρ used in the model [47]. It should further be noted that “failure energies” are obtained by summing areas of hysteresis loops: these may consist of “plastic” (for fast cycling) and “creep” (during stress relaxation) components. A further refinement allows for damage accumulation in the array of elements located in the cyclic plastic zone *before* the arrival of the crack tip. Details are given in the original paper [47] where it is shown that magnification factors up to 5 can apply to crack growth rates, depending on how many process zones ahead of the tip are taken as active.

Instability

Our interpretation of the creep-fatigue interaction for stage II cracking thus involves intergranular growth through pre-cavitated boundaries. Two conditions may be proposed for catastrophic failure [59,75,87,88]. The first is an effective reduction in stress by a factor $[1 - (p/\lambda)]$ where p/λ is cavity size/spacing ratio. Spacing is thought to be intimately related to the grain-boundary particle distribution. A more likely condition is that incorporated for example into Eq (22), that is

$$CTOD/2 = (\lambda - p) \quad (33)$$

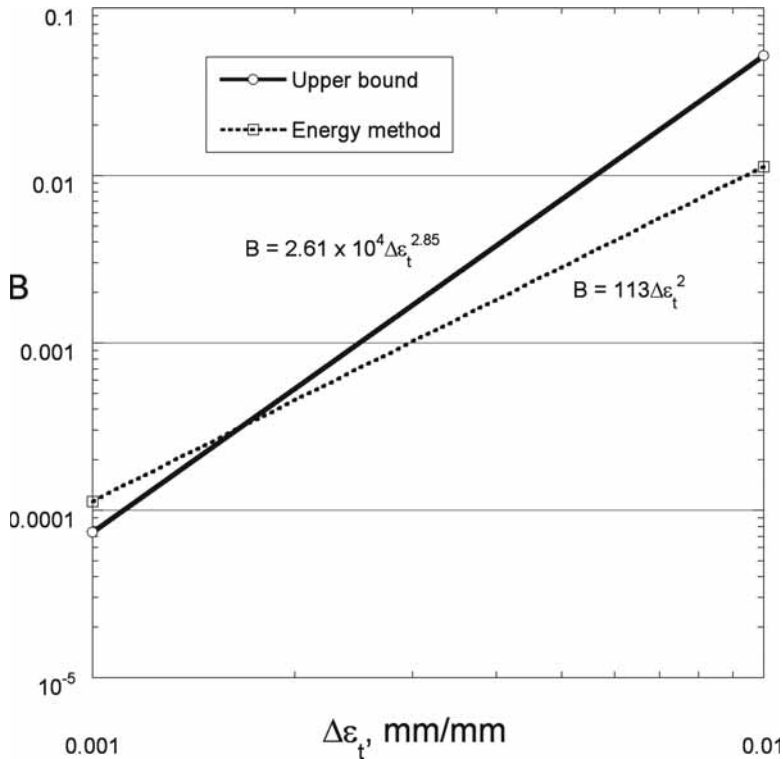


FIG. 17—Comparison of upper bound Eq (7) with predictions from energy model. After Ref [52].

There is no doubt that, at relatively high values of total strain range, unstable cracking at short depths can occur [6]. In a 10 h tension dwell test carried out on 316 steel at 570°C and total strain range of 1.5%, the results in Fig 18 show that after some cyclic hardening, rapid failure occurred across the whole section during the dwell in the 12th cycle, manifest by a large increase in the electrical potential-drop reading, accompanied by a rapid stress drop. A second test involved a bainitic 1CrMoV steel (the same steel was used for the data in Fig 1(b)) which had been service-exposed for 70,000 h at 570°C peak temperature. This was tested in LCF with a 16 h tension dwell at 565°C and total strain range of 0.68%. The starter crack was some 0.2 mm deep. The specimen failed completely after 23 cycles, during which the crack front had travelled 6.35 mm. In Fig 18 is shown the sudden fall off in the tension stress and accompanying increase in the crack potential-drop monitoring signal which occurred in the dwell during the 19th cycle. Catastrophic crack growth thus appeared to occur suddenly in this material over the whole gauge section. (A formal value of B calculated from Eq (5) is 5.5×10^{-1}). These results contrast with earlier work on this and other materials [4,6,8,9] where at lower total strain ranges crack growth increased in a steady manner with increasing crack depth, and for which Eq (7) continues to be a reliable upper bound.

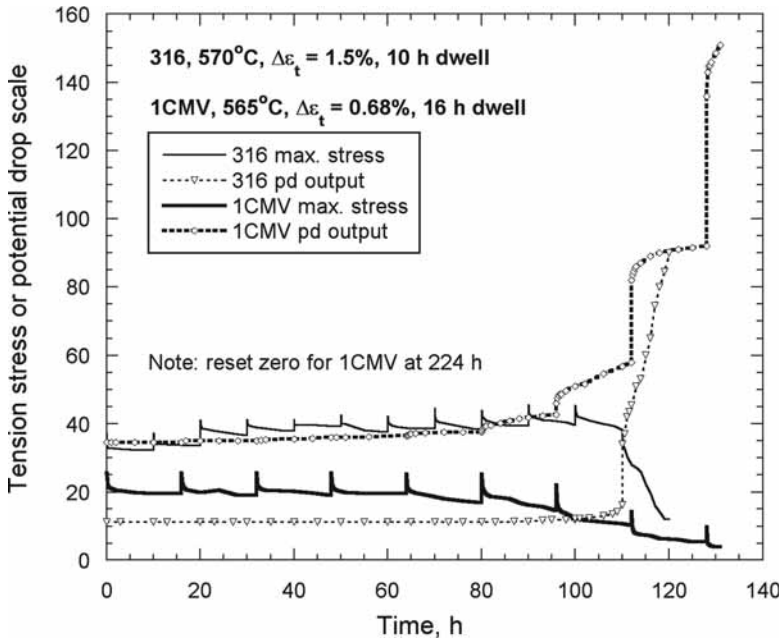


FIG. 18—Tracings from chart records (one “blip” = 1 cycle) showing unstable crack growth in an austenitic and a ferritic steel. Compiled from data in Ref [6].

Concluding Remarks

Several models for creep-fatigue damage during the growth of short cracks have been compared. All incorporate to differing extents a “damage factor” which takes into account the propensity for intergranular cavitation to accelerate the crack compared with continuous-cycling conditions. It is noted that basic models predict exponential growth, for which the term $Q = 1$ in Eq (5) whereas in practice values of $Q > 1$ are sometimes observed. Possible reasons for this effect (for example, crack tip re-welding in an inert environment) have been listed [1]. It has been surmised [47] that theories based solely on CTOD predict exponential growth, but for all other cases, the effective value of Q depends on the size of the assumed process zone, ρ , with respect to the cyclic plastic zone size r_p [89,90]. It is acknowledged that creep-fatigue growth rates may be expressed alternatively in terms, for example, of ΔJ [5,53,91], but the simpler approach adopted in this review has been to use the total surface strain range and specific crack depths in relations for comparison with a reference “upper bound” line given by Eq (7).

Finally it may be noted that this review has treated exclusively of the uniaxial case. Data have been reported for the growth of short cracks in reverse bending of Alloy 800 in the range 500–760°C [92] and $\frac{1}{2}\text{CrMoV}$ steel at 550°C [2]. Since there appeared to be little difference in the value of the empirical parameter B in Eq (5) compared with the uniaxial case, it may be concluded that the equations discussed in this paper may be with benefit applied for bending, given

the same surface strain range conditions. Equally, for large-radius-notched specimens of a 1 $\frac{1}{4}$ CrMoV casting steel in bending at 550°C [93] and for small-radius-notched uniaxial specimens of a forged 1CrMoV steel at a maximum temperature of 565°C [94], the same equations may be applied, provided the equivalent surface strain-range conditions have been identified.

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Hans-Jürgen Christ¹ and Valerij Bauer²

Effect of Creep and Oxidation on the Isothermal and Thermomechanical Fatigue Behavior of an Austenitic Stainless Steel

ABSTRACT: Austenitic stainless steels are often used for high-temperature applications under conditions where fatigue loading occurs in combination with varying temperatures, superimposed mean stresses or dwell times and environmental effects. In order to characterize and separate the various damage contributions on the deformation behavior and the damage evolution of AISI304L, isothermal and thermomechanical fatigue tests were carried out at temperatures ranging from room temperature to 800 °C. The test results in combination with microstructural observations were used for an adaption and application of a simple multi-component model to predict the stress-strain response under thermomechanical fatigue conditions solely from isothermal data. A very reasonable predictive accuracy was obtained and the cyclic stress-strain calculation was directly incorporated in fatigue life assessment models. Two models, which were found to be suitable for isothermal fatigue conditions, are presented. These models were extended and applied to non-isothermal conditions. A comparison of the experimentally observed fatigue life data with the predicted values revealed that thermomechanical fatigue life can reasonably be assessed from isothermal test results, if environmental effects are correctly taken into account. Thermomechanical fatigue tests employing dwell periods indicate that the life prediction models developed are robust and conservative.

KEYWORDS: austenitic stainless steel, high-temperature fatigue, thermomechanical fatigue, environmental effects, modeling of cyclic stress-strain behavior, TMF life prediction

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¹ Institut für Werkstofftechnik, Universität Siegen, D-57068 Siegen, Germany.

² Wieland-Werke AG, Graf-Arco-Straße, D-89070 Ulm, Germany.

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Introduction

Engineering components and structures that operate at elevated temperatures are often subjected not only to isothermal cyclic or static loading but also to thermal expansion and dilatation during the start-up and shut-down period resulting in plastic deformations in the highly stressed regions. The combination of cyclic mechanical and cyclic thermal loading leads to thermomechanical fatigue (TMF) which is considered to be the primary life-limiting factor for engineering components in many high-temperature applications. Depending on the location of the volume element considered and the cooling/heating situation, various types of strain-temperature phasing can arise. In laboratory testing, most commonly the two extreme cases, in-phase (IP) and out-of-phase (OP) TMF, are studied.

Due to the life-reducing effect of TMF it is important to investigate the TMF behavior of engineering materials in order to predict thermomechanical fatigue life correctly. However, TMF is known to cause a complex evolution of damage which can hardly be described in a unique, simple and accurate way. On the other hand, extensive low-cycle fatigue (LCF) data, which is traditionally used for design purposes, has been generated isothermally on several high-temperature materials, and thus, it is tempting to try to predict TMF life based mainly on isothermal LCF data. However, it is well known that depending on the actual loading conditions deformation behavior and damage evolution can be affected significantly by the varying temperature (e.g., [1,2]), and hence life prediction may well be non-conservative if the relevant damage mechanisms are not captured accurately. Creep, fatigue and environmental effects (often simply termed oxidation) are considered to be the main damage contributions. The individual extents of these constituents and their mutual and synergistic interactions depend very strongly on the loading conditions applied and the material considered.

A simple but helpful classification of the TMF life behavior of metallic engineering materials concerning the effect of strain-temperature phasing (i.e., IP and OP) was proposed by Nitta and Kuwabara [3] and is illustrated in Fig. 1. In case of Type I behavior, creep damage dominates. Therefore, the combined effect of high temperature and tensile stresses during IP loading is more detrimental as compared to OP, where high temperature occurs in compression. Environmental effects may lead to an embrittlement of the surface or the surface-near area giving rise to an early crack initiation if high tensile loads coincide with low temperatures. Consequently, cyclic life under OP conditions may drastically be reduced (Type O behavior). Finally, if neither environmental effects nor creep damage dominate, IP and OP loading should roughly lead to similar life (Type E behavior). A special case of Type E results, if creep is restricted to high stress levels (Type E').

The cyclic deformation behavior of austenitic stainless steels has frequently been studied during the last decades. Numerous publications can be found with respect to creep, crack propagation and isothermal fatigue, while TMF was dealt with in few investigations only, despite the fact that this class of material is very susceptible to TMF due to a low thermal conductivity and a high thermal

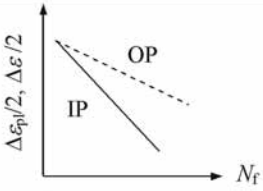
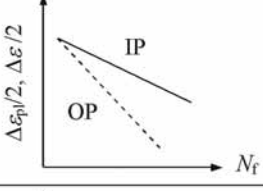
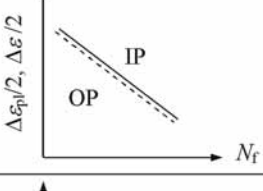
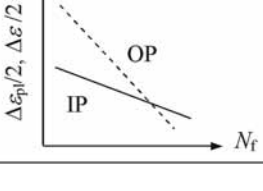
Behavior	Typ	predominant damage mode
	Typ I: IP-loading conditions cause earlier failure	creep damage
	Typ O: OP-loading conditions lead to shorter fatigue life	environmental effects
	Typ E: IP- and OP-loading conditions show nearly the same fatigue life	neither environmental effects nor creep damage dominate
	Typ E': IP- loading conditions lead to shorter fatigue life at high strain amplitudes	creep damage only at high loading amplitudes

FIG. 1—Classification of TMF life behaviour of engineering materials after Nitta and Kuwabara [3].

expansion coefficient. The intention of the study presented here was to summarize the experimental results, which were obtained in a series of doctoral studies carried out by Petry [4,5], Zauter [6,7] and Bauer [8,9] on AISI304L. On the basis of the extensive data base, a new concept for the calculation of the cyclic stress-strain behavior was developed and life prediction approaches were adapted to isothermal fatigue life data and expanded and applied to TMF conditions.

Material and Experimental Details

The as-received AISI304L (chemical composition in mass pct: Cr 18.3, Ni 9.3, Mn 1.82, and C 0.018) was solution annealed at 1100C for 1 h leading to an average grain diameter of about 100 μm. A subsequent aging for 1000 hrs at 600C aimed at a stable microstructure during cyclic loading. Specimens with a

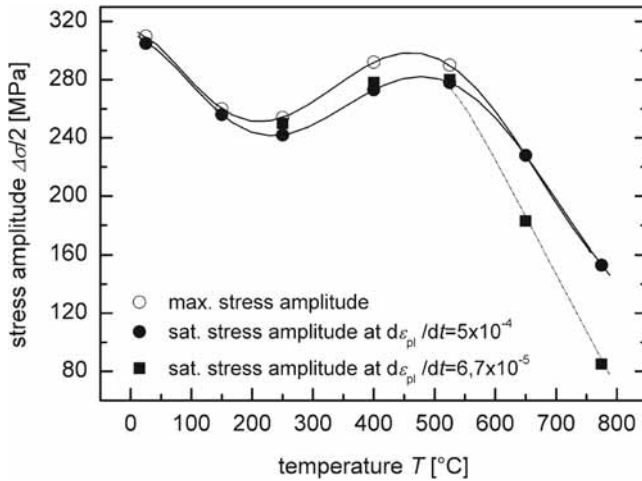


FIG. 2—Effect of temperature and plastic strain rate on the max and saturation stress amplitude at a plastic strain amplitude of $\Delta\epsilon_{pl}/2 = 0.5\%$ and two plastic strain rates.

cylindrical gauge length of 7 mm in diameter were tested in an electrohydraulic servocontrolled test system equipped with a vacuum chamber and a high frequency heating system. The isothermal fatigue tests were performed at seven different temperatures in the temperature regime between room temperature (RT) and 700°C both in air and vacuum. TMF tests were carried out in OP and IP mode using a temperature range in each test of 250°C within the temperature interval from 250°C to 800°C. True plastic strain control in combination with a triangular command signal was applied in all fatigue tests in such a way that the plastic strain amplitude was held constant at 0.5 pct and the (absolute) value of the plastic strain rate was adjusted to either $5 \cdot 10^{-4} \text{ s}^{-1}$ (mostly used for isothermal testing) or $6.7 \cdot 10^{-5} \text{ s}^{-1}$ (in the TMF tests). In additional tests, dwell times of three min each were superimposed both at maximum and minimum stress in each cycle.

Optical, scanning electron and transmission electron microscopy were applied, in order to characterize the dislocation arrangement, the relevant damage mechanisms and the corresponding damage evolution process as a function of testing mode and testing parameters used.

Experimental Results

Isothermal Conditions

The stress-strain response is independent of environment (laboratory air or vacuum) and hence results in similar values of the stress amplitude for both environments. Figure 2 shows the effect of temperature on the stress amplitude in cyclic saturation for both plastic strain rates used and on the maximum stress amplitude, which was observed before saturation for the higher plastic strain

rate. For all test temperatures above room temperature, a pronounced state of cyclic saturation was observed dominating the cyclic stress-strain response. This holds also true for the TMF conditions as will be shown later. The isothermal LCF behaviour of AISI304L is strongly affected by the occurrence of dynamic strain aging (DSA) in an intermediate temperature range. The influence of temperature on stress response can roughly be subdivided into three sections. At room temperature, fatigue life is affected by cyclic deformation-induced transformation of austenite into martensite [10] leading to a short crack initiation and propagation stage and a secondary hardening after a short state of saturation (not seen in Fig. 2). Between 250C and 525C, the stress amplitudes increase as a consequence of dynamic strain aging (DSA). DSA manifests itself in an inverse strain rate sensitivity of the stress response that is appreciable in this regime by comparing the saturation stress values of the two different plastic strain rates. At temperatures above 600C thermally activated dislocation motion gains significance. The creep contribution explains the further drop of the stress amplitude as well as the pronounced strain rate sensitivity of the saturation stress amplitudes.

DSA leads to a characteristic change of the saturation dislocation arrangement, as clearly seen in the TEM micrographs shown in Fig. 3. At low and high temperatures a cell (below DSA) and subgrain (above DSA) dislocation arrangement forms, respectively, whereas a disorderly planar-type arrangement was observed in the intermediate temperature range of DSA. Hence, the dislocation slip character changes with temperature from wavy to planar and back to wavy as a consequence of DSA.

Figure 4 documents the environmental effect on cyclic life. Surprisingly, the life reduction by air is rather strong despite the formation of a chromia scale on the surface. In both environments, the fatigue life N_f first increases strongly from the room temperature value and reaches a maximum at about 250C. This fatigue life extension can be attributed to the fact that secondary cyclic hardening as a consequence of martensitic transformation plays only a role at room temperature. Hence, a higher temperature leads to a smaller stress amplitude of cyclic saturation, which in combination with the constant plastic strain amplitude results in a life extension. As soon as the DSA temperature range is reached, N_f decreases with temperature, mainly because of the increase in stress amplitude. The crack initiation takes place at the surface often starting at oxidized grain boundaries. This finding explains the strong effect of environment. At 650C the crack propagation mode changes from mixed (trans-/intergranular) at the higher strain rate to intergranular at the lower strain rate. With further increasing temperature, intergranular crack propagation takes place at both strain rates [11].

Anisothermal Conditions

Figure 5 compares the stress response of two TMF tests performed in air under IP and OP conditions applying a temperature range of 400C–650C. The final parts of the corresponding tests performed in vacuum are also included in Fig. 5. The material shows a short period of cyclic hardening at the beginning

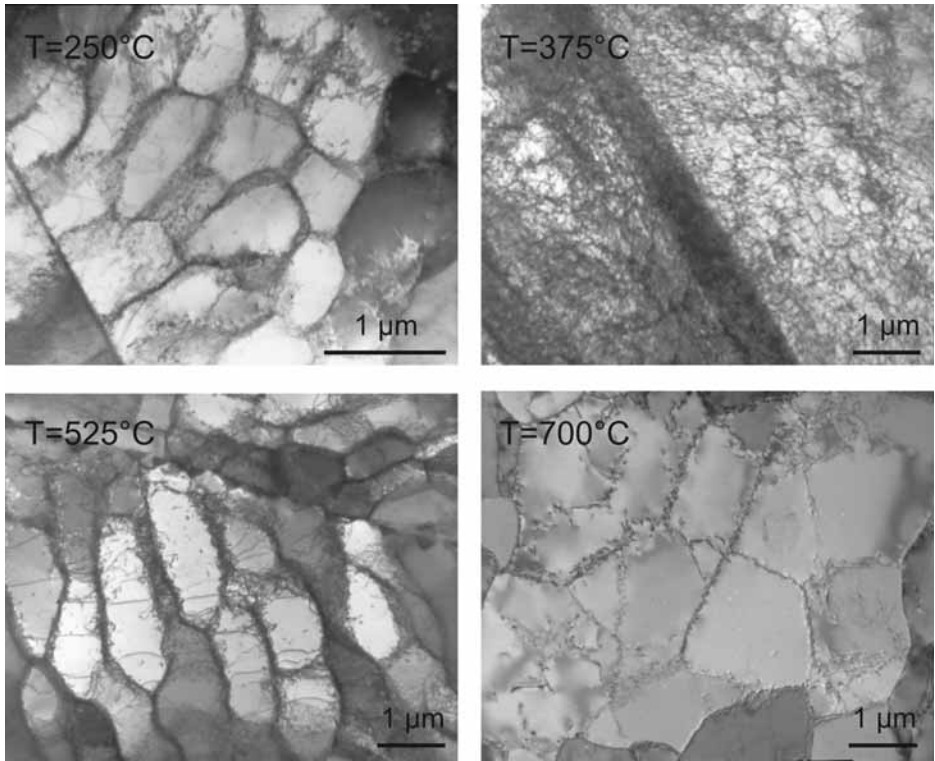


FIG. 3—Dislocation arrangements of isothermal fatigue tests at a plastic strain amplitude of $\Delta\epsilon_p/2 = 0.5\%$ and different temperatures.

before a very pronounced cyclic saturation state is established. Testing in vacuum significantly extends TMF life as compared to air. Since the flow stress at 400C is higher than the one at 650C, the IP testing mode leads to a negative (compressive) mean stress, while OP gives rise to a positive (tensile) mean stress.

TMF testing results in dislocation arrangements, which depend on the temperature applied, but are independent of phasing (OP or IP) and specific to TMF [12]. The permanent change in the dislocation mobility with temperature as a consequence of DSA hardens the material so that the stress response outside the DSA temperature range is higher as compared to isothermal conditions. Typical TMF-specific dislocation arrangements are shown in the TEM micrographs of Fig. 6.

The type of dislocation arrangement is dictated by the maximum temperature used in the TMF test. A schematic representation is shown in Fig. 7 mapping the various arrangements found experimentally by means of TEM investigations. If the maximum temperature $T_{\max}$ lies below a critical value, which is about 650C for the testing conditions used in this study, the impeded

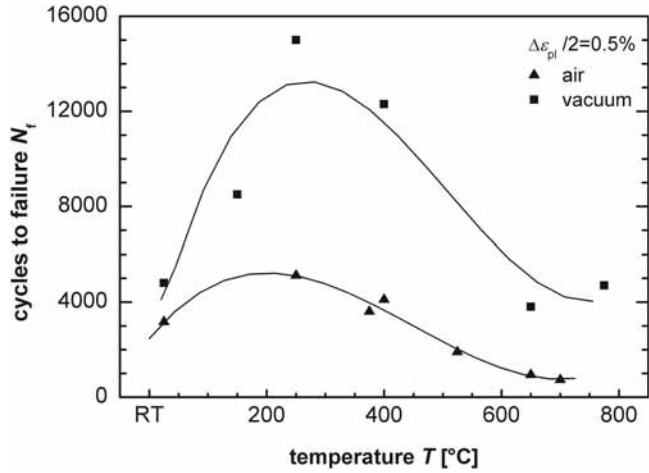


FIG. 4—Cyclic life in plastic strain controlled fatigue tests at a plastic strain amplitude of $\Delta\epsilon_{pl}/2=0.5\%$ and a plastic strain rate of $d\epsilon_{pl}/dt=5\times10^{-4}\text{ s}^{-1}$ as a function of temperatures in air and vacuum.

dislocation motion in the DSA temperature regime manifests itself in the formation of a layer/wall structure, labyrinth structure or dislocation cells (in the order of increasing values of T_{max}). The layer/wall arrangement seems to be a compromise between the planar arrangement of DSA and the wavy behavior at temperature outside the temperature regime of DSA. With increasing T_{max} the layer/wall and the labyrinth arrangement lose importance and a transition

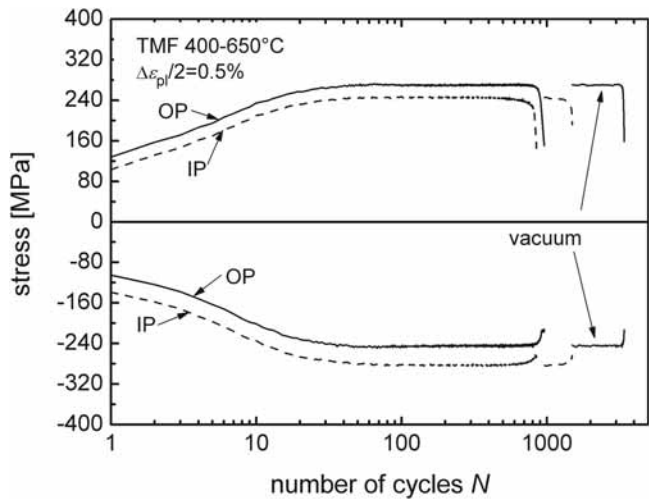


FIG. 5—Course of max and minimum stress per cycle for IP and OP TMF tests performed in the temperature range 400C–650C at $\Delta\epsilon_{pl}/2=0.5\%$ in air and vacuum.

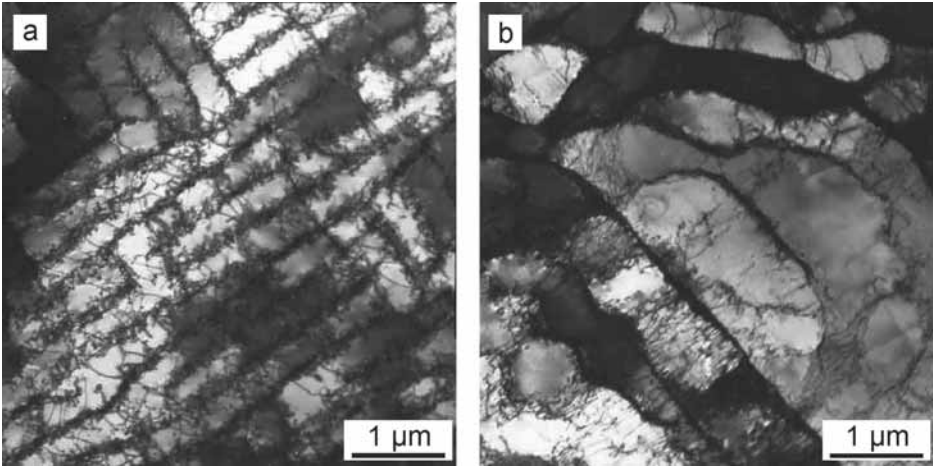


FIG. 6—Typical dislocation arrangements after TMF loading: (a) dislocation labyrinth formed in IP testing at 350C–600C and (b) transition to subgrains after IP loading at 400C–650C.

towards subgrains takes place. The subgrains are a consequence of the occurrence of thermally activated dislocation motion (climbing) which allows recovery processes. The subgrains, the boundaries of which are formed of dislocation networks, exhibit an elongated shape up to $T_{\max} < 700\text{C}$, since the walls are preferentially aligned parallel to the traces of $\{111\}$ planes. At higher values of $T_{\max}$ equiaxed subgrains develop.

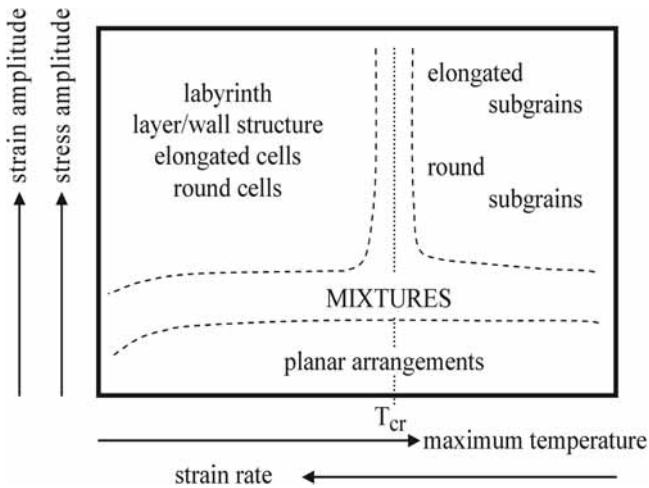


FIG. 7—Schematic representation of the correlation of TMF testing parameters and type of dislocation arrangement found after TMF testing.

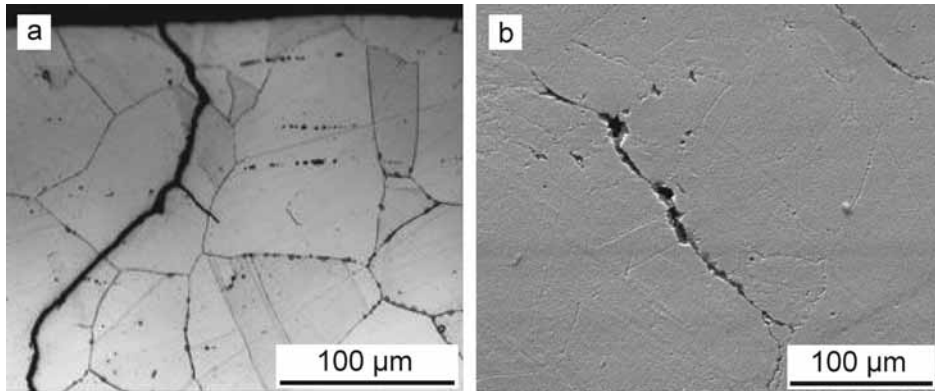


FIG. 8—Cracks found after TMF in air at $\Delta\epsilon_p/2 = 0.5\%$ and a plastic strain rate of $d\epsilon_p/dt = 6.7 \times 10^{-5} \text{ s}^{-1}$: (a) Transgranular crack propagation with crack initiation at a grain boundary at the surface (IP 250C–500C), and (b) crack formation along voids inside the material (IP 400C–650C).

Small strain amplitudes lead to small dislocation densities, and a planar dislocation arrangement prevails which appears to be rather independent of temperature. Bayerlein et al. [10] observed a planar arrangement of dislocations after room temperature LCF testing at a plastic strain amplitude of 0.02 pct. At 400C, DSA is responsible for a predominance of a planar glide mode. In TMF, the development of dislocation arrangements is strongly affected by DSA. Thus, at small plastic strain amplitudes, a planar dislocation arrangement is expected. Since cross slip is not taking place to a larger extent, the dislocations stay in their original slip planes.

In order to classify the TMF life data it is advisable to divide the behavior according to the maximum temperature of the TMF cycle applied into a low-temperature (LT) regime and a high-temperature (HT) regime. The boundary between LT and HT behavior is determined by the onset of creep damage and seems to be identical to the critical temperature T_{cr} shown in Fig. 7 (about 650C depending on the plastic strain rate used).

In the LT regime, IP and OP loading show very similar cyclic lives. This Type E behavior (see Fig. 1) holds true in air as well as in vacuum. However, life in vacuum is almost twice the life in air. As already mentioned above for the isothermal conditions, the reason seems to be mainly the premature crack initiation in air at grain boundaries at the surface, which is promoted by grain boundary oxidation (Fig. 8(a)).

In the HT temperature range, a strong difference between IP and OP loading exists. IP loading is much more detrimental than OP cycling, since creep damage exclusively develops under IP conditions because of the combination of high temperatures and tensile stresses favoring the formation of wedge-type cracks at grain boundary triple points and voids at grain boundaries (see Fig. 8(b)). The creep damage contribution is most pronounced in vacuum IP testing, since the surface crack initiation is delayed leading to a cyclic life extension by a

TABLE 1—Effect of environment (air and vacuum) on the TMF life and the TMF life type (according to Fig. 1).

	Air	Vacuum	$(N_f)_{\text{vac}}/(N_f)_{\text{air}}$
Low Temp. Regime	Type E	Type E	≈ 2
High Temp. Regime	Type E	Type I	1.5 (IP)...4 (OP)

factor of about 1.5. This factor is less than the one observed for IP testing in the LT range, because of the contribution of creep damage which is considered as unaffected by the environment. Consequently, the TMF behavior in the HT regime is of Type I (see Fig. 1) for testing in vacuum. However, oxidation in air reduces TMF life both for IP and OP conditions in such a strong way that creep damage is not developed significantly. Therefore; still Type E behavior prevails for air. The effect of environment is strongest for OP testing in air in the HT temperature range. These conditions exclude creep damage but promote oxidation because of the high maximum temperature. Life extension factors of up to four were obtained experimentally. The classification of the TMF behavior according to Ref. [3] and the life extension factor for vacuum testing are summarized in Table 1.

The experimental results document that the effect of environment must be included in TMF life prediction models. However, the oxidation mechanisms and their influence on cyclic life are obviously very similar in the isothermal and thermomechanical tests. Hence, a transfer of isothermally obtained life data to TMF conditions seems to be justified.

Modeling Cyclic Stress-Strain Behavior

It has been shown in an earlier work [13,14] that the so-called *pseudo-isothermal thermomechanical fatigue behavior* (PITMF behavior) can be calculated from the hysteresis loops of isothermal fatigue tests performed at different temperatures by applying a multi-component model originally proposed by Masing [15]. The basic assumption of this model is that a material consists of a parallel arrangement of elements which deform in an ideal elastic-plastic manner. Masing's model can be treated mathematically in a statistical manner (see Refs [16–18].). The distribution of the element-specific yield stresses σ_{if} of the elementary volumes is expressed by means of a yield stress distribution density function $f_p(\sigma_{if})$ which can be determined from a single branch of a hysteresis loop according to

$$f_p\left(\frac{\varepsilon_r E}{2}\right) = f_p(\sigma_{if}) = -\frac{2}{E^2} \frac{d^2 \sigma_r}{d\varepsilon_r^2} \quad (1)$$

σ_r and ε_r denote the relative stress and strain, respectively (i.e., relative to the last load reversal point), and E is the Young's modulus.

The stress-strain path can be calculated on the basis of the second integral function $G_p(\sigma_{if})$ of $f_p(\sigma_{if})$ according to

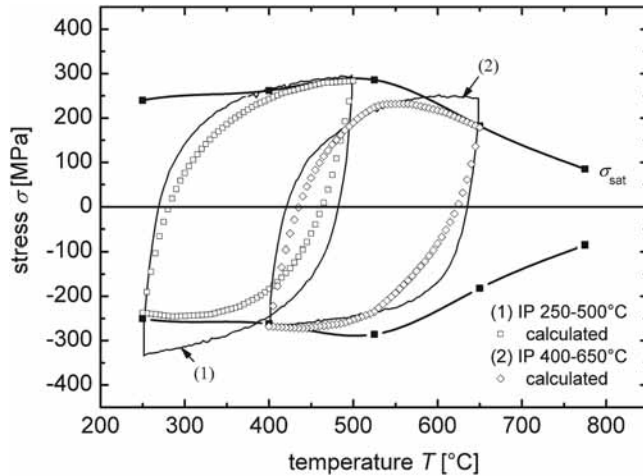


FIG. 9—Calculated and measured hysteresis loops of two IP TMF tests at $\Delta\epsilon_p/2 = 0.5\%$ compared to the isothermal saturation stress amplitude plotted as positive and negative values. (1) and (2) indicate the temperature regimes where the TMF behavior deviates significantly from the isothermal behavior.

$$\sigma_r = \epsilon_r E - 2G_p \left(\frac{\epsilon_r E}{2} \right) \quad (2)$$

In the case of anisothermal conditions, the effect of temperature on the yield stress distribution density function $f_p(\sigma_{if})$ must be included and the calculation according to Eq (2) is carried out in small strain and temperature increments. Pseudo-isothermal thermomechanical fatigue behavior means that the material is assumed to behave under varying temperature at each temperature as it does in the corresponding isothermal test.

Figure 9 shows the measured and the calculated (PITMF) hysteresis loops for two IP tests together with a representation of the saturation stress amplitudes of isothermal tests versus temperature. It can be seen that the peak and valley stresses of the calculated loops correspond very nicely to the saturation stress amplitude values σ_{sat} of the respective isothermal tests. However, outside the temperature range of DSA, the load reversal points of the experimentally observed loops deviate strongly from the σ_{sat} course. The microstructural reason for this behavior has been shown above by means of the results of the TEM examination of the dislocation arrangements formed. The impeded dislocation mobility in the DSA region strongly affects the establishing TMF dislocation arrangement and leads to higher strength both at higher and lower temperature as compared to the isothermal condition.

In order to improve the predictive capability of the composite model, additional tests were carried out with temperature changes from the saturation state within the DSA regime to a temperature above or below the DSA temperature range. Figure 10 shows the cyclic deformation curve of a test, during which the

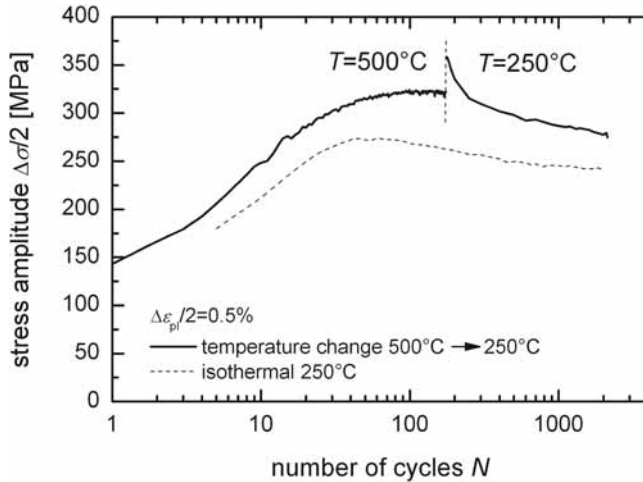


FIG. 10—Comparison of the stress response during a temperature change test (500C→250C) with the response at 250C from the beginning.

temperature was changed from 500C (i.e., from inside the DSA regime) to 250C (i.e., just below DSA), and compares the stress amplitude with the values from isothermal testing at 250C. The strengthening effect of predeformation in the DSA temperature range is clearly visible. This effect can be incorporated into the stress-strain response modeling by using the yield stress distribution density function $f_p(\sigma_{if})$ from hysteresis loops obtained in respective temperature change tests instead of those from loops of isothermal tests (without pre-cycling in the DSA regime).

Figure 11 compares the experimentally observed stress-strain data of cyclic saturation of an IP test applying a temperature range of 400C–650C with the predicted hysteresis loop and documents a reasonable agreement.

It should be noted that for higher maximum temperatures, low strain rates and superimposed dwell times, creep and stress relaxation have to be taken into account in the model. It is shown in Ref [8] that this can simply be done on the basis of Norton's power law of creep applying creep data from the literature. A comparison between the experimentally determined hysteresis loop at half life ($N_f/2$) of an IP TMF test carried out with a temperature interval of 525C–775C with the result of the plain pseudo-isothermal TMF calculation (dashed line) and the loop resulting from the expanded model (solid line) is shown in Fig. 12. The stresses obtained by the pseudo-isothermal calculation are too low at high temperatures (tension) as a consequence of the disregard of the DSA (pre-history) effect. Since the recovery of the microstructure at high temperature is also not considered, the calculated stresses are too high (in absolute values) in the low temperature part (compression). The expanded model considers the DSA influence and represents the recovery of the material at high temperature indirectly via the relaxation of the elementary yield stresses.

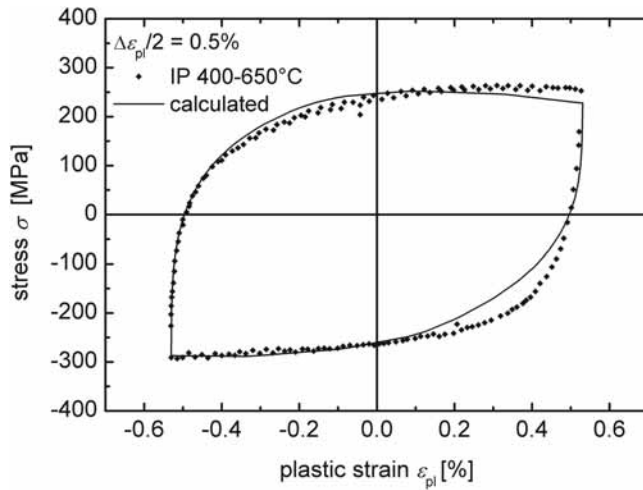


FIG. 11—Comparison of a measured hysteresis loop of an IP test at 400C–650C with a loop calculated by means of the expanded multi-component model that takes the DSA prehistory into account.

Hence, a much closer match of the real behavior can be expected, as shown in Fig. 12.

Modeling TMF Life

Three different approaches were used to predict TMF life of AISI304L solely from isothermal cyclic stress-strain and life data [8]. Here, a phenomenological energy-based approach [19] and a damage mechanics concept [9] are briefly introduced, while a fracture mechanics method is described in Ref [20].

A well-known damage parameter for high-temperature cyclic loading conditions is the *creep-fatigue damage parameter* D_{CF} proposed by Riedel [21]. This parameter can be considered to describe the deformation energy density per loading cycle, and therefore the respective life assessment approach belongs to the class of phenomenological, energy-based life prediction concepts. The definition of D_{CF} as used in this study is given in Eq (3)

$$D_{CF} = 2.9 \frac{(\Delta\sigma)^2}{2E} + 1.9\Delta\sigma(\Delta\varepsilon_{pl} + \Delta\varepsilon_{cr}) \left(1 + \frac{\Delta\varepsilon_{cr}}{\Delta\varepsilon_{pl}}\right)^{(1+n)} \quad (3)$$

Here, n is the cyclic hardening exponent, and $\Delta\sigma$, $\Delta\varepsilon_{pl}$ and $\Delta\varepsilon_{cr}$ are the ranges of stress, plastic strain and creep strain per cycle. $\Delta\varepsilon_{cr}$ can either be determined by means of additional creep tests or by comparing hysteresis loops observed at different plastic strain rates. According to the last-mentioned method, the creep strain range is determined as the difference of the total inelastic strain range observed at the plastic strain rate of interest and the plastic strain range

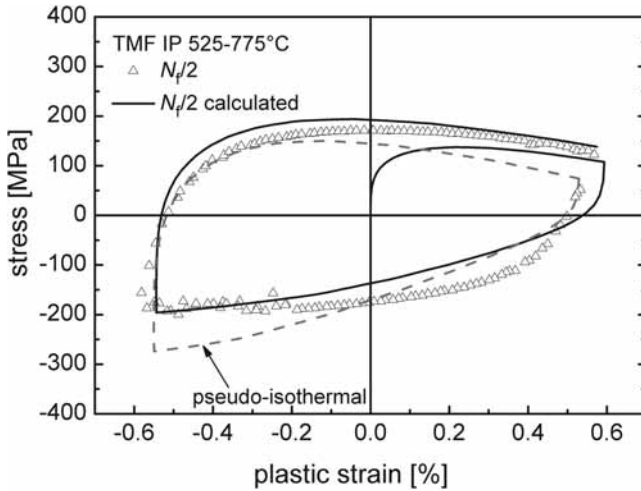


FIG. 12—Comparison of a measured hysteresis loop of an IP test at 525C–775C with two calculated loops. One loop represents the pseudo-isothermal TMF behaviour, while the other loop was calculated by means of the expanded multi-component model that takes the DSA prehistory and stress relaxation into account.

obtained during cycling at high strain rates (no creep contribution) applying the same stress range.

Figure 13 shows that the parameter D_{CF} correlates nicely with the isothermal fatigue life data. However, two separate lines exist for the two environments considered.

Clearly a simple power law describes the relation of D_{CF} with N_f in each case

$$N_f = L(D_{CF})^{-q} \quad (4)$$

The constants L and q can directly be taken from the lines shown in Fig. 13 and their values contain implicitly the effect of environment on fatigue life.

The so-called *loop reversion method* after Nitta et al. [3,22] was used to determine the creep strain range for the IP TMF cycles, while no creep contribution was assumed for OP cycling. Loop reversion means that the descending branch of the IP TMF hysteresis loop is mirrored, shifted to a common origin and compared with the ascending branch. The creep strain range is defined as the difference between the inelastic strain of the ascending branch at the stress of $T_{\max}$ and the plastic strain taken from the mirrored branch at the same stress (see Fig. 14).

TMF life was calculated using the constants L and q from the isothermal tests. Figure 15 compares the predicted number of cycles until failure with the experimentally determined ones. Taking the complexity of TMF loading and the simplicity of the model used into account, the agreement must be considered as very satisfactory. All data points lie within a band of a maximum deviation of a factor of two (dashed lines in Fig. 15).

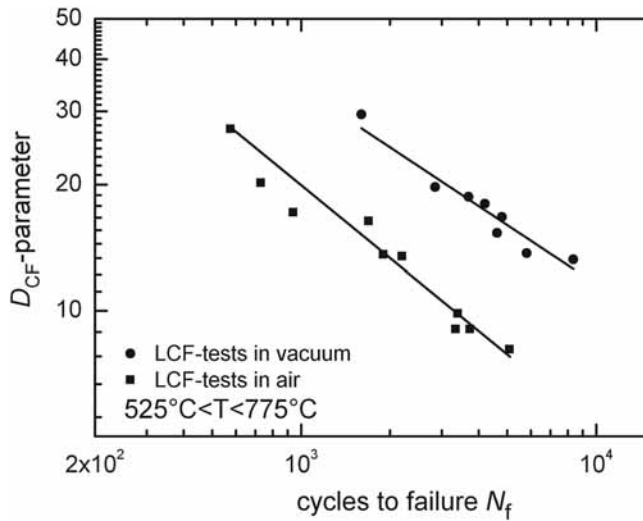


FIG. 13—Correlation between the number of cycles until failure of isothermal tests carried out in air and in vacuum and the corresponding value of the D_{cf} damage parameter.

The damage mechanics model used is based on an equation for the increase of the damage state variable ω proposed by Lemaitre and Chaboche [23]

$$d\omega = \left\{ \frac{(1-\omega)^{-\lambda}}{(\lambda+1) \cdot N_c} + \frac{[1 - (1-\omega)^{\beta+1}]^\alpha}{(\beta+1) \cdot N_f \cdot (1-\alpha) \cdot (1-\omega)^\beta} \right\} \cdot dN \quad (5)$$

The first term in braces represents the creep contribution to damage evolution, while the second term results from fatigue. λ denotes the creep damage tolerance factor and was determined from creep data according to

$$t_r = \frac{1}{\lambda+1} \cdot \frac{\sigma^{-n}}{A_0} \quad (6)$$

using the time until rupture t_r under creep conditions at creep stress σ and the two parameters stress exponent n and coefficient A_0 of Norton's creep law.

N_c represents the life in terms of the number of cycles to failure as a consequence of creep damage and can be calculated for cyclic loading from the cycle period t_{cyc} , the apparent activation energy of steady state creep Q , the stress exponent n and the coefficient A_0

$$\frac{1}{N_c} = (\lambda+1) \int_0^{t_{cyc}/2} \frac{\sigma(t)^n}{A_0} \cdot \exp\left(\frac{-Q}{RT}\right) dt \quad (7)$$

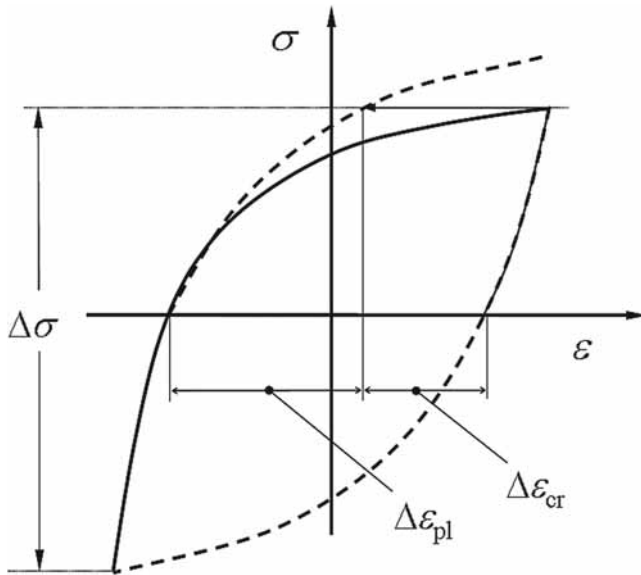


FIG. 14—Schematic representation of the "loop reversion method" used to determine the creep strain contribution of in-phase TMF loading.

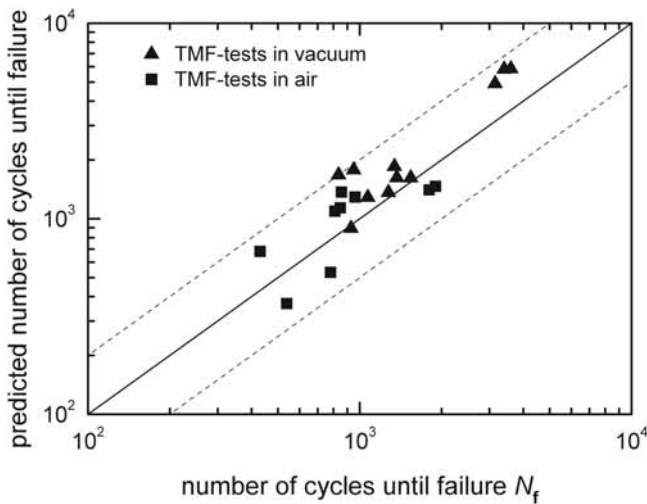


FIG. 15—Predicted TMF life calculated on the basis of the correlation between D_{cf} and N_f found for isothermal conditions (Fig. 13) in comparison to the experimentally observed TMF life data.

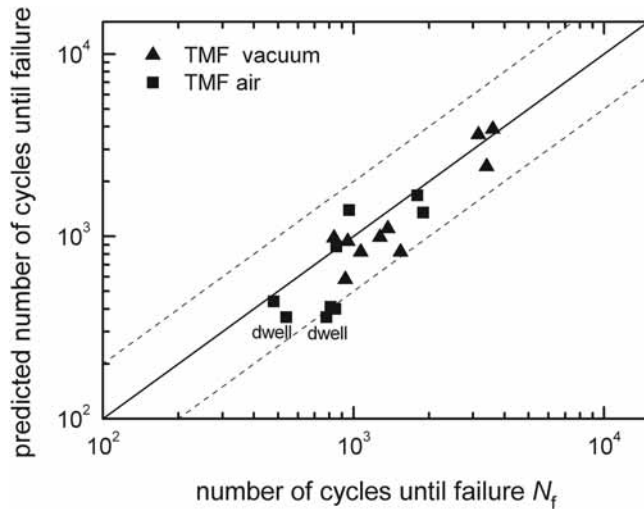


FIG. 16—Predicted TMF life calculated on the basis of the damage mechanics model according to Eqs (5)–(8) using solely isothermal results in comparison to the experimentally observed TMF life data.

The constants α and β of the fatigue term in Eq (5) were obtained from the isothermal fatigue life data, separately for air and vacuum [5]. Furthermore, the number of cycles to failure for plain fatigue N_f was calculated according to Ostergren [24]

$$N_f = L_1 \cdot (\Delta \varepsilon_{pl} \cdot \sigma_{max})^{-\eta} \cdot \nu^{1-k} \quad (8)$$

σ_{max} denotes the maximum stress and ν the test frequency. The constants L_1 , η and k were deduced from the isothermal fatigue lives, separately for the tests performed in vacuum and air. The values, which were obtained, are $L_1 = 1166$, $\eta = -1.86$, and $k = 0.76$ for vacuum [25] and $L_1 = 3577$, $\eta = -1.04$, and $k = 0.51$ for air [26].

Figure 16 compares the experimentally observed TMF life data with the predicted lives. A very reasonable agreement can be stated irrespectively of plastic-strain/temperature phasing (IP or OP) for both the vacuum tests and air tests. It is worth noting that also the effect of superimposed dwell times (see data points labeled “dwell”) leads to a conservative prediction.

Conclusions

The fatigue behavior of an austenitic stainless steel of type AISI304L under isothermal and thermomechanical conditions was investigated. The effect of creep and environment was characterized. For this purpose, tests were carried out both in vacuum and air, at different plastic strain rates and with and without

superimposed dwell times. In order to explain the observations, microstructural investigation were performed revealing the changes in the dislocation arrangement and the relevant damage contributions. Based on the experimental results the behavior under thermomechanical conditions was modeled. A simple multi-component model was used to describe the cyclic stress-strain response and an energy-based phenomenological approach and a damage mechanics model were applied to predict TMF life. The main results obtained can be summarized as follows:

- AISI304L shows a cyclic stress-strain behavior under TMF conditions that deviates characteristically from that observed under isothermal conditions. This phenomenon can be attributed to the occurrence of dynamic strain aging.
- The significance of environmental effects is moderate for the austenitic stainless steel studied. The main damage contribution through the interaction with the surrounding air as compared to vacuum is the preferential grain boundary oxidation leading to a premature crack initiation. This process is most important at high temperature and under conditions where the internally developing creep damage is negligible. Hence, OP loading with high maximum temperature promotes environmental effects.
- In principle, the influence of air is similar for isothermal and thermomechanical conditions so that isothermally obtained life data can successfully be applied to TMF life assessment.
- The multi-component model (i) helps to identify the TMF-specific behavior and (ii) can easily be expanded in such a way that these effects are considered. For this purpose additional fatigue tests applying temperature changes were carried out to capture the effect of cyclic deformation in the temperature regime of dynamic strain aging.
- For high maximum temperatures of the TMF cycle, superimposed dwell times or very low strain rates, creep and relaxation have to be taken into account as important contributions to the deformation of AISI304L. By means of a simple relaxation law based on Norton's power law creep equation the change of the yield strength resulting from relaxation was taken into account in each elementary volume leading to a very reasonable prediction of the stress-strain behavior.
- Both life prediction concepts used in this study were adapted to isothermally obtained fatigue life data and were found to provide a very satisfactory description. However, this adaption must be done separately for air and vacuum, in order to implicitly consider the environmental effect.
- The models were modified in such a way that the temperature variation of TMF is taken into account. This allows prediction of TMF life using solely isothermal test results.
- Both models show a very reasonable predictive capability for TMF conditions, if (i) the TMF-specific stress-strain behavior is considered by means of the expanded multi-component model and (ii) different model parameters are used for air and vacuum.

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R. A. Ainsworth,¹ D. W. Dean,² and P. J. Budden²

Creep Crack Growth Under Complex Loading

ABSTRACT: Fitness-for-service assessments of components operating at high temperature often require estimates of creep crack growth in service. Although methods for calculating creep crack growth are given in R5 and other codes, these methods are limited in terms of their range of application and can be over-conservative. This paper extends the current methods in a number of areas. First, a general expression is derived for relaxation of a reference stress for combined primary and secondary loading and significant in-service crack growth. Second, the reference stress history is used to develop an estimate of the transient creep crack tip parameter, $C(t)$, for elastic-plastic-creep material behaviour for combined loading. This both extends current methods to more complex loading and reduces some known conservatism in these current methods. The general expression for $C(t)$ then enables creep crack growth in service to be estimated using data collected from standard tests under mechanical loading. Finally, the paper presents some finite element calculations for defects in cylinders with welding residual stresses as validation for the new approaches.

KEYWORDS: creep, crack growth, residual stress, complex loading

Nomenclature

- a = defect size including any crack growth
- a_0 = initial defect size
- $\dot{a}$ = creep crack propagation rate
- A = constant of integration
- B = component dimension, specimen thickness
- $C(a/w)$ = elastic compliance
- $C' = dC/d(a/w)$

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¹ Faculty of Engineering and Physical Sciences, The University of Manchester, Manchester M13 9PL, UK (Corresponding author) e-mail: robert.ainsworth@manchester.ac.uk

² British Energy - Part of EDF Energy, Barnett Way, Barnwood, Gloucester GL4 3RS, UK.

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- $C^*, C(t)$ = creep crack growth parameters (noting C^* is evaluated for the primary loading only)
 D = constant in creep deformation law
 E = Young's modulus
 $f(L_r)$ = R6 failure assessment curve
 $F(a/w)$ = function used to calculate the primary reference stress
 $f' = df/d(a/w)$
 F = function used to calculate the total reference stress
 $F(t)$ = ratio of reference strain to initial reference strain
 $\dot{F}(t)$ = rate of change of $F(t)$
 $g(a/w)$ = function used to calculate the primary stress intensity factor
 $g' = dg/d(a/w)$
 $h(a/w)$ = function used to calculate the secondary stress intensity factor
 $h' = dh/d(a/w)$
 I_n = non-dimensional function of n
 J = elastic-plastic crack parameter
 J_0 = initial value of J
 K = stress intensity factor
 K^p = Stress intensity factor due to primary stresses
 K^s = Stress intensity factor due to secondary stresses
 L_r = load ratio used in R6
 m = limit load ratio
 $m' = dm/d(a/w)$
 M^p = primary resultant moment transverse to the crack
 M^s = secondary resultant moment transverse to the crack
 N = creep stress exponent
 N^p = primary resultant force transverse to the crack
 N^s = secondary resultant force transverse to the crack
 P = time exponent in primary creep law
 P = primary load (constant)
 P_L = limit load
 q = exponent in creep crack growth law
 Q = secondary load, which can relax
 $Q' = \partial Q / \partial (a/w)$
 $\dot{Q} = \partial Q / \partial t$
 R' = geometrical parameter
 t = time
 t_d = time for formation of defect in service
 $\dot{t}$ = dummy variable for time integrals
 V = plasticity correction for secondary stresses
 w = section width
 Z = elastic follow-up factor
 β = constant in plastic deformation power-law
 γ = scaling factor on K^s
 $\gamma_m(a)$ = factor accounting for relaxation of N^s due to the presence of a crack of size, a

- $\gamma_b(a)$ = Factor accounting for relaxation of M^s due to the presence of a crack of size, a
 $\gamma_h(a)$ = factor accounting for relaxation of σ_h^s due to the presence of a crack of size, a
 Δ^e = elastic displacement
 $\varepsilon, \varepsilon_{ij}$ = strain, strain tensor
 ε_c = creep strain
 ε_{ij}^0 = initial strain tensor
 $\tilde{\varepsilon}_{ij}(\theta, n)$ = non-dimensional strain field at crack tip
 ε_p = plastic strain
 ε_{ref} = reference strain
 $\varepsilon_{\text{ref}}^0$ = initial reference strain (total elastic-plastic strain at σ_{ref}^0)
 $\varepsilon_{\text{ref}}^e$ = elastic strain at reference stress
 $\varepsilon_{\text{ref}}^{e+p}$ = elastic-plastic strain at reference stress
 $\varepsilon_{\text{ref}}^{e+p+c}$ = elastic plus plastic plus creep strain at reference stress
 $\varepsilon_{\text{ref}}^c$ = creep strain accumulated under the total reference stress history
 $\varepsilon_{\text{ref}}^p$ = plastic strain at reference stress
 $\dot{\varepsilon}, \dot{\varepsilon}_{ij}$ = strain rate, strain rate tensor
 $\dot{\varepsilon}_c$ = creep strain rate
 $\dot{\varepsilon}_{c,\text{ref}}$ = creep strain rate at total reference stress
 $\dot{\varepsilon}_{c,\text{ref}}^p$ = creep strain rate at primary reference stress
 $\dot{\varepsilon}_{\text{ref}}$ = total strain rate at total reference stress
 μ = stress exponent in plasticity
 σ, σ_{ij} = stress, stress tensor
 σ_h^p = primary out-of-plane membrane stress
 σ_h^s = secondary out-of-plane membrane stress
 σ_{ij}^0 = initial stress tensor
 $\tilde{\sigma}_{ij}(\theta, n)$ = non-dimensional stress field at crack tip
 σ_{ref} = reference stress
 σ_{ref}^0 = initial value of reference stress
 σ_{ref}^p = primary load reference stress
 σ_y = yield stress
 $\dot{\sigma}_{\text{ref}}$ = rate of change of reference stress
 $\dot{\sigma}_{\text{ref}}$ = rate of change of $\dot{\sigma}_{\text{ref}}$
 τ = scaled time, $\tau = t^p$
 $\Phi(t)$ = non-dimensional function of time in $C(t)$ estimation formula
 $\dot{\Phi}(t)$ = rate of change of $\Phi(t)$
 (r, θ) = polar coordinates centred at the crack tip

Introduction

Methods for estimating creep crack growth in components operating at high temperature are contained in a number of fitness-for-service assessment procedures [1–5]. The most comprehensive advice is contained in R5 [1] which addresses combined primary and secondary loading, based on the results of [6]. In these approaches, the creep crack growth rate depends on the steady state

creep crack tip parameter C^* or its transient equivalent $C(t)$. A description of the background to these parameters is contained, for example, in Ref 7.

Estimates of the time-dependent relaxation of the transient parameter $C(t)$ to its steady state value C^* were first provided by Riedel and Rice [8] and Ohji et al. [9] for elastic-creep material behaviour in cracked components subjected to mechanical loading. Ohji and Kubo [10] then extended the result to elastic-plastic-creep materials with power-law plasticity and power-law creep, again for primary loaded components. Further extensions to more general material response and to combined primary and secondary loadings were then presented in Refs 6 and 11.

Despite these developments, there are limitations to the methods and the corresponding procedures in the fitness-for-service documents. First, the methods have been developed for essentially stationary cracks. However, some materials exhibit reduced ductility at low stresses and corresponding creep crack growth rates at low values of C^* in long-term tests which are higher than would be obtained from extrapolation of short-term creep crack growth data [12]. This may lead to estimates of significant creep crack growth in service. In addition, some welded components operating at high temperature have not been subjected to post-weld heat treatment (PWHT) or have been subjected to weld repair without PWHT. Thus, these components contain regions of high welding residual stress at start of life and significant creep crack growth is likely to be predicted for defects postulated in these regions. However, as these cracks grow out of the regions of high residual stress, the crack driving force reduces dramatically as does the corresponding creep crack growth rate. Therefore, to address these issues, the section "Relaxation of the Reference Stress for Combined Loading" of this paper presents a derivation for relaxation of a reference stress for combined primary and secondary loading and significant crack growth.

A second limitation of the current methods is that while elastic-plastic-creep material response is addressed for components subject to primary loading (using the background developments described above), the methods for combined primary and secondary loading are restricted to elastic-creep materials. Therefore, the section "Estimation of $C(t)$ For Elastic-Plastic-Creep Behaviour" of this paper uses the reference stress result of the section "Relaxation of the Reference Stress for Combined Loading" to derive an estimate of $C(t)$ for those more complete loadings for elastic-plastic-creep response. At the same time, a known conservatism introduced in Ref 6 is reduced.

The developments presented in the sections "Relaxation of the Reference Stress for Combined Loading" and "Estimation of $C(t)$ For Elastic-Plastic-Creep Behaviour" are essentially theoretical. In order to have confidence in the practical application of the methods, it is important to have supporting experimental and numerical validation. Experimental validation of the current fitness-for-service methods is contained, for example, in Ref 1. This is supplemented here by finite element validation of the new approaches in the section "Finite Element Validation."

Relaxation of the Reference Stress for Combined Loading

R5 [1] provides equations for estimating the relaxation of the reference stress with time for combined primary and secondary loading. However, the results

are restricted to stress relaxation arising solely from creep straining. Changes in reference stress arising from changes in crack length are assumed to be negligible. Some background information in R5 does cover the case of significant crack growth but restricts attention to pure displacement controlled loading and refers to Ref 6 for the background methodology for evaluating crack growth and associated stress relaxation for the more general case. The section “General Relaxation Equations” uses this background methodology to derive the general result for relaxation in the presence of creep straining and significant crack growth. The section “Explicit Advice for Situations Involving Significant Welding Residual Stresses” then gives specific advice for cases involving significant welding residual stresses.

General Relaxation Equations

It is assumed that the initial response is elastic so that the initial value of the reference stress, σ_{ref} , is

$$\sigma_{\text{ref}} = \sigma_{\text{ref}}^p (K^p + K^s) / K^p \quad (1)$$

although this assumption is relaxed in the section Estimation of $C(t)$ For Elastic-Plastic-Creep Behaviour to cover elastic-plastic behaviour on initial loading. In Eq 1, σ_{ref}^p is the reference stress for the primary loading and K^p, K^s are the stress intensity factors for the primary loading and the secondary loading, respectively. The primary load P is assumed fixed and if the corresponding limit load P_L is written as

$$P_L = m(a/w) B w \sigma_y \quad (2)$$

where:

$m(a/w)$ = non-dimensional function of the ratio of crack size, a , to section width, w ,

B = length (e.g., thickness for a test specimen), and

σ_y = yield strength used to define the limit load, then

$$\sigma_{\text{ref}}^p = \frac{P}{P_L} \sigma_y = \frac{P}{m(a/w) B w} \quad (3)$$

Then, it is possible to write

$$\sigma_{\text{ref}}^p = P f(a/w) \quad (4)$$

$$K^p = P g(a/w) \quad (5)$$

where:

$f(a/w)$ and $g(a/w)$ = limit load and stress intensity factor functions that depend on a/w and the constant geometry parameters w and B .

Note that any time dependence in Eqs 4 and 5 is via changes in crack size, a , alone. However, the secondary stresses may relax and it is assumed that the corresponding stress intensity factor can be written in a similar form to Eq 5 as

$$K^s = Q(t, a/w)h(a/w) \quad (6)$$

where:

Q = stress or load type term that can relax with both time, t , and crack extension, in general, and $h(a/w)$ is a stress intensity factor function.

It is assumed that Eq 1 not only defines the initial reference stress but also applies as relaxation of Q occurs. Then it is possible to differentiate Eq 1 to obtain the total relaxation rate. In the absence of crack extension, it is possible to define an elastic follow-up factor Z from R5, which gives the partial relaxation rate as

$$\frac{\partial \sigma_{\text{ref}}}{\partial t} = -E\dot{\epsilon}_c/Z \quad (7)$$

where:

E = elastic modulus and the creep strain rate $\dot{\epsilon}_c$ is either $\dot{\epsilon}_c = \dot{\epsilon}_c(\sigma_{\text{ref}}, \epsilon_c) - \dot{\epsilon}_c(\sigma_{\text{ref}}^p, \epsilon_c)$ or $\dot{\epsilon}_c = \dot{\epsilon}_c(\sigma_{\text{ref}}, \epsilon_c)$, depending on the definition of the elastic follow-up factor Z .

Combining the total relaxation rate, $\dot{\sigma}_{\text{ref}}$, obtained from differentiating Eq 1 with the partial relaxation rate, $\partial \sigma_{\text{ref}}/\partial t$ from Eq 7 gives the final relaxation equation

$$\frac{\dot{\sigma}_{\text{ref}}}{\sigma_{\text{ref}}} + \left(\frac{m'}{m}\right)\left(\frac{\dot{a}}{w}\right) + \frac{E\dot{\epsilon}_c}{Z\sigma_{\text{ref}}} - \frac{K^s \sigma_{\text{ref}}^p}{K^p \sigma_{\text{ref}}} \left(\frac{\dot{a}}{w}\right) \left[\frac{h'}{h} - \frac{g'}{g} + \frac{Q'}{Q}\right] = 0 \quad (8)$$

where:

$$\begin{aligned} m' &= [dm/d(a/w)], \\ g' &= [dg/d(a/w)], \\ h' &= [dh/d(a/w)], \text{ and} \\ Q' &= [\partial Q/\partial(a/w)]. \end{aligned}$$

Equation 8 may also be expressed in terms of the stress intensity factors K^p, K^s rather than the functions h, g , and the load quantity Q by using Eqs 2 and 4–6. The resulting equation, which is given below, may be more convenient for application purposes

$$\frac{\dot{\sigma}_{\text{ref}}}{\sigma_{\text{ref}}} + \left[\frac{K^s}{(K^p + K^s)} \left\{ \frac{\partial K^p/\partial(a/w)}{K^p} - \frac{\partial K^s/\partial(a/w)}{K^s} \right\} - \frac{\partial \sigma_{\text{ref}}^p/\partial(a/w)}{\sigma_{\text{ref}}^p} \right] \frac{\dot{a}}{w} + \frac{E\dot{\epsilon}_c}{Z\sigma_{\text{ref}}} = 0 \quad (9)$$

In the absence of significant crack growth, the basic relaxation equation in R5 is obtained by setting $\dot{a} = 0$ in either Eq 8 or Eq 9. Further details of the derivation of these equations are given in Appendix A, which also derives particular forms for pure primary loading and pure displacement controlled loading.

If Eq 9 is being used to predict relaxation of the reference stress, it is important to recognise that the secondary stresses and hence K^s will also relax (see Eq 6). From Eq 1, the current value of K^s is given by

$$K^s = \left(\frac{\sigma_{\text{ref}}}{\sigma_{\text{ref}}^p} - 1 \right) K^p \quad (10)$$

and hence $K^s \rightarrow 0$ as $\sigma_{\text{ref}} \rightarrow \sigma_{\text{ref}}^p$, i.e., as the reference stress approaches its primary load limit.

Explicit Advice for Situations Involving Significant Welding Residual Stresses

The use of Eq 1 to estimate the total reference stress, σ_{ref} , is appropriate for situations where stresses transverse to the crack are dominant and is most accurate for situations in which the transverse primary and secondary stress distributions are of similar form. Relaxation of the total reference stress is then estimated using Eq 8 or Eq 9. However, for situations involving significant welding residual stresses, for example, out-of-plane stresses are often higher than the in-plane stresses transverse to the crack and need to be considered in the estimation of σ_{ref} . For these cases, the use of Eq 1 can potentially underestimate the total reference stress and it is preferable to estimate the total reference stress at any time by using an appropriate reference stress solution in conjunction with the instantaneous values of the applied loadings or stresses. For example

$$\sigma_{\text{ref}} = F \left(N^p + \gamma_m(a)N^s, \quad M^p + \gamma_b(a)M^s, \quad \sigma_h^p + \gamma_h(a)\sigma_h^s, \quad a \right) \quad (11)$$

where:

N^p and M^p = primary transverse resultant force and moment, respectively,

σ_h^p = primary out-of-plane membrane stress, and

N^s , M^s , and σ_h^s = instantaneous values of the secondary transverse resultant force and moment and out-of-plane membrane stress, respectively, noting that the out-of-plane stresses are assumed to be constant through the thickness.

The factors $\gamma_m(a)$, $\gamma_b(a)$, and $\gamma_h(a)$ can be used to account for relaxation of the secondary loadings due to the presence of the crack (and are therefore dependent on the current crack size, a). This is illustrated later in the section "Finite Element Validation." The primary reference stress, σ_{ref}^p , is evaluated by removing the secondary stress terms from Eq 11. The functional form of F depends on the limit load solution used to derive the reference stress.

Estimation of $C(t)$ for Elastic-Plastic-Creep Behaviour

R5 presents expressions for the transient creep parameter $C(t)$ for both elastic response on initial loading and elastic-plastic response on initial loading. However, the latter case is restricted to steady creep loading, i.e., to pure primary loading. R5 generalises the solution to combined primary and secondary

loading only when initial loading leads to elastic response. Therefore, this section generalises the $C(t)$ expression to combined primary and secondary loading when initial loading leads to elastic-plastic response. The results presented in the section “Results for Stationary Cracks” are based on a stationary crack present at time $t = 0$; modifications for significant crack growth are discussed in the section “Estimation of $C(t)$ for Situations Involving Significant Crack Growth.” The approach here follows that in Refs 6 and 11 and assumes that plasticity and creep are described by power-law equations of the form $\varepsilon_p = \beta\sigma^\mu$ and $\dot{\varepsilon}_c = D\sigma^n$, with $\mu = n$. Although the derivation is complex, the main result can be written in a form that appears as a straightforward extension of the existing solutions in R5. Therefore only the main results are presented here, with algebraic details of the derivation contained in Appendix B.

Results for Stationary Cracks

For combined loading, a conservative estimate of $C(t)$ is given by Eq 74 as

$$\frac{C(t)}{C^*} = \left(\frac{\sigma_{\text{ref}}}{\sigma_{\text{ref}}^p} \right)^{n+1} \left[\frac{(\varepsilon_{\text{ref}}/\varepsilon_{\text{ref}}^0)^{n+1}}{(\varepsilon_{\text{ref}}/\varepsilon_{\text{ref}}^0)^{n+1} - (\sigma_{\text{ref}}^0/E\varepsilon_{\text{ref}}^0)} \right] \quad (12)$$

where:

σ_{ref}^p = reference stress for the primary loading,

σ_{ref} is the total reference stress as in the section “Relaxation of the Reference Stress for Combined Loading” above, and

ε_{ref} = total strain at σ_{ref} and time t .

In Eq 12, and throughout this paper, C^* is evaluated for the primary loading only. The initial value of the total reference stress is σ_{ref}^0 , defined by Eq 1 for elastic response on initial loading. In order to use Eq 12, it is necessary to estimate $\varepsilon_{\text{ref}}^0$, which is the total elastic-plastic strain corresponding to σ_{ref}^0 . For elastic behaviour on initial loading, this is simply σ_{ref}^0/E and Eq 12 then reduces to the result currently in R5. More generally, $\varepsilon_{\text{ref}}^0$ may be deduced from R6 [13] as described in the section “Estimate of Initial Strain $\varepsilon_{\text{ref}}^0$.” Equation 12 may also be expressed as

$$\frac{C(t)}{C^*} = \left(\frac{\sigma_{\text{ref}}\dot{\varepsilon}_{c,\text{ref}}}{\sigma_{\text{ref}}^p\dot{\varepsilon}_{c,\text{ref}}^p} \right) \left[\frac{(\varepsilon_{\text{ref}}/\varepsilon_{\text{ref}}^0)^{n+1}}{(\varepsilon_{\text{ref}}/\varepsilon_{\text{ref}}^0)^{n+1} - (\sigma_{\text{ref}}^0/E\varepsilon_{\text{ref}}^0)} \right] \quad (13)$$

where:

$\dot{\varepsilon}_{c,\text{ref}}$ and $\dot{\varepsilon}_{c,\text{ref}}^p$ = creep strain rates at σ_{ref} and σ_{ref}^p , respectively.

For elastic behaviour on initial loading, Eq 13 reduces to

$$\frac{C(t)}{C^*} = \left(\frac{\sigma_{\text{ref}}\dot{\varepsilon}_{c,\text{ref}}}{\sigma_{\text{ref}}^p\dot{\varepsilon}_{c,\text{ref}}^p} \right) \left[\frac{(E\varepsilon_{\text{ref}}/\sigma_{\text{ref}}^0)^{n+1}}{(E\varepsilon_{\text{ref}}/\sigma_{\text{ref}}^0)^{n+1} - 1} \right] \quad (14)$$

which is equivalent to the expression in R5.

The conservative estimate of $C(t)$ for combined loading given in Eq 12 was derived in Appendix B by taking the lower bound value of unity for a parameter $\Phi(t)$, which is defined as

$$\Phi(t) \equiv \frac{\dot{\epsilon}_{c,\text{ref}}^p - Z\dot{\sigma}_{\text{ref}}/E}{\dot{\epsilon}_{c,\text{ref}}^p - (Z-1)\dot{\sigma}_{\text{ref}}/E} = \frac{Z\dot{\epsilon}_{c,\text{ref}}}{(Z-1)\dot{\epsilon}_{c,\text{ref}} + \dot{\epsilon}_{c,\text{ref}}^p} \quad (15)$$

More accurate estimates of $C(t)$ are possible by allowing $\Phi(t)$ to vary with time. Appendix B details how the $C(t)$ estimation formula for combined loading can be refined in this way, leading to

$$\frac{C(t)}{C^*} = \left(\frac{\sigma_{\text{ref}}}{\sigma_{\text{ref}}^p} \right)^{n+1} \left[\frac{(\epsilon_{\text{ref}}/\epsilon_{\text{ref}}^0)^{n+1}}{\Phi(t)[(\epsilon_{\text{ref}}/\epsilon_{\text{ref}}^0)^{n+1} - 1] + (1 - \sigma_{\text{ref}}^0/E\epsilon_{\text{ref}}^0)} \right] \quad (16)$$

with $\Phi(t)$ given by Eq 15, noting that this results in $\Phi(t)$ lying between $Z/(Z-1)$ and unity. For secondary loading alone, with $Z > 1$, $\Phi(t) = Z/(Z-1)$ and, for primary loading alone, $\Phi(t) = 1$. Otherwise, $1 \leq \Phi(t) \leq Z/(Z-1)$ with $\Phi(t) \rightarrow 1$ at long times (when $\sigma_{\text{ref}} \approx \sigma_{\text{ref}}^p$). If $\Phi(t) = 1$, Eq 16 reduces to the conservative $C(t)$ estimate of Eq 12. Equation 16 may also be expressed as

$$\frac{C(t)}{C^*} = \left(\frac{\sigma_{\text{ref}}\dot{\epsilon}_{c,\text{ref}}}{\sigma_{\text{ref}}^p\dot{\epsilon}_{c,\text{ref}}^p} \right) \left[\frac{(\epsilon_{\text{ref}}/\epsilon_{\text{ref}}^0)^{n+1}}{\Phi(t)[(\epsilon_{\text{ref}}/\epsilon_{\text{ref}}^0)^{n+1} - 1] + (1 - \sigma_{\text{ref}}^0/E\epsilon_{\text{ref}}^0)} \right] \quad (17)$$

The detailed expressions for $C(t)$ of Eqs 16 and 17 are compared with $C(t)$ values from cracked body finite element analyses in the section “Finite Element Validation.” The more conservative estimates of $C(t)$ given by Eq 12 and 13 are, however, more likely to be used for practical applications

For elastic behaviour on initial loading, Eq 17 reduces to

$$\frac{C(t)}{C^*} = \left(\frac{\sigma_{\text{ref}}\dot{\epsilon}_{c,\text{ref}}}{\sigma_{\text{ref}}^p\dot{\epsilon}_{c,\text{ref}}^p} \right) \left[\frac{(\epsilon_{\text{ref}}/\epsilon_{\text{ref}}^0)^{n+1}}{\Phi(t)[(\epsilon_{\text{ref}}/\epsilon_{\text{ref}}^0)^{n+1} - 1]} \right] \quad (18)$$

which is equivalent to Eq 63 in Appendix B. For non-power law creep behaviour, a value of n in the above equations may be estimated as $n = q/(1-q)$, where q is the exponent in the creep crack propagation law, $\dot{a} \propto (C^*)^q$. Alternatively, for monotonic creep data expressed as $\dot{\epsilon}_c(\sigma, t)$, which is a product of separable functions of stress and time, n may be estimated as $\partial(\ln \dot{\epsilon}_c)/\partial(\ln \sigma)$ at the reference stress level at time t . For pure primary loading $\Phi(t) = 1$, and Eqs 12 and 16 can be written as

$$\frac{C(t)}{C^*} = \frac{(\epsilon_{\text{ref}}^{e+p+c}/\epsilon_{\text{ref}}^{e+p})^{n+1}}{(\epsilon_{\text{ref}}^{e+p+c}/\epsilon_{\text{ref}}^{e+p})^{n+1} - (\epsilon_{\text{ref}}^e/\epsilon_{\text{ref}}^{e+p})} \quad (19)$$

where superscripts e , $e+p$, and $e+p+c$ denote elastic, elastic-plastic, and elastic-plastic plus creep, respectively. This reduces to the equation in R5 [1] in

the absence of plasticity and is an alternative expression to the equation in R5, using strain rather than time, when plasticity occurs on initial loading.

It is clear from Eq 19 that relaxation of $C(t)$ towards C^* is controlled by the accumulated total (elastic + plastic + creep) strain at the reference stress. When the total strain is significantly larger than the initial strain, $\epsilon_{\text{ref}}^{e+p+c} \gg \epsilon_{\text{ref}}^{e+p}$, relaxation is essentially complete and $C(t) \approx C^*$. In addition, $C(t) \approx C^*$ when the initial elastic-plastic strain is significantly larger than the initial elastic strain, $\epsilon_{\text{ref}}^{e+p} \gg \epsilon_{\text{ref}}^e$.

Estimation of $C(t)$ for Situations Involving Significant Crack Growth

The $C(t)$ estimation formulae given in the section “Results for Stationary Cracks” have been derived for stationary cracks and care must be taken in applying these equations in situations where the total reference stress can change significantly as a result of increased crack size. In order to correctly apply Eqs 12–14 or Eqs 16–18 in these situations, it is necessary to evaluate the initial total reference stress, σ_{ref}^0 , and the initial elastic-plastic reference strain, ϵ_{ref}^0 . For situations where σ_{ref}^0 reduces with crack size, σ_{ref}^0 and ϵ_{ref}^0 should be based on the initial loading at $t = 0$ but at the current crack size, $a(t)$. For situations where σ_{ref}^0 increases with crack size, it is recommended that calculations should be carried out for:

- σ_{ref}^0 and ϵ_{ref}^0 based on the initial loading at $t = 0$ and the current crack size, $a(t)$; and
- σ_{ref}^0 and ϵ_{ref}^0 based on the initial loading at $t = 0$ and the initial crack size, a_0 .

The predicted crack growth should then be taken as the larger of the values obtained for assumption (a) or (b) above. In both cases, the total reference strain, ϵ_{ref} , used to estimate $C(t)$, is evaluated as the sum of the elastic-plastic reference strain, $\epsilon_{\text{ref}}^{e+p}$, and the reference creep strain accumulated over the total reference stress history, ϵ_{ref}^c . Thus

$$\epsilon_{\text{ref}} = \epsilon_{\text{ref}}^{e+p} + \epsilon_{\text{ref}}^c \quad (20)$$

The elastic-plastic reference strain is evaluated, noting that plastic strain is irrecoverable during stress relaxation, by adjusting the initial elastic-plastic reference strain, ϵ_{ref}^0 , to account for the reduced elastic reference strain at the current reference stress as follows:

$$\epsilon_{\text{ref}}^{e+p} = \epsilon_{\text{ref}}^0 - (\sigma_{\text{ref}}^0 - \sigma_{\text{ref}})/E \quad (21)$$

The reference creep strain is given by

$$\epsilon_{\text{ref}}^c = \int_0^t \dot{\epsilon}_{c,\text{ref}}[\sigma_{\text{ref}}(t'), t'] dt' \quad (22)$$

where:

$\dot{\epsilon}_{c,\text{ref}}[\sigma_{\text{ref}}(t'), t']$ = creep strain rate at the total reference stress, $\sigma_{\text{ref}}(t')$, and time t' .

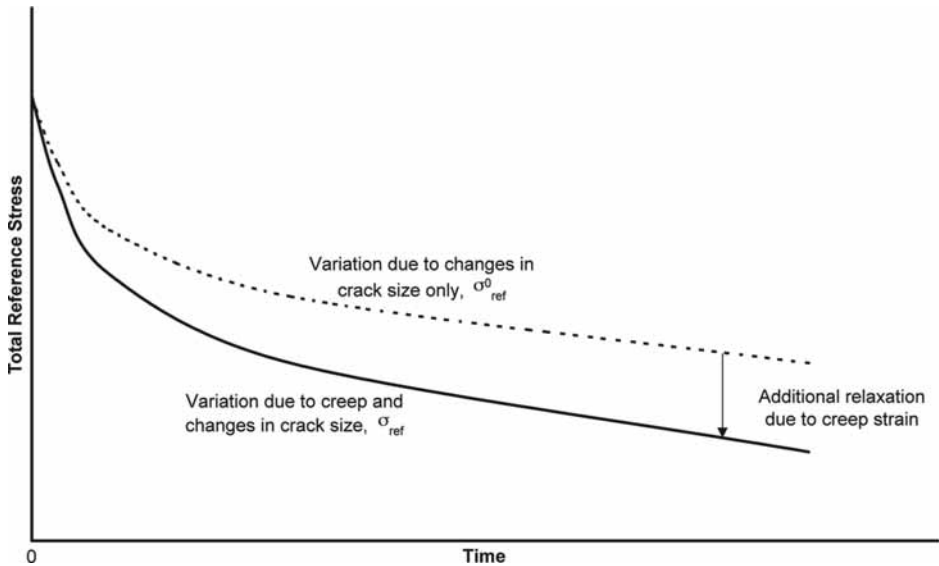


FIG. 1—Schematic variation of reference stress for a defect present at $t=0$.

For elastic behaviour on initial loading, $\epsilon_{\text{ref}}^0 = \sigma_{\text{ref}}^0/E$. Equations 20 and 22 are based on a defect present at the start of high temperature service at $t = 0$. Figure 1 shows schematically the variation of the total reference stress, σ_{ref} , with time as a result of changes in crack size and relaxation due to creep in this case. This figure also schematically shows the variation in σ_{ref}^0 with crack size for the case where σ_{ref}^0 reduces with increasing crack size, noting that the apparent sensitivity to time is simply due to the crack size increasing with time due to crack growth. However, this approach can also be applied to cases where the initial total reference stress increases with increasing crack size.

The advice given above can be extended to service-initiated defects with combined loading. Figure 2 shows schematically the variation of the total reference stress, σ_{ref} , with time for a service-initiated defect for the case where σ_{ref}^0 reduces with increasing crack size. The defect, of initial size a_0 , is assumed to form after high temperature service for a time t_d ; for times $0 \leq t < t_d$, it is assumed that the component is defect-free. Prior to commencing the creep crack growth calculations for the service-initiated defect, account can be taken of relaxation of the secondary stresses in the uncracked body in the period $0 \leq t < t_d$.

Finite Element Validation

Initial validation of the extended advice for combined loading presented in this paper involved comparing predictions of $J(t)$ and $C(t)$ (see Appendix B) with results from existing elastic-plastic-creep finite element analyses [14]. This showed that the approach generally leads to conservative predictions of $C(t)$,

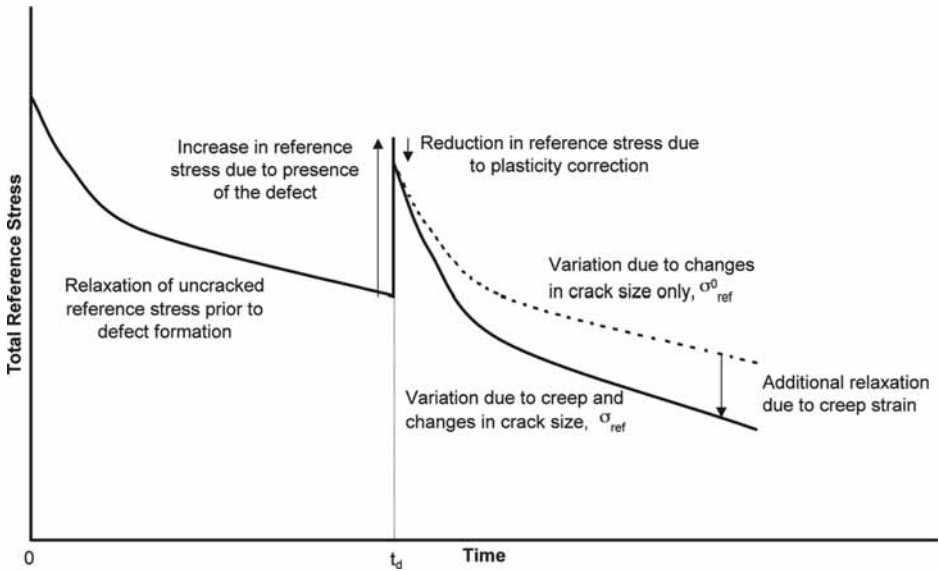


FIG. 2—Schematic variation of reference stress for a service-initiated defect formed at $t = t_d$.

although predictions are sensitive to the initial value of the total reference stress and the magnitude of the elastic follow-up factor, Z , for the cracked structure. Initial total reference stress estimates obtained using Eq 78 resulted in good predictions of the crack tip parameters $J = J(t)$ and $C(t)$ for an externally circumferentially cracked cylinder subjected to residual stress loading provided an appropriate value of Z was assumed. Conservative predictions of $J(t)$ and $C(t)$ were obtained for combined residual stress and mechanical loading, with conservatism increasing as plastic hardening capability increased. It was found that appropriate values of Z for the cracked geometries considered could be inferred by choosing values to optimise $J(t)$ estimates and that these values resulted in reliable predictions of $C(t)$.

Additional work examining the estimation of Z for cracked bodies [15] has shown that Z values can depend on the crack size, specimen geometry and creep stress exponent for structures subjected to displacement controlled loading. Displacement controlled loading situations allow elastic follow-up factors for cracked structures to be evaluated explicitly using the time variation of reference stress (i.e., based on the time variation of the reaction load and an appropriate collapse solution) in addition to inferring Z values by optimising $J(t)$ predictions. Although values of cracked body Z obtained using these two approaches differ in magnitude, in both cases follow-up factors increase with increasing crack size and with increases in the length scale over which the displacement is held constant. It was found that values of cracked body Z inferred by optimising $J(t)$ predictions led to more reliable $C(t)$ predictions than values obtained using the time variation of reference stress.

More recent validation studies [16] have examined a practically relevant case involving an externally circumferentially cracked cylinder subjected to either residual stress loading alone or combined residual stress and axial primary load. The residual stresses were taken from a finite element simulation of a girth weld in the cylinder and these were mapped onto an uncracked model into which cracks of various depths were introduced to allow $J(t)$ and $C(t)$ to be evaluated. If values of cracked body Z are inferred by optimising $J(t)$ predictions, this results in very conservative predictions of $C(t)$ when the initial total reference stress is estimated using Eq 1. However, if Eq 78 is used to estimate the initial value of total reference stress, $C(t)$ predictions are generally accurate although slightly non-conservative at very short times. In contrast to the work on displacement controlled loading [15], which showed that follow-up factors increase with increasing crack size, the residual stress dominated cases examined in Ref 16 gave Z values which generally reduced with increasing crack size. This suggests that the presence of a crack in a residual stress field does not lead to increased follow-up and that for these cases it may not be necessary to increase the uncracked values of Z by unity, as suggested in R5.

An approach that can be used to derive cracked body Z values for residual stress dominated situations is to average the uncracked values over the crack depth. This approach is illustrated in Fig. 3, which shows the variation of uncracked Z through the section for residual stress loading and the average Z values derived for crack size to section width, a/w , ratios of 0.1, 0.2, 0.3, and 0.4. The values derived for $a/w = 0.2, 0.3$, and 0.4 agree well with the values obtained by optimising $J(t)$ predictions [16]. However, the value derived for $a/w = 0.1$ is slightly lower than the value inferred from $J(t)$, which may be a result of the local peak in the uncracked Z value in the vicinity of the crack tip for this shallow defect.

The explicit advice on the treatment of combined loading for situations involving significant welding residual stresses (in the section "Explicit Advice for Situations Involving Significant Welding Residual Stresses") was first applied to the assessment of postulated extended reheat cracks in a plant weld [17]. This application used a reference stress solution for an edge cracked plate subjected to both in-plane and out-of-plane stresses together with secondary stress off-loading factors, $\gamma_m(a)$ and $\gamma_b(a)$, derived as the ratio between displacement and load controlled compliance factors for an edge cracked plate [18]. A cracked body $Z = 4$ was used in the assessments and it was pessimistically assumed that there was no off-loading of the out-of-plane secondary stresses due to the presence of the crack (i.e., $\gamma_h(a) = 1.0$). A similar approach is applied here to predict $C(t)$ values for the residual stress only cases from [16]. In order to predict the reference stress using the edge cracked plate solution for both in-plane and out-of-plane stresses, it is necessary to linearise the in-plane (axial) and out-of-plane (hoop) stresses. Figure 4 shows that linearising the axial stresses as a through-wall bending stress gives a reasonable or conservative idealisation of the stresses in the region of the defect locations on the outer surface of the cylinder. However, Fig. 5 shows that basing the hoop membrane stress on the value calculated over the full section results in a value of ~ 100 MPa, which is clearly not representative of the hoop stress levels in the

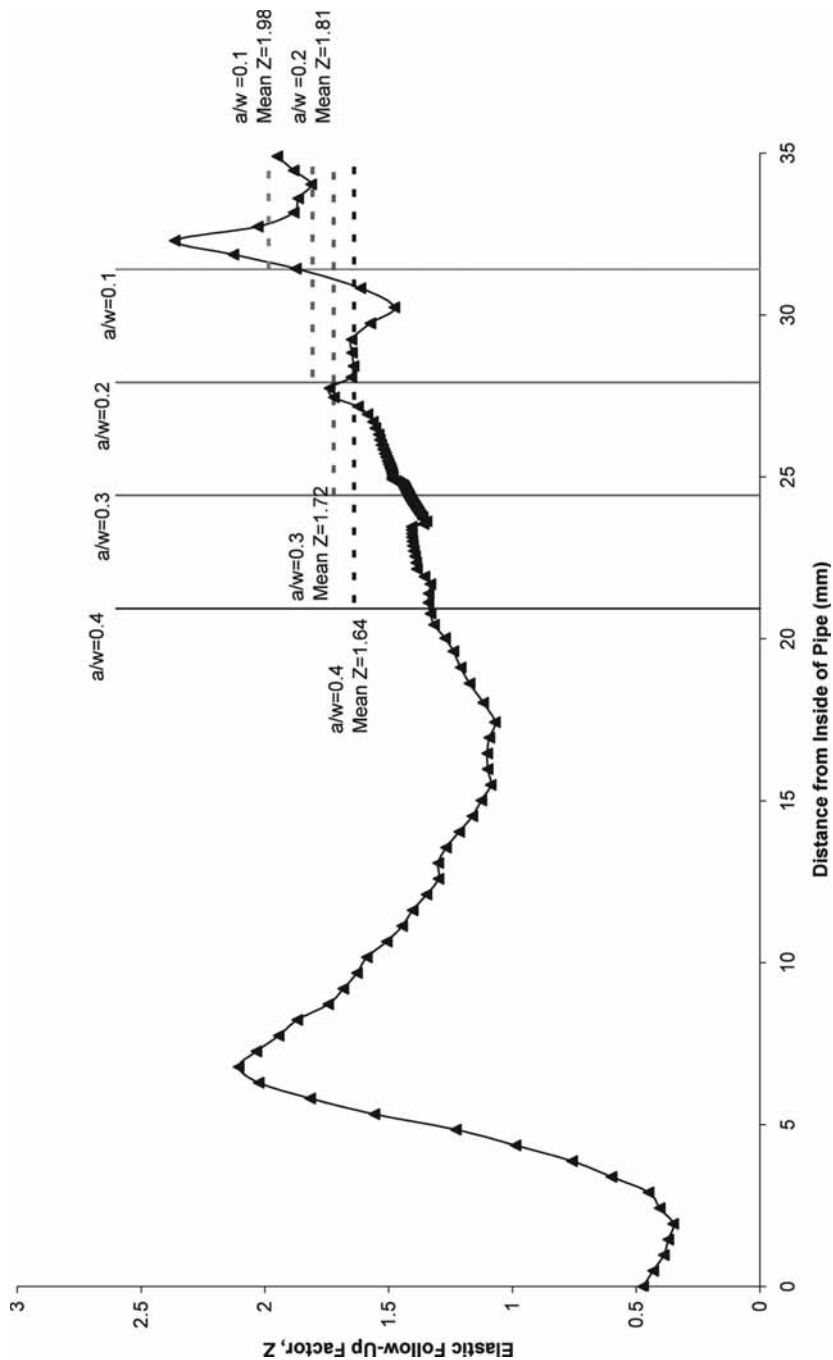


FIG. 3—Variation of uncracked body elastic follow-up factor, Z , through the section for a 35 mm thick cylinder subjected to welding residual stresses. Reference 16 presented $C(t)$ and J calculations for this geometry from cracked body finite element analyses.

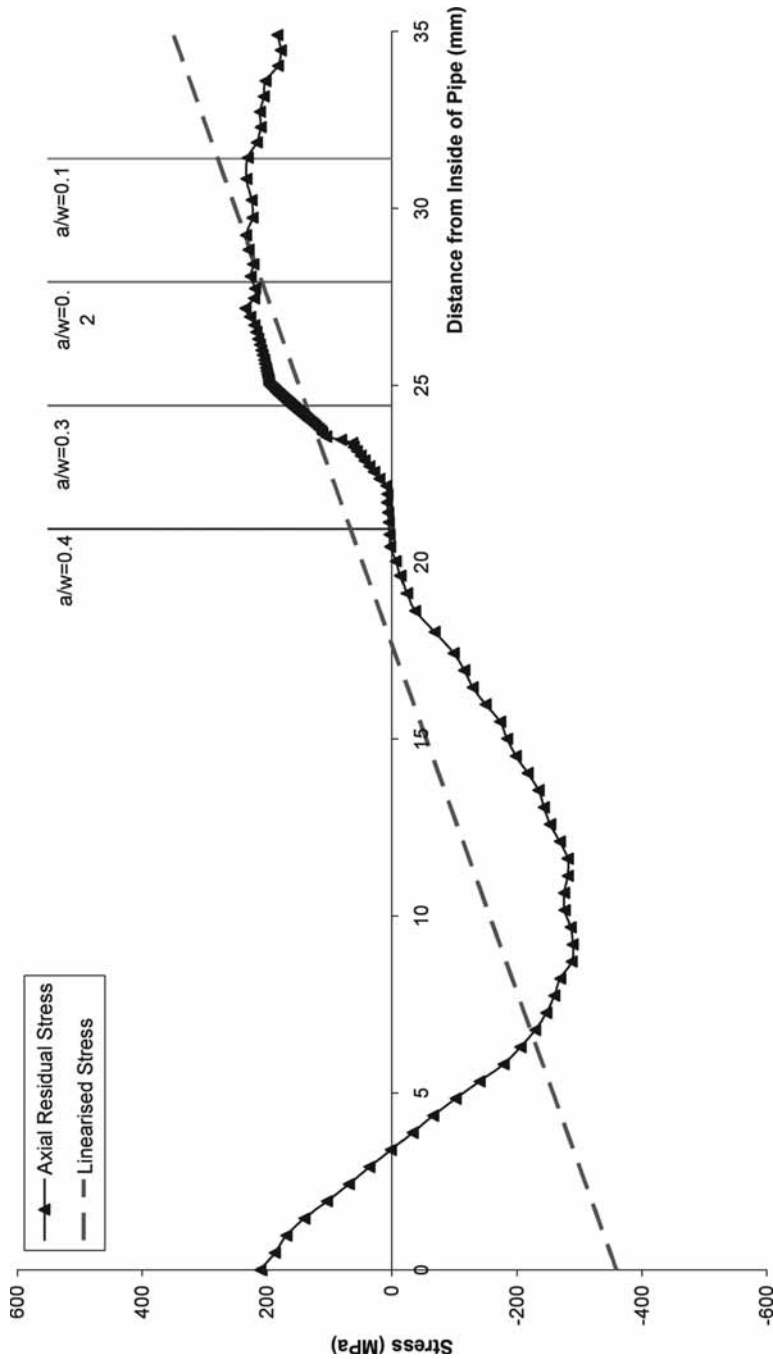


FIG. 4—Variation of axial residual stress through the section for a 35 mm thick cylinder subjected to welding residual stresses. Reference 16 presented $C(t)$ and J calculations for this geometry from cracked body finite element analyses.

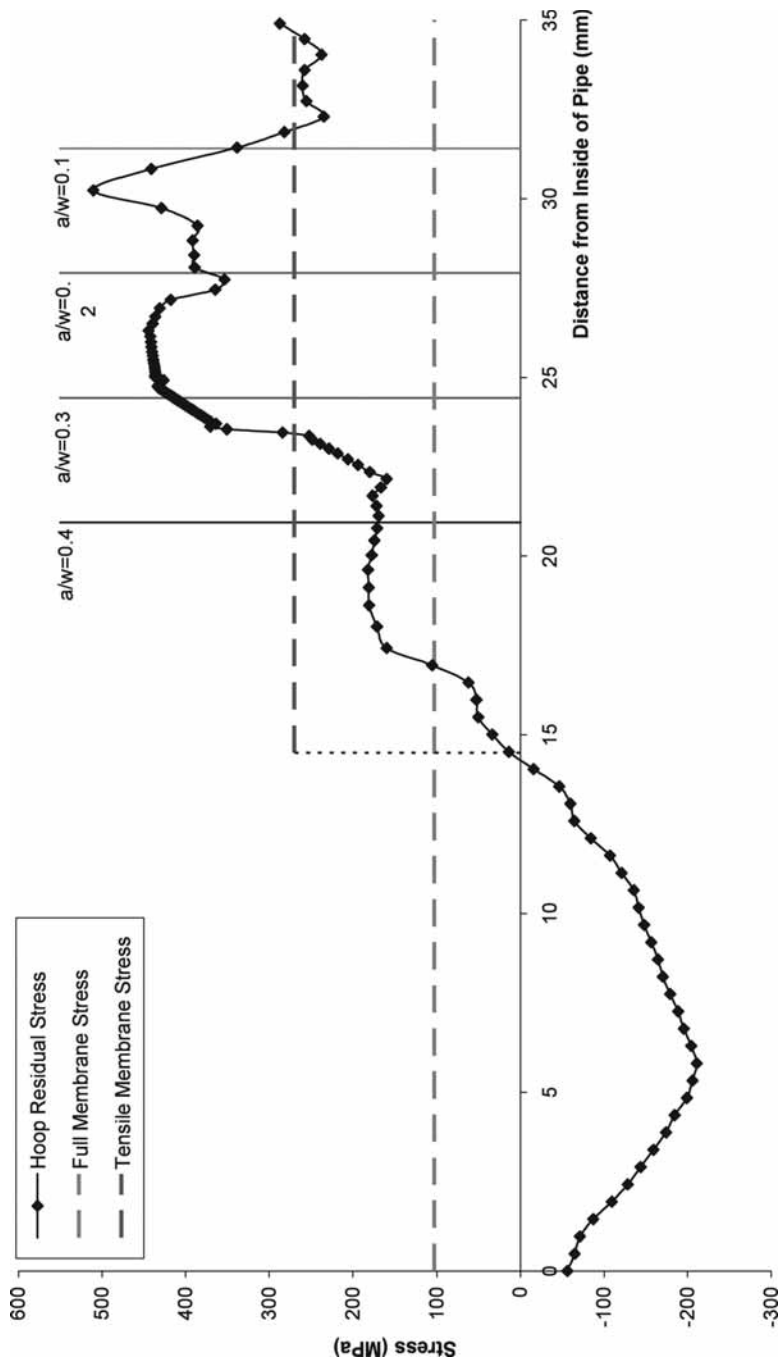


FIG. 5—Variation of hoop residual stress through the section for a 35 mm thick cylinder subjected to welding residual stresses. Reference 16 presented $C(t)$ and J calculations for this geometry from cracked body finite element analyses.

region of the defect locations on the outer surface of the cylinder. For this reason, the hoop membrane stress for this situation has been evaluated by integrating the hoop stress over the tensile outer wall region; as shown in Fig. 5, this leads to a more reasonable hoop membrane stress of ~ 270 MPa. These idealised stresses were then used in conjunction with the elastic follow-up factors from Fig. 3 to predict $C(t)$. As shown in Fig. 6, the $C(t)$ predictions for the residual stress only cases from [16] are conservative for both the two extreme defect sizes (corresponding to $a/w = 0.1$ and $a/w = 0.4$) and similar results were also obtained for intermediate defect sizes. For these cases, there are negligible differences between $C(t)$ predictions obtained using Eq 12, which is based on $\Phi(t) = 1$, and Eq 17, which allows $\Phi(t)$ to vary with time according to Eq 15.

A more recent application of the advice for situations involving significant welding residual stresses involved the assessment of postulated semi-elliptical manufacturing defects in an austenitic pipe butt weld containing a weld repair [19]. It was first demonstrated [20] that it is conservative to use thin cylinder limit load solutions [21] to estimate the total reference stress for situations where the radial stress is positive. As secondary stress off-loading factors were not readily available for semi-elliptical defects, it was pessimistically assumed that $\gamma_m(a) = \gamma_b(a) = \gamma_h(a) = 1.0$. Z values were required for crack depths in the range 5–20 mm and these were derived by averaging the uncracked Z values over crack depths of 5 mm ($Z = 2.00$) and 20 mm ($Z = 2.35$) and then taking the average of these values ($Z = 2.18$) for all crack depths in the range 5–20 mm as shown in Fig. 7. $C(t)$ predictions for stationary 5 mm and 20 mm deep semi-elliptical defects were then compared with values obtained from 3D cracked body finite element analyses of the cracked structure subjected to combined residual stress and mechanical loading [22]. As shown in Fig. 8, $C(t)$ values for the deepest points of both the 5 mm and 20 mm defects are conservatively predicted by Eq 13, which is based on $\Phi(t) = 1$. The use of Eq 17, which allows $\Phi(t)$ to vary with time according to Eq 15, results in more accurate predictions of $C(t)$, although these become slightly non-conservative at longer times, particularly for the shallower 5 mm crack.

Discussion

It is clear from the section “Finite Element Validation” that there has been a significant amount of work carried out aimed at validating the extended advice for combined loading given in this paper. This has primarily focussed on comparing $C(t)$ predictions made using these methods with results from cracked body finite element analyses and generally supports the extended advice outlined above. The new explicit advice for situations involving significant welding residual stresses represents an improvement for these cases, for which out-of-plane stresses are significant and are often higher than the in-plane stresses transverse to the crack. However, there are a number of areas where further refinements of the existing advice are needed:

- A more robust procedure is needed for evaluating the appropriate value of elastic follow-up factor, Z , for the cracked structure. The advice to

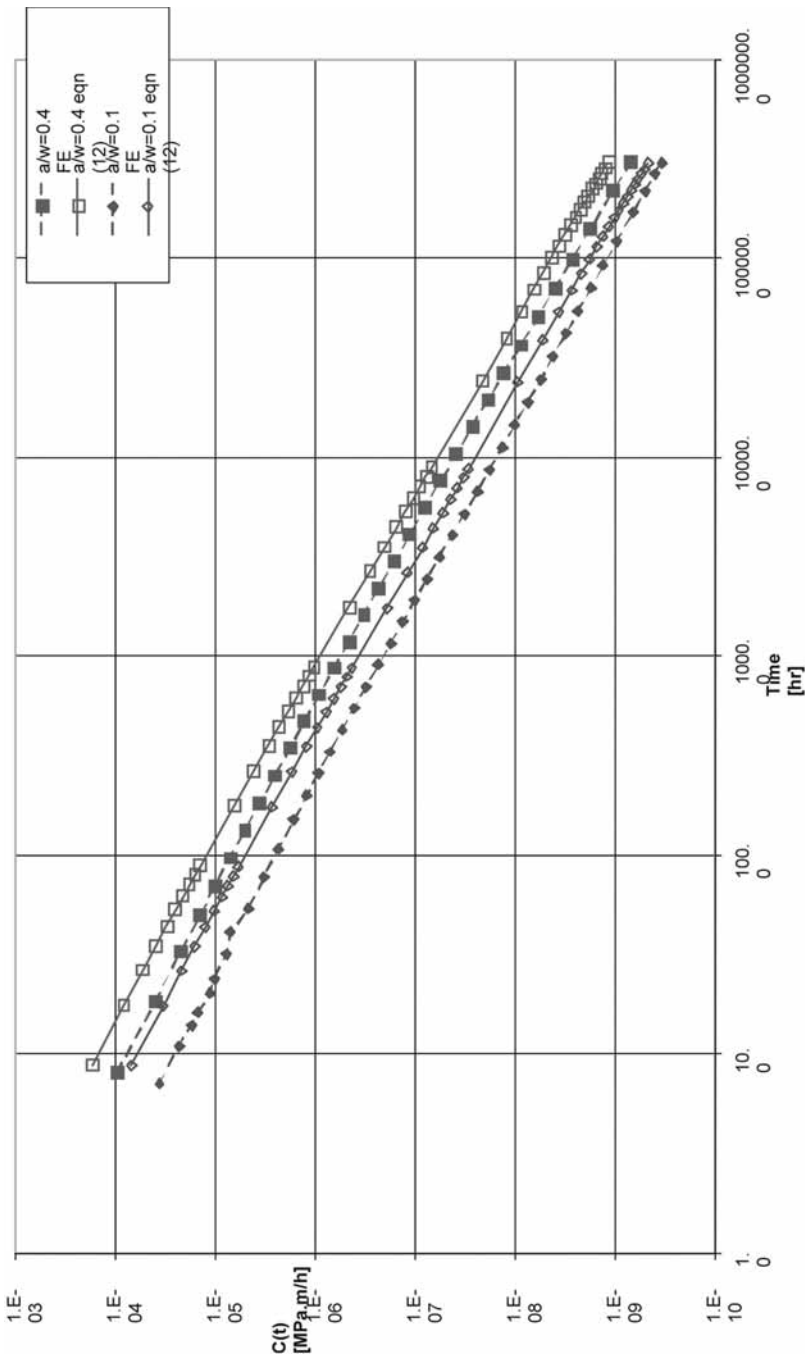


FIG. 6—Comparison of $C(t)$ predictions with values obtained from cracked body finite element analyses for a 35 mm thick cylinder subjected to welding residual stresses [16].

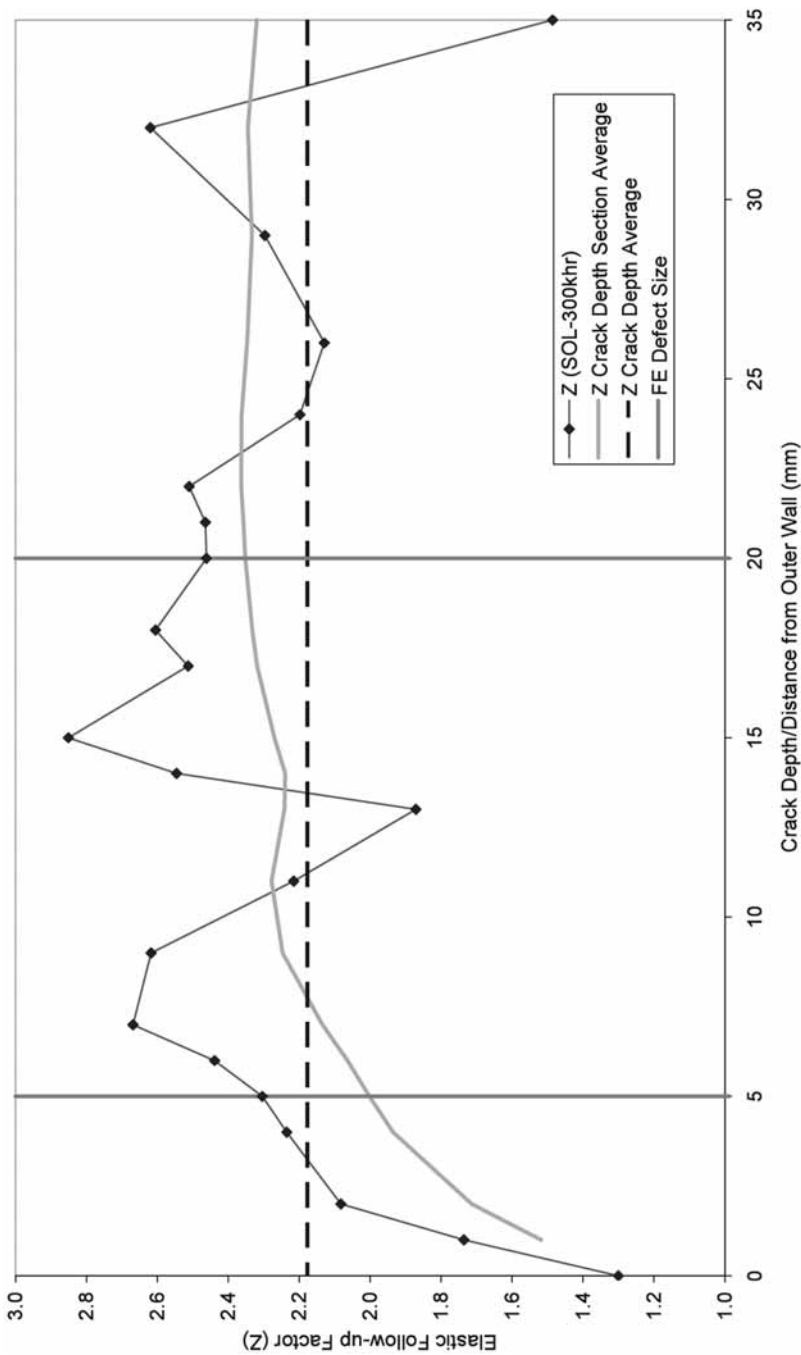


FIG. 7—Variation of uncracked elastic follow-up factor, Z , through the section for the Esshete Butt Weld [19]. The value used for cracks in the range 5–20 mm was derived by averaging the uncracked Z values, between the start of life and 300 kh, over crack depths of 5 mm ($Z=2.00$) and 20 mm ($Z=2.35$) and then taking the average of these values ($Z=2.18$).

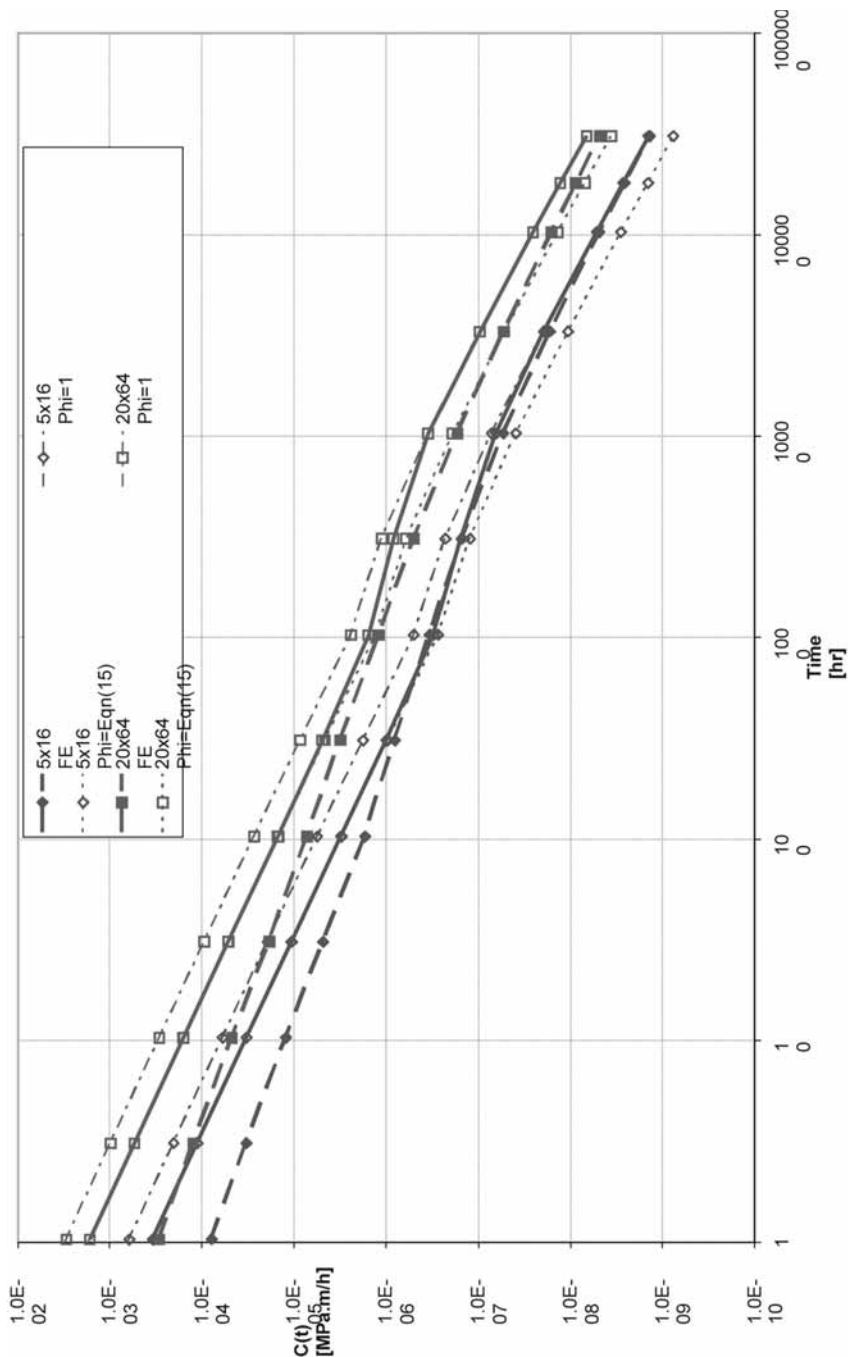


FIG. 8—Comparison of $C(t)$ predictions at the deepest point of 5x16 mm and 20x64 mm cracks with values obtained from 3D cracked body finite element analyses [19].

increase the uncracked value by unity remains a reasonable pragmatic approach in some cases although it can underestimate Z for displacement controlled tension situations [15] and may be overly conservative for situations involving significant welding residual stresses.

- Improved reference stress solutions are required for residual stress dominated situations as currently available collapse solutions often require simplifying assumptions to be made as they cannot accommodate out-of-plane bending stresses, for example.

Concluding Remarks

The current advice in R5 and other fitness-for-service assessment procedures on the calculation of creep and creep-fatigue crack growth for combined loading has been extended in the following areas.

- A general expression to calculate relaxation of the (total) reference stress for cases involving combined loading and significant crack growth has been derived;
- Advice has been provided on the estimation of $C(t)$ for elastic-plastic-creep behaviour for both primary and combined loading;
- Additional advice is given on the estimation of $C(t)$ for situations where crack growth is significant;
- A method is presented for assessing service-initiated defects under combined loading;
- Explicit advice is included on the treatment of combined loading for situations involving significant welding residual stresses; and
- Finite element validation of the extended advice has been given.

Acknowledgments

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Appendix A—Further Details on Relaxation of the Reference Stress for Combined Loading with Significant Crack Growth

Following the presentation of Eqs 1–6 in the section “Relaxation of the Reference Stress for Combined Loading,” it is assumed that Eq 1 also applies as relaxation of Q occurs. Then, differentiating Eq 1 recognising that P is fixed leads to

$$\frac{\dot{\sigma}_{\text{ref}}}{\sigma_{\text{ref}}} = \left(\frac{f'}{f} \right) \left(\frac{\dot{a}}{w} \right) - \left(\frac{g'}{g} \right) \left(\frac{\dot{a}}{w} \right) + \frac{K^p \left(\frac{g'}{g} \right) \left(\frac{\dot{a}}{w} \right) + K^s \left[\left(\frac{h'}{h} \right) \left(\frac{\dot{a}}{w} \right) + \frac{\dot{Q}}{Q} + \frac{Q'}{Q} \frac{\dot{a}}{w} \right]}{(K^p + K^s)} \quad (23)$$

where $f' = df/d(a/w)$, $g' = dg/d(a/w)$, $h' = dh/d(a/w)$, $Q' = \partial Q/\partial(a/w)$, $\dot{Q} = \partial Q/\partial t$ and $\dot{\sigma}_{\text{ref}} = d\sigma_{\text{ref}}/dt$.

Equation 23 can be written

$$\frac{\dot{\sigma}_{\text{ref}}}{\sigma_{\text{ref}}} = \left(\frac{f'}{f}\right) \left(\frac{\dot{a}}{w}\right) + \left(\frac{K^s}{K^p + K^s}\right) \left[\left(\frac{h'}{h} - \frac{g'}{g} + \frac{Q'}{Q}\right) \left(\frac{\dot{a}}{w}\right) + \frac{\dot{Q}}{Q} \right] \quad (24)$$

and hence

$$\dot{\sigma}_{\text{ref}} = \left(\frac{f'}{f}\right) \left(\frac{\dot{a}}{w}\right) \sigma_{\text{ref}} + \frac{K^s \sigma_{\text{ref}}^p}{K^p} \left[\left(\frac{h'}{h} - \frac{g'}{g} + \frac{Q'}{Q}\right) \left(\frac{\dot{a}}{w}\right) + \frac{\dot{Q}}{Q} \right] \quad (25)$$

Now, from Eqs 1, 6 and 7

$$\frac{\partial \sigma_{\text{ref}}}{\partial t} = \left(\frac{\sigma_{\text{ref}}^p}{K^p}\right) \left(\frac{\dot{Q}}{Q}\right) K^s = -E \dot{\varepsilon}_c / Z \quad (26)$$

Combining Eqs 25 and 26

$$\dot{\sigma}_{\text{ref}} = \sigma_{\text{ref}} \left(\frac{f'}{f}\right) \left(\frac{\dot{a}}{w}\right) - \frac{E \dot{\varepsilon}_c}{Z} + \frac{K^s \sigma_{\text{ref}}^p}{K^p} \left(\frac{\dot{a}}{w}\right) \left[\frac{h'}{h} - \frac{g'}{g} + \frac{Q'}{Q}\right] \quad (27)$$

Now, from Eqs 2–4

$$f = [mBw]^{-1} \quad (28)$$

i.e.,

$$f' / f = -m' / m \quad (29)$$

where:

$$m' = dm/d(a/w).$$

Therefore, Eq 27 can be written as the relaxation Eq 8, after substituting for Eq 29.

Primary Load Only

For primary loading only, Eq 8 reduces to

$$\frac{\dot{\sigma}_{\text{ref}}}{\sigma_{\text{ref}}} + \frac{m'}{m} \left(\frac{\dot{a}}{w}\right) = 0 \quad (30)$$

since $Z \rightarrow \infty$ and $K^s = 0$. This simply expresses the increase in reference stress with increasing crack size (noting that m' is negative).

Displacement Controlled Loading

Displacement controlled loading acting alone is described by

$$P \rightarrow 0, \quad h = g, \quad (K^S \sigma_{\text{ref}}^p / K^p) = \sigma_{\text{ref}}$$

with Q then equal to the load point reaction. Whence Eq 8 becomes

$$\frac{\dot{\sigma}_{\text{ref}}}{\sigma_{\text{ref}}} + \left(\frac{m'}{m} \right) \left(\frac{\dot{a}}{w} \right) + \frac{E \dot{\epsilon}_c}{Z \sigma_{\text{ref}}} - \left(\frac{\dot{a}}{w} \right) \left(\frac{Q'}{Q} \right) = 0 \quad (31)$$

Under displacement control, the elastic displacement Δ^e can be written in terms of a compliance function $C(a/w)$ as

$$\Delta^e = \frac{QC}{EB}$$

Since there is no change in displacement under displacement controlled loading, $[\partial \Delta^e / \partial (a/w)] = 0$; therefore $(Q'/Q) + (C'/C) = 0$ where $C' = dC/d(a/w)$. Then, Eq 31 becomes

$$\frac{\dot{\sigma}_{\text{ref}}}{\sigma_{\text{ref}}} + \left(\frac{\dot{a}}{w} \right) \left[\frac{m'}{m} + \frac{C'}{C} \right] + \frac{E \dot{\epsilon}_c}{Z \sigma_{\text{ref}}} = 0 \quad (32)$$

which is a special case given in R5. Thus, the general relaxation Eq 8 reduces to the current approach in R5 for pure displacement controlled loading.

Appendix B—Estimation Formulae for $C(t)$ Under Elastic-Plastic-Creep Conditions

The approach described below assumes a stationary crack and follows that set out in Refs 6 and 11, which may be consulted for more details. It is assumed that plasticity and creep are described by power-law equations of the form $\epsilon_p = \beta \sigma^\mu$ and $\dot{\epsilon}_c = D \sigma^n$, with $\mu = n$.

General Expressions

Initially, the stress and strain fields at a short distance ahead of the crack are defined by the HRR fields (see Refs 7 and 8)

$$\sigma_{ij}^0 = [J_0 / \beta I_n r]^{1/(n+1)} \tilde{\sigma}_{ij}(\theta, n) \quad r \rightarrow 0 \quad (33)$$

$$\epsilon_{ij}^0 = \beta [J_0 / \beta I_n r]^{n/(n+1)} \tilde{\epsilon}_{ij}(\theta, n) \quad r \rightarrow 0 \quad (34)$$

where:

(r, θ) = polar coordinates at the crack tip,

$I_n, \tilde{\sigma}_{ij}$, and $\tilde{\epsilon}_{ij}$ = known functions of the stress exponent, n , and

J_0 = initial value of the J -integral, for the combined loading.

This could, for example, be evaluated using the methods in R6 [13] and some simplified estimates of J_0 are examined below in the section "Estimate of Initial Strain ϵ_{ref}^0 ."

With creep, Eqs 33 and 34 are replaced by

$$\sigma_{ij} = [C(t)/DI_n r]^{1/(n+1)} \tilde{\sigma}_{ij}(\theta, n) \quad r \rightarrow 0 \quad (35)$$

$$\dot{\epsilon}_{ij} = D[C(t)/DI_n r]^{n/(n+1)} \tilde{\epsilon}_{ij}(\theta, n) \quad r \rightarrow 0 \quad (36)$$

for a given value of $t > 0$, where $C(t)$ is the transient crack tip parameter, which is a function of time, t . Combining Eqs 34 and 36, the total strain near the crack tip is

$$\begin{aligned} \epsilon_{ij} &= \left\{ \beta [J_0/\beta I_n r]^{n/(n+1)} + \int_0^t D[C(t')/DI_n r]^{n/(n+1)} dt' \right\} \tilde{\epsilon}_{ij}(\theta, n) \\ &= \frac{\tilde{\epsilon}_{ij}}{(I_n r)^{n/(n+1)}} \left[\beta \left[\frac{J_0}{\beta} \right]^{n/(n+1)} + D^{1/(n+1)} \int_0^t C(t')^{n/(n+1)} dt' \right], \quad r \rightarrow 0 \end{aligned} \quad (37)$$

Noting that $\sigma_{ij}^0 \epsilon_{ij}^0 = (J_0/I_n r) \tilde{\sigma}_{ij} \tilde{\epsilon}_{ij}$, Eqs 34 and 36 essentially define a time-dependent J -integral, $J(t)$, which may be deduced from $\sigma_{ij} \epsilon_{ij} = \tilde{\sigma}_{ij} \tilde{\epsilon}_{ij} (J/I_n r)$. Whence

$$J(t) = C(t)^{1/(n+1)} \left[(\beta/D)^{1/(n+1)} J_0^{n/(n+1)} + \int_0^t C(t')^{n/(n+1)} dt' \right] \quad (38)$$

Differentiating Eq 38,

$$dJ/dt = C(t) + \frac{J(t)}{(n+1)C(t)} \frac{dC}{dt} \quad (39)$$

Integrating

$$J^{n+1}/C = (n+1) \int_0^t J(t')^n dt' + A \quad (40)$$

where:

$A = \text{constant.}$

From Eq 38 for $t \rightarrow 0$,

$$J^{n+1}/C \rightarrow (\beta/D) J_0^n \equiv A$$

whence

$$C(t) = \frac{J(t)^{n+1}}{\left[(n+1) \int_0^t J(t')^n dt' + \beta J_0^n / D \right]} \quad (41)$$

This equation is solved in the section “Pure Mechanical Loading” for pure primary loading and in the section “General Loading” for more general loading. For the latter case, the input value of J_0 is discussed in the section “Estimate of Initial Strain $\varepsilon_{\text{ref}}^0$.”

Pure Mechanical Loading

For mechanical loading only, the value of J can be estimated as [11]

$$J = J_0 + C^*t \quad (42)$$

Then Eq 41 becomes

$$C(t) = \frac{C^*[1 + C^*t/J_0]^{n+1}}{[1 + C^*t/J_0]^{n+1} - (1 - \beta C^*/DJ_0)} \quad (43)$$

which is essentially the result given in R5. Neglecting small-scale yielding corrections, the initial value of J can be estimated as (see Ref 13)

$$J_0 = \frac{K^2}{E} [1 + E\varepsilon_{\text{ref}}^p / \sigma_{\text{ref}}]$$

where $\varepsilon_{\text{ref}}^p = \beta \sigma_{\text{ref}}^n$, and similarly the value of C^* can be estimated as [7]

$$C^* = D\sigma_{\text{ref}}^{n-1}K^2$$

Then

$$1 - \beta C^*/DJ_0 = \frac{\sigma_{\text{ref}}/E}{\sigma_{\text{ref}}/E + \varepsilon_{\text{ref}}^p} = \frac{\varepsilon_{\text{ref}}^e}{\varepsilon_{\text{ref}}^{e+p}}$$

where superscripts e and $e + p$ denote elastic and elastic-plastic, respectively. Similarly,

$$1 + C^*t/J_0 = \left[\frac{\sigma_{\text{ref}}/E + \varepsilon_{\text{ref}}^p + \varepsilon_{\text{ref}}^c}{\sigma_{\text{ref}}/E + \varepsilon_{\text{ref}}^p} \right]$$

where $\varepsilon_{\text{ref}}^c \equiv \dot{\varepsilon}_{\text{ref}}^c t$. Whence

$$1 + C^*t/J_0 = \frac{\varepsilon_{\text{ref}}^{e+p+c}}{\varepsilon_{\text{ref}}^{e+p}}$$

where superscript $e + p + c$ denotes elastic-plastic plus creep. The final result for $C(t)$ then follows from Eq 43 as

$$\frac{C(t)}{C^*} = \frac{(\varepsilon_{\text{ref}}^{e+p+c} / \varepsilon_{\text{ref}}^{e+p})^{n+1}}{(\varepsilon_{\text{ref}}^{e+p+c} / \varepsilon_{\text{ref}}^{e+p})^{n+1} - (\varepsilon_{\text{ref}}^e / \varepsilon_{\text{ref}}^{e+p})} \quad (44)$$

This reduces to the result given in R5 in the absence of plasticity and Eq 44 may be considered as a more general form of that expression.

General Loading

For more general loading, the approach in Refs 6 is followed with the estimates (see Ref 13)

$$J = \sigma_{\text{ref}} \varepsilon_{\text{ref}} R' \quad (45)$$

and

$$J_0 = \sigma_{\text{ref}}^0 \varepsilon_{\text{ref}}^0 R' \quad (46)$$

where the geometrical parameter $R' = (K^p / \sigma_{\text{ref}}^p)^2$ and J_0 , and hence σ_{ref}^0 , may be estimated from R6 [13] for combined loading, with

$$\sigma_{\text{ref}}^0 = \sigma_{\text{ref}}^p (K^p + K^s) / K^p$$

as in Eq 1, for elastic response. Now, following initial loading, stress relaxation occurs according to Eq 7 with

$$\dot{\sigma}_{\text{ref}} = -(E/Z) [\dot{\varepsilon}_{c,\text{ref}} - \dot{\varepsilon}_{c,\text{ref}}^p] \quad (47)$$

The corresponding total strain rate is

$$\dot{\varepsilon}_{\text{ref}} = \dot{\varepsilon}_{c,\text{ref}} + \dot{\sigma}_{\text{ref}}/E = \dot{\varepsilon}_{c,\text{ref}}^p - (Z-1)\dot{\sigma}_{\text{ref}}/E \quad (48)$$

Note that there is no plastic contribution to the total strain rate in Eq 48 as plastic strains are assumed to be irrecoverable and $\dot{\sigma}_{\text{ref}} < 0$. For a secondary creep law (constant creep strain rate), this can be integrated to give

$$\varepsilon_{\text{ref}} = \varepsilon_{\text{ref}}^0 + \dot{\varepsilon}_{c,\text{ref}}^p t - (Z-1)(\sigma_{\text{ref}} - \sigma_{\text{ref}}^0)/E \quad (49)$$

Combining this expression with Eqs 45 and 46

$$\left(\frac{J}{J_0}\right) = \left(\frac{\sigma_{\text{ref}}}{\sigma_{\text{ref}}^0}\right) \left[1 + \frac{\dot{\varepsilon}_{c,\text{ref}}^p t}{\varepsilon_{\text{ref}}^0} - (Z-1) \frac{(\sigma_{\text{ref}} - \sigma_{\text{ref}}^0)}{E \varepsilon_{\text{ref}}^0}\right]$$

and for use with Eq 41

$$\left(\frac{J}{J_0}\right)^n = \frac{\left[\dot{\varepsilon}_{c,\text{ref}}^p - \frac{Z\dot{\sigma}_{\text{ref}}}{E}\right]}{D(\sigma_{\text{ref}}^0)^n} \left[1 + \frac{\dot{\varepsilon}_{c,\text{ref}}^p t}{\varepsilon_{\text{ref}}^0} - \frac{(Z-1)(\sigma_{\text{ref}} - \sigma_{\text{ref}}^0)}{E \varepsilon_{\text{ref}}^0}\right]^n \quad (50)$$

Now,

$$\dot{\varepsilon}_{c,\text{ref}}^p - \frac{Z\dot{\sigma}_{\text{ref}}}{E} \geq \dot{\varepsilon}_{c,\text{ref}}^p - \frac{Z\dot{\sigma}_{\text{ref}}}{E} + \frac{\dot{\sigma}_{\text{ref}}}{E}$$

as $\dot{\sigma}_{\text{ref}} \leq 0$. Also

$$(Z-1)\dot{\varepsilon}_{c,\text{ref}}^p - Z(Z-1)\dot{\sigma}_{\text{ref}}/E \leq Z\dot{\varepsilon}_{c,\text{ref}}^p - Z(Z-1)\dot{\sigma}_{\text{ref}}/E$$

as $\dot{\varepsilon}_{c,\text{ref}}^p \geq 0$. Combining these inequalities

$$1 \leq \frac{\dot{\varepsilon}_{c,\text{ref}}^p - Z\dot{\sigma}_{\text{ref}}/E}{\dot{\varepsilon}_{c,\text{ref}}^p - (Z-1)\dot{\sigma}_{\text{ref}}/E} \leq \frac{Z}{Z-1} \quad (51)$$

as in Refs 6. Writing the ratio in Eq 51 as $\Phi(t)$ ($1 \leq \Phi \leq Z/(Z-1)$), the above equations are solved for the elastic-creep case in the section “Elastic-Creep Case” and for elastic-plastic-creep behaviour in the section “Elastic-Plastic-Creep Case.”

Elastic-Creep Case

For elastic-creep behaviour, $\beta = 0$ and Eq 41 can be simplified to

$$C(t) = \frac{J^{n+1}(t)}{(n+1) \int_0^t J^n(t') dt'} = \frac{J_0 (J(t)/J_0)^{n+1}}{(n+1) \int_0^t (J(t')/J_0)^n dt'} \quad (52)$$

Equations 45 and 46 and 49 give

$$\frac{J(t)}{J_0} = \left(\frac{\sigma_{\text{ref}}(t)}{\sigma_{\text{ref}}^0} \right) F(t) \quad (53)$$

where

$$F(t) \equiv 1 + \frac{\dot{\varepsilon}_{c,\text{ref}}^p t - (Z-1)(\sigma_{\text{ref}} - \sigma_{\text{ref}}^0)/E}{\varepsilon_{\text{ref}}^0} = \frac{\varepsilon_{\text{ref}}}{\varepsilon_{\text{ref}}^0} \quad (54)$$

and hence

$$\dot{F}(t) = \frac{\dot{\varepsilon}_{c,\text{ref}}^p - (Z-1)\dot{\sigma}_{\text{ref}}/E}{\varepsilon_{\text{ref}}^0} \quad (55)$$

Then

$$\dot{F}(t) \geq 0, \quad F(t) \geq F(0) = 1 \quad (56)$$

Equations 50 and 51 then give

$$\int_0^t (J(t')/J_0)^n dt' = \frac{\varepsilon_{\text{ref}}^0}{D(\sigma_{\text{ref}}^0)^n} \int_0^t \Phi(t') \dot{F}(t') F^n(t') dt' \quad (57)$$

where

$$\Phi(t) \equiv \frac{\dot{\varepsilon}_{c,\text{ref}}^p - Z\dot{\sigma}_{\text{ref}}/E}{\dot{\varepsilon}_{c,\text{ref}}^p - (Z-1)\dot{\sigma}_{\text{ref}}/E} = \frac{Z\dot{\varepsilon}_{c,\text{ref}}}{(Z-1)\dot{\varepsilon}_{c,\text{ref}} + \dot{\varepsilon}_{c,\text{ref}}^p} \quad (58)$$

as used in inequality Eq 51. Note that for combined loading

$$\Phi(\infty) = 1, Z/(Z-1) > \Phi(t) > 1 \quad \text{since} \quad \dot{\varepsilon}_{c,\text{ref}}^p > 0, \dot{\sigma}_{\text{ref}} \leq 0, Z \geq 1 \quad (59)$$

$\Phi(t) \equiv 1$ for pure primary loading ($Z = \infty$) and $\Phi(t) \equiv Z/(Z-1)$ for pure secondary loading with $Z > 1$. The case of pure secondary loading with $Z = 1$, that is relaxation under constant applied strain, is a special case which is not considered here.

Substituting Eqs 53 and 57 into Eq 52 gives

$$C(t) = \left[\frac{J_0 D(\sigma_{\text{ref}}^0)^n}{\varepsilon_{\text{ref}}^0} \right] \left[\frac{(\sigma_{\text{ref}}/\sigma_{\text{ref}}^0)^{n+1} F^{n+1}(t)}{(n+1) \int_0^t \Phi(t') \dot{F}(t') F^n(t') dt'} \right] \quad (60)$$

It is now assumed that $\dot{\Phi}(t) \leq 0$, which is equivalent to assuming that the rate of stress relaxation reduces with increasing time (i.e., $\ddot{\sigma}_{\text{ref}} \geq 0$). Then, since $F(t) \geq 1$, the integral in the denominator of this equation can be bounded below by

$$\begin{aligned} (n+1) \int_0^t \Phi(t') \dot{F}(t') F^n(t') dt' &= [\Phi(t) F^{n+1}(t) - \Phi(0)] + \int_0^t (-\dot{\Phi}(t') F^{n+1}(t')) dt' \\ &\geq \Phi(t) F^{n+1}(t) - \Phi(0) + \int_0^t (-\dot{\Phi}(t')) dt' \\ &= \Phi(t) [F^{n+1}(t) - 1] \end{aligned} \quad (61)$$

Using the reference stress approximations for J_0 and C^* [7], it can be shown that

$$\frac{J_0 D(\sigma_{\text{ref}}^0)^n}{\varepsilon_{\text{ref}}^0} = C^* \left(\frac{\sigma_{\text{ref}}^0}{\sigma_{\text{ref}}^p} \right)^{n+1} \quad (62)$$

A conservative estimate of $C(t)$ is obtained by replacing the inequality in Eq 61 by an equals sign. Hence it follows that:

$$\frac{C(t)}{C^*} = \left(\frac{\sigma_{\text{ref}}}{\sigma_{\text{ref}}^p} \right)^{n+1} \left[\frac{(\varepsilon_{\text{ref}}/\varepsilon_{\text{ref}}^0)^{n+1}}{\Phi(t) [(\varepsilon_{\text{ref}}/\varepsilon_{\text{ref}}^0)^{n+1} - 1]} \right] \quad (63)$$

Now consider the limiting cases of long and short times. In the long time limit for the case where $\sigma_{\text{ref}}^p > 0$, $\sigma_{\text{ref}}/\sigma_{\text{ref}}^p \rightarrow 1$, $\Phi(t) \rightarrow 1$, $\varepsilon_{\text{ref}}(t)/\varepsilon_{\text{ref}}^0 \rightarrow \infty$ as time $t \rightarrow \infty$ so that from Eq 63, $C(t) \rightarrow C^* > 0$ as expected. In the short time limit for $t \rightarrow 0$, $\varepsilon_{\text{ref}}/\varepsilon_{\text{ref}}^0 \rightarrow 1$, $\sigma_{\text{ref}}/\sigma_{\text{ref}}^p \rightarrow \sigma_{\text{ref}}^0/\sigma_{\text{ref}}^p$, $\Phi(t) \rightarrow \Phi(0)$. However, the behaviour of the strain ratio, that is the function F (see Eq 54), depends on the creep law. For a secondary creep law, F can be expanded in a Taylor series, so that

$$F(t) = 1 + \dot{F}(0)t + \dots \quad (64)$$

and hence

$$\frac{C(t)}{C^*} \approx \left(\frac{\sigma_{\text{ref}}^0}{\sigma_{\text{ref}}^p} \right)^{n+1} \left[\frac{1}{\Phi(0)(n+1)\dot{F}(0)t} \right] \quad (65)$$

Equation 65 is the short time limit for $C(t)$ for the case of secondary creep, noting that, in the case currently considered by R5 [1], $\sigma_{\text{ref}}^0/\sigma_{\text{ref}}^p = K^0/K^p$ where K denotes the stress intensity factor. The result is consistent with R5 for the case of pure primary load, where $\sigma_{\text{ref}}^0 = \sigma_{\text{ref}}^p$, $\Phi(0) = 1$, $\dot{F}(0) = E\dot{\varepsilon}_{c,\text{ref}}^p/\sigma_{\text{ref}}^p = EC^*/(K^p)^2$.

For a primary creep law of the form

$$\varepsilon_c(\sigma, t) = D\sigma^n t^p \quad (66)$$

with $n > 1$ and $0 < p < 1$, the substitution $\tau = t^p$ leads to the secondary creep law

$$\dot{\varepsilon}_c = D\sigma^n \quad (67)$$

where here the dot denotes the derivative with respect to τ rather than t . Equation 65 can then be written as

$$\frac{C(t)}{C^*} \approx \left(\frac{\sigma_{\text{ref}}^0}{\sigma_{\text{ref}}^p} \right)^{n+1} \left[\frac{1}{\Phi(0)(n+1)\dot{F}(0)t^p} \right] \quad (68)$$

but again with the derivative $\dot{F}(0) \equiv dF/d\tau(\tau = 0)$.

Elastic-Plastic-Creep Case

In this case, Eq 41 implies that

$$C(t) = \frac{J^{n+1}(t)}{(n+1) \int_0^t J^n(t') dt' + \beta J_0^n/D} = \frac{J_0 \left(J(t)/J_0 \right)^{n+1}}{(n+1) \int_0^t \left(J(t')/J_0 \right)^n dt' + \beta/D} \quad (69)$$

Hence it follows from earlier that $C(t)$ can be bounded above by

$$\frac{C(t)}{C^*} \leq \left(\frac{\sigma_{\text{ref}}}{\sigma_{\text{ref}}^p} \right)^{n+1} \left[\frac{(\varepsilon_{\text{ref}}/\varepsilon_{\text{ref}}^0)^{n+1}}{\Phi(t)[(\varepsilon_{\text{ref}}/\varepsilon_{\text{ref}}^0)^{n+1} - 1] + \beta(\sigma_{\text{ref}}^0)^n/\varepsilon_{\text{ref}}^0} \right] \quad (70)$$

Replacing the inequality in Eq 70 with an equals sign and noting that the plastic strain at σ_{ref}^0 is given by $\beta(\sigma_{\text{ref}}^0)^n = \varepsilon_{\text{ref}}^0 - \sigma_{\text{ref}}^0/E$

$$\frac{C(t)}{C^*} = \left(\frac{\sigma_{\text{ref}}}{\sigma_{\text{ref}}^p} \right)^{n+1} \left[\frac{(\varepsilon_{\text{ref}}/\varepsilon_{\text{ref}}^0)^{n+1}}{\Phi(t)[(\varepsilon_{\text{ref}}/\varepsilon_{\text{ref}}^0)^{n+1} - 1] + (1 - \sigma_{\text{ref}}^0/E\varepsilon_{\text{ref}}^0)} \right] \quad (71)$$

which reduces to the elastic-creep case (Eq 63) when $\varepsilon_{\text{ref}}^0 = \sigma_{\text{ref}}^0/E$. Note also that Eq 71 leads to a finite value of $C(t)$ at $t = 0$ of $(\sigma_{\text{ref}}^0/\sigma_{\text{ref}}^p)^{n+1}/(1 - \sigma_{\text{ref}}^0/E\varepsilon_{\text{ref}}^0)$.

For widespread initial plasticity, $\varepsilon_{\text{ref}}^0 \gg \sigma_{\text{ref}}^0/E$, Eq 71 becomes

$$\frac{C(t)}{C^*} = \left(\frac{\sigma_{\text{ref}}}{\sigma_{\text{ref}}^p} \right)^{n+1} \left[\frac{(\varepsilon_{\text{ref}}/\varepsilon_{\text{ref}}^0)^{n+1}}{\Phi(t)[(\varepsilon_{\text{ref}}/\varepsilon_{\text{ref}}^0)^{n+1} - 1] + 1} \right] \quad (72)$$

which reduces to

$$C(t)/C^* = 1 \quad (73)$$

for the case of pure primary loading ($\Phi = 1$), as expected from R5.

Equation 72 is the general equation for elastic-plastic-creep conditions. As earlier, n is the exponent of stress for either secondary creep (as used in the derivation of the equations) or for primary creep.

A conservative estimate of $C(t)$ follows from the lower bound to Φ in Eq 71, i.e.,

$$\frac{C(t)}{C^*} = \left(\frac{\sigma_{\text{ref}}}{\sigma_{\text{ref}}^p} \right)^{n+1} \left[\frac{(\varepsilon_{\text{ref}}/\varepsilon_{\text{ref}}^0)^{n+1}}{(\varepsilon_{\text{ref}}/\varepsilon_{\text{ref}}^0)^{n+1} - (\sigma_{\text{ref}}^0/E\varepsilon_{\text{ref}}^0)} \right] \quad (74)$$

This is identical to Eq 44 for $\sigma_{\text{ref}} = \sigma_{\text{ref}}^p$, reduces to the result in R5 for elastic behaviour on initial loading and generalises that expression to initial elastic-plastic response.

Estimate of Initial Strain $\varepsilon_{\text{ref}}^0$

In order to use Eq 71 or Eq 74, it is necessary to estimate $\varepsilon_{\text{ref}}^0$. This can be obtained from Eq 46 using R6 to estimate J_0 . This may be written

$$\frac{K^p + VK^s}{\sqrt{EJ_0}} = f(L_r) \quad (75)$$

where:

V = parameter in R6 for treating interactions between primary and secondary stress, and

$f(L_r)$ = failure assessment diagram (Option 1, 2,...).

Then

$$J_0 = \frac{(K^p + VK^s)^2}{Ef^2(L_r)} \quad (76)$$

From Eq 46

$$\sigma_{\text{ref}}^0 \varepsilon_{\text{ref}}^0 (K^p / \sigma_{\text{ref}}^p)^2 = \frac{(K^p + VK^s)^2}{Ef^2(L_r)} \quad (77)$$

or

$$\sigma_{\text{ref}}^0 \varepsilon_{\text{ref}}^0 = \frac{(\sigma_{\text{ref}}^p)^2 (1 + VK^s / K^p)^2}{Ef^2(L_r)} \quad (78)$$

This may be used to define $\varepsilon_{\text{ref}}^0$ if the shape of the stress-strain curve is known. For elastic behaviour, $V \rightarrow 1$ and $f(L_r) \rightarrow 1$ so that

$$\sigma_{\text{ref}}^0 = \frac{(K^p + K^s)}{K^p} \sigma_{\text{ref}}^p; \quad \varepsilon_{\text{ref}}^0 = \sigma_{\text{ref}}^0 / E$$

as in Eq 1.

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Zhigang Wei,¹ Fulun Yang,² Henry Cheng,² and Kamran Nikbin³

Probabilistic Prediction of Crack Growth Based on Creep/Fatigue Damage Accumulation Mechanism

ABSTRACT: In this paper a deterministic creep/fatigue accumulation model from simple sequential and repetitive creep and fatigue crack growth mechanisms for creep-brittle materials is developed. In the model the maximum stress intensity factor and stress intensity factor range can be used to describe, respectively, the creep and fatigue crack growth behaviors. The probabilistic behavior of the combined creep-fatigue can be calculated by direct integration of a joint log-normal probability density function or other methods. The Monte Carlo simulation method is used in this paper to analyze the probabilistic behavior of the derived deterministic creep-fatigue model and to estimate the reliability of a creep-fatigue correlation as introduced by the uncertainties in both creep and fatigue lives. Predicted results from this new crack growth method are compared with the bi-linear creep-fatigue interaction models as adopted by ASME Code Case N-47 and API579/ASME FFS, and the results are discussed.

KEYWORDS: probabilistic analysis, creep-fatigue interaction, crack growth, damage

Nomenclature

A = material parameter in damage model
 a_0 = initial crack length
 a = current crack length
 a_f = crack length at failure
 C = creep accumulation damage

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¹Tenneco Automotive, Grass Lake, MI 49240 (Corresponding author).

²Tenneco Automotive, Grass Lake, MI 49240.

³Imperial College, London SW7 2AZ, UK.

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- C_C = creep crack growth coefficient
 C_F = fatigue crack growth coefficient
 D = damage
 D_0 = initial damage
 D_f = damage at failure
 D_{TT} = generalized total creep damage
 dtn = dimensionless time fraction
 D_{fm} = fatigue damage at the intersection point of bilinear creep-fatigue diagram
 D_{cm} = creep damage at the intersection point of bilinear creep-fatigue diagram
 ΔD_N = generalized fatigue damage for each cycle
 D_{TN} = generalized total fatigue damage
 F = fatigue accumulation damage
 $g(C)$ = performance function
 $K_{\max}$ = maximum stress intensity factor
 ΔK = stress intensity factor range
 L = total gauge length of a specimen
 m = material constant in fatigue crack growth
 N = cycle number
 N_f = cycle to failure
 $\bar{N}$ = equivalent cycle number for a given damage produced by a complete creep-fatigue cycle
 N_0 = mean cycle to failure
 n = material constant in creep crack growth
 N_i = cycles to failure at stress level σ_i
 n_i = cycles at stress level σ_i
 P = failure possibility
 R = stress ratio
 T = total creep time
 T_j = creep time to failure at stress level σ_j
 T_0 = mean creep time to failure
 T_C = total rupture time under pure creep loading condition
 t = time
 Δt = hold time in a cycle
 t_j = creep time at stress level σ_j
 Z = normal distribution parameter
 $\bar{\Delta t}$ = equivalent time for a given damage produced by a complete creep-fatigue cycle
 η_T = standard deviation of time to failure
 η_N = standard deviation of cycles to failure
 σ = stress
 $\Delta\sigma$ = stress range
 $\sigma_{\min}$ = minimum stress
 $\sigma_{\max}$ = maximum stress
 β = material parameter in damage model
 $\text{erf}()$ = error function

- μ = mean value in normal distribution
- δ = perturbation parameter
- λ_N = fatigue damage ratio
- λ_T = creep damage ratio

Introduction

Creep-fatigue damage has long been recognized as one of the principal mechanisms in many high-temperature applications such as power plant components, turbine engines, and exhaust systems in vehicles. Start-up and shut-down of thermal systems will result in transient temperature gradients in materials. Different thermal expansion during each transient cycle results in a thermally induced cyclic stress, which together with the superimposed normal operating stress leads to creep-fatigue damage. Fatigue and creep fracture should not happen during the service time of a component, therefore, the basic understanding of the creep-fatigue phenomena and subsequent engineering design are very important both economically and for safety reason. Accordingly, creep-fatigue damage has been the focus of extensive studies, and a variety of predictive models for creep-fatigue failure have been proposed and several industry guidelines have been developed by these efforts.

The ASME Case N-47 [1] is the most widely used engineering code for the creep-fatigue design and life assessment of components. More recently, API579/ASME FFS 2 provides a similar guideline for creep-fatigue resistance design. Other alternative creep-fatigue modeling methods include strain range partitioning, frequency separation etc. 3. ASME code Case N-47 is based on the concept that an interaction diagram exists for fatigue and creep loadings. In its most general form, this model assumes the fatigue damage F to be given by Miner's law, as $F = \sum_{i=1}^k (n_i/N_i)$, where n_i is the number of cycles at strain range $\Delta\epsilon_i$ or stress range $\Delta\sigma_i$, and N_i is the number of cycles to failure at this stress or strain range. Likewise, for piecewise-constant stress histories, the creep damage C is assumed to be given by the time fraction, or Robinson's rule, in the form of $C = \sum_{j=1}^l (t_j/T_j)$, where t_j is the creep time at stress level σ_j , and T_j is the time to failure at this stress level. The damage summation rule then postulates the existence of a performance or limit-state function $F = g(C)$ which defines the creep-fatigue failure envelope on C, F -axes. The combinations of F and C for which $F < g(C)$ would not lead to creep-fatigue failure, whereas, failure would occur whenever $F > g(C)$. $g(0) = 1$ and $g(1) = 0$ are the failure relationship of Miner's and Robinson's laws in the case of pure fatigue, and pure creep, respectively. $g(C)$ is determined from experimental creep-fatigue failure data and is usually found to be nonlinear, i.e., $F + C = \sum_{i=1}^k (n_i/N_i) + \sum_{j=1}^l (t_j/T_j) < 1$, which is assumed to be more damaging than the so called linear accumulation damage, i.e., $\sum_{i=1}^k (n_i/N_i) + \sum_{j=1}^l (t_j/T_j) = 1$. In engineering design, bi-linear creep-fatigue interaction curves are usually recommended [1,2]. In the ASME Boiler and Pressure Vessel Code, the intersection of the two linear segments of the failure function is assumed to be at a point where the creep damage is equal to the fatigue damage. In API579/ASME FFS 2 the intersection point of these two straight lines is set as $D_{fm} = D_{cm} = 0.15$ for Carbon steels, low alloy steels, and

alloy 800H; whereas $D_{fm} = D_{cm} = 0.30$ for Type 304 SS and Type 316 SS. Therefore, the bi-linear curves are symmetric about the intersection point and in a concave shape. It is generally believed that the so called nonlinear behavior of the damage is caused by the creep-fatigue interaction, which usually accelerates the crack growth damage. Many experimental and analytical works have been done on the so called “creep-interaction effect” 4.

Most of creep-fatigue models available are deterministic, and do not incorporate the uncertainties in the values of the parameters that affect creep and fatigue life. However, it is well known that both pure fatigue and pure creep failure behavior show remarkable amounts of scatter in the number of cycles to failure at a given strain/stress range, and in the time to failure at a given stress level, respectively. Therefore, it is reasonable to believe that failure data under creep-fatigue loading would show a similar degree of scatter 5. Harlow and Delph 5 presented such a probabilistic model for creep-fatigue analysis based on the bi-linear creep-fatigue failure function as adopted in the ASME Boiler and Pressure Vessel Code 1. In 5, the probability of failure was estimated for a given deterministic loading history, with the material creep life and fatigue life as random variables 5. Several approaches, such as direct integration, Monte Carlo simulation and Second Order Reliability Method (SORM), have been introduced to analyze the probabilistic behavior of creep-fatigue interaction 5. The bi-linear failure function used in 5 and the ASME code has the drawback of being non-smooth. The SORM is inapplicable at the intersection of the two linear segments of the failure function, because SORM requires the existence of continuous second derivatives at this point. Harlow and Delph 5 suggested that this problem may be avoided by fitting the test results with smooth function. Mao and Mahadeven 6 developed a generalized smooth function, which needs not to be symmetric with respect to the intersection point. Mao and Mahadevan 6,7 also extend probabilistic analysis by additionally including the randomness in the loading history and demonstrate the predictability of the First Order Reliability Method (FORM) in probabilistic analysis of creep-fatigue interaction. More related topics can be found in Ref 8.

In this paper, a new creep-fatigue damage model is developed by considering the periodical and sequential fatigue and creep crack growth for a given materials with initial and final damage states. It demonstrates how the so called nonlinear creep-fatigue damage accumulation rule can be achieved without introducing any creep-fatigue interaction. Next, the Monte Carlo simulation is implemented in the proposed deterministic model to predict the probabilistic behavior of the creep-fatigue phenomenon. It reveals how the scatter in fatigue and creep failure data may be incorporated into the creep-fatigue failure model to yield the probability of creep-fatigue failure.

A Crack Growth Based Deterministic Creep-Fatigue Model

Consider a very simple case, where the creep-fatigue failure is assumed to be controlled by crack growth only. This is usually the case for compact tension loading specimen with a sharp pre-crack, see Fig. 1.

We first define a damage parameter

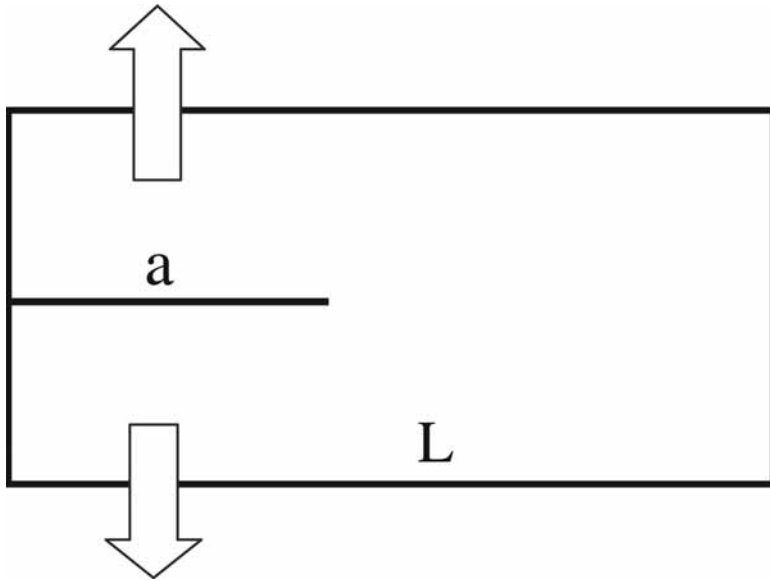


FIG. 1—Crack growth in a compact specimen.

$$D = \frac{a}{L} \quad (1)$$

where:

a = current crack length, and

L = total gauge length of a specimen.

If the initial and the final lengths of the crack at failure are a_0 and a_f , respectively, then the initial and the final damage can be written as

$$D_0 = a_0/L \quad (2)$$

and

$$D_f = a_f/L \quad (3)$$

The definitions in Eqs 1–3 are consistent with traditional damage definition in terms of classic damage mechanics context 9.

The loading profile is assumed to be periodical and sequential fatigue-creep loading, in which the loading (the red color in Fig. 2) and unloading time is assumed to be very rapid and can be negligible. The creep crack is assumed to propagate only in the holding time period (the green color in Fig. 2) under constant stress σ . In addition, the fatigue crack growth is assumed to grow only in the loading stage and nothing happens in the holding time period and unloading stage. This assumption is consistent with the resharpening and blunting mechanisms as observed in metallic materials under uniaxial loading 10. For simplicity and without losing generality, the stress ratio $R = \sigma_{\text{Min}}/\sigma_{\text{Max}}$ is

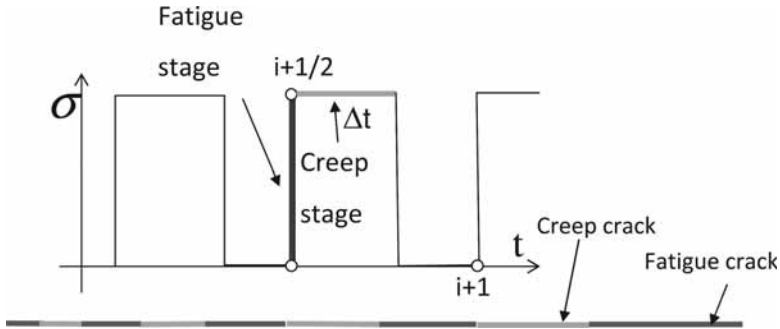


FIG. 2—Sequential creep-fatigue loading and crack pattern.

assumed to be zero in this paper, therefore, $\Delta\sigma = \sigma$. Finally, the crack growth is assumed to propagate in a “self-similar” manner, i.e., collinear propagation.

For a periodical and sequential creep-fatigue loading in this model, a cycle consists of two parts: the first part is from i to $i + 1/2$ as shown in Fig. 2, and it is controlled purely by fatigue crack growth; the second part, from $i + 1/2$ to $i + 1$ as shown in Fig. 2, is controlled purely by creep crack growth. $1/2$ here stands for the border between fatigue and creep processes of a complete cycle and does not represent any value. It is noted that the unloading part and the low stress level part ($\sigma = 0$ here) do not contribute to fatigue and creep damage. Therefore, for the sequential creep-fatigue loading, the creep damage obtained from previous cycle can be used as the initial damage value for the current fatigue damage calculation; in turn, the final fatigue damage in current loading cycle can be used as the initial value for the creep damage calculation in the second part of this cycle. This head-tail process continues recursively. In this way, the crack length a and damage D at any given time can be calculated with a given initial crack length a_0 or damage value D_0 , and the calculation process will be stopped when the total crack length or damage reaches a certain critical value a_f or D_f .

Without losing generality, the creep-fatigue crack growth of creep-brittle materials 11 will be considered here for simplicity. For a creep-brittle material, the maximum stress intensity factor and stress intensity factor range can be used for creep crack growth and fatigue crack growth, respectively. For creep-ductile materials, path independent parameter, such as C^* , can be used for fatigue and creep crack growth 11, and therefore it will not be addressed here. However, even though there are some differences between creep-brittle and creep-ductile materials, the conclusions drawn here from creep-brittle materials can be directly applied to creep-ductile or other materials.

Fatigue crack growth (Paris law)

The Paris law for the simplest loading and geometry, such as that shown in Fig. 1 can be expressed as

$$\frac{da}{dN} = C_F(\Delta K)^m$$

$$\Delta K = \Delta \sigma \sqrt{\pi a}$$

where:

K = stress intensity factor, and

C_F and m = fatigue crack growth related constants.

With the damage definition of Eq 1, Eq 4 can be rewritten as

$$\frac{dD}{dN} = C_F (\Delta \sigma)^m \pi^{(m/2)} L^{(m/2-1)} D^{(m/2)} \quad (5)$$

or

$$D_{i+1}^{(1-(m/2))} - D_i^{(1-(m/2))} = \left(1 - \frac{m}{2}\right) C_F (\Delta \sigma)^m \pi^{m/2} L^{((m/2)-1)} \Delta N \quad (6)$$

where:

$$\Delta N = N_{i+1} - N_i.$$

An obvious link between the Paris law and fatigue S-N curve can be obtained when considering the life of a cracked specimen having an initial damage D_0 , and final damage D_f under the assumptions of constant remote stress range $\Delta \sigma$. Based on Eq 6 and for $m > 2$ (it is the case for most of metal and composite materials), the S-N curve can be derived as

$$N_f = \frac{D_f^{(1-(m/2))} - D_0^{(1-(m/2))}}{\left(1 - \frac{m}{2}\right) C_F (\Delta \sigma)^m \pi^{m/2} L^{((m/2)-1)}} \quad (7)$$

Substituting Eq 7 into Eq 6 we have

$$D_{i+1}^{(1-(m/2))} - D_i^{(1-(m/2))} = [D_f^{(1-(m/2))} - D_0^{(1-(m/2))}] \frac{\Delta N}{N_f} \quad (8)$$

N_f is the cycles to failure under pure fatigue loading condition. With $\Delta N = 1$ we get

$$\frac{\Delta D_N}{D_{TN}} = \frac{1}{N_f} \quad (9)$$

where:

$\Delta D_N = D_{i+1}^{(1-(m/2))} - D_i^{(1-(m/2))}$ = generalized fatigue damage for each cycle, and

$D_{TN} = D_f^{(1-(m/2))} - D_0^{(1-(m/2))}$ = generalized total fatigue damage.

Clearly, the crack growth based damage produced in each cycle as shown in Eq 9 is different from the conventional linear damage rule as defined in Miner's rule because there is a exponent $1 - (m/2)$ in the damage term of Eq 8.

Therefore, the actual damage caused by fatigue crack growth mechanism is actually nonlinear in nature in terms of Miner's definition.

Creep Crack Growth

For a creep-brittle material at a certain given constant temperature, the creep crack growth can be expressed as 12

$$\begin{aligned}\frac{da}{dt} &= C_c (K_{\max})^n \\ K_{\max} &= \sigma_{\max} \sqrt{\pi a}\end{aligned}\quad (10)$$

where:

C_c and n = constants.

For stress ratio $R = \sigma_{\min}/\sigma_{\max} = 0$, $\sigma_{\max} = \sigma$, which will be used in the following calculations for simplicity. With the damage definition of Eq 1, Eq 10 can be rewritten as

$$\frac{dD}{dt} = C_c \sigma^n \pi^{(n/2)} L^{((n/2)-1)} D^{(n/2)} \quad (11)$$

or

$$D_{i+1}^{(1-(n/2))} - D_i^{(1-(n/2))} = \left(1 - \frac{n}{2}\right) C_c (\sigma)^n \pi^{n/2} L^{((n/2)-1)} \Delta t \quad (12)$$

where:

Δt is hold time, which is assumed to be a constant in this paper. Then, the following creep stress rupture relationship can be derived from the crack growth formula, Eq 12

$$T_C = \frac{D_f^{(1-(n/2))} - D_0^{(1-(n/2))}}{\left(1 - \frac{n}{2}\right) C_c \sigma^n \pi^{n/2} L^{((n/2)-1)}} \quad (13)$$

$$D_{i+1}^{(1-(n/2))} - D_i^{(1-(n/2))} = [D_f^{(1-(n/2))} - D_0^{(1-(n/2))}] \frac{\Delta t}{T_C} \quad (14)$$

T_C is the total rupture time under pure creep loading condition. Equation 14 can be written in the following form:

$$\frac{\Delta D_T}{D_{TT}} = \frac{\Delta t}{T_C} \quad (15)$$

where:

$\Delta D_T = D_{i+1}^{(1-(n/2))} - D_i^{(1-(n/2))}$ = generalized creep damage for each cycle, and

$D_{TT} = D_f^{(1-(n/2))} - D_0^{(1-(n/2))}$ = generalized total creep damage.

Once again, the crack growth based damage produced in a given hold time as shown in Eq 15 is different from the conventional linear damage rule as defined in Robinson's rule because there is an exponent $1 - (n/2)$ in the damage term in Eq 14. Therefore, the actual damage caused by creep crack growth mechanism is nonlinear in terms of Robinson's definition.

Combined Creep-Fatigue Crack Growth

As already mentioned before, we assume that fatigue crack growth only occurs in the loading stage and is cycle dependent, whereas the creep crack growth is assumed to occur only in the hold stage at high stress level and is time dependent. In addition, without losing generality, for each creep-fatigue cycle we assume that the fatigue crack growth occurs first and then followed by the creep crack growth. Then, we have the following damage evolution equations for fatigue, Eq 16-1, and subsequent for creep crack growth, Eq 16-2, respectively.

$$D_{j+1/2}^{(1-(m/2))} - D_i^{(1-(m/2))} = [D_f^{(1-(m/2))} - D_0^{(1-(m/2))}] \frac{\Delta N}{N_f} \text{ for fatigue}$$

$$D_j^{(1-(n/2))} - D_{i+1/2}^{(1-(n/2))} = [D_f^{(1-(n/2))} - D_0^{(1-(n/2))}] \frac{\Delta t}{T_C} \text{ for creep} \quad (16)$$

Substituting $D_{1+1/2}$ in Eq 16-1 into Eq 16-2 and rearranging Eq 16-2, we have the following non-dimensional iterative algebraic equation, Eq 17.

$$D_{i+1} = \left\{ \left[D_i^{(1-(m/2))} + \frac{D_f^{(1-(m/2))} - D_0^{(1-(m/2))}}{N_f} \Delta N \right]^{((2-n)/(2-m))} + \left[D_f^{(1-(n/2))} - D_0^{(1-(n/2))} \right] \frac{\Delta t}{T_C} \right\}^{(2/(2-n))} \quad (17)$$

For a complete creep-fatigue cycle, $\Delta N = 1$. Equation 17 is the main result of the deterministic part of this paper. With Eq 17, the damage at any time and cycle can be calculated easily in a recursive process for a given initial and final damages D_0, D_f , material properties m, n , life information from pure fatigue and pure creep test $\Delta N, T_C$, and creep loading condition Δt . The most important information we can get from Eq 17 is that the performance function $F = g(C)$ can be obtained by solving Eq 17 with varying Δt or ΔN . In calculating $F = g(C)$, the final damage D_f should be reached in calculation. Furthermore, for a calculated damage caused by a complete loading cycle with fatigue as the first part and creep as the second part, the equivalent cycle/relative damage and equivalent hold time/relative damage for pure fatigue and pure creep can be derived: Substituting Eq 17 into Eq 8 and Eq 14 we have the relative damage for fatigue and creep respectively

$$\frac{\overline{\Delta N}}{N_f} = \frac{1}{D_f^{(1-m/2)} - D_0^{(1-m/2)}} \left\{ \left\{ \left[D_i^{(1-(m/2))} + \frac{D_f^{(1-(m/2))} - D_0^{(1-(m/2))}}{N_f} \right]^{((2-n)/(2-m))} \right. \right. \\ \left. \left. + \left[D_f^{(1-(n/2))} - D_0^{(1-(n/2))} \right] \frac{\Delta t}{T_C} \right\}^{((2-m)/(2-n))} - D_0^{(1-(m/2))} \right\} \quad (18)$$

$$\frac{\overline{\Delta t}}{T_C} = \frac{1}{D_f^{(1-m/2)} - D_0^{(1-m/2)}} \left\{ \left\{ \left[D_i^{(1-(m/2))} + \frac{D_f^{(1-(m/2))} - D_0^{(1-(m/2))}}{N_f} \right]^{((2-n)/(2-m))} \right. \right. \\ \left. \left. + \left[D_f^{(1-(n/2))} - D_0^{(1-(n/2))} \right] \frac{\Delta t}{T_C} \right\}^{((2-m)/(2-n))} - D_0^{(1-(n/2))} \right\} \quad (19)$$

$\overline{\Delta N}$ and $\overline{\Delta t}$ are, respectively, the equivalent cycle and equivalent hold time under pure fatigue loading and pure creep loading for a given damage calculated from a complete cycle with a fatigue part and a creep part.

Results of the Deterministic Model

Since Eq 17 is non-dimensional it is very easy to conduct parametric study. The cycles to failure $i = N_f$ can be obtained when $D \geq D_f$. Then, $T = N\Delta t$ will be the

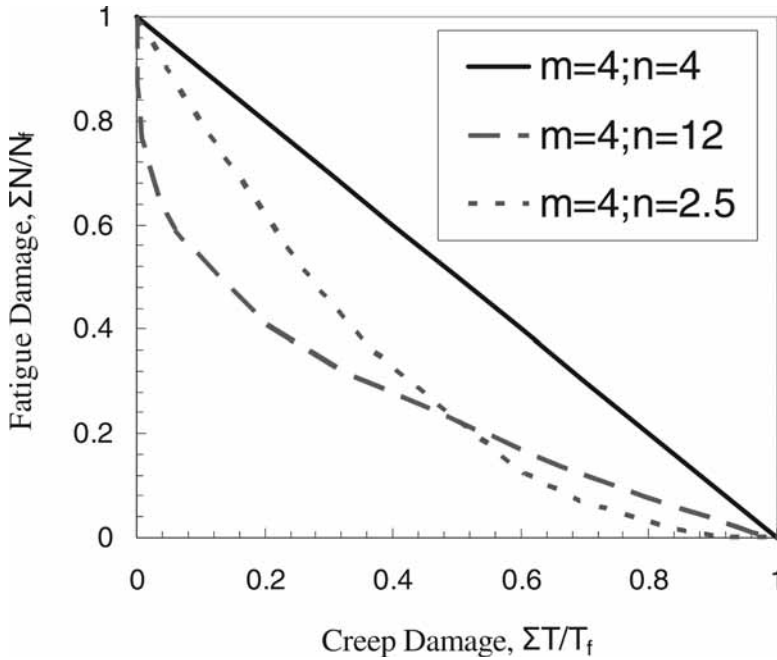


FIG. 3—Creep-fatigue failure function without introducing any interaction in the proposed model.

actual total hold time consumed for total cycle N . Therefore, the defined linear fatigue damage can be expressed as $\sum(N/N_f)$, and the ASME defined linear creep damage can be expressed as $\sum(T/T_c)$.

In the following damage calculations, $D_0 = 10^{-5}$, $D_f = 1.0$ and $N_f = 10\,000$ are used as an example to illustrate the general trend of the damage process. In damage mechanics the final damage is usually set in the range of $0.2 < D_f \leq 1$. However, the selection of the values does not make a noticeable change in life prediction because the damage process is always an accelerating process, especially at the late stage of failure. $m = 4$ is used here to represent the typical fatigue properties of engineering metallic materials. $n = 12$ is a typical value for creep behavior of power plant materials 12. The values other than $n = 12$ are also evaluated to see the influence of material properties on the so called ASME “creep-fatigue interaction” diagram.

Figure 3 shows three cases: $m = 4$, $n = 12$, $m = 4$, $n = 4$, and $m = 4$, $n = 2.5$. It is interesting to note that the creep-fatigue curve calculated from the numerical experiments are always concave, i.e., $\sum(N/N_f) + \sum(T/T_c) \leq 1$ except $m = 4$, $n = 4$, in which $\sum(N/N_f) + \sum(T/T_c) = 1$. The cases $m > n$ and $m < n$ only change the shape of the curve but do not change the concavity, Fig. 3. Overall, the failure envelope is asymmetric, which is different from conventional engineering assumptions 1,2. Furthermore, extensive numerical tests have confirmed the linear pattern for more general case of $m = n$ and the concave shape of the curves

for $m \neq n$. It should be noted that these phenomena are observed without introducing any creep-fatigue interaction.

It should also be noted that the conventional creep-fatigue interaction as adopted in ASME Case N-47 gives a different but physically inaccurate interpretation about the creep-fatigue curves and data: the black line in Fig. 3 is called the linear damage assumption, while any other curves or data below the black line are believed to be caused by the creep-fatigue interaction, e.g., $\sum(N/N_f) + \sum(T/T_C) = 0.6$ at the intersection point in Fig. 3. From Fig. 3 it is found that with the new crack growth based creep-fatigue model $\sum(N/N_f) + \sum(T/T_C) = 0.6$ can be approximately achieved without any assumption of creep-fatigue interaction for the given $m = 4$, $n = 12$.

Theoretically, if the above observations are true for any general cases, it could be proved mathematically. However, it is very difficult, if not impossible, because there is no closed-form solution for function $g(C)$, and the cycle number is implicitly included in Eq 17. However, some properties of the solution can be demonstrated by examining the Eqs 18 and 19 for a complete creep-fatigue cycle.

First, from Eqs 18 and 19 it is clear that $\overline{\Delta N}/N_f = 1/N_f$ and $\overline{\Delta t}/T_C = \Delta t/T_C$ for $m = n$, which indicates that the conventional linear summation rule holds for any individual complete creep-fatigue cycles, and eventually, the whole creep-fatigue life. In other words, the conventional linear summation rule can be observed for the case when $m = n$.

Second, by the perturbation theory, we assume that $n = m + \delta$, δ can be infinitesimal or any real values for the purpose in this paper. By this way, we can calculate the influence of the δ on the value of the ratio $(\overline{\Delta N}/N_f)(m, \delta)$ and $(\overline{\Delta t}/T_C)(m, \delta)$ as shown in Eqs 18 and 19. The influence of any infinitesimal change of δ on the shape of the performance function $g(C)$ can be examined by evaluating $\partial[\overline{\Delta t}/T_C]/\partial\delta|_{\delta=0^+}$, $\partial[\overline{\Delta t}/T_C]/\partial\delta|_{\delta=0^-}$, $\partial[\overline{\Delta N}/N_f]/\partial\delta|_{\delta=0^+}$, and $\partial[\overline{\Delta N}/N_f]/\partial\delta|_{\delta=0^-}$. Once again, it is analytically difficult if not impossible to get the analytical solutions even though the closed-form solution may exist. Numerical calculations listed below can shed some lights on the properties of the model.

Figure 4(a), 4(b) show, respectively, the fatigue damage ratio λ_N at $D_i = 0.00002$ and $D_i = 0.001$. The fatigue damage ratio λ_N is defined as the ratio between $1/\overline{N}$ and $1/N_f + \Delta t/T_C$. In Fig. 4, $dtn = \Delta t/T_C$ is used to describe the dimensionless time fraction. The equivalent cycle number $\overline{N}$ is calculated from pure fatigue mechanism for a given damage produced by a complete creep-fatigue cycle. Therefore, $1/\overline{N}$ can be considered as a unit damage caused by $\overline{N}$. $1/N_f + \Delta t/T_C$ is the unit damage caused by a complete creep-fatigue cycle according to the conventional damage accumulation rule. It is found that in almost all cases, the ratio $\lambda_N > 1$ for $m \neq n$, implicitly indicating that the combined damage caused by creep-fatigue is greater than that of the linear accumulation rule predicted. Figure 4(c), 4(d) show the creep damage ratio λ_T at $D_i = 0.00002$ and $D_i = 0.001$, respectively. The creep damage ratio λ_T is defined as the ratio between $1/\overline{\Delta t}$ and $1/N_f + \Delta t/T_C$. The equivalent hold time $\overline{\Delta t}$ is calculated from pure creep mechanism for a given damage produced by a complete creep-fatigue cycle. Therefore, $1/\overline{\Delta t}$ can be considered as a unit damage

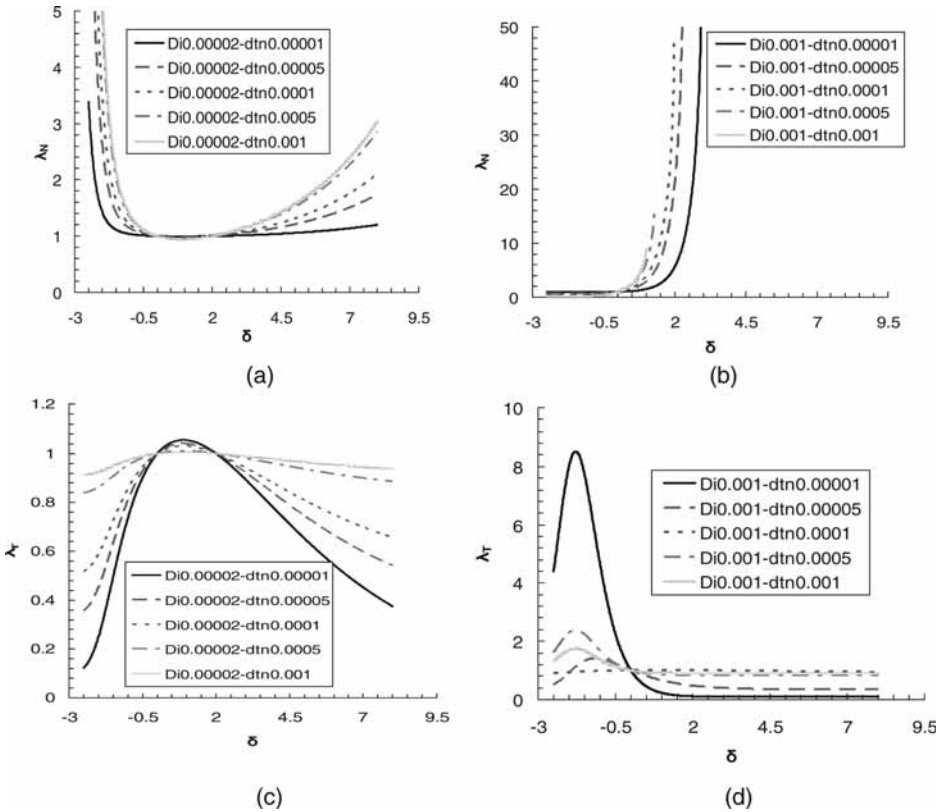


FIG. 4—The equivalent pure fatigue and creep damage for a complete creep-fatigue cycle. (a) Fatigue damage ratio λ_N at $D_i=0.00\ 002$; (b) fatigue damage ratio λ_N at $D_i=0.001$; (c) creep damage ratio λ_T at $D_i=0.00\ 002$; and (d) creep damage ratio λ_T at $D_i=0.001$

caused by $\overline{\Delta t}$. It is found that, generally, the ratio $\lambda_T < 1$ for $m \neq n$ except in higher damage stage, Fig. 4(d), for $m > n$. However, the influence of δ on the ratios is weaker than that shown in Fig. 4(a) and 4(b). In all of these cases, $1/N_f + \Delta t/T_C = 1/\overline{\Delta t} = 1/\overline{N}$ for $m = n$, indicating that the conventional linear damage accumulation rule holds for this case.

The major reason why conventional linear damage accumulation rule fails in the interpretation of creep-fatigue is that the crack growth for both fatigue and creep are not uniformly distributed over the life time of a component with $m \neq n$, see Fig. 2. A crack always grows slow at early stage and fast in the late stage, usually in exponential manner with time. Therefore, any attempts to use the so called “linear fatigue damage” $\sum(N/N_f)$ and “linear creep damage” $\sum(T/T_f)$ to interpret the creep-fatigue interaction simply does not give consistent prediction and is misleading.

Finally, this new creep-fatigue model is based only on crack growth mechanism relevant to creep-brittle materials. This assumption significantly simplify

the mathematical model and make a deep physical insight possible because both creep and fatigue phenomena can be described by only one parameter: the stress intensity parameter and, therefore, the creep-fatigue damage mechanism revealed by the model can be understood much easier than a more realistic but complex model. For real materials, especially the ductile materials, the damage is usually caused by the combination of crack initiation and growth. Crack initiation sometimes plays the major role in the failure of plain structures without notch effect and pre-cracks. However, like the crack growth phenomenon, crack initiation is also a continuous material degradation process, in which the damage evolves slowly first and then very fast at the late stage. Actually, alternative damage definitions considering the nonlinear phenomena for total life approach have been developed, e.g., $D = A(N_r/N)^\beta$ 13,14, where A and β are two constants and were found to be $A = 0.35$ and $\beta = 0.25$ for high strength steels. Similar non-linear damage for creep can be also found in literature. Therefore, the same trend observed here is expected to occur if crack initiation is further considered in the creep-fatigue model.

It is further noted that the creep-fatigue diagram adopted by ASME Case N-47 1 and API579/ASME FFS 2 is essentially based on ductile materials and total life, in which crack initiation may be more important. Therefore, the conclusions drawn from this paper should not be directly applied to these ductile materials without thorough benchmark test and evaluation. In the light of this paper, a sophisticated creep-fatigue damage model for ductile materials will need to be developed, as the present predictions are likely to be over conservative when applied to ductile situations.

Probabilistic Analysis of the Creep-Fatigue Model

Experimental data for creep and fatigue and creep-fatigue exhibit considerable amounts of scatter. One may then seek to calculate the probability that failure will occur after a total creep time T and after a total number of fatigue cycle N for a given materials and loading conditions. Analytically, it is simpler to deal with the survival probability, which is related to the failure probability by $P\{\text{survival}\} = 1 - P\{\text{failure}\}$. The survival probability is equal to the probability that the points C, F lie within the region, called A, below the performance function curve in Fig. 3. The probability of survival can be computed directly in terms of the underlying random variables T_j and N_i by [5]

$$\begin{aligned} P\{\text{Survival}\} &= P\{(C, F) \in A\} = P\{0 < C \leq 1, 0 < F \leq g(C)\} \\ &= P\left\{0 < C = \sum_{j=1}^l \frac{t_j}{T_j} \leq 1, 0 < C = \sum_{i=1}^k \frac{n_i}{N_i} \leq g\left(\sum_{j=1}^l \frac{t_j}{T_j}\right)\right\} = \int f(x_1 \cdots x_n) dx_1 \cdots dx_n \end{aligned} \quad (20)$$

where:

$f(x_1, \dots, x_n)$ = joint probability density function (pdf).

For independent variables, $f(x_1, \dots, x_n) = f(x_1) \cdots f(x_n)$. For creep and fatigue data, the log-normal distribution provides a good estimation of the damage. Thus, it is appropriate to assume that the joint pdf for the creep failure times and fatigue failure times at differing stress levels is jointly log-normally distributed. The general expression for the second-order pdf for creep-fatigue life with random variables T and N is

$$f(T, N) = \frac{1}{2\pi\eta_T\eta_N\sqrt{1-\rho_{TN}^2}} \exp \left\{ -\frac{1}{2(1-\rho_{TN}^2)} \left[\left(\frac{\ln T - \mu_T}{\eta_T} \right)^2 + \left(\frac{\ln N - \mu_N}{\eta_N} \right)^2 - \frac{2\rho_{TN}(\ln T - \mu_T)(\ln N - \mu_N)}{\eta_T\eta_N} \right] \right\} \quad (21)$$

where:

ρ_{TN} = correlation coefficient defined as $\rho_{TN} = E[(\ln T - \mu_T)(\ln N - \mu_N)] / \eta_T\eta_N$.

If $\rho_{TN} = 0$, $\ln T$, and $\ln N$ are statistically independent because Eq 21 may be factorized to give $f(T, N) = f(T)f(N)$.

In general, the solution of the multi-dimensional integral Eq 20 is complicated. The conditions for a multi-dimensional integral be reduced to simple integral equations are (1) all the random variables have a standard distribution function, (2) all the random variables are independent of each other, and (3) the performance function is a hyper plane, which means that in two-dimension (two random variables) the performance function is represented by a line, and in three dimension by a plane [15]. Since the performance function developed in this paper is non-linear, the direct integration is difficult and the Monte Carlo simulation will be used in this paper for probabilistic analysis.

First, based on the observation of calculations, the performance function $g(C)$ can be approximately represented by the following nonlinear function for $m < n$:

$$g(c) = 1 - C^\alpha \quad (22)$$

α can be obtained using the least square curve fitting method. The solution is $\alpha = \sum x_i y_i / \sum x_i^2$, where $x_i = \log(C_i)$ and $y_i = \log[1 - g(C_i)]$. In curve fitting the data related to $C = 0$ and $C = 1$ have to be deleted because of singularity introduced. Finally, $\alpha = 0.2837$ is obtained for $m = 4$ and $n = 12$, and $\alpha = 0.3$ is used in the following analysis. The fit curve and initial data curve are plotted in Fig. 5, and a good correlation is observed even though only one parameter α is used in the performance function. For $m = n$, we have $\alpha = 1$. While any mathematical function may be used to fit test data, the formula of the proposed performance function here has a simple form to model the creep-fatigue damage. $g(c) = (1 - C)^\alpha$ is found to provide a proper fit to the data for $m > n$.

Monte Carlo Simulation

Monte Carlo method has an attractive feature that it completely avoids numerical evaluation of integrals but instead works directly with the basic equation

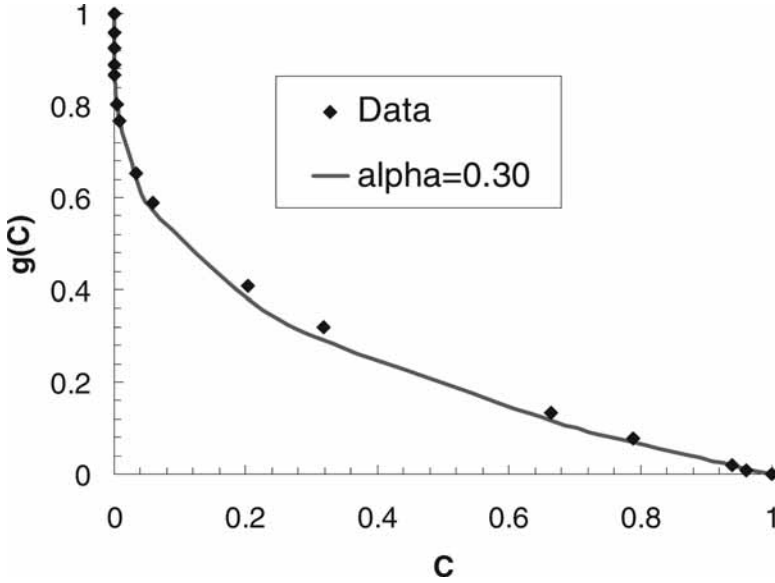


FIG. 5—Comparison between the original data curve and the predicted with $g(C)=1-C^{0.3}$.

defining the creep-fatigue failure envelope $F = g(C)$ [5]. For the general problem shown in Eq 20, in a single Monte Carlo simulation, one generates a total of k random cycle-to-failure N_i and l random failure times T_i , using a random number generator. For given values of n_i and t_j , failure occurs whenever.

$$\sum_{i=1}^k \frac{n_i}{N_i} > g\left(\sum_{j=1}^l \frac{t_j}{T_j}\right) \quad (23)$$

This process is repeated a large number of times, with the probability of failure being estimated as the number of instances of failure divided by the total number of simulations [5].

In application of Monte Carlo method to present problem, it is necessary to generate values that are normally distributed. There are many ways to do it and the most straightforward method is based on the inverse normal cumulative distribution function (cdf). However, there is no closed-form solution for the inverse normal cdf, Eq 24

$$\Phi\left(\frac{x-\mu}{\eta}\right) = \frac{1}{2} \left[1 + \operatorname{erf}\left(\frac{x-\mu}{\sqrt{2}\eta}\right) \right] \quad (24)$$

where, the error function $\operatorname{erf}()$ has to be calculated numerically that is very inconvenient for present problem. A number of approximate functions for a cumulative normal distribution have been reported, such as

$\Phi(Z) = 1/2\{1 + [1 - \exp(-2Z^2/\pi)]^{1/2}\}$ 16. In this paper the Box-Muller transform 17 is found to be more appropriate to generate the normally distributed random data. The basic idea of the Box-Muller transform is that a pair (Z_N and Z_T) of independent standard normally distributed (zero mean and unit variance) random numbers can be obtained by using a pair (U and V) of independent random numbers uniformly distributed on interval (0,1]. The uniformly distributed numbers can be generated in any related software. Then, the two standard normally distributed random variables Z_N and Z_T can be obtained by solving Eq 25

$$\begin{aligned} Z_N &= \sqrt{-2 \ln U} \cos(2\pi V) \\ Z_T &= \sqrt{-2 \ln U} \sin(2\pi V) \end{aligned} \quad (25)$$

Z_N and Z_T will both have the standard normal distribution, and will be independent.

The parameters in normal pdf $f(x) = 1/\sqrt{2\pi}\eta \exp\{-1/2[(x - \mu)/\eta]^2\}$ can be obtained from the standard normal pdf $f(Z) = 1/\sqrt{2\pi} \exp\{-(1/2)Z^2\}$ through the transformation $Z = (x - \mu)/\eta$. Then, the fatigue life and creep rupture life can be related to the normal distribution parameters by $x_N = \ln(N_R)$, $\mu_N = \ln(N_0)$, $x_T = \ln(T_R)$, and $\mu_T = \ln(T_0)$. Where N_0 and T_0 represent the mean cycles to failure and mean rupture time for pure fatigue and pure creep, respectively, for a given stress/strain range and a stress level. The subscript R here represents the random numbers.

Finally, we have

$$\begin{aligned} N_R &= \exp[Z_N \eta_N + \ln(N_0)] \\ T_R &= \exp[Z_T \eta_T + \ln(T_0)] \end{aligned} \quad (26)$$

With a given pair of mean, standard deviation in log form, and pseudo-random number generated, the random cycle and time can be calculated from Eq 26. Finally, the failure probability can be calculated for any fatigue cycle N and creep time T by comparing N/N_R with $g(T/T_R)$. The flow chart of the probabilistic analysis process is listed in Fig. 6.

The calculation parameters used in Monte Carlo simulations are summarized in Table 1.

For comparison purpose, the ASME adopted bi-linear creep-fatigue envelope is also studied in this paper with Monte Carlo method for the same parameters except the performance function, which is expressed as

$$g(C) = \begin{cases} 1 - 7/3C & \text{for } C \leq 0.3 \\ 3/7(1 - C) & \text{for } C \geq 0.3 \end{cases} \quad (27)$$

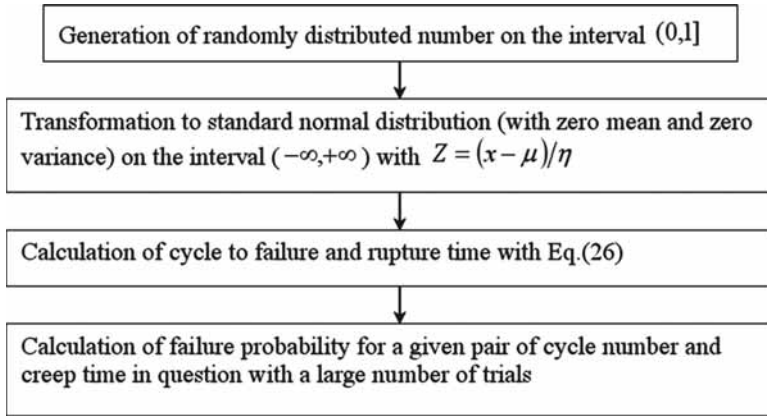


FIG. 6—Flow chart of creep-fatigue damage analysis procedure using Monte Carlo method.

The results of ASME adopted bi-linear model and the crack growth based model proposed in this paper are shown in Fig. 7(a) and 7(b), respectively. The probability of failure is plotted against the creep time T and total number of cycles N . Each point on the plots was the results of 10000 Monte Carlo trials. In Fig. 7, the maximum value limit for cycle is set as 15 000 while the maximum of creep failure time is set as 2000. For pure fatigue and pure creep, the Monte Carlo simulated failure probabilities at $N_0 = 10000$ and $T_0 = 1000$ are all close to 50 %. Both cases in Fig. 7 clearly show that the combined effects of creep and fatigue act to substantially increase the failure probability in comparison to the case of pure creep or pure fatigue.

It is expected that the probabilistic behaviors of the bi-linear and the nonlinear model is different because their deterministic counterparts are very different. This can be observed at low creep and low cycle regimes in Fig. 7, which is consistent with the deterministic observations that an introduction of a little bit creep can decrease the fatigue life significantly for $m < n$. These differences can be clearly distinguished by examining these curves in a more detailed comparison.

Figure 8(a) and 8(b) plot the failure probability comparison between the bi-linear model and the nonlinear model as a function of cycle number and creep time, respectively. Three cases with different creep time, $T = 1$, $T = 101$, and $T = 201$ are chosen for comparison in Fig. 8(a) while $N = 1$, $N = 751$, and $T = 1501$ are selected for comparison in Fig. 8(b). It is obvious that the nonlinear model predicts higher failure probability in these low cycle number and creep time regimes. For example, at $T = 1$, the failure probability predicted by the nonlinear model is 0.0056 for $N = 5251$, whereas, the predicted by using bi-linear model is 0.0012, which is almost five times lower than that predicted by the nonlinear model, for the same conditions. These observations are consistent with the deterministic observations. In summary, as the difference in deterministic models increases the difference in failure probability increases as well.

TABLE 1—The parameters used in Monte Carlo simulation.

Monte Carlo Trials	Pure Fatigue Parameters			Pure Creep Parameters		
	Mean Cycle to Failure N_0	Standard Deviation of Cycles to Failure η_N	Fatigue Crack Growth Exponent m	Mean Creep Time to Failure T_0	Standard Deviation of Time to Failure η_T	Creep Crack Growth Exponent n
10,000	10000	0.2	4	1000	0.3	12

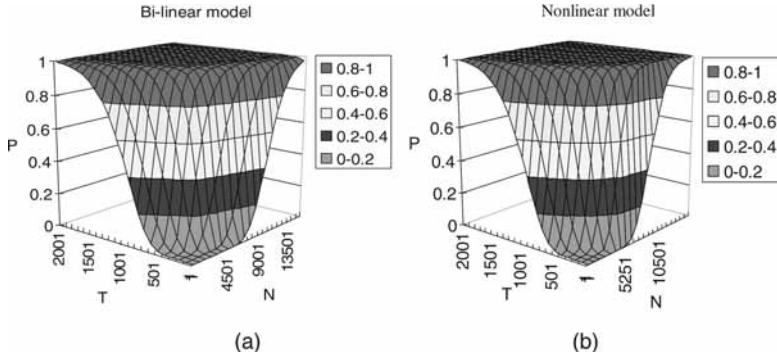


FIG. 7—Comparison of failure surfaces between ASME bi-linear performance function and the nonlinear function for one creep period and cycling at one stress range.

Discussions

At elevated temperatures, the processes of fatigue damage and creep damage can interact. The presence of creep damage is believed to accelerate the growth rate of the fatigue damage, and the presence of fatigue damage increases the growth rate of creep damage. From Fig. 3 we can see that the accelerated damage is more pronounced in the regimes near $\sum_{i=1}^k (n_i/N_i) = 0$ and $\sum_{j=1}^l (t_j/T_j) = 0$, indicating that even a little bit creep ($m < n$) or fatigue ($m > n$) can significantly increase the damage. This kind of effect has not been systematically investigated in experiments. Available data indicate that the “interaction effect” is not as significant as predicted in this paper. In some cases, data with $\sum_{i=1}^k (n_i/N_i) + \sum_{j=1}^l (t_j/T_j) > 1$ have been observed 6. All of these facts indicate that the real “creep-fatigue interaction” could retard the damage evaluation rather than only accelerate damage as commonly thought.

We have presented here a probabilistic lifetime assessment methodology for high temperature that takes into account the very scatter presented in both creep and fatigue life data. This scatter is usually sufficient large as to render existing deterministic techniques for life time prediction highly problematic and probabilistic life prediction techniques should be used 18. There are several techniques available for carrying out such calculations, but the Monte Carlo technique is especially attractive for this application, primarily due to its ease of use. However, the basic Monte Carlo method becomes prohibitively expensive and time consuming to achieve high accuracy in estimating a low failure probability, especially for complicated systems that may need a large amount of computational effort for each deterministic analysis. As a variety of numerical methods of solving structural reliability problems have already been proposed, an attempt is being made to improve the calculation times while retaining the acceptable accurate results for systematic probabilistic creep-fatigue analysis. Special attention will be paid to the FORM and SORM.

In this paper, the uncertainties are assumed to be from fatigue and creep lives, which can be easily obtained by standard tests and are usually available in

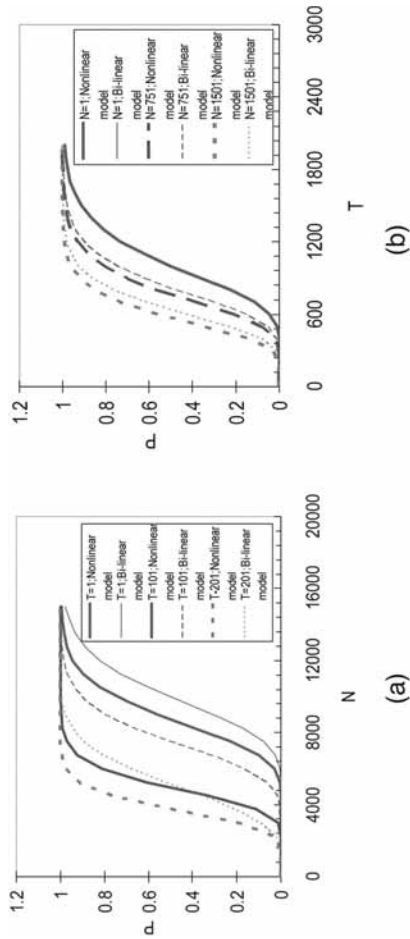


FIG. 8—Comparison of failure surfaces between ASME bi-linear performance function and present nonlinear function for some creep-fatigue combinations.

literature and reports. The uncertainty analysis could also be applied to Eq 17 directly. In this case, at least three parameters, i.e., m , n and D_0 can be subjected to uncertainty and can also be properly treated. The influence of D_0 on reliability is the most significant not only because it is usually unknown but also the final results are very sensitive to the value of D_0 , while the effect of D_f is negligible. In addition, only crack growth is investigated in this paper. In reality, both crack initiation and growth are subjected to uncertainties and should be considered in future works. Finally, the deterministic and the corresponding probabilistic analysis of creep-fatigue-oxidation should be investigated in the future because of its significance in high-temperature applications. Certainly, the time for calculation is more demanding for multiple variables.

Conclusions

In this paper, a new probabilistic creep/fatigue damage model procedure is developed to predict creep and fatigue crack growth in creep-brittle materials for sequential and periodical fatigue and creep loadings. It is found that the experimentally observed and ASME Case N-47 documented the so called "interaction diagram" could be obtained with the new proposed model or similar ones without introducing any creep-fatigue interaction effect. Uncertainties in creep and fatigue life of the material are considered and evaluated in this paper using the Monte Carlo simulation. The Monte Carlo simulation is an efficient way to derive results for probabilistic analysis assuming a log-normal relationship of the variables and has been implemented in the creep-fatigue model. It is recommended that sensitivity analysis using a deterministic calculation or a full probabilistic analysis would result in increased confidence in engineering design and fitness-for-service.

Acknowledgments

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PART II

**CREEP-FATIGUE INTERACTIONS
IN NI-BASE ALLOYS**

R. W. Neu¹ and D. J. Smith²

Modeling Creep-Fatigue Deformation of Ni-Base Superalloys Using Crystal Viscoplasticity

ABSTRACT: Crystal viscoplasticity enables one to correctly capture the anisotropic inelastic behavior of single crystal and directionally solidified alloys as well as the microstructural sensitivity in polycrystalline alloys. This paper reviews several physics-based crystal viscoplasticity models that have been established to predict the deformation response of Ni-base superalloys. The attributes of crystal viscoplasticity models necessary for predicting the creep-fatigue deformation are identified. Several new developments in constitutive modeling of Ni-base superalloys are discussed.

KEYWORDS: crystal plasticity, Ni-base superalloy, deformation mechanisms, finite element analysis

Introduction

Modeling the deformation response of Ni-base superalloys used in hot section components with the aim of predicting crack formation is challenging for several reasons. The response of these components depends on the cyclic loading and temperature cycles that often include extended dwells, and the interaction between environmental reactions and diffusion that affect the local composition and microstructure. These superalloys are often either a single crystal or directionally solidified (DS), and therefore both the elastic and plastic response is highly anisotropic. In addition, the heterogeneity of the microstructure includes local crystal

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¹The George W. Woodruff School of Mechanical Engineering, Georgia Institute of Technology, Atlanta, GA and School of Materials Science and Engineering, Georgia Institute of Technology, Atlanta, GA.

²The George W. Woodruff School of Mechanical Engineering, Georgia Institute of Technology, Atlanta, GA.

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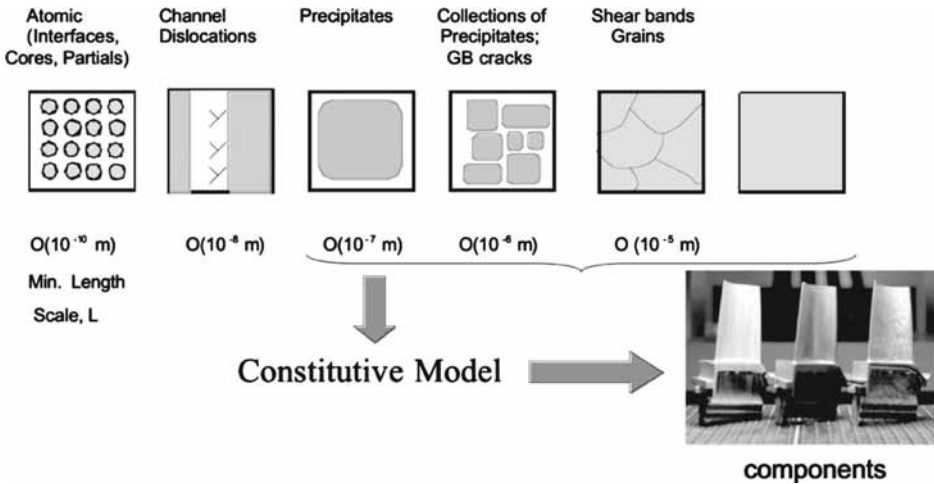


FIG. 1—Levels of hierarchy for γ - γ' Ni-base superalloys 1.

orientation of DS alloys and distributions of the precipitate phase coupled with compositional variations between the dendritic and interdendritic regions.

This is a multiscale problem as illustrated in Fig. 1. The elastic response depends on atomic arrangement and orientation while the inelastic response depends on the resistance of dislocation movement through the microstructure. The direction and planes of dislocation movements depend on the lower length scales while the coherent γ' precipitates provide resistance to shearing. In DS and polycrystalline alloys, the grain boundaries provide further resistance to dislocation movement. Ideally, all of these influences in dislocation movement should be captured, either explicitly or in a phenomenological manner.

One of the key microstructural features of Ni-base superalloys is the presence of small coherent L_{12} ordered (face-centered cubic (fcc)-like) precipitates γ' (primarily Ni_3Al , depending on composition) embedded in fcc γ matrix, mostly Ni. Mobile dislocations bypass γ' particles either by cutting or looping around them. Particle cutting occurs at lower temperatures and Orowan looping occurs at higher temperatures with deformation confined to the γ channels or looping followed by γ' shearing after dislocation density in the matrix increases significantly [2,3]. Therefore the flow stress inversely scales with the particle spacing and their morphology will also affect the response. Polycrystalline alloys also often contain carbides (MC and $M_{23}C_6$) where M is Ti, Cr, Ta, etc. that can help stabilize grain boundaries.

The small γ' particles are generally coherent with the γ matrix with small misfit strains that lead to additional flow resistance. Larger γ' , particular the primary γ' in fine-grained polycrystalline alloys like IN100, are not coherent. The flow stress in γ' increases with temperature up to about 800°C while that of γ tends to decrease as shown in Fig. 2. Therefore, alloys with a mixture of γ and γ' will exhibit the benefits of both phases. The amount of γ' in typical superalloys used in hot section components ranges from 50 to 70 %, which provides nearly uniform

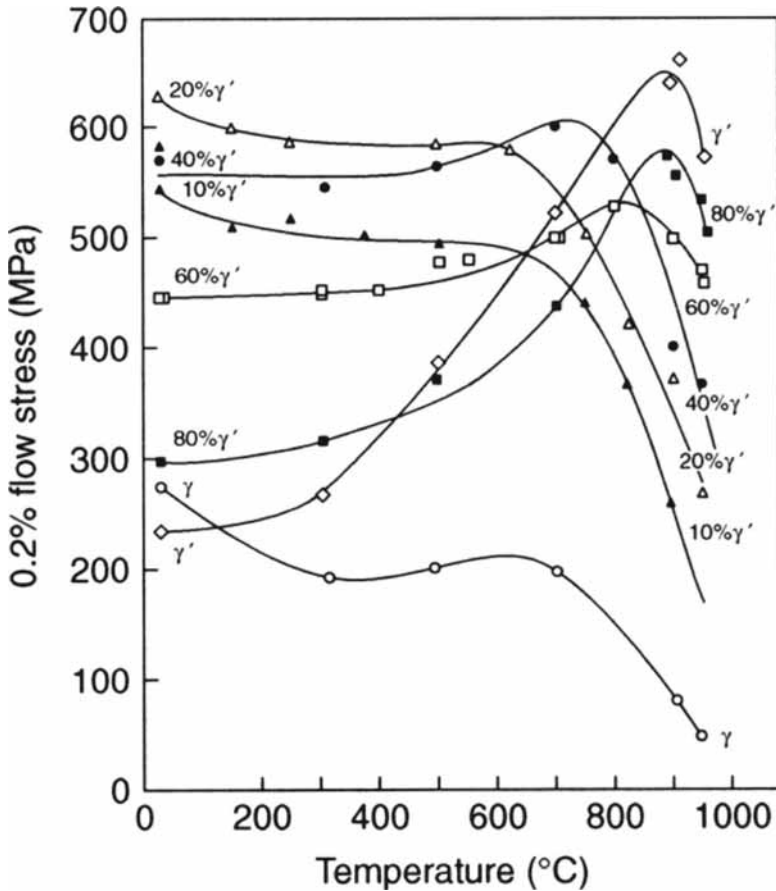


FIG. 2—Flow stress (here demarcated by the 0.2 % offset yield strength) as a function of temperature for different volume fractions of γ' in Ni-Cr-Al ternary alloys 3.

strength from room temperature to 800°C. These alloys are heat treated to give a bi- or tri-modal distribution of sizes of precipitate particles to optimize strength.

Superalloys are often exposed to sufficiently high temperatures such that the microstructure continues to evolve and age, influencing the constitutive response. One important mechanism is directional coarsening, which is also called rafting, that results in coarsened γ' particles with morphologies that depend on the direction of loading. The process in part is driven by the coherency misfit stresses resulting from negative lattice mismatch [4]. Material transport is driven by gradient in elastic strain energy density. Rafts parallel to the loading axis tend to decrease flow stress while rafts perpendicular to the loading axis increase flow stress for temperatures below 700°C [5]. The effect is reversed at higher temperatures. Generally, rafts do not form if temperature is 800°C or lower [4]. At 950°C, raft formation in CMSX-4 leads to increase in creep rate

due to cutting of γ' [4]. However, at 1100°C, raft formation leads to decrease in creep rate due to looping of γ' .

The aim of this paper is to review models and identify the important features of these models for predicting creep-fatigue response of Ni-base superalloys while capturing the influence of important microstructural attributes.

Framework of Continuum Crystal Viscoplasticity

Crystal viscoplasticity (CVP) became a feasible computational tool in the 1980s [6]. At each material point, the orientation of the crystal lattice is prescribed to track the collective dislocation motion on each slip system, explicitly defined by a slip plane and slip direction. For Ni-base superalloys, two families of slip systems may be activated in γ and γ' in fcc. As shown in Fig. 3, these include the 12 octahedral slip systems $\langle 110 \rangle \{111\}$ and six cube slip systems $\langle 110 \rangle \{001\}$. The octahedral slip occurs on the close-packed planes and is active over the entire temperature range. Cube slip is only active at higher temperatures.

Conventional CVP models including temperature dependence, which is needed for non-isothermal simulations, assume that the deformation gradient $\mathbf{F}$ with respect to the reference configuration, which maps particles $\mathbf{X}$ in the reference configuration to their positions $\mathbf{x}$ in the deformed configuration, can be decomposed into three parts [7]

$$\mathbf{F} = \frac{\partial \mathbf{x}}{\partial \mathbf{X}} = \mathbf{F}^e \mathbf{F}^p \mathbf{F}^\theta \quad (1)$$

where:

$\mathbf{F}^e$ = deformation gradient associated with elastic deformation and rigid body rotations,

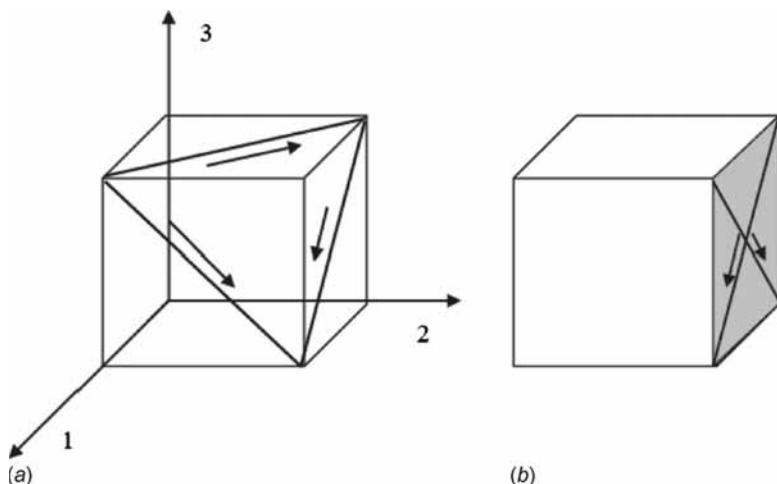


FIG. 3—(a) One of four octahedral slip planes showing slip directions and (b) one of three cube planes showing slip directions.

$\mathbf{F}^p$ = deformation gradient associated with inelastic deformation, here considering all mechanisms, and

$\mathbf{F}^0$ = deformation gradient resulting from temperature change due to thermal expansion and contraction.

The velocity gradient, the time derivative of the deformation gradient, is given by

$$\mathbf{L} = \dot{\mathbf{F}}\mathbf{F}^{-1} \quad (2)$$

In continuum crystal plasticity formulations, the $\mathbf{F}^p$ component is determined from the macroscopic inelastic velocity gradient which is the aggregate sum of the inelastic shear strain rate $\dot{\gamma}^\alpha$ on all possible slip systems α

$$\mathbf{L}^p = \dot{\mathbf{F}}^p \mathbf{F}^{p-1} = \sum_{\alpha=1}^{N_{\text{slip}}} \dot{\gamma}^\alpha (\mathbf{s}_o^\alpha \otimes \mathbf{n}_o^\alpha) \quad (3)$$

where:

$\mathbf{s}_o^\alpha$ and $\mathbf{n}_o^\alpha$ = slip direction and slip plane normal unit vectors, respectively, defined in the reference configuration for each slip system, and

N_{slip} = number of possible active slip systems.

For Ni-base superalloys, most models homogenize the fcc γ and L1₂ ordered γ' . For fcc, the octahedral slip systems are active over the entire temperature range. When loaded in the $\langle 001 \rangle$ direction, the Schmid factors on eight of the twelve octahedral slip systems are equivalent with magnitude of 0.408. As will be shown later when specific models are described, it is possible to take into account the effects of cross slip as temperature increases to account for the anomalous temperature dependence on flow stress as well as the tension/compression asymmetry in the flow behavior.

At higher temperatures (650°C and above), the macroscopic deformation can also have the appearance of flow along the cubic planes particularly in orientations where the Schmid factor on the octahedral planes is small [8]. There are six possible cube slip systems. For example, three of the six cube slip systems have high Schmid factor of 0.471 in the $\langle 111 \rangle$ direction, while the maximum Schmid factors on any of the octahedral slip systems is 0.272. The current thinking is that slip is still occurring on the octahedral slip systems via a zig-zag mechanisms in the γ channels between the γ' particles [8] though macroscopically, it appears slip occurs along the cubic planes.

Specific Flow Rules and Evolution Equations

The challenge of modeling creep-fatigue is the need for the constitutive model to capture the response of vastly different loading rates accurately, often over a range of temperatures, to simulate conditions of typical applications. Most CVP models employ a flow rule that relates the shear strain rate on each slip system to the viscous overstress defined by the resolved shear stress minus the back stress and threshold stress. The specific functional form is aimed to capture the

experimentally observed response over the required temperature range of applicability for a given state of the microstructure which is tracked by the internal state variables (ISVs).

The Shenoy et al. [9] CVP model captures many of the elements useful for modeling Ni-base superalloys and therefore will be discussed in some detail. The flow rule is assumed to depend on the viscous overstress, defined as

$$\tau_v^\alpha = |\tau^\alpha - \chi^\alpha| - \kappa^\alpha \quad (4)$$

where:

τ^α = resolved shear stress acting on slip system α

$$\tau^\alpha = \sigma : (\mathbf{s}^\alpha \otimes \mathbf{n}^\alpha) \quad (5)$$

where:

$\mathbf{s}^\alpha$ = vector along slip direction, and

$\mathbf{n}^\alpha$ = vector normal to the slip plane which are related to the respective unit vectors defined previously by $\mathbf{s}^\alpha = \mathbf{F}^e \mathbf{s}_o^\alpha$ and $\mathbf{n}^\alpha = \mathbf{n}_o^\alpha (\mathbf{F}^e)^{-1}$.

Both κ^α and χ^α are ISVs that are individually tracked on each slip system α . The threshold stress κ^α represents the resistance to plastic flow arising from statistical strengthening mechanisms associated with an increase of dislocation density, solid solution strengthening, Kear-Wilsdorf locks in γ' , and so on. The back stress χ^α captures the directional dependence on hardening. These features include internal stresses that develop from dislocation pile-up at obstacles such as γ' and other inclusions, grain or phase boundaries, dislocation cell walls, the collective elastic repulsion due to the split of dislocations due to the high antiphase boundary energy (APBE), cross slip, and so on.

The viscoplastic flow rule for each slip system

$$\dot{\gamma}^\alpha = \dot{\gamma}_o \Theta(T) \left\langle \frac{\tau_v^\alpha}{D^\alpha} \right\rangle^n \exp \left\{ B_o \left\langle \frac{\tau_v^\alpha}{D^\alpha} \right\rangle^{n+1} \right\} \text{sgn}(\tau^\alpha - \chi^\alpha) \quad (6)$$

includes two factors, a power law to capture creep-dominated flow and an exponential factor to capture power-law breakdown, which occurs when either strain rate increases or temperature decreases. This flow rule is an extension of the flow rule given in McDowell [10], where $\Theta(T)$ is a function that relates to the diffusivity, $\dot{\gamma}_o$ is a reference shearing strain rate coefficient that relates to the maximum strain rate associated with the limit on dislocation velocity, and D^α is the drag strength (or lattice friction) on each slip system. The parameters n and B_o control the steady creep rate and power law breakdown regimes. The Macauley brackets $\langle f \rangle$ represent $\langle f \rangle = f$ if $f \geq 0$, $\langle f \rangle = 0$ if $f < 0$. A plot of this flow rule function showing the influence of n is shown in Fig. 4. The n represents the power law creep exponent at lower stresses. As n increases, the power law creep regime becomes less active and the response approaches the near rate independent response.

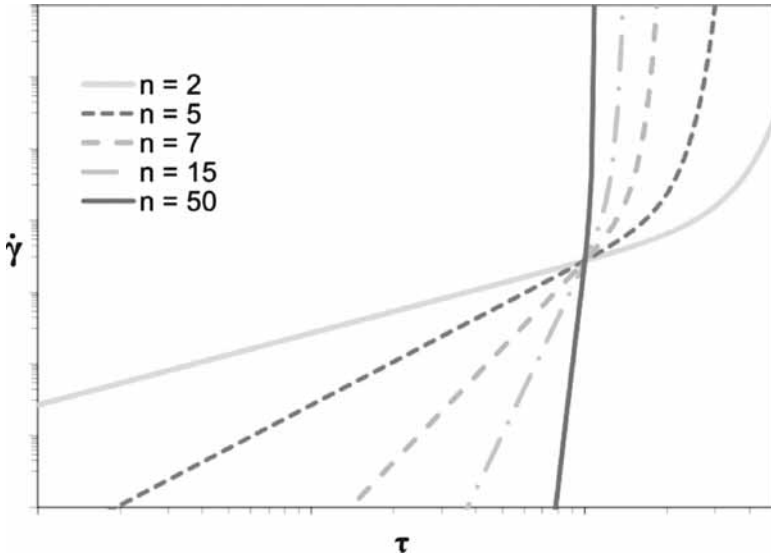


FIG. 4—Plot of Eq 6 for different values of n .

The diffusivity function generally has an Arrhenius form

$$\Theta(T) = \exp\left(-\frac{Q_o}{RT}\right) \quad (7)$$

where:

Q_o = activation energy of the barrier associated with the deformation mechanism,

R = universal gas constant, and

T = absolute temperature.

To cover a large temperature regime, this thermal activation factor often needs to be more complex to account for changes in the deformation mechanisms with temperature. The influence of temperature on the inelastic shear strain rate is shown in Fig. 5.

The drag strength D^z can also be treated as an ISV and evolve with changes in lattice friction resistance. Here, the drag strength scales to the temperature dependence of shear modulus in the simplest formulations

$$D^z = D_o \frac{\mu}{\mu_o} \quad (8)$$

where:

D_o = drag strength (or lattice friction) at absolute zero.

In the Shenoy et al. [9] model, both κ^z and χ^z evolve independently on each slip system with inelastic strain increments and thermal activation. The current

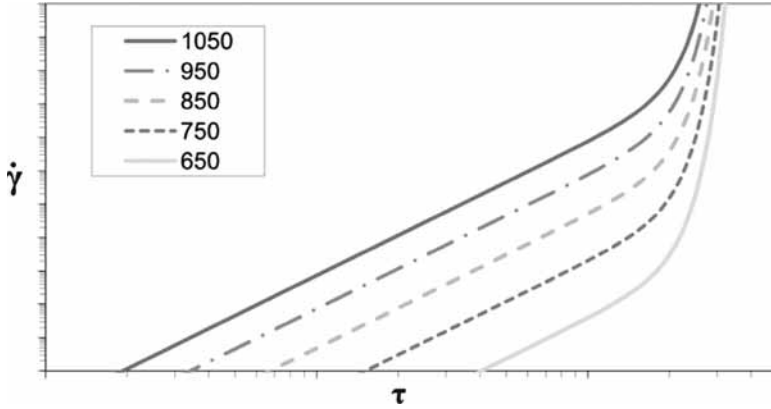


FIG. 5—Plot of temperature dependence of Eq 6, achieved through Eq 7, using temperature-independent parameters given in Ref 9 and $n=4$.

threshold stress involves two components scaled by the current shear modulus μ normalized by the shear modulus at absolute zero temperature μ_o

$$\kappa^\alpha = \frac{\mu}{\mu_o} (\kappa_c^\alpha + \kappa_e^\alpha) \quad (9)$$

The first component

$$\kappa_c^\alpha = \kappa_o^\alpha + h_{pe} \tau_{pe}^\alpha + h_{se} \tau_{se}^\alpha + h_{cb} |\tau_{cb}^\alpha| \quad (10)$$

captures the temperature-dependent resistance to dislocation movement. The last three terms of Eq 10 capture the non-Schmid effects that result in an increased resistance of dislocation movement through the γ' [11,12]. Here, the influence of shear stress acting on other cube τ_{cb}^α and secondary octahedral τ_{se}^α slip systems in the γ' lead to the formation of Kear-Wilsdorf locks of dislocations particularly as temperature increases as seen in Fig. 2. This functional form captures the anomalous increase in yield strength as temperature is increased [13,14]. The κ_e^α term in Eq 9 evolves according to a competition between dislocation storage, rearrangement, and annihilation mechanisms

$$\dot{\kappa}_e^\alpha = h_o \sum_{\beta=1}^{N_{slip}} q^{\alpha\beta} |\dot{\gamma}^\beta| - h_{kd} \kappa^\alpha \sum_{\beta=1}^{N_{slip}} |\dot{\gamma}^\beta| - h_s \Theta_{KS}(T) (\kappa^\alpha - \kappa_{th})^{r_s} \quad (11)$$

The first hardening term captures dislocation storage and includes the influence of dislocation interaction through the interaction matrix $q^{\alpha\beta}$ with $\alpha = \beta$ representing self hardening and $\alpha \neq \beta$ representing latent hardening due to interactions of dislocations on different slip systems. More complex forms of the interaction matrix are evaluated in Ref [15, though it appears not to be critical for modeling Ni-base superalloys for two reasons: At lower temperatures

strength is dominated by particle shearing and hence slip tends to remain planar with little interaction [1]. At higher temperatures, diffusion tends to mitigate any interactions. The second term represents dynamic recovery which controls the saturation of hardening controlled by the competition between storage and annihilation. The third term represents diffusion-controlled static thermal recovery that controls the longer term creep and stress relaxation response. Several material parameters (κ_o^α , h_{pe} , h_{se} , h_{cb} , h_o , $q^{\alpha\beta}$, h_{kd} , h_s , κ_{th} , r_s) and the thermal activation for static recovery $\Theta_{ks}(T)$ need to be defined, often as a function of temperature, in the full implementation of this model. The initial value of κ_e^α also needs to be defined.

The evolution equation for back stress is given by

$$\dot{\chi}^\alpha = h_\chi |\dot{\gamma}^\alpha| \text{sgn}(\tau^\alpha - \chi^\alpha) - h_{\chi d} \chi^\alpha |\dot{\gamma}^\alpha| + \left(\frac{1}{R_\chi} \frac{\partial R_\chi}{\partial T} + \frac{1}{h_{\chi d}} \frac{\partial h_{\chi d}}{\partial T} \right) \chi^\alpha \dot{T} - h_{\chi s} \Theta_{\chi s}(T) |\chi^\alpha|^{r_{\chi s}} \quad (12)$$

where the first term accounts for increased resistance of shear on the slip system, the secondary term is the dynamic recovery controlling the saturation, the third term is a temperature-dependent term to ensure flow stress scales correctly with temperature changes, and the last term is a static thermal recovery term. In this expression, parameters that need to be established include h_χ , $h_{\chi d}$, $h_{\chi s}$, and $r_{\chi s}$, the thermal activation for static recovery of the back stress, $\Theta_{\chi s}(T)$. The ratio $R_\chi = h_\chi/h_{\chi d}$ represents the saturation level of the back stress and h_χ controls the rate of hardening. The initial value of χ^α also needs to be defined. This equation captures the Bauschinger effect primarily due to dislocation pile-ups at the γ - γ' interfaces and is important for capturing slip reversibility for creep-fatigue loading. The static thermal recovery term is important for modeling the stress relaxation during long-term dwells.

To illustrate the significance of orientation on the response with and without dwells, the hysteresis responses for four orientations predicted by the model using constants for GTD-111 at 871°C given in Ref [9] are shown in Fig. 6. For comparison, all were strain-control, with the strain amplitude selected so that the response in $\langle 001 \rangle$ is nearly elastic. Clearly, the influence of the elastic anisotropy is critical in controlling the amount of cyclic inelastic strain. Since these parameters were calibrated to a DS alloy, there is no direct comparison to experiments, but the trends in yield behavior are consistent with single crystal experiments in these orientations [3,16]. Adding a compressive creep dwell increases the cyclic inelastic strain. The increase is particularly notable in the $\langle 001 \rangle$ orientation which is nearly elastic under continuous cycling. Also, because the strength in the $\langle 001 \rangle$ is larger, the maximum stress for that case increases when a compressive dwell is imposed.

Alternative Flow Rules

An alternative flow rule to capture creep-fatigue response [1,17] is based on the two-term flow rule given by Chaboche [18]

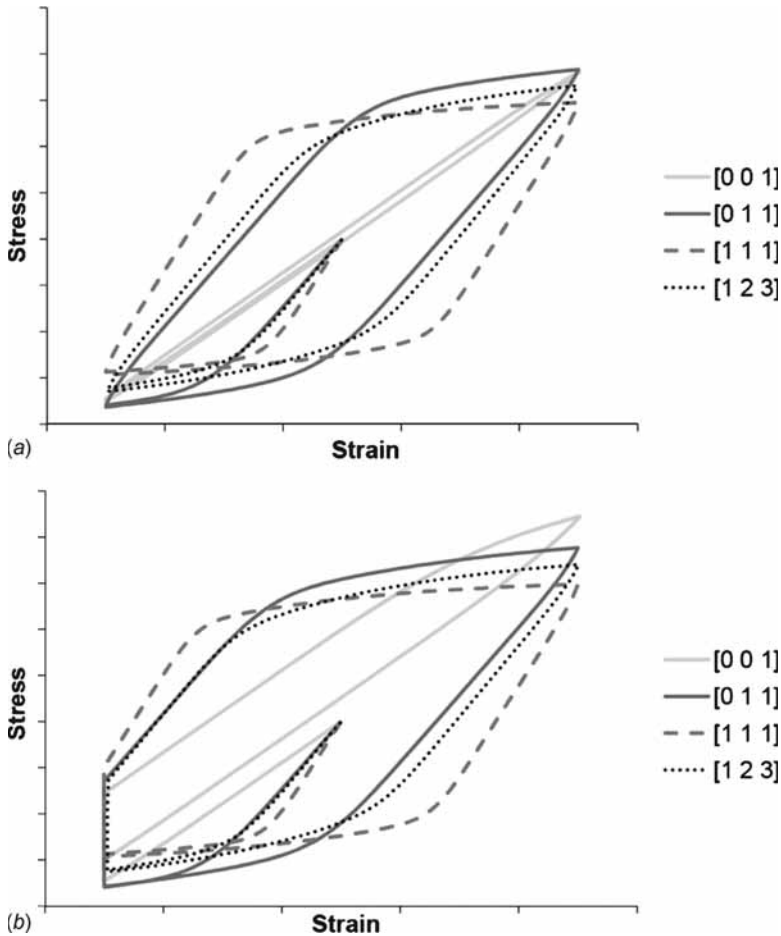


FIG. 6—Influence of orientation on strain-controlled constant amplitude response (a) without dwells; (b) with a dwell in compression.

$$\dot{\gamma}^{\alpha} = \left[\dot{\gamma}_o \left\langle \frac{|\tau^{\alpha} - \chi^{\alpha}| - \kappa^{\alpha}}{D^{\alpha}} \right\rangle^{n_1} + \dot{\gamma}_1 \left\langle \frac{|\tau^{\alpha} - \chi^{\alpha}|}{D^{\alpha}} \right\rangle^{n_2} \right] \text{sgn}(\tau^{\alpha} - \chi^{\alpha}) \quad (13)$$

Here, the first term is aimed at capturing the shorter term cyclic behavior where the threshold stress relates to the yield strength. The second term is intended to capture the longer term creep and stress relaxation response associated with dwells. With both terms, cyclic loadings with dwells can be modeled.

The exponents control the slope of the inelastic shear strain rate—shear stress response. In Fig. 7, the ratio n_1/n_2 is held fixed at 1.67 and the magnitude of the exponent is varied. Increasing the exponent reduces the time-dependent response. If instead, the long-term creep term is active with $n_2 = 3$ and n_1 is

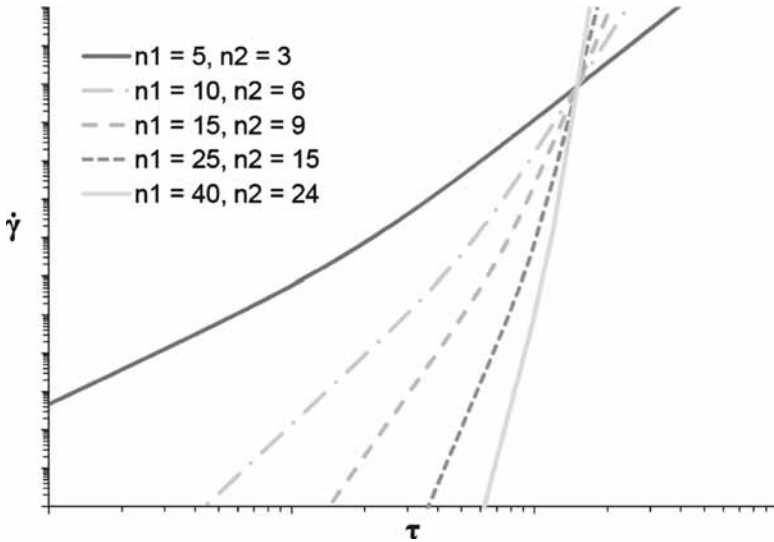


FIG. 7—Plot of Eq 13 for different values of n_1 for $n_1/n_2=1.67$.

increased, the transition from power law creep to near rate-independent response becomes more distinct as shown in Fig. 8.

When B_o and $\dot{\gamma}_1$ are set to zero in Eqs 6 and 13, respectively, both flow rules reduce to the single power law flow rule used by Meric et al. [16], which was one of the first CVP models for modeling the higher temperature response of single crystal Ni-base superalloys.

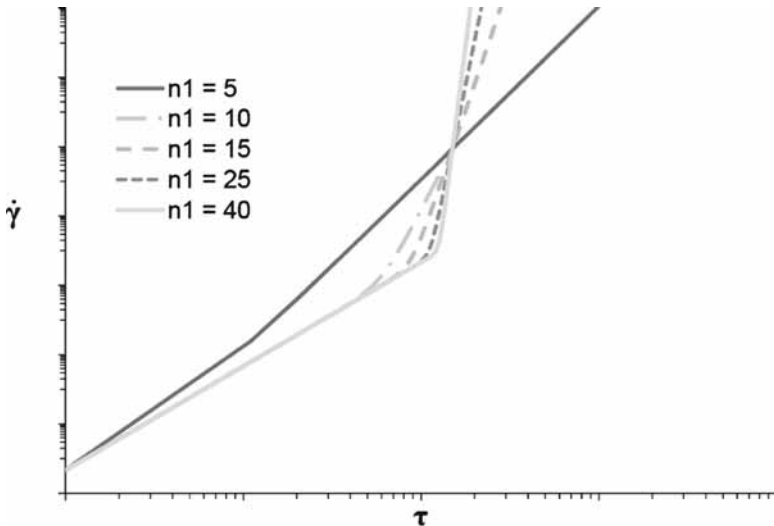


FIG. 8—Plot of Eq 13 for different values of n_1 for $n_2=3$.

Another CVP flow rule that has been used to model Ni-base superalloys [19] is based on the quasi-physical model proposed by Busso and McClintock [20]

$$\dot{\gamma}^{\alpha} = \dot{\gamma}_o \exp \left\{ -\frac{F_o}{kT} \left\langle 1 - \left\langle \frac{|\tau^{\alpha} - \chi^{\alpha}| - \kappa_o^{\alpha} \frac{\mu}{\mu_o}}{D^{\alpha}} \right\rangle^p \right\rangle^q \right\} \text{sgn}(\tau^{\alpha} - \chi^{\alpha}) \quad (14)$$

where:

p and q = flow exponents.

This flow rule is based on a stress-dependent activation energy instead of a multiplicative diffusivity parameter. F_o is the total free energy necessary to overcome the lattice resistance without the aid of external work (i.e., from thermal fluctuations). This model also has two variables that depend on the current structure, the threshold stress, κ^{α} , representing the athermal component of the flow stress, and the back stress, χ^{α} .

Alternative Evolution Equations

In “quasi-physical” models, the dislocation density ρ^{α} is used as one of the ISV, often replacing the threshold stress as the ISV [1,17,19,20]. In this approach, the slip resistance depends on the initial reference critical resolved shear stress (CRSS) κ_o^{α} and hardening is attributed to interactions of statistically stored dislocations (SSDs) ρ^{α} along each slip system typically following a generalized Taylor relation:

$$\kappa^{\alpha} = \kappa_o^{\alpha} + \lambda \mu_{\text{eff}} b^{\alpha} \left(\sum_{\beta=1}^{N_{\text{slip}}} q^{\alpha\beta} \rho^{\beta} \right)^{1/2} \quad (15)$$

where:

λ = material statistical parameter that accounts for spatial arrangements of the dislocation population, and

b^{α} = magnitude of the Burgers vector in addition to parameters already defined.

When planar slip is prevalent, the interactions between slip on different systems is minimal, and Eq 15 reduces to [1]

$$\kappa^{\alpha} = \kappa_o^{\alpha} + \lambda \mu_{\text{eff}} b^{\alpha} \sqrt{\rho^{\alpha}} \quad (16)$$

The slip in Ni-base superalloys tends to be planar in nature particularly at lower temperatures with strength dominated by particle shearing.

Since κ_o^{α} is related to the initial reference CRSS, it strongly depends on microstructure attributes such as the average grain size, precipitate volume fraction and size distribution. A possible functional form for lower temperature (i.e., 650°C) behavior is given by Shenoy et al. [1]. For IN100, it is a function of the volume fraction and size of the primary γ' , the volume fraction and size of

the secondary γ' , and volume fraction and size of the tertiary γ' , and the grain size. Details of this microstructure-sensitive model are described in Ref [1.

One form of the evolution of the dislocation density is given by [17]

$$\dot{\rho}^{\alpha} = \sum_{\beta=1}^{N_{\text{slip}}} h_o q^{\alpha\beta} (Z_o + k_1 \sqrt{\rho^{\beta}} - k_2 \rho^{\beta}) |\dot{\gamma}^{\beta}| - h_s \Theta_{ps}(T) [\rho^{\alpha} - \rho_{th}^{\alpha}]^{r_s} \quad (17)$$

where Z_o depends inversely on the spacing of the precipitates (see Ref [17 for details) and the terms $k_1 \sqrt{\rho^{\beta}}$ and $k_2 \rho^{\beta}$ represent the SSD generation and recovery, respectively. The second static recovery term describes thermally assisted diffusional recovery needed to model longer term creep and stress relaxation response [10]. Parameters h_o , k_1 , k_2 , h_s , r_s , and ρ_{th}^{α} need to be determined from experiments.

The back stress is a phenomenological ISV that captures the effect of the heterogeneous distribution of dislocations often located at γ' precipitate interfaces among other influences of the arrangement of dislocations. One possible form of the back stress evolution is given by

$$\dot{\chi}^{\alpha} = C_{\chi} [\eta \mu_{\gamma} b \sqrt{\rho^{\alpha}} \operatorname{sgn}(\tau^{\alpha} - \chi^{\alpha}) - \chi^{\alpha}] |\dot{\gamma}^{\alpha}| \quad (18)$$

where:

η = phenomenological variable that represents a measure of the rate of generation of geometrically necessary dislocations (GNDs) to the total rate of dislocation density (SSDs and GNDs).

It is given by [17]

$$\eta = \frac{\eta_o Z_o}{Z_o + k_1 \sqrt{\rho^{\alpha}}} \quad (19)$$

where:

η_o = constant that controls the saturation level of back stress.

To better represent the higher temperature response of Ni-base superalloys, the back stress evolution described by Eq 18 will likely need to be further modified to include static thermal recovery and possibly microstructure evolution.

Enhanced Models to Capture Additional Mechanisms and Microstructural Features

Coupled CVP and Mass Diffusion Model

The microstructure near the surface can also evolve due to the formation of corrosion compounds that deplete the near surface layers of oxide forming metals such as Al, shown in Fig. 9. Since Al is a primary component in forming the ordered γ' phase, the depletion of Al results in a γ' -depleted zone near the surface. The depth of this depleted zone depends on the diffusion kinetics, shown in Fig. 10. Clearly, if the volume fraction of γ' substantial changes, the deformation

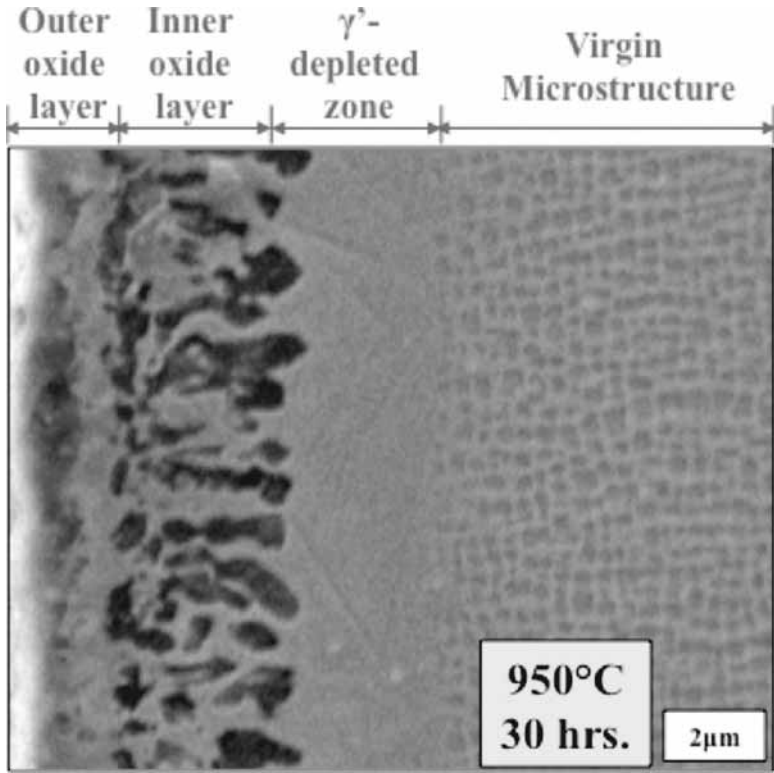


FIG. 9—Surface layers of CM247LC after high temperature exposure 21.

response changes in the surface layer. One way to deal with this is to couple CVP with a mass diffusion model to determine the γ' -depleted region [22,23].

Equation 14 is used as the flow rule. The back stress follows the standard hardening-dynamic recovery form, e.g., Eq 12. The threshold stress variable is “quasi-physical” written in terms of the dislocation density of the form given by Eq 15. The evolution of dislocation density includes hardening and dynamic recovery terms

$$\dot{\rho}^{\alpha} = \frac{1}{b^{\alpha}} \left(\frac{1}{l_{app}^{\alpha}} - h_d \rho^{\alpha} \right) |\dot{\gamma}^{\alpha}| \tag{20}$$

The rate of hardening increases with a reduction in the apparent mean dislocation spacing (dislocation mean free path)

$$l_{app}^{\alpha} = \left(\sum_{\beta=1}^{N_{slip}} q^{\alpha\beta} \rho^{\beta} \right)^{-1/2} \hat{f}_{\kappa}(D, v_f, T) \tag{21}$$

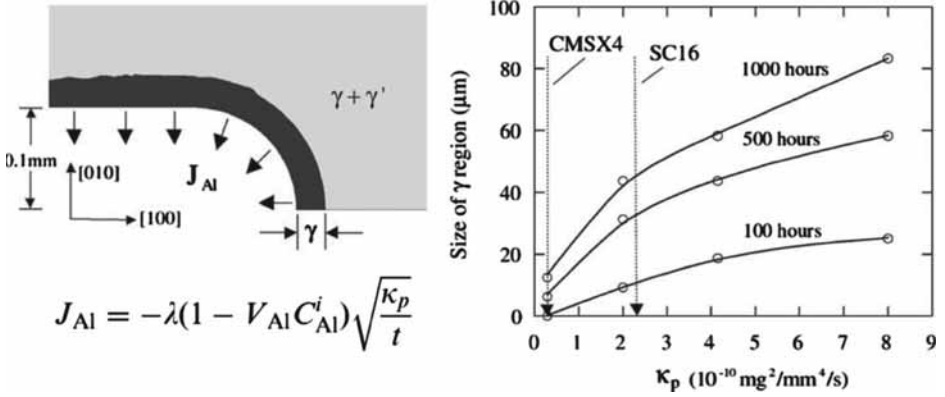


FIG. 10—Model description for the formation of a γ' -depleted layer 23.

which depends on the dislocation density on each slip system modified by a function $f_{\kappa}(D, v_f, T)$ to account for changes in the size D and volume fraction v_f of γ' due to mass diffusion. The initial dislocation density also depends on the size and volume fraction of γ'

$$\rho_o^z = \hat{\rho}_o^z(D, v_f, T) \quad (22)$$

The functional forms providing the diffusion influence are obtained either experimentally or through micromechanics modeling [22]. The use of computational thermodynamics and diffusional simulations such as ThermoCalc and DICTRA [24] to account for composition changes due to mass diffusion also shows promise of being a viable method to obtain these functional forms for a coupled analysis including prediction of directional coarsening [25].

Diffusion-Mediated Microtwinning “Creep”

Recently it has been observed [26,27] that creep in Ni-base superalloys when the stress remains below the macroscopic yield may be due to a diffusion-mediated reordering of the crystal structure that then promotes microtwinning leading to creep strain. This creep mechanism only operates at low to intermediate stress well below the macroscopic yield and ceases to operate if ever the material is loaded beyond the macroscopic yield because of the extreme disorder due to dislocation density increase. To capture this mechanism, Song and McDowell [28] added a second term to the flow rule

$$\dot{\gamma}^z = \dot{\gamma}_{vp}^z + \dot{\gamma}_{mt}^z \quad (23)$$

where the first term represents a conventional viscoplastic deformation mechanism involving dislocation nucleation, movement, and rearrangements (e.g., Eq 6, Eq 13, or Eq 14), while the second term captures the microtwinning “creep” mechanism associated with the octahedral slip systems. The details of

the flow rule are given in Ref [28]. At higher strain rates and stress levels in the vicinity of the macroscopic flow stress and above, where threshold stress and back stress evolve substantially, the first term dominates. So this term is only important if stresses in component always remain well below the macroscopic flow stress.

Explicit modeling of microstructure features

The influence of the size and distribution of the γ' particles has a first order influence on the response, yet they are at a length scale often considerably smaller than the grain scale. Homogenization procedures have been found to be successful. Explicit models of the γ matrix and the γ' precipitates in single crystal Ni-base superalloys have been used to predict the influence of these interactions [17,22,29–31]. These models tend to have simplifying assumptions, so they are not generally practical for component modeling. Instead, they are useful to explore effects of variations of microstructure features and attributes while maintaining other attributes fixed, something that is challenging to do experimentally due to the coupling of influences, and would likely be extremely costly if even feasible. Extreme variations of microstructure can be studied to determine how they will affect the extreme mechanical behavior and ultimately impact variation in performance. For example, Kumar et al. [31] determined the distribution of cyclic strain accumulation in the γ – γ' structure to isolate the influence of the attributes of the precipitates (e.g., their size and volume fraction) on the intrinsic fatigue behavior of the microstructure. In addition, these studies can provide functional forms for higher length scales, such as those used in establishing the influence of microstructure parameters in Eqs 21 and 22 [22].

At these length scales, the higher order gradients associated with GNDs can be studied to explore the role of γ and γ' size and distributions on the deformation response [29,30,32]. For example, Forest et al. [30] showed that lattice curvature in the matrix channel is one explanation for the strengthening of the γ – γ' . However, modeling γ – γ' structure explicitly with finite elements is problematic when the interfaces artificially become serrated due to the finite elements.

In addition, other features in the microstructure, such as non-metallic inclusions [33] and pores [22], embedded in a CVP matrix can be explicitly modeled to determine how they affect the localization of the inelastic strain. The fatigue drivers are related to the magnitude of this local cyclic inelastic strain.

Directionally Solidified and Polycrystalline Alloys

For DS and polycrystalline alloys, the grain interactions play a critical role in the deformation response. The polycrystalline alloy is modeled by applying a deformation history to a representative volume element containing several grains. For capturing just the macroscopic response of polycrystalline alloys while keeping the simulation time small, the Taylor constraint can be used to approximate the intergranular interactions [9]. For the latter, several single element simulations with different orientations corresponding to the orientation distribution are run assuming all of the elements experience the same

deformation and the stress depends on the aggregate response. For DS alloys, 30 grains were found to be sufficient to represent the macroscopic response [9]. On the other hand, to explicitly model the grain interactions including size and morphology effects, the individual grains need to be represented in a finite element mesh using CVP. With a meshed microstructure, the influence of intergranular constraints on the local response can be captured to obtain a real picture of the internal distribution of inelastic strain and stress state [34,35] and the model parameters for DS and polycrystalline alloys can be refined [9]. The predicted response is always considerably more heterogeneous even in an alloy that is macroscopically isotropic. The intergranular constraints strongly contribute to the kinematic hardening behavior [36]. In addition, these intergranular constraints as well as the presence of large carbides or other nonhomogeneous microstructural features suggest that the driver for crack formation is related to the localized accumulation of ratchet strain [37,38] that may be enhanced under creep-fatigue conditions.

The elastic properties of Ni-base superalloys are highly anisotropic. For example, the uniaxial modulus varies from a minimum of 124 GPa in the $\langle 001 \rangle$ orientation and a maximum of 310 GPa in the $\langle 111 \rangle$ orientation for PWA 1480 [39], which is clearly observed in the uniaxial response shown in Fig. 6. This leads to formation of internal stresses due to the intergranular constraints even when stresses remain below the threshold for dislocation movement. Therefore, it is critically important to use the correct crystal-scale elastic properties since these internal stresses contribute to τ^z , the driver for inelastic deformation. When homogenizing γ and γ' , it is usually sufficient to specify elastic constants C_{11} , C_{12} , and C_{44} , derived from single crystal measurements, and if nonisothermal simulations are desired, they need to be specified as a function of temperature.

Implementation in Finite Element Codes

Since CVP formulations are rooted in continuum mechanics, they can readily be adopted and used in finite element codes, for example through User MATERIAL subroutines (UMATs) in ABAQUS [40]. The UMAT is called at each integration point and is given the current state of stress, strain, and temperature (if non-isothermal), current values of the ISVs, and the deformation gradient and time increment. It then returns updated values of the stress, ISVs, and the Jacobian matrix $\partial\Delta\sigma/\partial\Delta\epsilon$ used by the finite element code to help obtain global convergence. Generally, a fully implicit integration scheme is used, coupled with a Newton–Raphson method to find the roots of the equations representing the shearing rates on each slip system [41,42]. Sub-incrementation schemes to ensure convergence are also often employed. Detailed descriptions of typical algorithms and implementation procedures are provided in Refs [29, 43–45].

Finite element simulations using CVP can be used to determine the influence of microstructure in fatigue crack formation [46]. They have been used to predict the location and likelihood of cracks forming at notches in single crystal [47,48] and DS alloys [49]. They have been used to determine crack tip fields to predict

creep-fatigue crack growth of single and large grained alloys [50]. And they have utility in designing creep-fatigue resistant single crystal turbine blades [51].

Parameter Determination

Deriving and hypothesizing the theoretical relationships is only one aspect of implementation. The other, and perhaps more challenging, is the determination of the parameters. Most of the physics-based parameters cannot be measured directly. Hence, an inverse approach must be employed. For isothermal models, this is usually feasible through a judicious iterative process. However, if one desires a temperature-dependent model, then the challenge of this task increases by an order of magnitude.

In order to calibrate both the octahedral and cube slip systems, experiments must be conducted in different orientations. First, the deformation response when only the octahedral slip systems are activated are conducted and used to calibrate those parameters. For single crystal alloys, the $\langle 001 \rangle$ orientation is suitable and for DS alloys, the longitudinal orientation will work. To determine the cube slip parameters, an orientation where the Schmid factors are considerably higher on the cube slip systems compared to the octahedral system, such as $\langle 111 \rangle$, needs to be selected. For DS alloys, loading transverse to the grain growth direction will activate both cube and octahedral slip.

Experiments need to be conducted at multiple temperatures since the deformation mechanisms are highly temperature dependent. If the aim is to use the model for low cycle creep-fatigue, then at each temperature cyclic loadings should be applied at different strain rates with sufficient inelastic strain to clearly identify yielding and hardening response. An example of a calibration experimental program is shown in Fig. 11. This protocol will often work with Ni-base superalloys that are cyclically stable, but may need to be modified for materials that exhibit considerable cyclic hardening or softening. An iterative process coupled with using optimization software such as Isight [52] or Model-Center [53] is used to identify the material parameters [28,54].

Reduced Order Models

Because of the need to evaluate multiple flow rules at each integration point, one for each active slip system, the computational expense for CVP is considerably greater than conventional J_2 -based viscoplasticity or unified creep-plasticity models that have been successfully implemented for modeling the creep-fatigue response of polycrystalline Ni-base superalloys [18,55,56].

If one is interested in knowing how changes in the microstructural attributes, such as grain size and orientation distribution, γ' size and distributions, affect the response, different microstructure realizations can be explicitly explored using a CVP model. Then the output responses of these simulations can be used to calibrate the parameters of either a conventional J_2 -based viscoplasticity model for polycrystalline alloys or a transversely isotropic viscoplasticity model (e.g., Ref [57] for DS alloys. The relationships between the

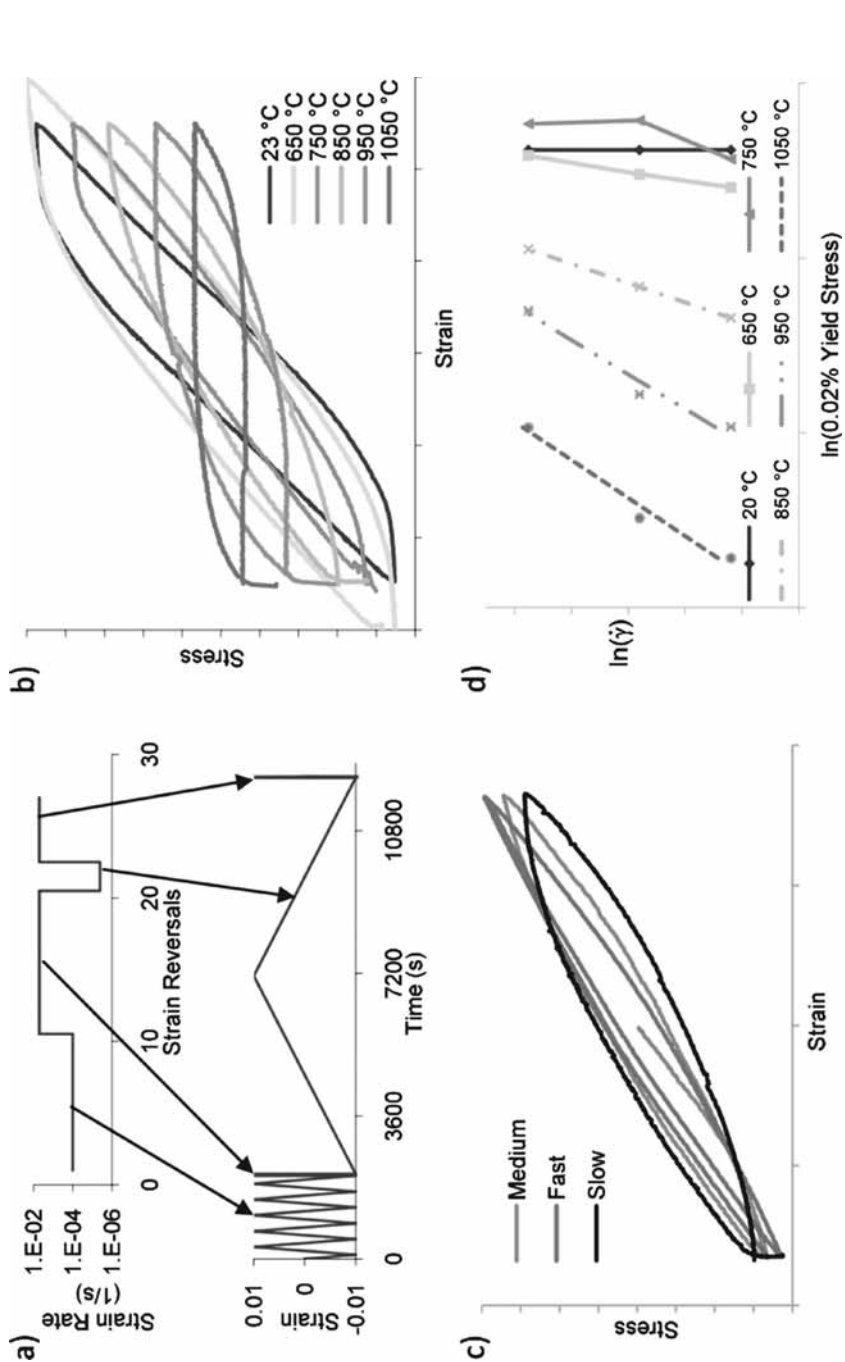


FIG. 11—An example of (a) a test protocol for generation of calibration data and the resulting response data showing (b) effect of temperature during the slow strain rate cycle, (c) effect of strain rate at 850 °C, and (d) the viscoplastic flow.

microstructural attributes and the material parameters of the reduced order model are determined through either an influence function or artificial neural network approach [1,58].

Summary and Prospects

Creep-fatigue and thermomechanical fatigue life prediction requires highly accurate cyclic deformation models. The development and improvements in CVP models will enable the engineer and designer to predict the influence of the material microstructure directly in the life analysis. Much progress has been made in the past decade. Enhancements to capture the aging and changes in response due to temperature exposures and cycling, such as directional coarsening, are needed. Effort must continue to be placed on reducing the time and effort to calibrate these physics-based and microstructure-sensitive constitutive models. Protocols for calibration, which may include a hierarchy of experiments coupled with lower length scale simulations, need to be established. Improvements in the efficiency of their implementation including code parallelization continue to be needed. Moreover, CVP simulations may be used to develop influence functions or train artificial neural networks for including microstructure-sensitive parameters in more efficient reduced-order, homogenized viscoplasticity models.

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Olena Trunova,¹ Tilmann Beck,¹ and Lorenz Singheiser

Influence of Protective Coatings on Damage and Lifetime of Alloy 247 DS in Thermomechanical Fatigue and Bending Tests

ABSTRACT: The present work was focused on the damage mechanisms and lifetime of a directionally solidified, oxidation protection coated Ni-base superalloy under out-of-phase thermomechanical fatigue (OP TMF) loading. The main scope of the work was the analysis of crack formation and propagation in the coating depending on its fracture strain. Two coating systems applied by low pressure plasma spraying were investigated: (i) MCrAlY-coating and (ii) duplex-coating consisting of the same MCrAlY coating and a NiAl top coat. The fracture strains of the coatings were determined at $T = 20^{\circ}\text{C}$ up to 600°C by four-point bending with acoustic emission analysis. The MCrAlY-coating did not influence significantly the TMF lifetime of the superalloy, whereas duplex-coatings reduced it drastically. This correlates with the fracture strains of the coatings that are for the MCrAlY-coating distinctly higher and for the duplex coating lower than the mechanical strain range of the TMF tests. Thus, the damage process was controlled by propagation of cracks formed after few TMF cycles in the duplex coated material and by slower initiation and propagation of fatigue cracks in the MCrAlY coated alloy. Moreover, crack branching at the interface between coating and substrate alloy affects crack penetration into the substrate. In addition to measurement of TMF life, the early stages of crack initiation and propagation in the duplex coating were studied in more detail after TMF tests that were interrupted after defined cycle numbers as well as during in situ bending tests in the scanning electron microscope. Furthermore, the influence of pre-oxidation on crack propagation is discussed.

KEYWORDS: thermomechanical fatigue, crack growth, coatings, microstructure

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¹ IEK-2, Forschungszentrum Juelich GmbH, 52425 Juelich, Germany.

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Introduction

The continuous increase of operation temperatures of advanced gas turbines during the last decades required application of protective coatings against high temperature oxidation and corrosion of the components. Advanced engine designs as well as the development of innovative processing technologies led to the evolution of many new and improved coating systems in a variety of engine component applications [1–3]. One of the commonly used types of coatings consists of MCrAlY-alloys applied by low-pressure plasma spraying (LPPS), where M = Ni or/and Co. In the past, many variations of NiCoCrAlY coatings containing additions of tantalum, tungsten, titanium, niobium, rhenium, zirconium, etc., singly or in combination, have been developed for specific purposes [4,5].

During operation, the hot section components of gas turbines undergo complex cyclic thermal and mechanical loadings. As a result, thermomechanical fatigue (TMF) is one of the main lifetime limiting factors especially for turbine blades. The performance of the coatings under TMF is determined by the local thermal and mechanical strains in the coating and the underlying substrate as well as by microstructural changes due to oxidation and due to interdiffusion at the substrate/coating interface [6,7]. For reliable design and life prediction of oxidation protection coated turbine components, it is therefore important to quantitatively analyze the influence of the coating on crack formation and propagation during TMF.

The aim of the present study is to identify the lifetime and damage mechanisms of a directionally solidified Ni-based superalloy (Alloy 247 DS) coated with (i) MCrAlY-oxidation protection coating and (ii) duplex (MCrAlY + NiAl)-oxidation protection coating under out-of-phase TMF loading and to correlate the results with microstructure and fracture strain of the coating as well as with changes in the coating-substrate interface obtained by an additional annealing treatment.

Experimental

The directionally solidified Ni-base superalloy Alloy 247 LC DS (Ni-Cr8-Co9-Ta-Al-W-Mo) was provided in two batches. Two coating systems, both with a total thickness of 350 μm , were applied using LPPS technology: (i) MCrAlY-coating (Ni-25Co-17Cr-10Al-0.3Y) and (ii) duplex-coating consisting of the same MCrAlY-coating and a NiAl-topcoat. Micrographs of the microstructures in as coated condition are represented in Fig. 1. The duplex-coating was divided in two batches due to some microstructural differences: duplex I with strongly pronounced interface between both layers and duplex II with smooth transition (Fig. 1(b) and 1(c)). All coatings were manufactured at Institute IEK-1 of Forschungszentrum Juelich. After coating, the samples were heat treated for 2 h at 1080°C and 24 h at 870°C in vacuum.

For TMF tests solid cylindrical specimens with a gage length of 18 mm and a diameter of 10 mm within the gage length were manufactured in longitudinal $\langle 001 \rangle$ -orientation of the DS Ni-base substrate. For bending tests samples of dimension 3 mm $\times$ 4 mm $\times$ 50 mm, also oriented in $\langle 001 \rangle$ direction were used.

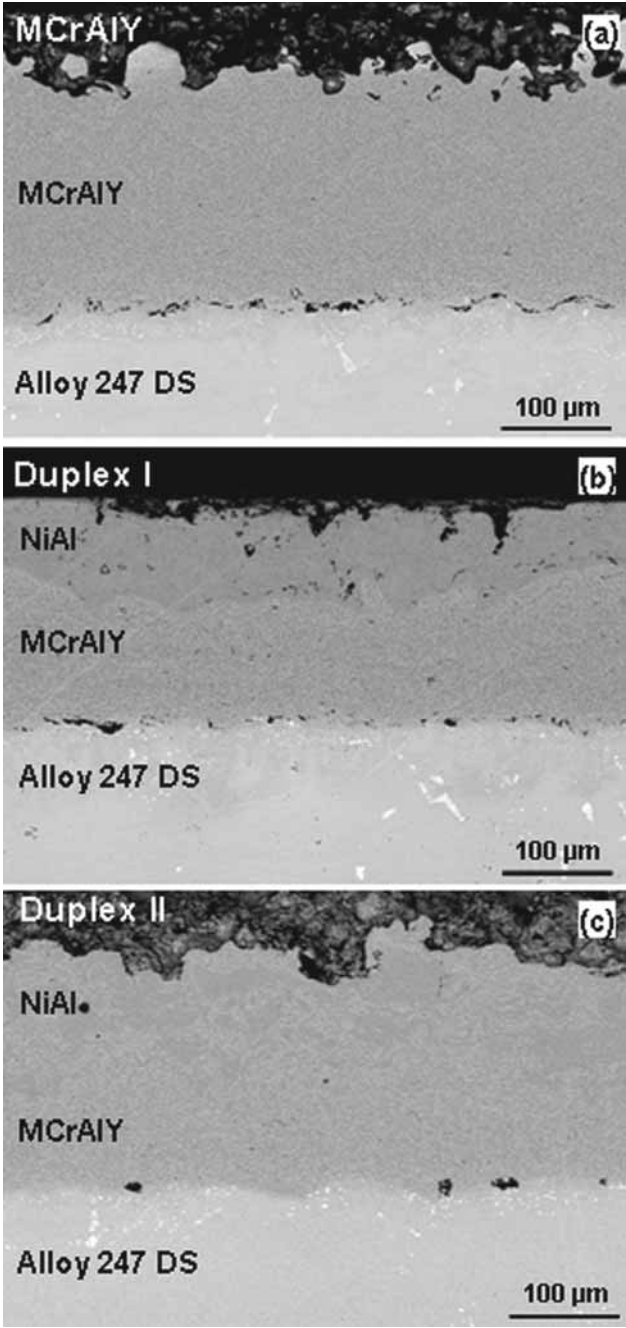


FIG. 1—Microstructure of coated Alloy 247 DS in as received state: (a) with MCrAlY coating, (b) with duplex I coating, (c) with duplex II coating.

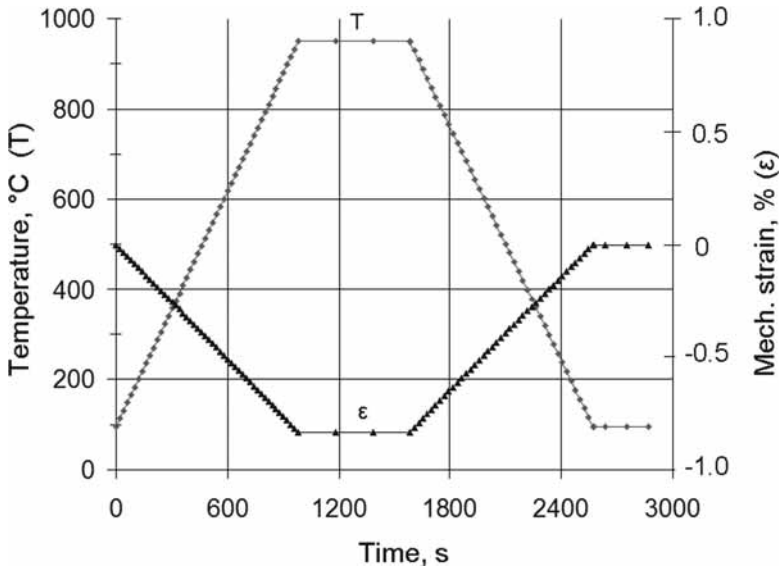


FIG. 2—OP TMF cycle with 0.85 % mechanical strain range and 95–950°C temperature range.

Before the tests, all uncoated as well as some coated specimens were polished to a surface roughness of $R_a < 1 \mu\text{m}$. The rest of the coated specimens was tested with rough “as-sprayed” surface with $R_a = 8 \mu\text{m}$.

The TMF tests were conducted using a servohydraulic fatigue testing machine (Instron, 8802) with a 100 kN load cell equipped with an induction heating system (Hüttinger Elektronik). The temperature was controlled using a ribbon type Pt/Pt-Rh-thermocouple wrapped around the specimen at the middle of the gage length. An additional reference thermocouple was spot-welded at the specimen shoulder. The specimen deformation was monitored using high temperature extensometer (MTS systems). For observation of the crack initiation at the specimen surface, a CCD-camera was used.

The out-of-phase (OP) TMF tests were carried out under strain control with a mechanical strain range of 0.85 % and 0.51 % and a temperature range of 95°C to 950°C. The TMF cycle consisted of heating with a rate of 0.87 K/s followed by a dwell time of 10 min at the maximum temperature, subsequent cooling to minimum temperature with 0.87 K/s and a final dwell time of 5 min at minimum temperature. To keep the cooling rate constant, an external compressed air flow was applied. Figure 2 shows the TMF cycle with a mechanical strain range of 0.85 %. The tests were started at the minimum temperature of 95°C and a mechanical strain of zero. The experiments were carried out according to the European Code-of-Practice for strain controlled TMF testing [8].

The quasistatic four-point bending tests were carried out using an electro-mechanical universal test machine with a 50 kN load cell equipped with a resistance furnace. The temperature was controlled using a Pt/Pt-Rh-thermocouple

TABLE 1—Summary of OP TMF tests on uncoated and coated Alloy 247 DS

Number	Coating/surface finish	Mech. strain, %	Norm. number of cycles	Comments
Batch I				
1	Uncoated/polished	$\Delta\epsilon_{\text{mech}} = 0.85 \%$	$N_f = 1$	
2	Duplex I/as sprayed	$\Delta\epsilon_{\text{mech}} = 0.85 \%$	$N_f = 0.64$	
3	Duplex I/as sprayed	$\Delta\epsilon_{\text{mech}} = 0.85 \%$	$N = 0.015$	Stopped
4	Duplex I/as sprayed	$\Delta\epsilon_{\text{mech}} = 0.85 \%$	$N = 0.15$	Stopped
5	Duplex I/polished	$\Delta\epsilon_{\text{mech}} = 0.85 \%$	$N_f = 0.58$	
6	Duplex I/polished	$\Delta\epsilon_{\text{mech}} = 0.51 \%$	$N_f = 1.46$	
7	Duplex I/polished	$\Delta\epsilon_{\text{mech}} = 0.51 \%$	$N_f = 1.30$	
Batch II				
8	Uncoated/polished	$\Delta\epsilon_{\text{mech}} = 0.51 \%$	$N = 4.2$	
9	Uncoated/polished	$\Delta\epsilon_{\text{mech}} = 0.85 \%$	$N_f = 0.80$	
10	MCrAlY/as sprayed	$\Delta\epsilon_{\text{mech}} = 0.85 \%$	$N_f = 0.73$	
11	Duplex II/as sprayed	$\Delta\epsilon_{\text{mech}} = 0.85 \%$	$N_f = 0.31$	
12	Duplex II/as sprayed	$\Delta\epsilon_{\text{mech}} = 0.85 \%$	$N_f = 0.29$	
13	Duplex II/as sprayed	$\Delta\epsilon_{\text{mech}} = 0.51 \%$	$N_f = 2.33$	

attached to the sample. The displacement during bending was recorded using an LVDT-Sensor. For damage detection, an acoustic emission (AE) monitoring system (Vallen Systems) was applied. The samples were tested with a 4-point-bending device with 20-mm inner span and 40-mm outer span at a crosshead displacement rate of $44 \mu\text{m s}^{-1}$. The specimens were loaded with the coating in tension to correlate the determined fracture strains with mechanical strains in TMF because under OP TMF, the maximum tensile strain is applied at low temperatures, where coating ductility is relatively low. Additionally, in situ three-point bending tests were carried out at room temperature in a scanning electron microscope (SEM).

The microstructural investigations were performed after metallographic preparation using optical microscopy (Leica – MEF4M) as well as scanning electron microscopy (LEO 1530 – ‘Gemini’).

Results

TMF tests

OP TMF tests were carried out on uncoated and coated specimens of Alloy 247 LC DS up to fracture or a maximum stress drop of 20 %. An overview of the tests is given in Table 1. Some TMF tests were conducted with the same parameters on coated specimens with polished ($R_a < 1 \mu\text{m}$) and as-sprayed surface ($R_a = 8 \mu\text{m}$). All results were normalised to the number of cycles to failure of the uncoated specimen 1.

Figure 3 shows the maximum and minimum stress versus the normalised cycle number for the uncoated specimens from both batches as well as for

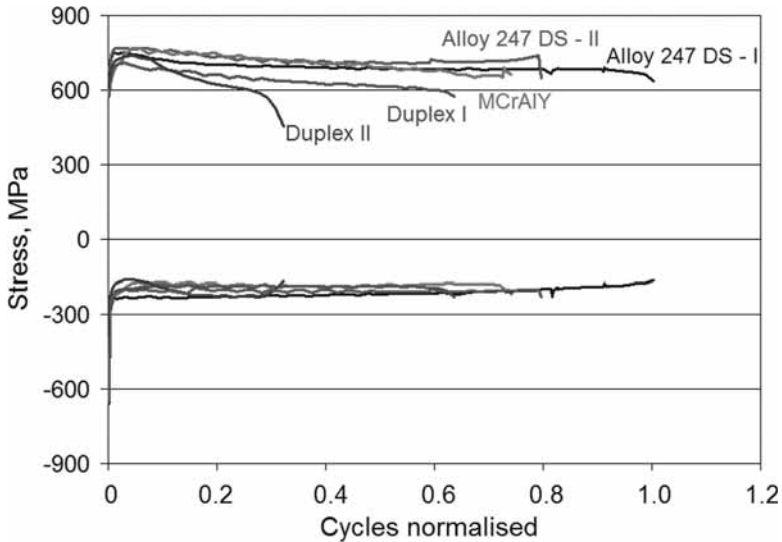


FIG. 3—Evolution of maximum and minimum stresses under TMF with 0.85 % mechanical strain range and 95–950°C temperature range.

coated specimens under OP TMF with 0.85 % mechanical strain range. The stresses were calculated using the sample cross section including the thickness of the coating. There was no significant influence of surface roughness on the TMF lifetime observed; therefore only results for as-sprayed condition for each coating type are displayed in the diagram.

The cyclic deformation curves of the uncoated samples from both batches show almost identical deformation behaviour and similar lifetime within usual scatter. Comparing these results with coated samples revealed that MCrAlY-coating has no significant effect on TMF lifetime. However, the duplex-coatings from batches I and II drastically reduced the number of cycles to failure to ~40 % and ~60 % of the lifetime of the uncoated material, respectively (also see Table 1).

In Fig. 4, the maximum and minimum stress of duplex coated and uncoated sample during TMF with a mechanical strain range of 0.51 % is shown as a function of cycle number. Compared to a mechanical strain range of 0.85 % TMF life of the uncoated Alloy 247 DS increased by more than a factor of 4. This experiment was stopped because of very long testing time, but some cracks on the sample surface were already detected, indicating that failure probably would have taken place during less than 100 cycles after the end of the test. The stress drop observed after a normalised cycle number of ~2.9 (Fig. 4) occurred due to a test interruption because of thermocouple failure.

For the same mechanical strain range, the duplex I coated sample had ~60 % shorter life than the duplex II-coated one. At the same time, reduction of mechanical strain range from 0.85 % to 0.51 % led to an increase in lifetime by a factor of 2 (duplex I) and by a factor of 7 (duplex II). A summary of the measured TMF lifetimes as a function of mechanical strain range is given in Fig. 5.

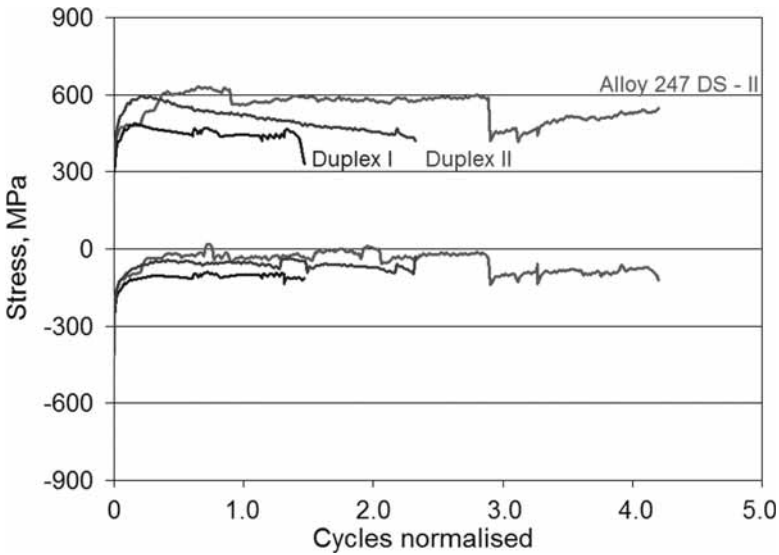


FIG. 4—Evolution of maximum and minimum stresses under TMF with 0.51 % mechanical strain range and 95–950°C temperature range.

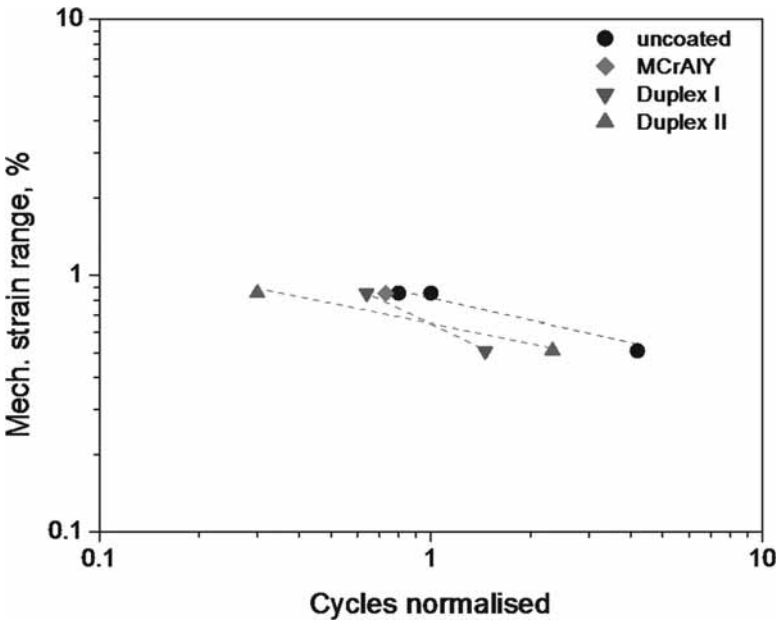


FIG. 5—Dependence of TMF lifetime on mechanical strain range.

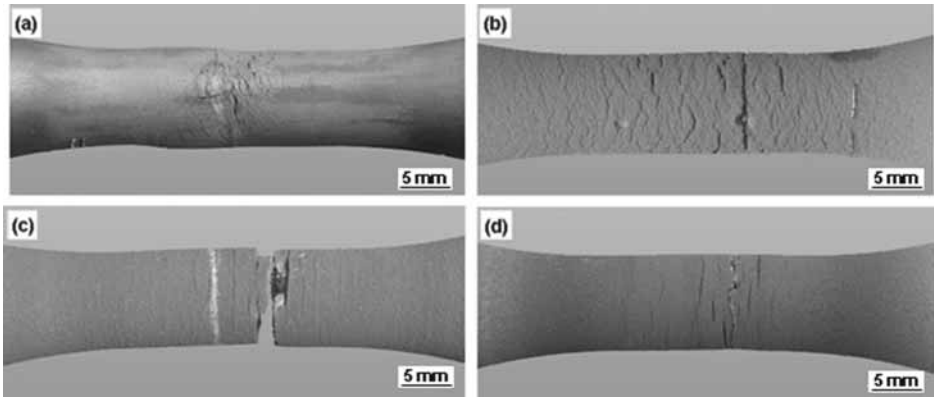


FIG. 6—Overview of the gage length after TMF with 0.85 % mechanical strain range: (a) uncoated Alloy 247 DS, (b) with MCrAlY coating, (c) with duplex I coating, (d) with duplex II coating. The surface for all coatings is in as sprayed condition.

Damage evolution of uncoated and coated Alloy 247 DS during TMF testing was observed using a CCD camera. Surface pictures were taken after each cycle at minimal temperature. For the uncoated samples, first cracks were detected after ~ 50 % of the lifetime. With increasing number of cycles, the number and length of cracks continuously increased. The sample with MCrAlY coating showed a qualitatively similar damage evolution, but first cracks in the coating occurred after ~ 20 % of the lifetime and thus earlier than for the uncoated alloy.

In contrast, for duplex I coating after already ~ 2 % of lifetime transversal cracks were observed to form in regular longitudinal intervals of about 1 mm. The number of cracks remained constant up to failure. The sample with duplex II coating revealed a kind of mixed degradation mode: similar to the MCrAlY-coating first cracks appeared after 45 % of the lifetime, but the shape and amount of cracks is comparable with duplex I coating.

Comparing damage evolution of duplex-coated samples during TMF with mechanical strain ranges of 0.51 % and 0.85 %, it is clearly seen that the number of cracks strongly decreases with decreasing strain range. An overview of the cracks formed after the TMF tests is given in Figs. 6 and 7 for mechanical strain ranges of 0.85 % and 0.51 %, respectively.

Detailed microstructural investigations of fatigue cracks formed after OP TMF with mechanical strain ranges of 0.85 % and 0.51 % were conducted on polished longitudinal cross sections, as shown in Figs. 8 and 9, respectively. In the uncoated material, relatively short secondary cracks perpendicular to the direction of the applied force could be found (Fig. 8(a)). For the MCrAlY-coated material, cracks initiated at the surface and propagated toward the coating-substrate interface. At the interface, most of them branched along the interface before penetration into the substrate occurred. In case of the duplex I coating (Fig. 8(c)), cracks show a similar branching effect at the MCrAlY/substrate interface. In contrast, cracks in the duplex II coated alloy showed no branching

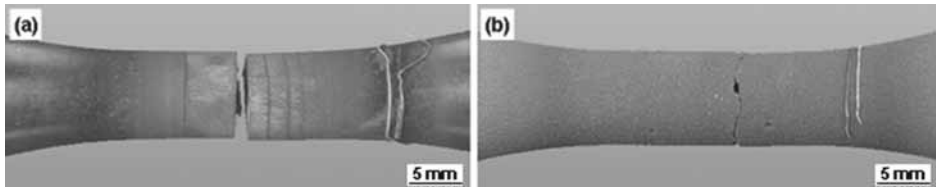


FIG. 7—Overview of the gauge length after TMF with 0.51 % mechanical strain range: (a) with duplex I coating (polished), (b) with duplex II coating (as sprayed).

at the MCrAlY/substrate interface (Fig. 8(d)). This is also evident for the duplex-coated samples after TMF with a mechanical strain range of 0.51 %, where crack branching was observed only for the duplex I coating (Fig. 9(a) and 9(b)).

The initial stages of damage in the duplex I-coated samples were investigated in more detail after TMF tests with predefined cycle numbers at a mechanical strain range of 0.85 %. Figure 10 shows optical micrographs of longitudinal sections after 2 % and 20 % of total life. After 2 % of the lifetime, the radial cracks propagated through the NiAl coating and reached the NiAl/MCrAlY interface (Fig. 10(a)). Further TMF cycling promoted crack propagation through the MCrAlY coating up to the MCrAlY/substrate interface. At the interface crack branching (Fig. 10(b)) decelerating further crack penetration into the substrate material was observed.

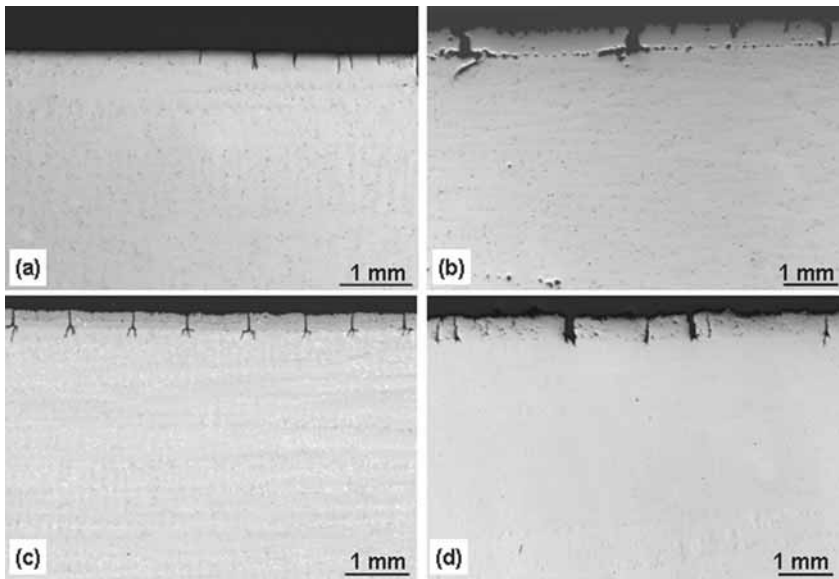


FIG. 8—Light microscopy image showing longitudinal cross sections of samples after TMF with 0.85 % mechanical strain range: (a) uncoated Alloy 247 DS, (b) with MCrAlY coating, (c) with duplex I coating, (d) with duplex II coating.

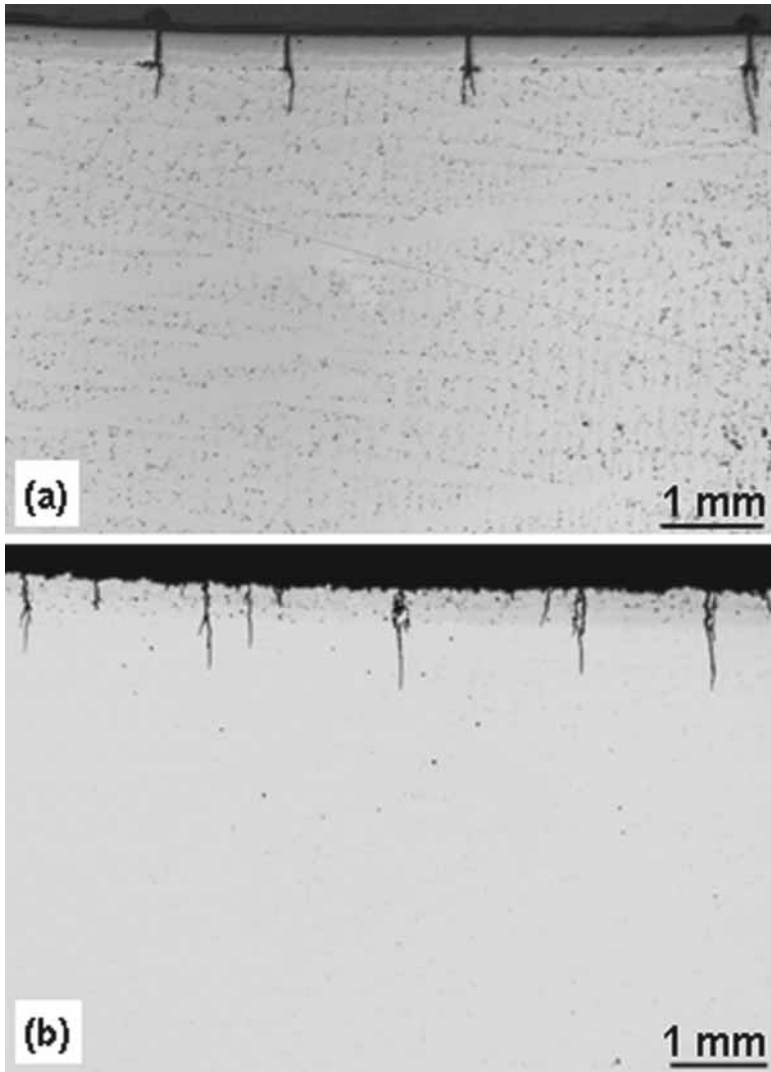


FIG. 9—Light microscopy image showing longitudinal cross sections of samples after TMF with 0.51 % mechanical strain range for coated Alloy 247 DS: (a) with duplex I coating, (b) with duplex II coating.

Bending tests

The four-point bending tests coupled with acoustic-emission analysis were carried out up to a maximum strain of 1.7 % in the temperature range between room temperature (RT) and 600°C.

In case of the MCrAlY coating, no cracks were detected up to reaching the maximum strain in the whole temperature range. This result was confirmed by

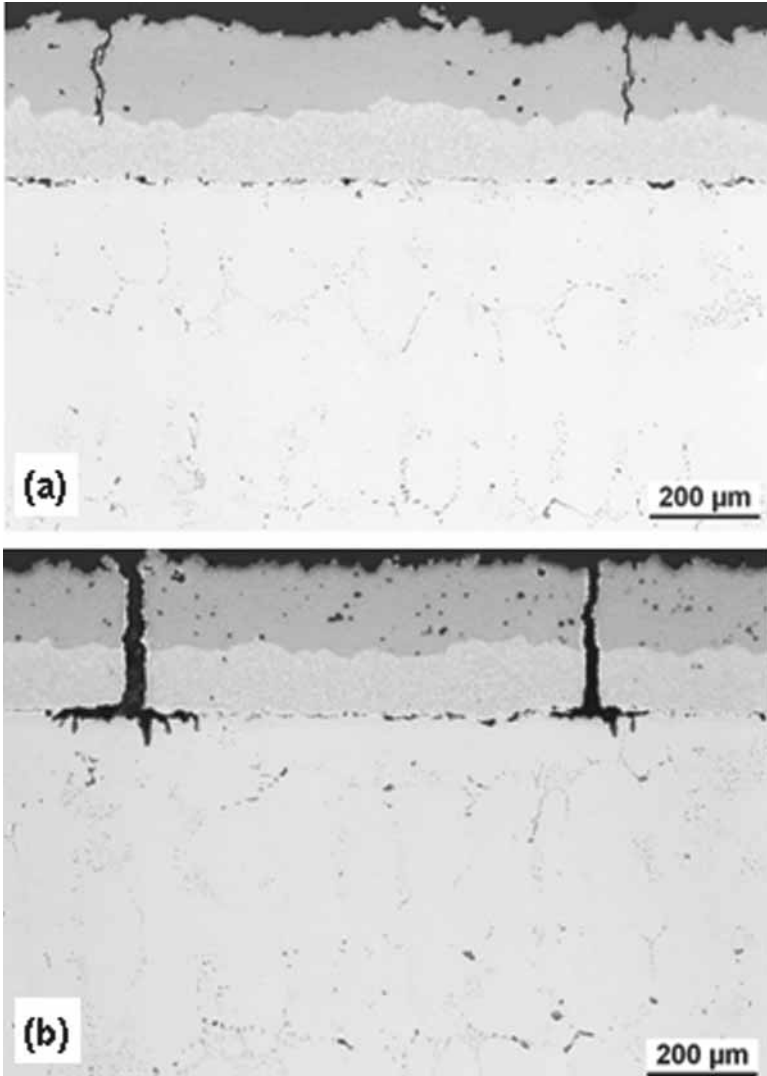


FIG. 10—*Light microscopy image showing crack development in duplex I coating during TMF with 0.85 % mechanical strain range after (a) 0.015 normalised cycles and (b) 0.15 normalised cycles.*

acoustic-emission analysis, as shown in Fig. 11(a). To determine a critical strain of the coating, some additional tests up to 3.5 % of maximum strain were carried out at room temperature. Within some insecurity caused by the high plastic deformations of substrate and coating obtained in this case, the fracture strain was between 3 and 3.5 %. Microstructural analysis of coating revealed presence of microcracks, as shown in the SEM image of a tested sample in Fig. 11(b).

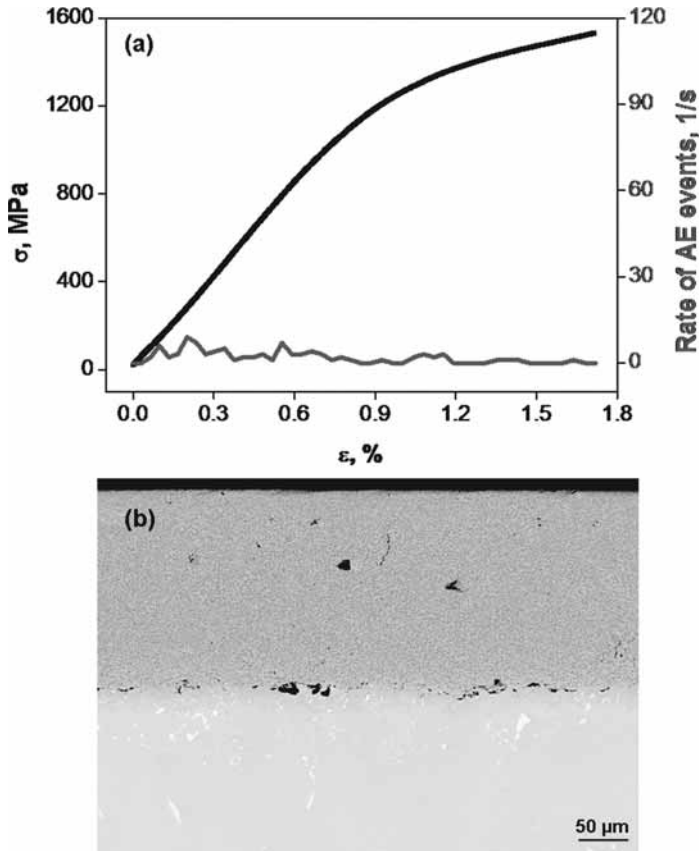


FIG. 11—Loading curve with AE analysis and SEM image of the MCrAlY coating after 4-point bending test at RT.

In the duplex I-coated samples, cracks were formed according to the acoustic-emission analysis after about 0.65 % of strain at $T = 100^\circ\text{C}$ (Fig. 12(a)). Cracks started in the NiAl-coating and propagated to the NiAl/MCrAlY interface (Fig. 12(b)). The first peak in the rate of AE events (Fig. 12(a)) at the beginning of the bending test was not attributed to cracking. This was proven by an additional experiment with 0.5 % maximum strain, in which no damage of the coating was detected. As expected, increasing temperature led to an increase of the critical strains and increasing distance between the cracks. At 600°C , only some microcracks within the NiAl-coating were found. However, in the temperature range from 100°C to 500°C , the single cracks show similar propagation behaviour to the NiAl/MCrAlY interface, as shown in Fig. 12(b).

For the duplex II coating, critical strains could be determined only at room temperature (about 0.85 %) and at 100°C (about 0.95 %). At higher temperatures, no cracks were observed after reaching the maximum strain of 1.7 %. In Fig. 13(a) and 13(b), the loading curve plotted along with the rate of AE signals

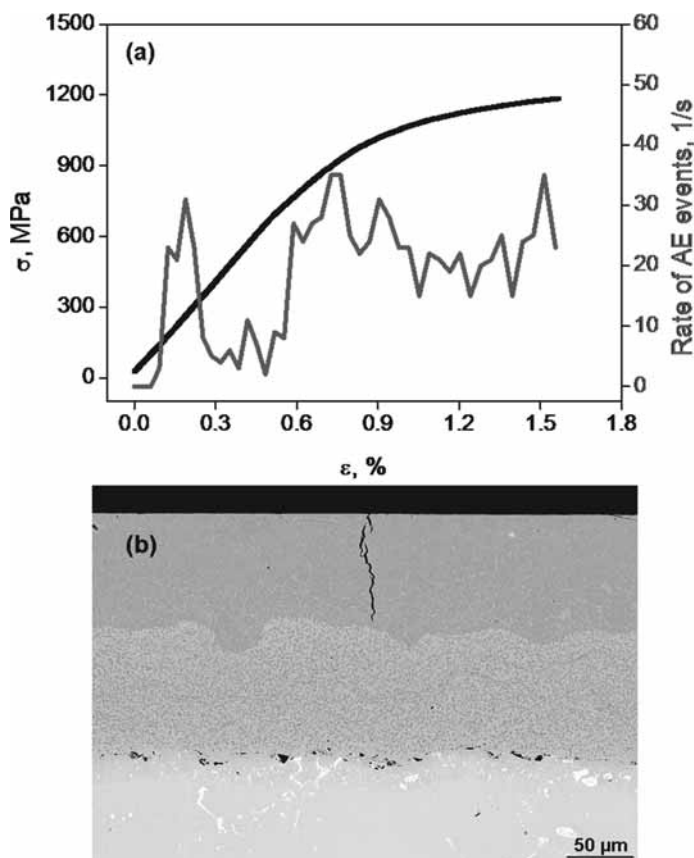


FIG. 12—Loading curve with AE analysis and SEM image of the duplex I coating after 4-point bending test at 100°C.

and a SEM picture of the cracks after bending test at room temperature are represented. In contrast to the duplex I coating, the crack propagated through the whole coating up to the MCrAlY/substrate interface. A summary of the critical strains for all coating systems determined from four-point bending tests by AE analysis as a function of temperature is given in Fig. 14.

To get more information about crack development in the duplex I coating during bending, in situ three-point bending tests in SEM were conducted. Two tests were performed at ambient temperature: (i) in as sprayed condition and (ii) after additional exposure at 950°C for 100 h. The high temperature exposure resulted in the formation of an interdiffusion zone at the interface between MCrAlY layer and substrate material.

These tests both on as-sprayed and exposed samples were interrupted several times to document the main steps of crack development:

- I. Crack initiation in the NiAl-coating.
- II. Propagation to the NiAl/MCrAlY interface.

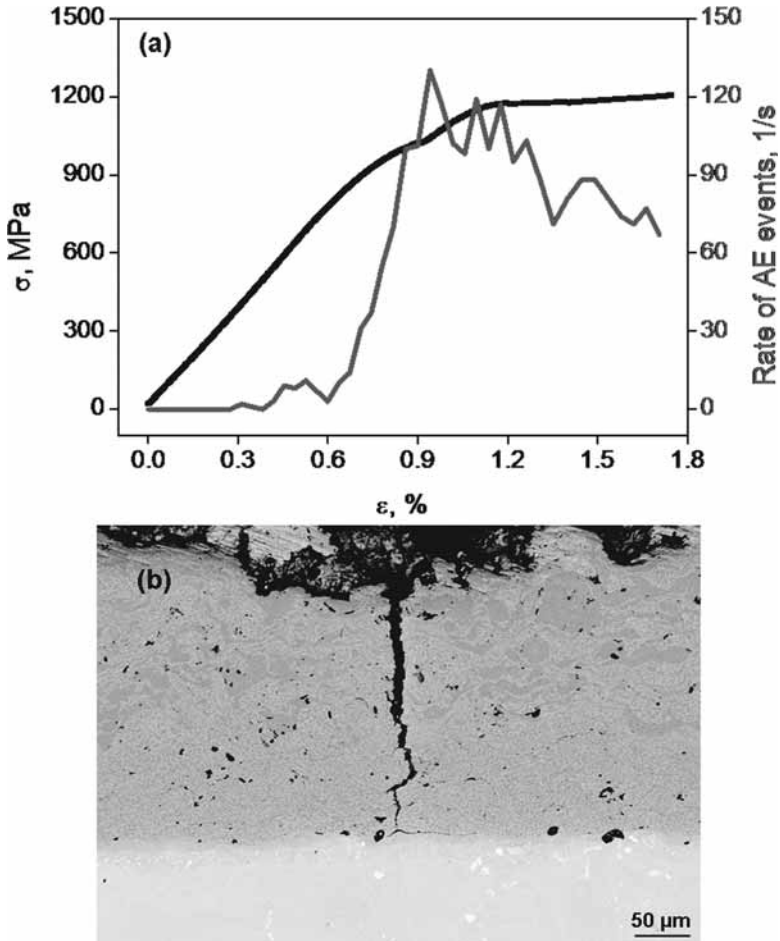


FIG. 13—Loading curve with AE analysis and SEM image of the duplex II coating after 4-point bending test at RT.

III. Penetration into the MCrAlY-coating and propagation to the MCrAlY/substrate interface.

IV. Transition to lateral crack growth at the MCrAlY/substrate interface.

V. Lateral crack growth along the MCrAlY/substrate interface.

Figure 15 displays these steps in the measured load-displacement curves. For both specimens (as sprayed and annealed) crack growth behaviour up to the MCrAlY/substrate interface (steps I to III) is very similar. However, the transition to lateral crack growth (step IV) occurred significantly earlier for the as-sprayed specimen than for the exposed one. SEM pictures of the crack tips at the MCrAlY - substrate interface (Fig. 16) revealed remarkable plastic deformation in the interdiffusion zone between coating and substrate where the intermetallic β -NiAl-phase in the coating was dissolved resulting in locally

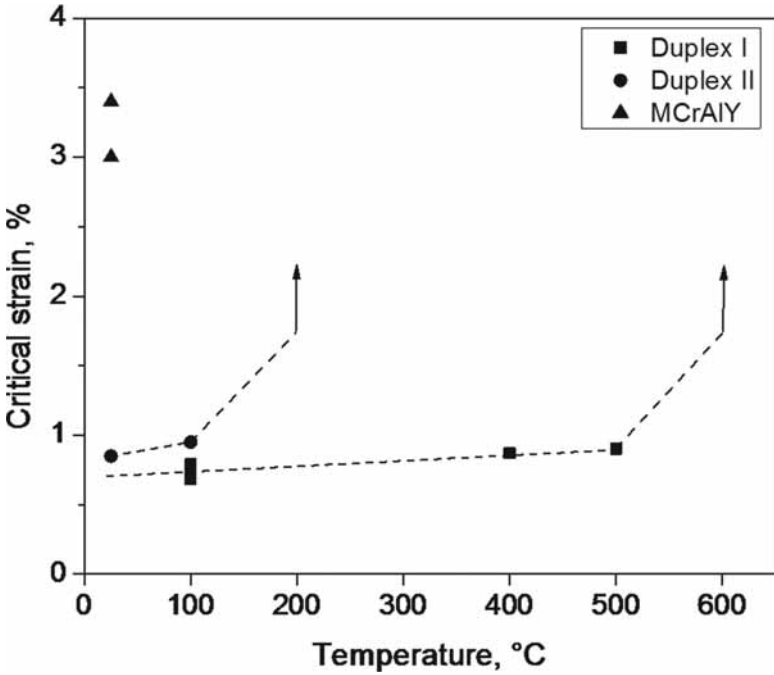


FIG. 14—Fracture strains of coatings determined from 4-point bending tests by AE method as a function of the temperature. Vertical arrows indicate that no cracks occurred up to the applied strain of 1.7 %.

increased ductility. Consequently, a large plastic zone formed at the crack tip which reduced the stress intensity and resulted in delayed crack propagation compared to the as-sprayed state.

Discussion

Damage and lifetime of oxidation protection coated Alloy 247 DS during OP TMF strongly depends on the applied mechanical strain range and on the fracture strains of the coatings at minimum temperature of the TMF cycle.

During OP TMF with 0.85 % mechanical strain range, the high tensile stresses developing at low temperatures led to crack initiation during the first TMF cycles in the duplex I-coated specimens and, consequently, distinct reduction of TMF life. This correlates with the fact that the fracture strains of this coating at temperatures between 100°C and 400°C are below the mechanical strains applied in TMF. Thus the damage process was controlled by brittle cracking of the intermetallic top layer of the coating followed by crack propagation during the low temperature stage of the TMF cycle. Therefore, the reduction of TMF life by ~40 % induced by duplex I coating is attributed to accelerated crack formation in the NiAl top layer. The delay in crack propagation induced by crack branching at the interface between MCrAlY sublayer and

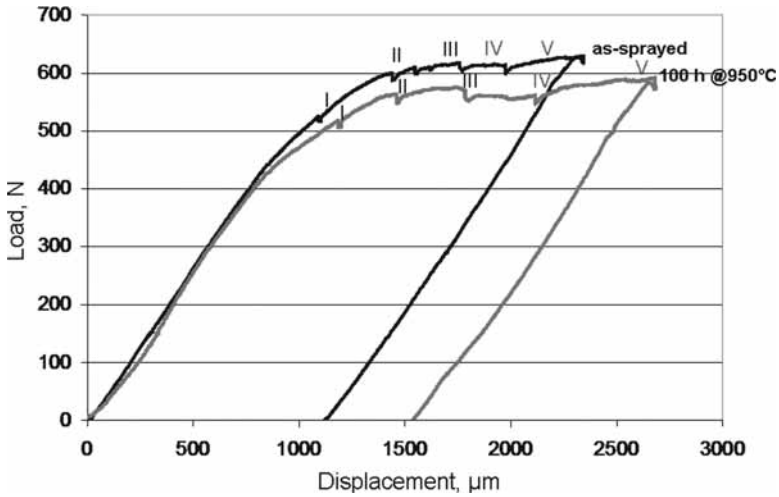


FIG. 15—Loading curves for duplex I coating for in situ 3-point bending test in SEM at RT.

substrate prevents an even more pronounced life reduction than it would be expected from the very early crack formation in the top layer.

In contrast, for duplex II and MCrAlY-coated specimens, no cracks were observed during the first 20% of TMF life. Due to the higher fracture strains of these coatings, cracks were formed continuously as a result of fatigue processes in the coatings. It should be noted that both coatings show earlier surface crack formation than the uncoated material. However, distinct lifetime reduction was only caused by the duplex II coating, whereas the MCrAlY coating did not change TMF life significantly compared to the uncoated state. This is due to the fact that crack branching at the interface to the substrate delays the crack propagation in the MCrAlY coated material; this is not the case for the duplex II coating (see Fig. 8).

Furthermore, the presence of an interdiffusion zone with increased local ductility at the MCrAlY/substrate interface after heat treatment results—at least in bending tests—in significant crack delay compared to an as-sprayed coating (see Figs. 15 and 16). Further research will address the crack growth behavior of annealed samples during TMF loading.

The comparison of surface cracks after OP TMF with 0.85 mechanical strain range and bending tests revealed similar regular intervals between the cracks. The results for both duplex coatings are summarized in Fig. 17, where the crack intervals are represented as a function of temperature or cycle number for bending tests and TMF tests, respectively. In an additional Y-axis, the fracture strains are given as dotted areas. It is clearly seen in Fig. 17 that with an increase of coating fracture strains the crack intervals after TMF as well as after bending tests increased. Furthermore, if the fracture strains are below the mechanical strains applied in TMF, as in case of duplex I coating, the degradation process is controlled by early crack formation in the coating followed by crack

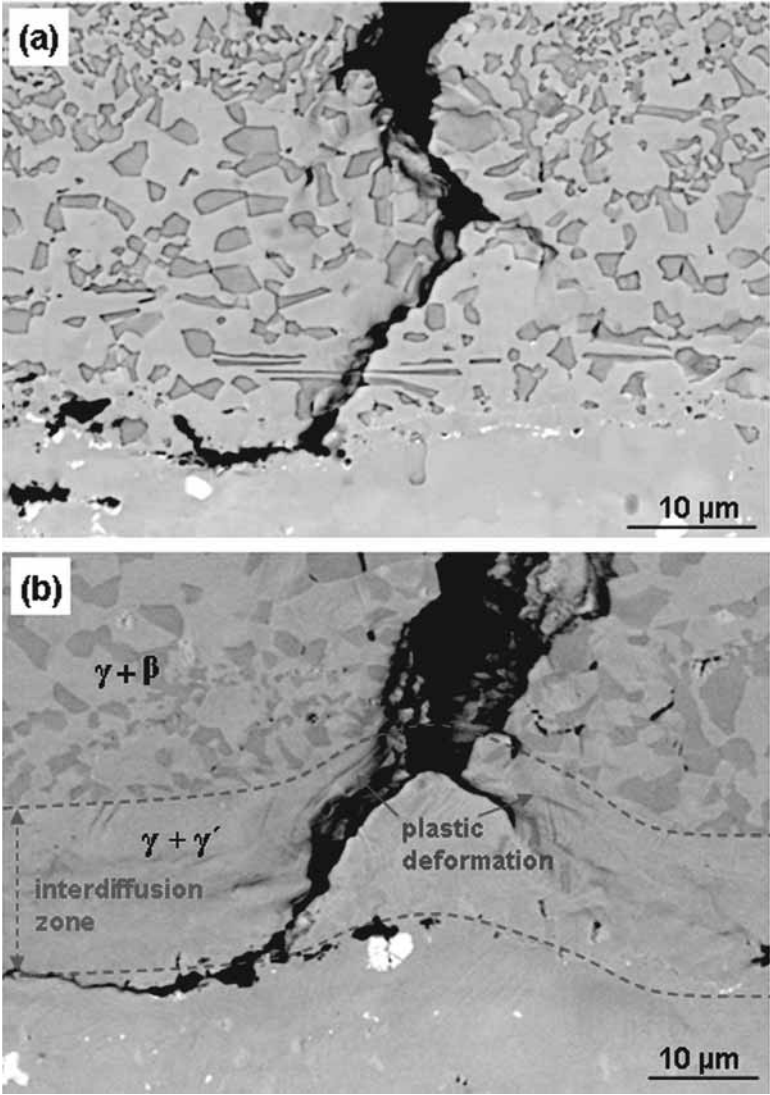


FIG. 16—SEM images of crack tips at the MCrAlY/substrate interface during *in situ* 3-point bending test: (a) as sprayed and (b) annealed sample.

growth as described above. For TMF tests with mechanical strain range below the coating fracture strains, fatigue crack formation and growth is the dominant damage mechanism. Thus, the correlation between coating fracture strains and crack growth mechanisms could be useful for design of oxidation protective coated components, which experience high thermomechanical loadings under service conditions, such as turbine blades.

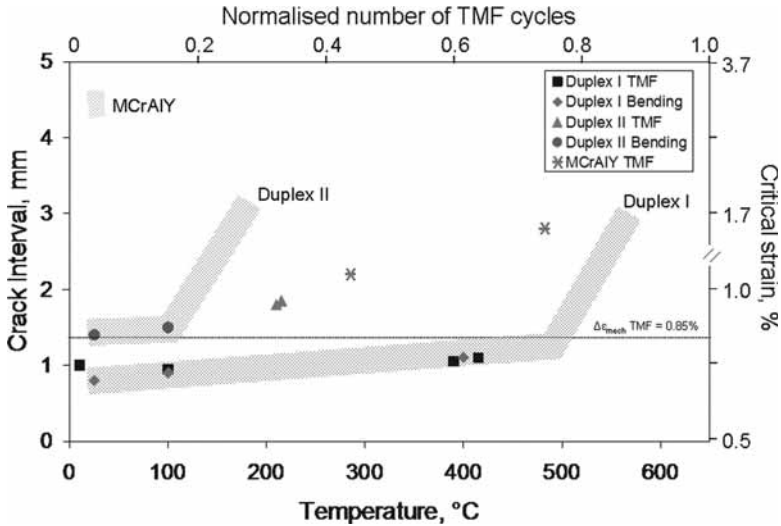


FIG. 17—Summary of results for the crack formation after TMF and bending tests: crack intervals are represented as a function of temperature and cycle number for bending tests and TMF tests, respectively. The fracture strains correlated to the right Y axis are given as dotted areas.

It is believed that the correlation between degradation mode under OP TMF and coating fracture strains is not valid for other TMF cycle forms, e.g., in-phase or diamond type, because cracks were found to form under tensile strains at low temperatures where coating ductility is relatively low. Investigations of the influence of oxidation protection layers on TMF life of Ni-alloys in the case of cycle types other than OP are subject to further research.

Conclusions

The effect of two oxidation protection coating systems, i.e., a MCrAlY single layer and a MCrAlY/NiAl duplex coating on TMF lifetime of the directionally solidified Ni-base superalloy Alloy 247 LC DS was investigated. It was observed that early crack initiation in the coating during tensile loading at low temperatures in the OP TMF cycle may significantly reduce TMF life. On the other hand, life reduction is limited due to crack branching at the interface between protection layer and substrate. The influence of both effects can be summarized as follows:

- MCrAlY coating: Reduction of cycles to crack initiation is compensated by crack delay due to branching at the interface. Consequently, no life reduction occurs compared to the uncoated material.
- Duplex I coating: Brittle cracking of the NiAl top layer induces crack formation during the first cycles. Pronounced crack branching limits life reduction to 40 % of total life of the uncoated material.

- Duplex II coating: Early crack formation in the layer is not compensated by crack branching. This results in distinct life reduction of 60 % compared with the uncoated alloy.

TMF tests at strain ranges below the fracture strain of the coatings resulted in similar lifetimes as for the uncoated material. Therefore, the following rough conclusion for the influence of oxidation protection coatings on OP TMF life can be drawn:

- TMF with mechanical strain range above the critical strain of coating: Significant life reduction due to crack formation at very early stages of lifetime, growth kinetics controlled by coating microstructure, i.e., stopping/branching effect at the interfaces.
- TMF with strain range below critical strain: No significant influence of the coating on TMF life, crack formation and propagation is controlled by fatigue processes.

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Ken Wackermann,¹ Ulrich Krupp,² and Hans-Jürgen Christ¹

Effects of the Environment on the Crack Propagation Behavior of IN718 in the Temperature Range of the Dynamic Embrittlement

ABSTRACT: IN718 is a standard nickel-base alloy for high temperature usage, e.g., as forged gas turbine disks. At elevated temperatures IN718 is prone to cracking by dynamic embrittlement. Dynamic embrittlement is caused by oxygen diffusion along the grain boundaries, resulting in a time-dependent intergranular decohesion. The result is a fast brittle intercrystalline crack propagation. In contrast, vacuum testing shows a drastic reduction of the crack propagation rate and a ductile, transcrystalline fracture morphology. Objective of this investigation is to gain a better understanding of the dynamic embrittlement process. In the present study the crack growth rate has been measured with an alternate current potential drop system (ACPD) and a far-field microscope during dwell-time testing at 650 °C, a positive stress ratio and two different load schemes.

KEYWORDS: nickel-base alloy, IN718, dynamic embrittlement, dwell-time cracking, fatigue

Introduction

IN718 is a commonly used nickel-base alloy for high temperature usage, e.g., in gas and steam turbines. At elevated temperatures IN718 is prone to the failure mechanism “dynamic embrittlement.” Dynamic embrittlement can be

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¹Institut für Werkstofftechnik, Universität Siegen, Paul-Bonatz-Strasse 9-11 57076 Siegen, Germany (Corresponding author), e-mail: wackermann@ifwt.mb.uni-siegen.de

²Faculty of Engineering and Computer Science, University of Applied Sciences Osnabrück, Albrechtstr. 30 49076 Osnabrück, Germany.

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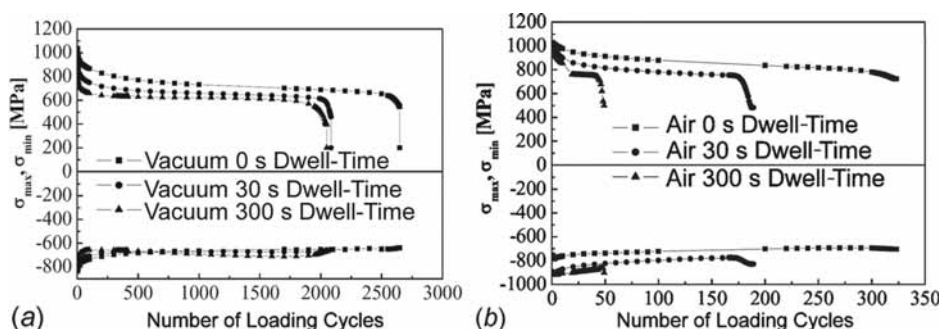


FIG. 1—Strain-controlled fatigue tests ($\Delta\epsilon/2=0.7\%$, $T=650^{\circ}\text{C}$) with 0s, 30s, and 300s dwell-time: (a) Cyclic deformation curve in vacuum and (b) air 3.

considered as a kind of stress corrosion cracking, driven by tensile stress controlled oxygen grain boundary diffusion. Elemental oxygen embrittles the grain boundaries by weakening the grain boundary cohesion and results in a fast brittle intercrystalline crack propagation (CP). In contrast, testing in vacuum results in an increase of the fatigue life by orders of magnitudes [1,2], cf. Fig. 1 [3]. When the crack growth proceeds, oxygen from the environment attacks the new crack surface and builds up a thermal oxide layer. Articles describing the generic mechanism of dynamic embrittlement were published by Krupp, Kane, McMahon Jr. and Pfaendtner [1,2,4–7]. Dynamic embrittlement of nickel-base superalloys was reviewed by Krupp [8].

Another explanation for brittle intergranular crack propagation at elevated temperatures was given by the SAGBO mechanism (stress-assisted grain boundary oxidation). Similar to dynamic embrittlement, SAGBO is governed by stress-assisted oxygen grain boundary diffusion. The atomic oxygen reacts to form a bulk oxide at the grain boundaries resulting in a brittle cracking of the oxide scale [9]. As both mechanisms provide explanations for the appearance of an oxide layer after fracture on the new developed fracture surface, it is hardly possible to clearly attribute brittle cracking solely to either dynamic embrittlement or SAGBO. However, SAGBO assumes a detectable oxide layer within the grain boundaries ahead of the crack tip, which was not yet found by means of optical or electron microscopy. Therefore, the dynamic embrittlement seems to be the most likely mechanism for the oxygen induced dwell cracking of nickel-base material [10,11].

The mechanism of dynamic embrittlement is not limited to fatigue loading conditions, where dwell-times are superimposed, but appears also during static creep and load relaxation. In load relaxation tests, the force decrease is much slower in vacuum as in laboratory air as it was shown in [6,12]. Creep tests of IN718 at 700°C reveal a fast brittle crack propagation in laboratory air, 100 kPa pure oxygen and moist argon. In contrast, in a pure argon atmosphere the creep rate is about four orders of magnitude slower [13]. Therefore, the embrittling species has not necessarily to be oxygen, but might also be the water vapor in a gaseous atmosphere.

Similar results were obtained for fatigue tests of nickel-base material showing a brittle crack propagation in water vapor atmosphere [14], in laboratory air and in pure oxygen [15]. Again, fatigue tests in pure argon yielded very low crack propagation rates [15]. Therefore, oxygen is one reason for dynamic embrittlement, but water vapor dissolved in laboratory air might support the embrittlement process as a superimposed attacking agent. In a similar way, the oxidation rate of nickel-base alloys was reported to be higher in humid air than in dry air [16].

The mechanism of the dynamic embrittlement manifests itself on the fracture surfaces and depends on the dwell-time duration and/or the test frequency. Under vacuum conditions in the temperature range of dynamic embrittlement, the fracture morphology is always transcrystalline for any test frequency or dwell-time. In laboratory air in the temperature range of dynamic embrittlement, the fracture morphology can either be ductile transcrystalline or brittle intercrystalline in dependence on the test mode. For short dwell-time durations or a high test frequencies there is a ductile transcrystalline fracture mode with lower CP rates. Sometimes also fatigue striations were observed. With increasing dwell-time duration or decreasing test frequency the crack propagation rate increases and there is a change to a brittle intercrystalline fracture mode showing deformation-less grain boundary facets instead of fatigue striations within the fracture surface. Therefore, in the temperature range of dynamic embrittlement the effects of the mechanism are time dependent as the different fracture modes for the different test modes show [3,17–19].

In the literature, the dynamic embrittlement mechanism is comprehensively discussed and understood in many aspects. However, the grain boundary diffusion rate and the direct measurement of atomic oxygen concentration in the grain boundaries are still under investigation, as solid results have not yet been obtained according to the knowledge of the authors.

One explanation might be the very small diffusion depth [7,20]. Furthermore, there are results from crack propagation rate measurements which may not be solely explained by the theory of dynamic embrittlement. Lynch et al. [18,19] performed fatigue crack propagation tests on IN718 and Waspaloy on compact tension specimens in the temperature range of 500–700°C. The tests were carried out at a test frequency of 1/60 Hz or as dwell tests applying a dwell-time of 60 s (see Fig. 2). A faster CP rate was observed for the test with the test frequency of 1/60 Hz.

The theory of dynamic embrittlement would predict a faster CP rate for the 60 s dwell test in comparison to the 1/60 Hz test. The reason is that long dwell-time durations at maximum tension should lead to high oxygen grain boundary diffusion, resulting in a fast CP rate (compare also Fig. 1). In the case of the 1/60 Hz test, the average tension in the loading cycle is smaller which should cause lower oxygen grain boundary diffusion resulting in a lower CP rate. The results are explained by Lynch et al. [18,19] by a pronounced crack branching and the formation of a thicker oxide layer in the 60 s dwell test leading to crack closure. A similar result was obtained by Andersson et al. [17] on IN718 specimens with a rectangular cross sections showing the same CP rate for a 600 s dwell test and a test with a frequency of 0.5 Hz. Therefore, besides the sole explanation of the

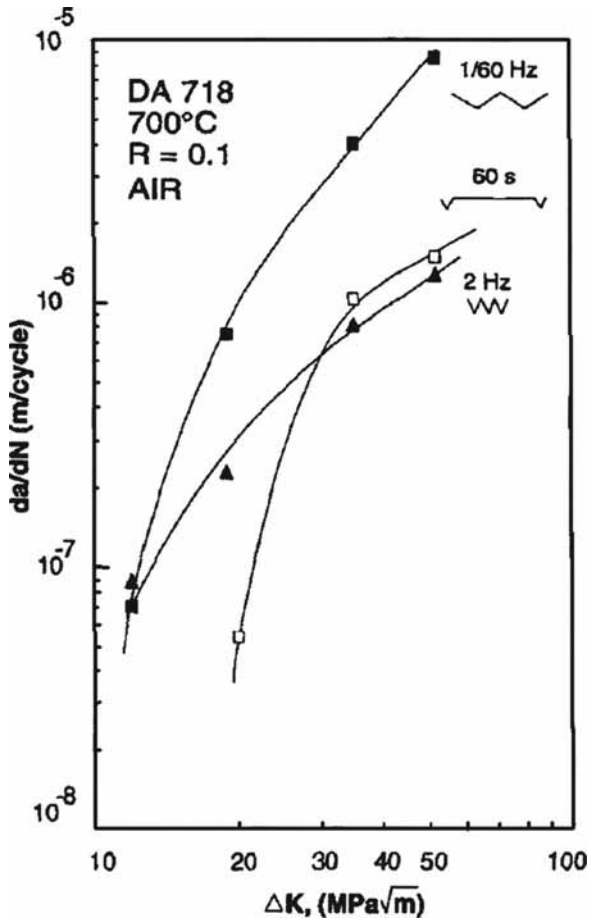


FIG. 2—Crack propagation rates for IN718 at 700°C on compact tension specimens for various load modes performed in load control after [18].

crack propagation acceleration by dynamic embrittlement also other aspects like crack tip closure, formation of an oxide layer, fracture mechanics parameters (such as crack tip shape) and creep might influence the CP rate of nickel-base material. Hypothetically the formation of an oxide scale at the crack tip in dwell-time test might lead to a crack closure and thus hinder the further oxygen grain boundary diffusion. Nevertheless, most CP measurements of other authors are consistent with the theory of the dynamic embrittlement, compare for example Refs 21–24.

The CP in a dwell-time test can take place in the loading/unloading ramps, during the dwell-time duration or in any part of the loading cycle. A load relaxation test of Pfaendtner and McMahon [7] shows that there should be CP in the dwell-time duration. According to the knowledge of the authors there is no published research giving an answer to the question when in the loading cycle of a

TABLE 1—*Chemical composition in wt %.*

Element	Ni	Fe	Cr	Nb	Mo	Ti	Al	Co	Si	Mn	C	B
wt %	Bal. (~ 56.3)	15.10	17.01	5.60	2.98	1.00	0.60	0.25	0.17	0.09	0.036	0.0028

dwel-time test CP appears. Therefore, this paper concentrates on the question, in which of the different segments of a dwell-time test the fatigue crack grows.

Experimental Methods

The tests were performed on the nickel-base superalloy IN718 applying corner crack (CC) specimens with a quadratic cross section of $7 \times 7 \text{ mm}^2$. The nominal chemical composition of the studied material is given in Table 1:

The applied standard heat treatment consists of:

- (1) Solution annealing for 20 min at 1080°C ,
- (2) Water quenching,
- (3) 12 h ageing at 718°C followed by slow cooling for 12 h to 650°C , and
- (4) Air cooling.

After the heat treatment the specimens were manufactured by turning and milling. Afterwards the surface was mechanically polished with a finishing using a $1\text{-}\mu\text{m}$ suspension. The pre-cracks were grown from a 0.25 mm eroded starter notch by cyclic loading at room temperature to a length of about 1.3 zmm. The subsequent crack propagation tests were stopped at a crack length at about 4–4.6 mm because of two reasons. First, the potential drop crack length monitoring is only working reasonably for cracks of a length of up to 5.3 mm. Second, with increasing crack size, the type of the range of the stress intensity factor ΔK at the crack tip changes to a mixed mode condition, aggravation a comparison of the CP on the basis of the ΔK .

The tests were performed on a servohydraulic test machine under load control. The temperature was established using an induction heating system. For the crack length measurements an alternate current potential drop technique (ACPD) was used with one measuring channel applied to the crack faces and a reference channel applied to the crack free material on the opposite specimen site. The setting of the ACPD was adjusted to work without skin effect to measure the bulk information of the crack. To minimize measurement deviations caused by small changes in the applied current and temperature the signal of the crack channel was divided by that of the reference channel to obtain a normalized ACPD signal. Nevertheless the normalized ACPD signal is influenced by massive changes in the temperature and the load on the specimen. The normalized ACPD signal is linearly connected with the crack length. For the presented tests the absolute reading of the normalized ACPD signal can be different for the same crack length but the increase of the normalized ACPD signal for 1 mm crack increase is identical for all tests.

In addition to that, the surface crack length was documented on one side with a Questar QM1 far-field microscope. Surface crack length means the visible crack length on the surface of the specimen which might behave differently to the

TABLE 2—Start and end bulk crack length and number of loading cycles for the three performed tests.

Load Mode Test	Starter Crack Length	Final Crack Length	Number of Loading Cycles
2s-296s-2s	1.32	4.67	29
296s-2s-2s	1.33	3.74	85
2s-56s-2s	1.1	3.68	485

bulk crack size. For example the CP rate and the crack length of the bulk crack measured by the ACPD can be different to the surface crack observed by the far-field microscope. Similar to the definition of the surface crack is the surface crack propagation, which means the propagation of the visible surface crack.

The CP tests were performed isothermally at 650°C at a stress ratio of $R = 0.1$ with a time per cycle of 300 and 60 s. The cycle time was divided in the loading period, the dwell-time duration at maximum load and the unloading period. For the cycle duration of 300 s two load modes were used. The first has an increasing loading of 2 s up to maximum force of 22.22 kN, a dwell-time duration of 296 s at maximum force and a load decrease of 2 s to the minimum load of 2.22 kN (this test will be called: 2s-296s-2s). Afterwards the cycle starts from the beginning. The second load mode has a load increase of 296 s up to maximum force of 18.89 kN, a dwell-time duration of 2 s at maximum force and a load decrease of 2 s to the minimum load of 1.89 kN (this test will be called: 296s-2s-2s). In case of the cycle time of 60 s a load scheme of 2 s loading up to the maximum force of 22.22 kN, 56 s dwell-time duration at maximum force and 2 s unloading to the minimum force of 2.22 kN (2s-56s-2s) is used. The test was performed in the described order. In the first test the CP rate was fast from the very beginning of the test, please compare the result part. Therefore, in the second test, the 296s-2s-2s load mode test, the minimum and maximum load was altered to have a lower starting value of the range of the stress intensity (ΔK) to achieve more loading cycles until the end of the test for a better analysis.

The calculation of the stress intensity factor is done according to the draft prEN 3873 [25] and for the average crack length and the maximum load in one loading cycle.

The far-field microscope was used to document the surface crack length of the 2s-296s-2s and the 2s-56s-2s load mode test. For the 2s-296s-2s test the pictures were taken at 0, 70, 140, and 210 s of the dwell-time, meaning the time in seconds at maximum load for the dwell-time of 296 s. For the 2s-56s-2s the pictures are made at 0 s and every 5 s afterwards.

Results

A detailed overview of the starter crack length and the final crack length for the different tests is given in Table 2.

Fig. 3 shows the increase of the normalized ACPD signal of the two tests with a cycle time of 300 s. Part a shows the increase for the load mode test

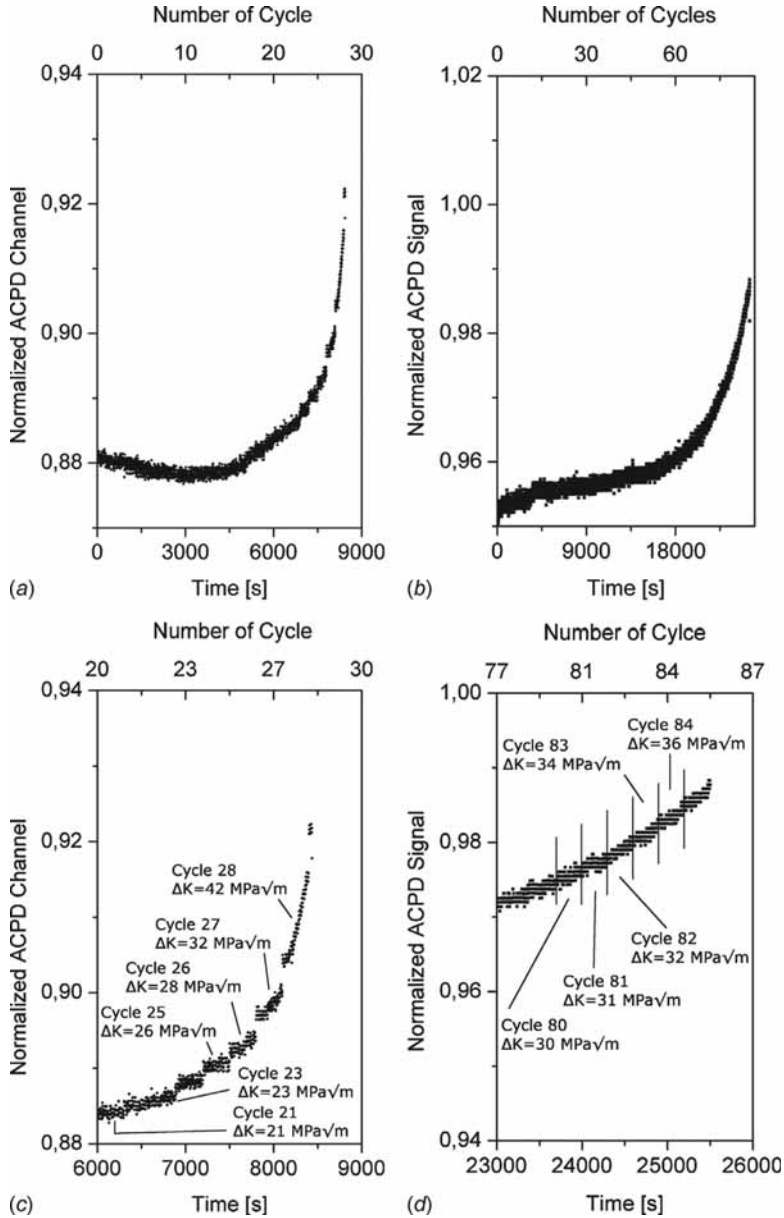


FIG. 3—Increase of the normalized ACPD channel signal over the testing time and cycle for (a) the total 2s-296s-2s test; (b) the total 2s-296s-2s test; (c) a selected part of the 2s-296s-2s test; and (c) a selected part of the 296s-2s-2s test.

TABLE 3—Crack propagation rate and ΔK for selected cycles of the tests with 300s cycle time.

Load Mode 2s-296s-2s			Load Mode 296s-2s-2s		
Cycle Number	ΔK [MPa $\sqrt{m}$]	da/dN [m/cycle]	Cycle number	ΔK [MPa $\sqrt{m}$]	da/dN [m/cycle]
21	21	8.6×10^{-4}	77	27	6.6×10^{-5}
22	22	8.4×10^{-4}	78	28	7.7×10^{-5}
23	23	7.5×10^{-5}	79	29	9×10^{-5}
24	24	1.8×10^{-4}	80	30	1.1×10^{-4}
25	26	1.9×10^{-4}	81	31	1.1×10^{-4}
26	28	2.1×10^{-4}	82	32	1.3×10^{-4}
27	32	4.3×10^{-4}	83	34	1.5×10^{-4}
28	42	8.3×10^{-4}	84	36	1.7×10^{-4}

2s-296s-2s and part b for the load mode test 296s-2s-2s. As can be seen, the increase of the ACPD signal is mostly smooth, but for the last 3000 s in the 2s-296s-2s test there is a step-wise behavior. For a better representation of this behavior, this part of the test is enlarged in Fig. 2(c) including the ΔK value for these cycles. The same representation of the test 296s-2s-2s is shown in Fig. 2(d).

The step length corresponds to the dwell-time duration because the steps have a time of 300s which perfectly fits to a cycle time of 300 s. For the cycle numbers 21 to 24, which correspond approximately to $21 \text{ MPa}\sqrt{m} < \Delta K < 28 \text{ MPa}\sqrt{m}$, the increase of the ACPD signal during the dwell-time is very small in comparison to the increase between two dwell-times. For higher cycle numbers (27, 28) and higher ΔK ($32 \text{ MPa}\sqrt{m} < \Delta K < 42 \text{ MPa}\sqrt{m}$) there is a stepwise increase between two dwell-time durations and a large increase during the dwell-time duration. For lower cycle numbers ($N < 20$) and therefore also lower ΔK ($\Delta K < 21 \text{ MPa}\sqrt{m}$) no stepwise increase was detected in the ACPD signal, as the signal always has a smooth increase over time, shown in Fig. 3(a). For the whole test time of the 296s-2s-2s load mode test there was a smooth increase of the normalized ACPD signal measured as shown in Figs. 3(b), 3(d). Please refer to Table 3 for the individual ΔK and the CP rate for the selected loading cycles, to compare the CP between the 2s-296s-2s and the 296s-2s-2s load mode test.

A normal crack propagation plot with the crack propagation rate plotted versus ΔK is shown in Fig. 4. The CP for the 2s-296s-2s and the 296s-2s-2s load mode test are plotted according to the measured absolute values of the ACPD but the CP for the 2s-56s-2s test is averaged over ten loading cycles to minimize the noise influence. For the load mode of 296s-2s-2s the CP rate is slower than for the load mode of 2s-296s-2s and the CP rate of the 2s-56s-2s load is the slowest.

The pictures of the far-field microscope of the 2s-296s-2s test document the surface crack and can be used as additional information to supplement the

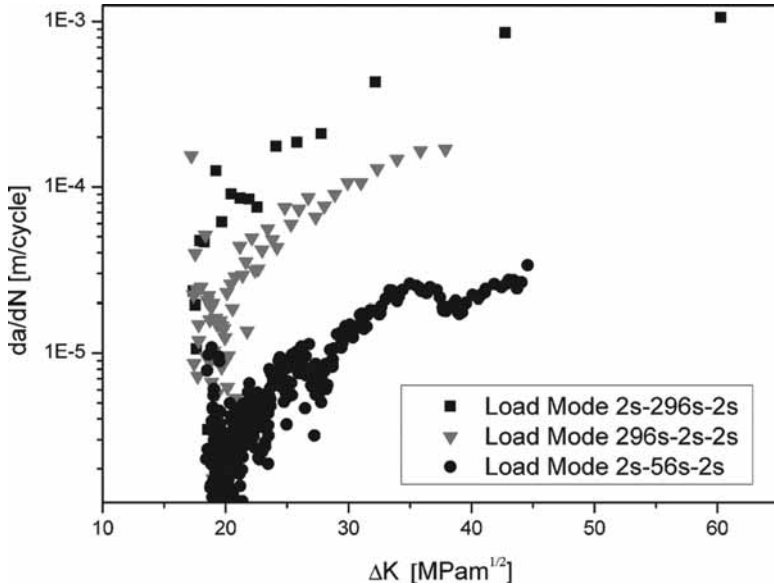


FIG. 4—Crack propagation rate for IN718 at 650°C for three load modes under load control.

results obtained by means of the ACPD technique. An observation of the crack propagation through the far-field microscope was only possible for high cycle numbers (> 20). However, when the crack tips at the end of the 22nd loading cycle (210s after beginning of dwell-time duration, 86 s before the end of the dwell-time duration) and the beginning of the dwell-time duration of the 23rd loading cycle (0 s) are compared, there is a small movement of the crack tip, cf. Fig. 5. During the dwell-time duration of 22nd cycle there was no surface CP. Please be aware that the surface CP between cycle 22 and 23 in Fig. 4 is much smaller than the measured CP of the bulk crack, Table 3. For all other cycle numbers there was also CP in the dwell-time duration in the far-field microscope pictures, or the crack tip was not recognizable.

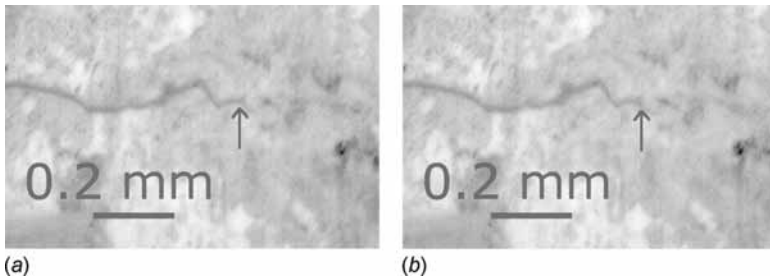


FIG. 5—Crack at (a) the end of the 22nd loading cycle; (b) in the beginning of the 23rd loading cycle.

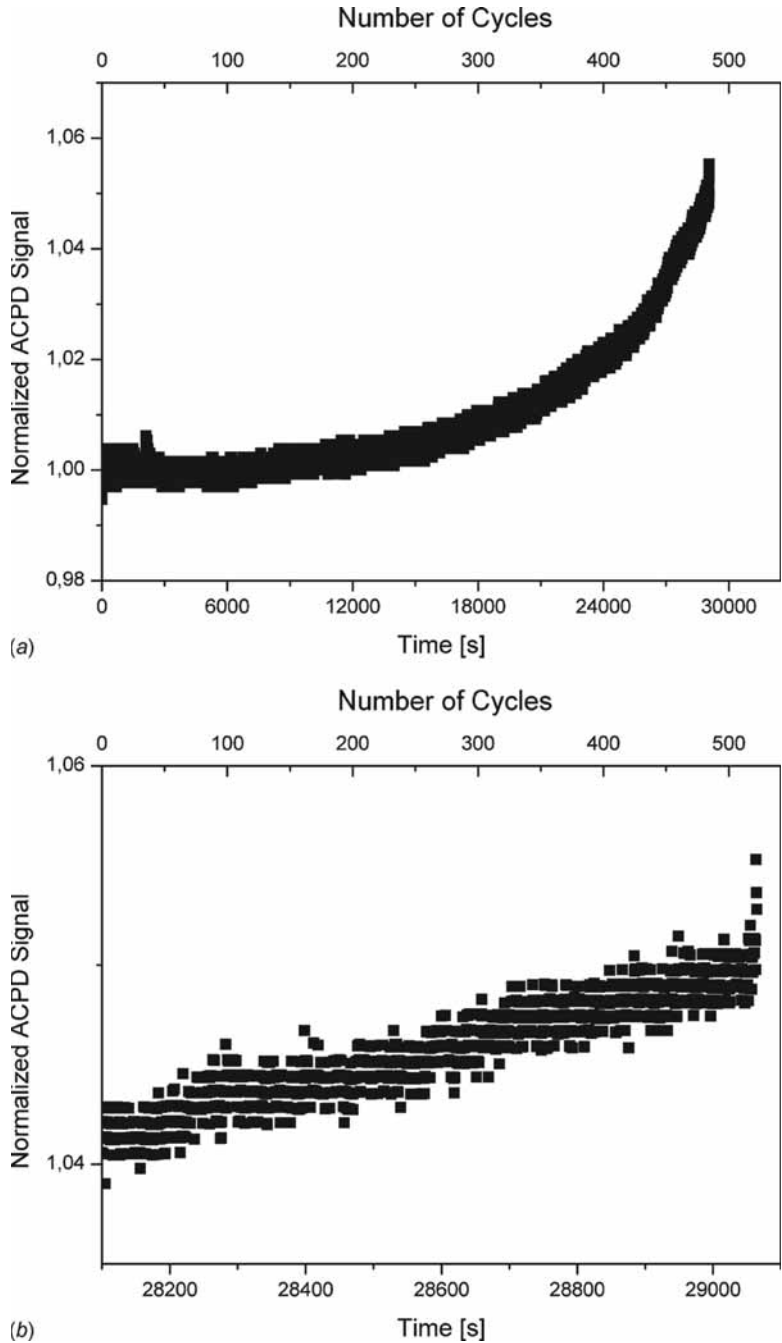


FIG. 6—Increase of the normalized ACPD channel signal versus testing time for the 2s-56s-2s load mode: (a) Whole test; (b) last 1000s of test.

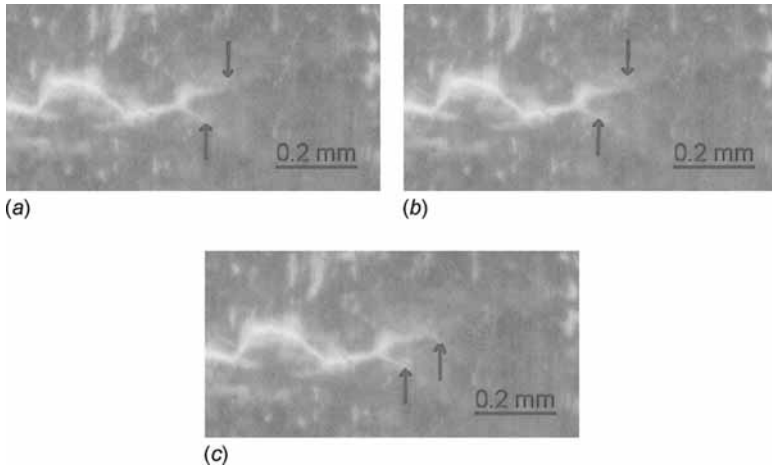


FIG. 7—Crack length for the load mode 2s-56-2s load at (a) 471st cycle at 0s of dwell, (b) 471st cycle at 55 s of dwell; and (c) 472nd cycle at 0 s of dwell.

Fig. 6(a) shows the increase of the ACPD signal of the 2s-56s-2s load mode test. For the whole test preferentially only a smooth increase was detected as the enlargement of the last 1000 s of this test supports, cf. Fig. 5. A representation of the CP rate over ΔK is plotted in Fig. 4.

The far-field microscope was applied to study the fatigue crack propagation during individual cycles for the 2s-56s-2s test. The picture quality of the 2s-56s-2s load mode was better as for the 2s-296s-2s load mode test. Therefore, for some cycles a clear observation of the crack tip was possible. In Fig. 7 the crack tips of the 471st and 472nd loading cycle are depicted. Clearly, the crack tip moves between the end of the 471st and the beginning of the 472nd cycle. At the beginning of the 471st loading cycle (0s, Fig. 6(a)) the crack tip is in the same position as at the end of the 471st loading cycle (55 s, Fig. 6(b)) but there is a crack elongation visible at the beginning of the 472nd loading cycle (0 s, Fig. 6(c)). Similar observations were made for the loading cycles 470 to 475 ($\Delta K = 40 \text{ MPa}\sqrt{\text{m}}$ to $42 \text{ MPa}\sqrt{\text{m}}$) which rather showed CP between two dwell-time durations than during the dwell-time duration. For all other load cycles, there was either a CP in any part of the loading cycle or it was not possible to correlate the CP to any part of the loading cycle.

Discussion

The results correspond to the theory of the dynamic embrittlement. In case of the 2s-296s-2s load mode test the CP rate is faster than in the 296s-2s-2s load mode test showing that the 2s-296s-2s load mode is more detrimental to the material which can be explained by the dynamic embrittlement, cf. Fig. 4. The average stress intensity in one loading cycle in the 296s-2s-2s is smaller than in the 2s-296s-2s load mode test. As the oxygen grain boundary diffusion and

embrittlement is tension activated [1,2,4-7] a higher average stress intensity results in a higher average tension in the crack tip which results in a higher oxygen grain boundary diffusion and a faster embrittlement with faster CP. The same explanation applies to the 2s-56s-2s load mode test with the slowest CP rate. The dwell-time duration is shorter than for the two tests with 300 s cycle time which reduces the oxygen grain boundary diffusion because less time is given for diffusion [3,7].

In the introduction, results of Lynch et al. [18,19] and Andersson et al. [17] were presented which show a faster CP rate for tests with low test frequencies in comparison to test with the same cycle time but long dwell-time durations. The long dwell-time durations were obtained by rather fast loading and unloading rates between maximum and minimum load. The authors explained their findings by a pronounced crack branching and the formation of a thicker oxide layer in dwell-time tests in comparison to low frequency testing. The comparison of the 2s-296s-2s and the 296s-2s-2s test of the present paper does not support this observation. One possible explanation might be that the tests of this paper are performed under load control in contrast to Lynch et al. and Andersson et al., who performed the test under strain control. During the dwell-time duration in a strain controlled test there is a decrease of the load when the maximum strain is reached and the crack propagates or the material deforms plastically by creep, whereas in load control tests the stress intensity stays the same or increases when the crack propagates or specimen elongation appears by creep. This different behavior in strain and load controlled tests could explain the results of Lynch et al. and Andersson et al. considering the lower resulting oxygen diffusion in the dwell-time test.

The main aspect of this paper was to analyze, in which part of the loading cycle in a dwell-time test CP appears. For the cycle numbers 21 to 24 ($21 \text{ MPa}\sqrt{\text{m}} < \Delta K < 28 \text{ MPa}\sqrt{\text{m}}$) in the 2s-296s-2s load mode test the ACPD increases step-like. As these steps are about 300 s long, they correspond to the dwell-times. Hence, it is a reasonable assumption that the CP appears mainly in the unloading and loading parts between two dwell-time durations for these cycle numbers of the test. For higher cycle numbers (26 to 28, $28 \text{ MPa}\sqrt{\text{m}} < \Delta K < 42 \text{ MPa}\sqrt{\text{m}}$) there is an increase of the normalized ACPD signal in the dwell-time duration, besides the step-wise increase between two dwell-time durations, meaning that the crack propagates in any part of the loading cycle. For lower cycle numbers ($N < 20$, $\Delta K < 21 \text{ MPa}\sqrt{\text{m}}$) it was not possible to identify in which part of the load cycle the increase of the ACPD appears. The reason might be that the noise level of the ACPD system was higher than the corresponding crack-induced increase of the ACPD signal in these cycles.

The surface crack length at the end of the 22nd and at the beginning of the 23rd cycle of the 2s-296s-2s test observed by the far-field microscope support the results of the ACPD, cf. Fig. 4. However, other cycles showed CP in any part of the loading cycle or the crack tip was not recognizable, and hence the results of the ACPD could not be confirmed unambiguously. Especially as the CP of the surface crack in Fig. 4 between cycle 22 and 23 is very different to the bulk crack propagation shown in Table 3, the question arises, if the surface CP and the bulk CP are correlated or independent from each other. The surface crack could

behave differently due to a different stress state than the bulk crack. Furthermore the surface crack only enables the analysis of a single surface grain instead of an average CP rate for a high number of grains at the crack tip which are analyzed by the ACPD. Therefore, the results obtained by the far field microscope support the measurements of the ACPD only in such a way that they show that there are conditions in which the CP appears mainly in the unloading and loading parts.

In case of the 296s-2s-2s load mode the increase of the normalized ACPD signal is smooth and does not show any step-wise character. The obvious interpretation would be that this shows CP in any part of the loading cycle. Nevertheless the reader should be aware that the normalized ACPD signal sees influences through massive load which definitely applies to this load mode, as the difference of 17 kN between maximum and minimum load is applied in the load cycle. This hardens the correlation of the CP to the three parts (loading, dwell-time duration and unloading) of the loading cycle, but still the most likely interpretation of the signal is that the CP appears smoothly in the loading cycle because there is no step-wise increase of the ACPD signal after the unloading or the short dwell-time duration which would be a sign for CP. As conclusion the CP appears in the loading part, as this covers most of the time for one loading cycle.

The 2s-56s-2s load mode test also shows a smooth increase of the normalized ACPD signal. Maybe the CP rate in one loading cycle is too slow for a correlation to the loading cycle segments as this information is lost in the noise level of the ACPD. Otherwise the CP could really appear in any part of the loading cycle. However, the far-field microscope pictures of the 2s-56s-2s test have a better quality than for the 2s-296s-2s test and show for the cycle number 470 to 475 clearly mainly CP in the unloading and loading between two dwell-time durations for the cycle number 470 to 475. For all other cycles there was either a CP in the dwell-time duration or an observation of the crack tip was not possible. Therefore, the far-field microscope pictures of the 2s-56s-2s load mode support basically the observation of the ACPD for the 2s-296s-2s load mode test that there are conditions of main CP in the unloading and loading part.

Obviously, there are two different CP conditions. First those which have the main CP in the unloading and loading part between two dwell-time durations and second those which have CP in any part of the loading cycle. The following part will propose a possible explanation why there are different CP behaviors. The detrimental effect of the dynamic embrittlement is the oxygen grain boundary diffusion, which appears in the dwell-time duration. This can be seen in a load relaxation test of Pfaendtner and McMahon [7]. At laboratory air under constant displacement they observed a drastic reduction of the load which is not found in vacuum. Hence, when a load is applied there is a detrimental effect on nickel-base material. In case of the long dwell-time duration of the 2s-296s-2s load mode test for low ΔK there is huge oxygen grain boundary diffusion in the dwell-time duration. As a result the material might embrittle in the dwell-time duration, which might support the cracking in the unloading and loading. The stress intensity in the dwell-time duration might be too small for CP although the detrimental effect through the oxygen diffusion appears. For

higher values of ΔK the stress intensity in the dwell-time duration is high enough for CP also in the dwell-time duration. For very low ΔK the dwell-time duration the CP rate is too small to correlate the CP to the loading cycle with the used ACPD system. Therefore, the ΔK might influence in a dwell-time test in which part of a loading cycle the main CP appears.

For short dwell-time durations as in the 2s-56s-2s load mode the oxygen diffusion and the grain boundary embrittlement in the dwell-time duration is smaller than in long dwell-time durations. Therefore, the CP in the unloading and loading part is smaller as less embrittlement appeared and it might be too small for detection with the used ACPD.

Finally, more research must be carried out to identify the dependence of the CP in the different parts (loading, dwell-time, unloading) of a dwell-time test on the dwell-time duration, the stress intensity and the range of the stress intensity.

Summary

The effect of dynamic embrittlement on the crack propagation rate of IN718 at elevated temperatures in fatigue tests with and without dwell-time was dealt with and the influence of frequency was considered. From the literature it is well-known that in general an increase of the dwell-time duration or a reduction of the test frequency increases the crack propagation rate. However, in the case of dwell-time testing it is not known, where within the individual cycle the crack propagation takes place. Therefore, tests were carried out at 650°C aiming at a correlation between the crack growth behavior with the loading cycle segments. The bulk crack size was measured by an ACPD and the surface crack size at one of the crack surfaces of the CC specimen used was monitored by means of a far-field microscope.

The results obtained support basically the classical idea of dynamic embrittlement that as a consequence of oxygen penetration along grain boundaries long times with an open crack lead to an accelerated crack propagation. The analysis of the ACPD measurements observations show that for the lower range of the stress intensity factor in tests applying long dwell-times there is pronounced crack propagation in the unloading and loading parts of the cycle and only minor crack propagation occurs during the dwell-time (i.e., constant stress). However, tests at higher values of the stress intensity factor range indicate that crack extension takes place within all segments of the complete cycle leading to appreciable crack growth also during the stress holds. The observation with the far-field microscope basically supports these findings.

Hence, the combination of dwell-time duration and the value of the corresponding stress intensity factor and the range of the stress intensity appears to determine the crack growth behavior within an individual fatigue loading cycle.

Acknowledgments

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Sachin Shinde¹ and Philip Gravett¹

The Effects of Dwell on the LCF Behavior of IN617

ABSTRACT: Much has been studied on the individual effects of creep and fatigue on alloy life. However, not much is known of the combined effects of these two mechanisms. Therefore, a study was put into place to determine the effects of dwell on fatigue life and deformation mechanisms, for IN617, a solid-solution strengthened Ni-base alloy used widely in the power generation and aerospace industries. Low cycle fatigue (LCF) tests were conducted from 649–982 °C with either tensile or compressive dwell. Fracture surfaces of the test specimens as well as longitudinal and transverse sections were examined via scanning electron microscopy to determine the damage and failure mechanisms. Test results confirmed tensile dwell lives that were significantly lower than those seen in compressive dwell. The mechanics for the reduction in cyclic life for tensile dwell was attributed to creep damage accumulation at grain boundaries that led to widespread intergranular cracking and failure. Tensile dwell life reductions were largest in tests at moderate (649–760 °C) temperatures. The failed specimens for this temperature range showed the most evidence of grain boundary cavitation and intergranular cracking. At higher test temperatures, the tensile dwell sensitivity for IN617 was significantly reduced or almost entirely eliminated at high temperatures (871–982 °C). This was attributed to the lower stresses that developed at these temperatures for a given strain range. The LCF testing and subsequent analysis indicated that a substantial tensile stress during dwell time coupled with moderate temperatures, to allow for diffusion creep, lead to grain boundary damage that can reduce cyclic life.

KEYWORDS: LCF, Dwell, Fatigue, Intergranular, Transgranular

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¹ Siemens Energy Inc., Orlando, FL.

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Introduction

Strain-controlled LCF has long been used to help predict fatigue behavior of structural components in service. To better represent actual service conditions, dwells at constant strain loading can be applied during LCF testing. The combination of these hold or “dwell” times and an elevated temperature environment can lead to creep-fatigue interactions that often cause material failures more quickly than would be predicted by summing the individual effects of creep and fatigue [1]. Factors that have been shown to affect dwell time LCF behavior include test temperature, environment, dwell time duration, and the nature of dwell (tensile versus compressive) [2,3]. The reduction in lifetime due to dwell at elevated temperatures is believed to come as a result of the introduction of different deformation modes than those observed in LCF testing without dwell time. Traditional creep phenomena, such as grain boundary cavitation and sliding as well as oxidation damage, facilitate the lower cyclic lifetimes when dwells are present [3]. Dynamic strain aging can also occur during dwell time LCF testing and can lead to serrated yielding. This process can be described as the re-pinning of dislocations by interstitial atoms after initial yielding, where dwell times at elevated temperature allow time for atoms to diffuse to dislocations [4].

The purpose of this study is to examine the effects of dwell time on the LCF behavior of IN617, a solid-solution strengthened Nickel-base alloy with good oxidation resistance. It is a common alloy used in land based gas turbine engine transitions connecting the combustor and turbine sections. Nickel-base alloys in particular have demonstrated a pronounced sensitivity to dwell time effects [2]. A significant reduction in LCF lifetimes is observed when a tensile dwell is applied as compared to no dwell LCF or compressive dwell test conditions. This is in contrast to the general trend that compressive dwells lead to lower lifetimes due to the development of a positive mean stress. The lifetime reduction in IN617 was initially attributed to serrated yielding observed in stress-strain hysteresis loops for the tensile dwell tests. The work presented here involved re-analysis of previous test results, conducting new trials to gain further understanding of serrated yielding in IN617, and investigate whether or not it is the primary cause of lifetime reductions in tensile dwell LCF tests.

Experimental Procedures

Materials

The IN617 material was supplied by Haynes International in the form of 1 in. thick solution annealed plate, with the chemical composition shown in Table 1.

Specimens

Cylindrical samples were machined with low stress grinding and polishing of the gage sections to eliminate machining marks.

TABLE 1—Chemical composition in wt. % of IN617 Alloy plate used in this study.

Ni	Balance
Cr	22.10
Co	12.30
Mo	9.49
Fe	1.13
Al	1.07
Ti	0.38
C	0.080
Mn	0.07
Si	0.06
Cu	0.02
P	0.002
B	<0.002
S	<0.002

Dwell Time Fatigue Test Approach

LCF testing was carried out at Cincinnati Testing Laboratories and Siemens Energy Inc., Casselberry Labs on MTS servo-hydraulic test frames. A high temperature extensometer provided the strain control to run the tests. All of the tests were performed in air with a strain rate of 6%/min. Tests were performed at 649°C, 760°C, 871°C and 982°C with several different strain ranges and temperatures, while the additional tests were performed primarily at one strain range and a temperature of 760°C. The focal point of the tests at 760°C was to investigate the serrated yielding observed in this alloy and to gain confidence in dwell sensitivity and failure mechanics.

Twenty minute dwell times were introduced at maximum strain for tensile dwell testing and minimum strain for compressive dwell testing. Positive strains ($R\epsilon = 0$) were applied for tensile dwell testing and negative strains ($R\epsilon = -\text{inf.}$) were

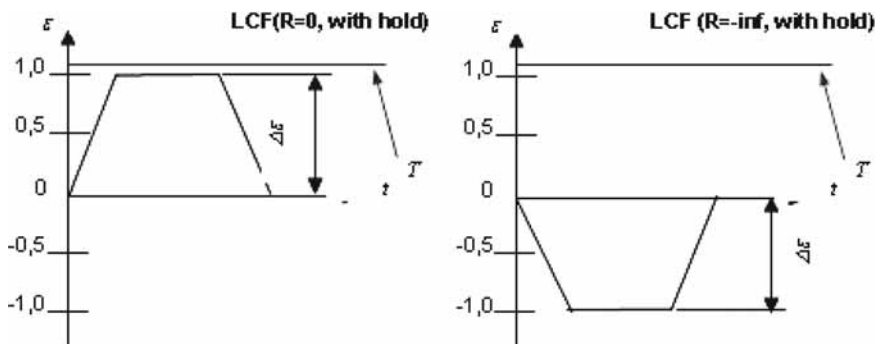


FIG. 1—Schematic of strain life cycles for the tensile and compression dwell.

used for compressive dwell testing. Figure 1 represents the strain cycles applied. Tests were run until specimen failure by fracture, the number of cycles to initiation (Ni) was determined to be at the point where the maximum load dropped below the stabilized maximum load and number of cycles to failure (Nf) was determined to be at the point where the maximum load dropped below the stabilized maximum load.

Post Test Inspections

A number of test specimens underwent fractographic and microstructural analysis after testing to reveal damage mechanisms and to characterize failures. Fracture surfaces were cleaned with acetone and inspected via scanning electron microscope (SEM). Longitudinal and transverse sections were prepared by sectioning down the length of the gage section, polishing, and etching with a 10 % oxalic acid aqueous solution. Metallographic specimens were examined via optical microscopy and SEM.

Experimental Results and Discussion

During the evaluation of the dwell fatigue test results, and post test inspections, it was observed that there are two distinct behaviors that seem to be influenced by temperature. This influence is segregated as “moderate” and “high” temperature behavior in the following paragraphs.

Effects of Dwell Time on Fatigue Life (Moderate Temperatures)

At moderate temperatures, 649°C and 760°C, the effect of dwell fatigue is quite pronounced when measured on strain range versus cycles. As seen in Fig. 2 a significant reduction in life is observed at both temperatures with the application of the dwell time when compared to standard no dwell LCF tests. There appears to be a 60 % reduction in life due to a compressive dwell, and nearly 90 % reduction in life due to a tensile dwell. As measured by strain range, the fatigue strength reduction is 30 % on compressive strain and 60 % on tensile strain. Although the amount of testing is limited, there appears to be consistent trend on fatigue strength reduction throughout the test range.

Effects of Dwell Time on Fatigue Life (High Temperatures)

At high temperatures, 871°C and 982°C, the effect of dwell fatigue is much less pronounced when measured on strain range versus cycles. As seen in

Figure 3, a minor reduction in life is observed at 871°C with the application of the dwell time, when compared to standard no dwell LCF tests. There appears to be a 45 % reduction in life due to a compressive dwell, and 60 % reduction in life due to a tensile dwell. The test results at 982°C indicate there is no influence on the life from either the compressive or tensile dwell.

Stress History and Evolution

Mean Stress.—Table 2 shows a summary of LCF results from recent 760° C testing at Siemens Energy Inc., Casselberry Labs. It is clear that both tests that

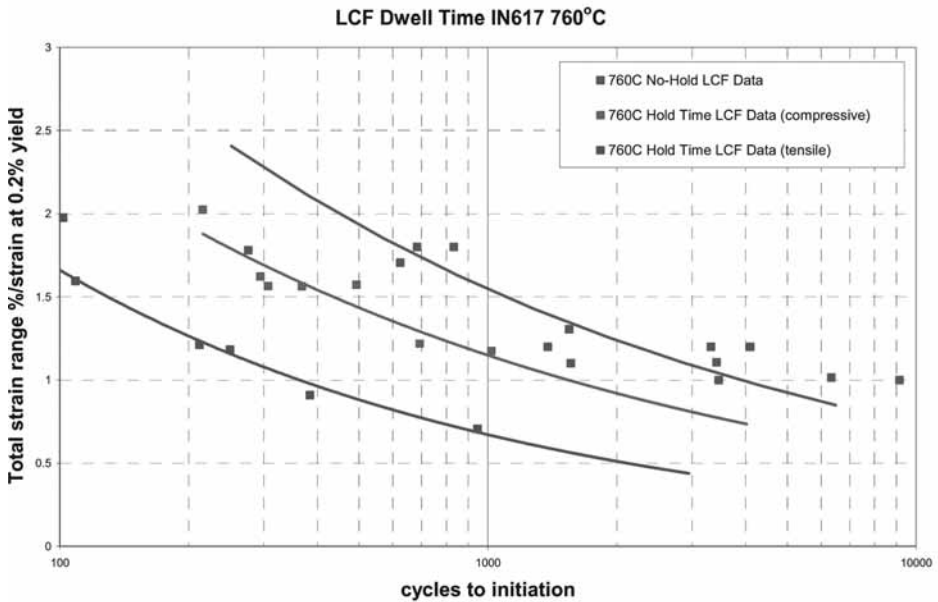
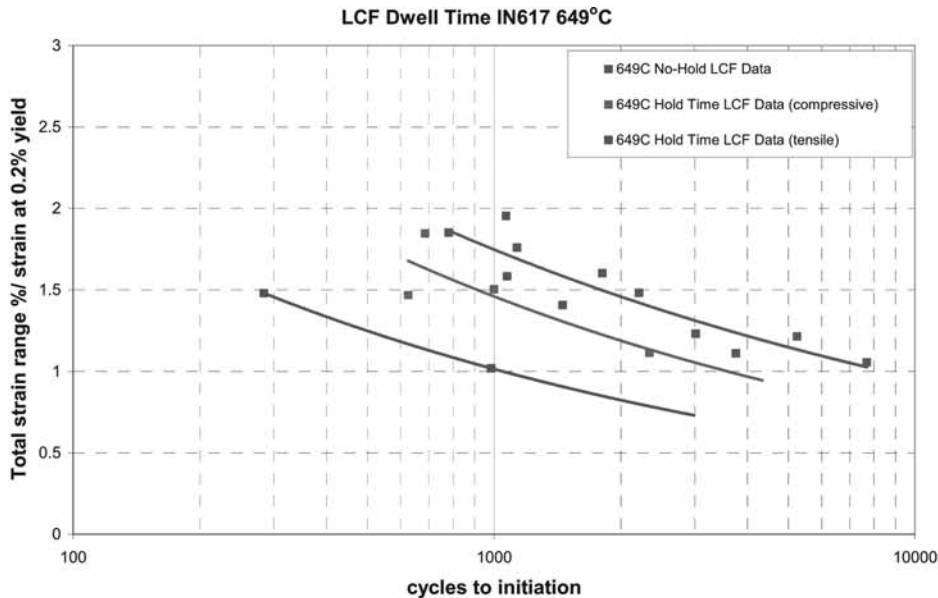


FIG. 2—Dwell fatigue effects at moderate temperatures.

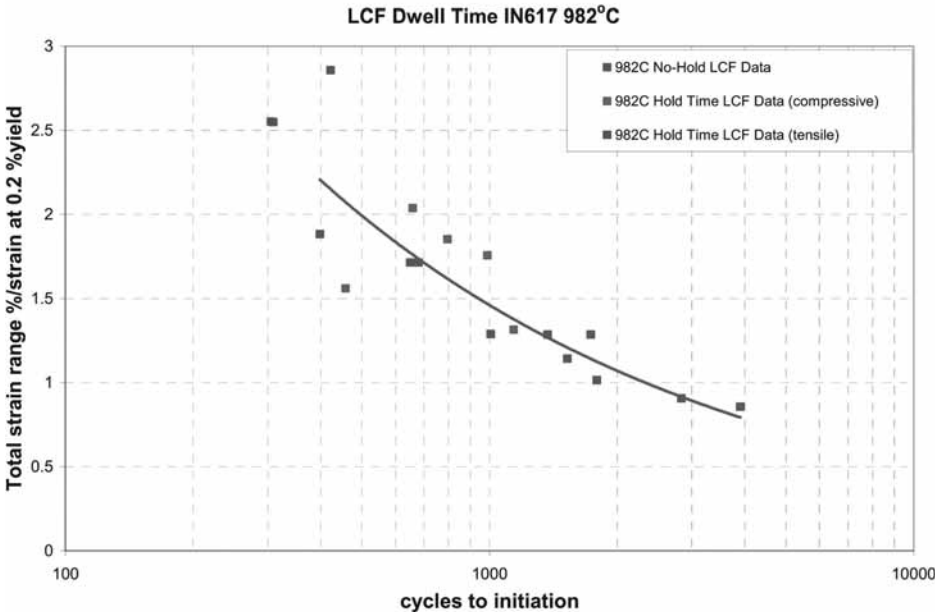
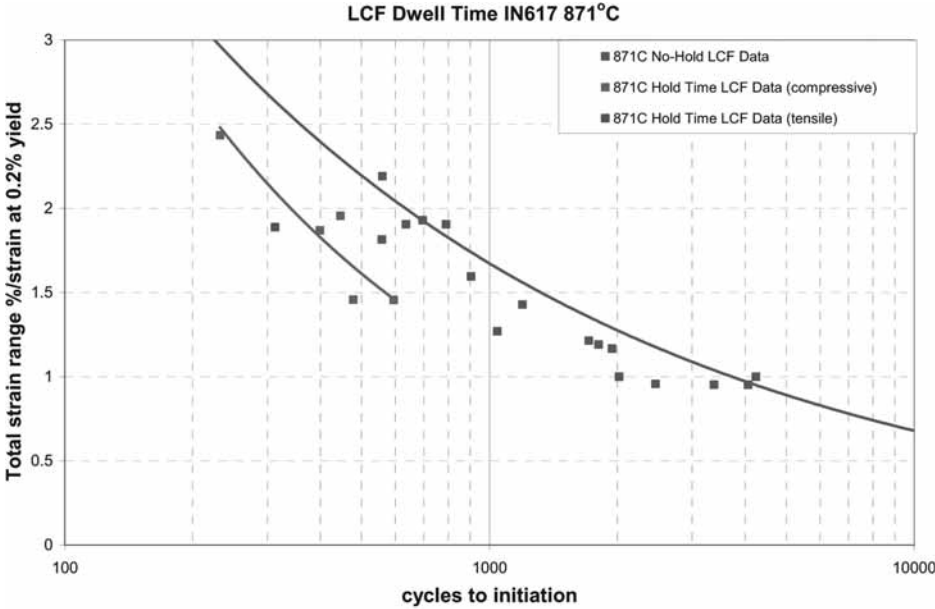


FIG. 3—Dwell fatigue effects at moderate temperatures.

TABLE 2—Summary of recent LCF with hold time testing at 760°C.

Specimen	Hold Type	Normalized Total Strain Range	Mid-life mean stress (MPa)	N _i	N _f
LCF-1	Tensile	1	− 32.50	109	122
LCF-2	Tensile	1	− 32.80	80	85
LCF-3	Compressive	1	17.55	307	321
LCF-4	Compressive	1	22.55	368	401

underwent compressive dwells had significantly longer lifetimes than the tensile dwell tests. Positive mean stresses in the compressive dwell tests and negative mean stresses in the tensile dwell tests developed and are shown in Fig. 4 at the mid-life ($N \sim N_i/2$) of the trial. The behavior shown here does not follow the general rule that development of positive mean stresses reduces LCF lifetimes compared to negative mean stresses.

Stress Relaxation

The stress relaxation at mid-life comparison shown in Fig. 5 illustrates the similarities in stress response during tensile and compressive dwell time. The mid-

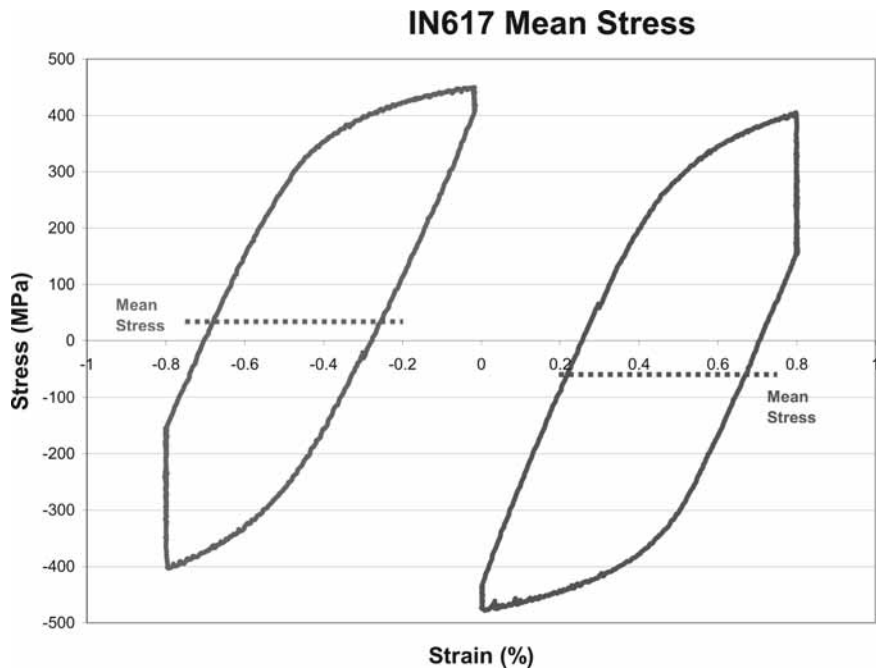


FIG. 4—Mean stress response from compressive dwell and tensile dwell tests at mid life.

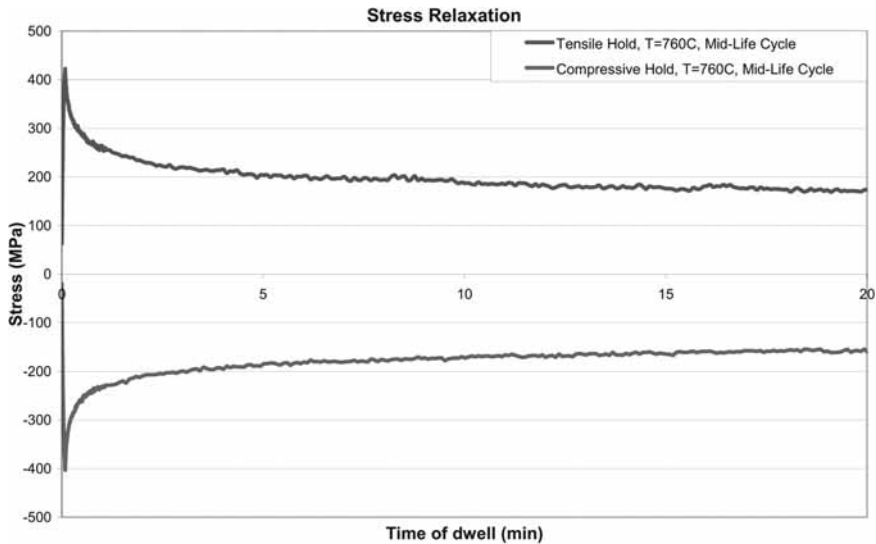


FIG. 5—The stress relaxation at mid-life comparison for tensile and compressive dwell at 760°C.

life stress relaxation is 245 MPa during the tensile dwell, and 252 MPa during the compressive dwell.

The stress relaxation at mid-life comparison shown at moderate and high temperatures in Fig. 6 illustrates the stress response during tensile dwell time.

The shorter lifetimes in tensile dwell testing are likely due to differences in creep damage accumulation. The primary cause of creep damage and shorter lifetimes at moderate temperatures is that at these two temperatures IN617 has a higher yield stress and experiences more work hardening than at higher temperatures. The likely cause of the increase in yield stress near 700°C is thermally activated cross slip of dislocations. Dislocations become active on more slip planes and interfere with each other to work harden the material [3,4]. A material with a higher yield stress and higher levels of work hardening will require more stress to reach a specific strain in LCF testing.

As seen in Fig. 6, the relaxed maximum stress during the tensile dwell at 649°C is 343.6 MPa, or is 54.6% of the ultimate tensile strength for IN617 at that temperature. The relaxed maximum stress during the tensile dwell at 760°C is 170 MPa, or is 35% of the ultimate tensile strength for IN617 at that temperature. The relaxed maximum tensile dwell stress in the 760°C test is a smaller fraction of the ultimate tensile strength than the 649°C test. The shorter lifetime at 760°C can be explained by the fact that more creep contributing diffusion processes can occur at 760°C than at 649°C. Creep deformation is a function of both temperature (environment) and stress. It appears that for IN617 a substantial tensile stress during the dwell coupled at moderate temperatures high enough to allow for diffusion leads to creep damage at grain boundaries that

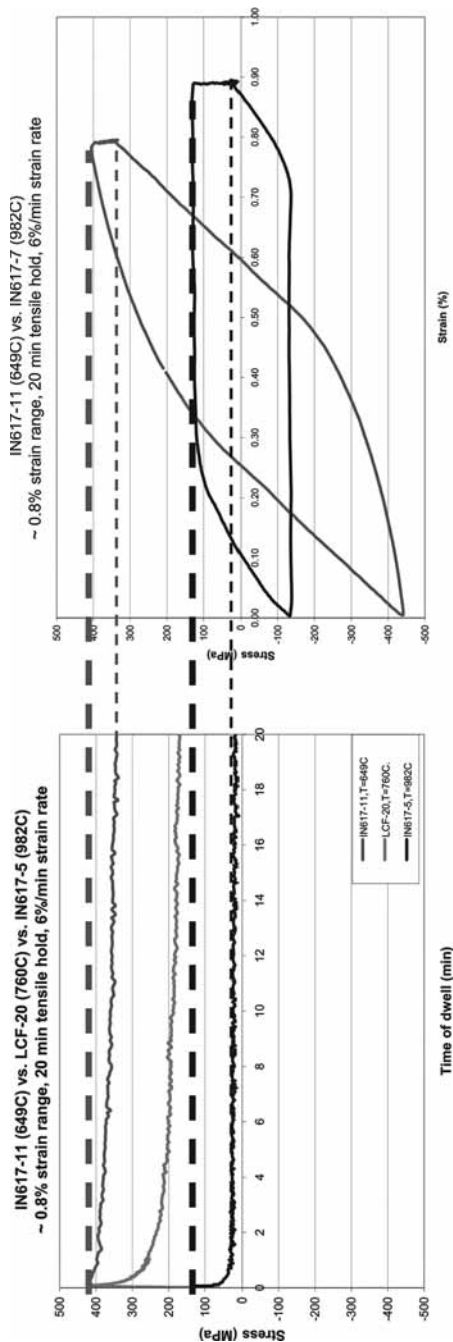


FIG. 6—The stress relaxation at mid-life comparison for dwell at 649°C, 760°C and 982°C.

can reduce cyclic lifetimes. The shorter lifetimes in tensile dwell at moderate temperatures are likely due to differences in creep damage accumulation.

The relaxed maximum stress during the tensile dwell at 871°C is 32.5 MPa, or is 11.5 % of the ultimate tensile strength for IN617 at that temperature. For testing at 982°C, on the other hand, the relaxed maximum stress during the tensile dwell is 19.3 MPa, which is only 14.1 % of the ultimate tensile strength at 982°C. It appears that for IN617 a substantial tensile stress is not retained during the dwell at high temperatures; therefore at higher temperatures, cyclic plasticity is more dominant than creep, see Fig. 6.

Serrated Yielding

Dynamic strain aging can also occur during dwell time LCF testing and can lead to serrated yielding. This process can be described as the re-pinning of dislocations by interstitial atoms after initial yielding, that result in a nearly instantaneous drop of the measured stress in strain control. If the serrated yielding occurs faster than the strain control capabilities of the test rig can react, a brief increase in strain can result as the control attempts to return to the specified strain.

Hysteresis loops for a compressive and tensile dwell tests indicate significant serrated yielding occurred during the first cycle of both loadings, shown in Fig. 7. Hysteresis loops for subsequent cycles; however, do not show significant serrated yielding for either compressive or tensile hold cycles shown in Fig. 8. This is an indication that serrated yielding is unlikely the primary reason for fatigue lifetime differences observed since it occurs during such a small portion of the test cycles.

Failure Mechanism (Moderate Temperatures)

Fracture surfaces, longitudinal and transverse sections were characterized to determine failure modes and the extent of creep damage accumulated in tensile dwell and compressive dwell tests. The fracture surfaces for tensile dwell tests at the moderate temperatures shown in Fig. 9 indicate a clear inter-granular fracture commonly associated with creep effects [5]. This failure mode typically occurs due to damage accumulation at grain boundaries that can lead to the motion of boundaries and grain boundary cavitation, see

Figure 10, under tensile loads for long periods of time. With this grain boundary damage, no singular crack formation site was observed due to multiple cracks nucleating at grain boundaries and linking up to lead to final fracture. This widespread creep damage results from the tensile dwell time applied during LCF testing of IN617 at these temperatures. This effect has also been observed by other researchers, and has been identified as the primary cause of reduced LCF lifetimes for IN617 when tensile dwells are applied [2,6].

Analysis of specimens that underwent compressive dwell LCF testing revealed a distinctively different failure mechanism than tensile dwell LCF. As seen in Figure 11, failure occurred by crack formation at the surface followed by crack growth leading to overload and failure of the specimen. Microstructural

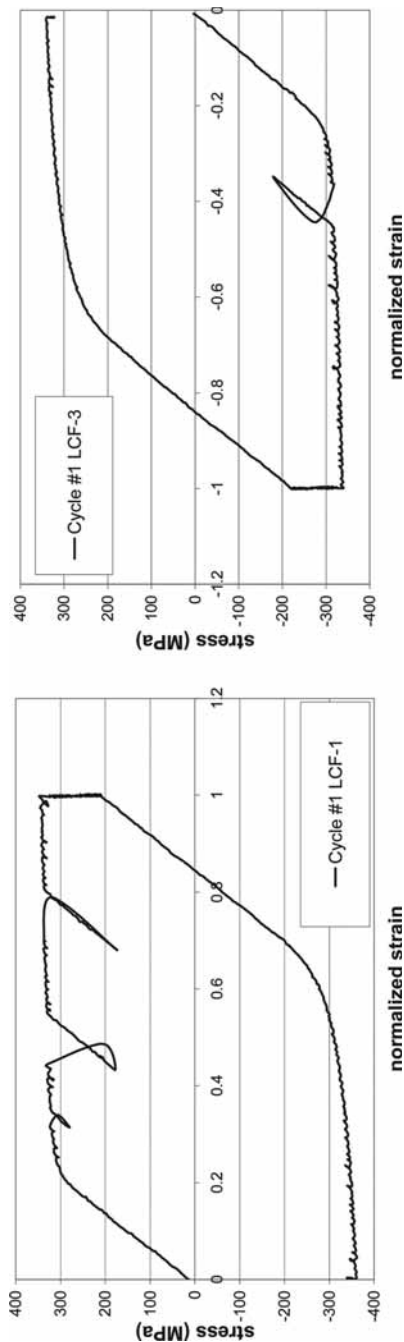


FIG. 7—Serrated yielding observed on the first cycle of tensile and compressive dwell cycles.

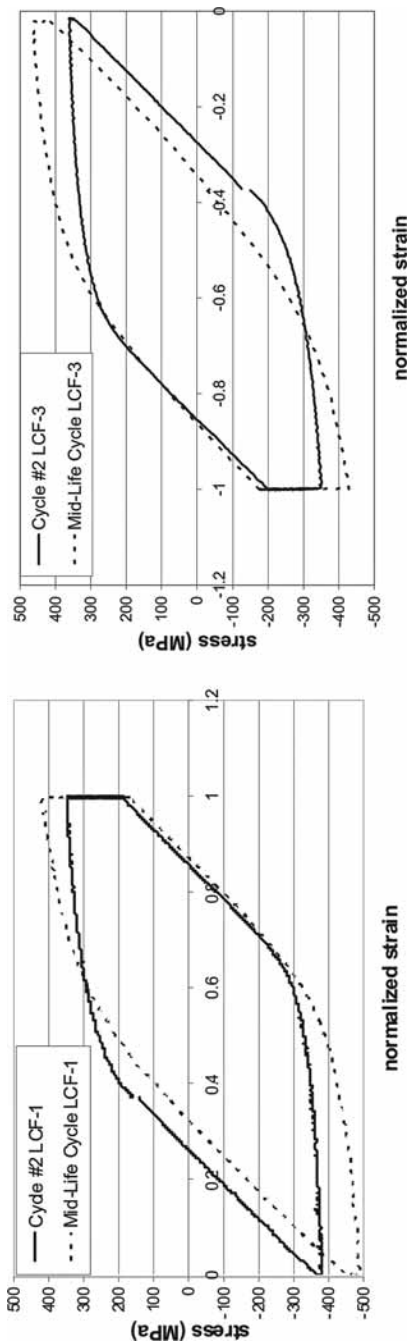
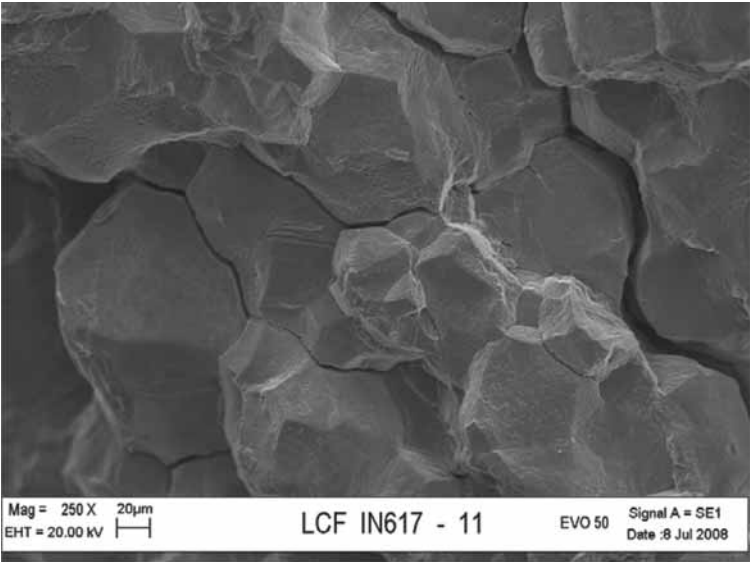
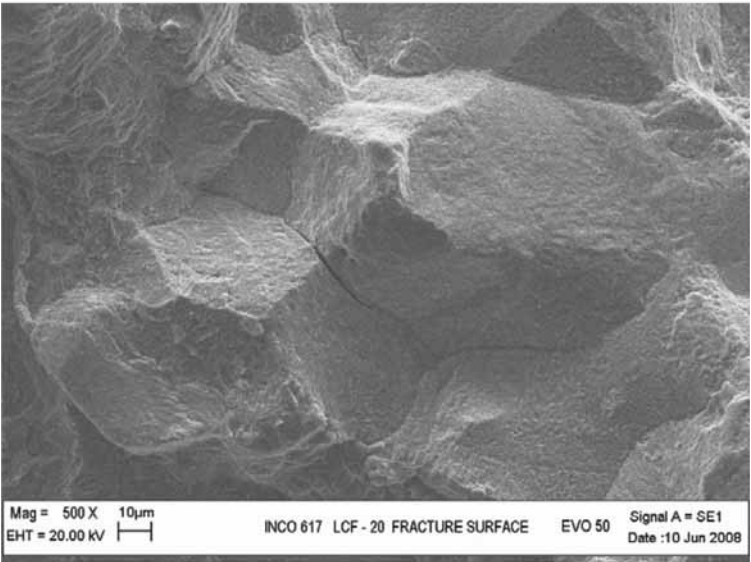


FIG. 8—Hysteresis loops for subsequent cycles; however, do not show significant serrated yielding for tensile and compressive or dwell cycles.

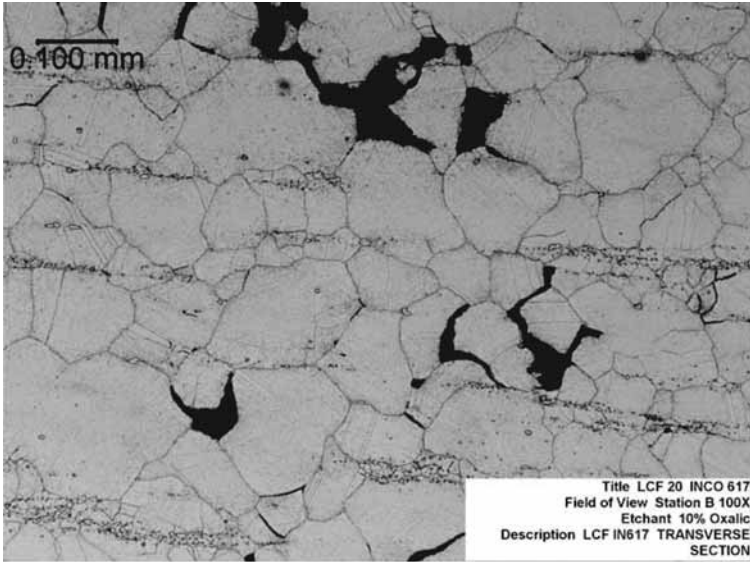


a.) 649°C

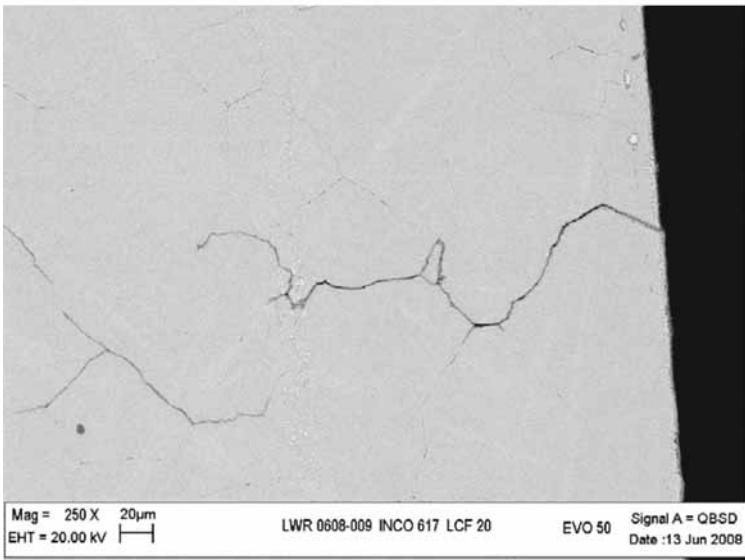


b.) 760°C

FIG. 9—Tensile dwell fracture surface.



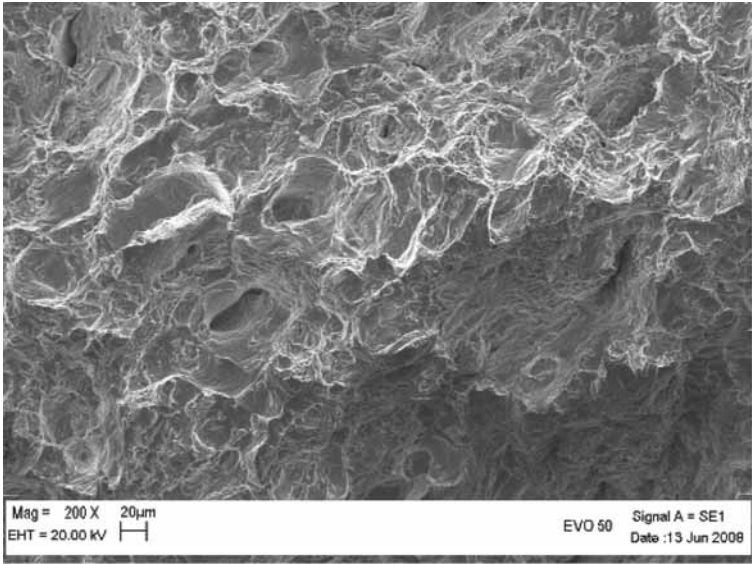
a.) grain boundary cavitation



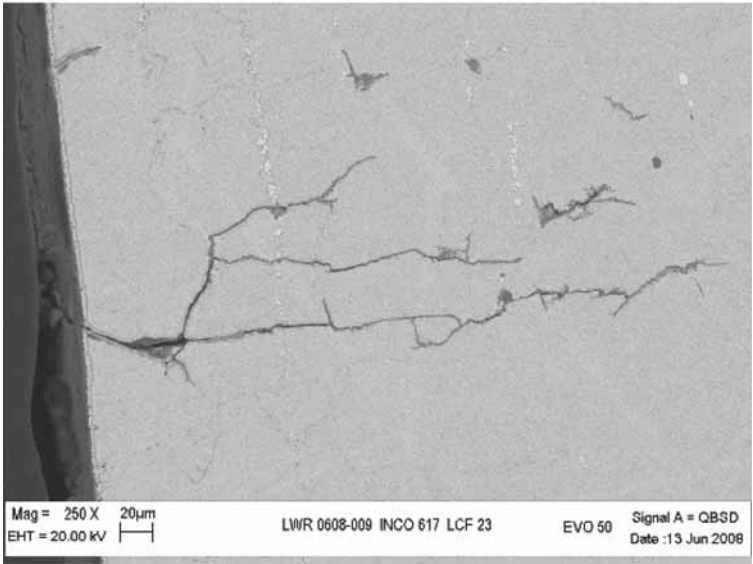
b.) inter-granular cracking

FIG. 10—Tensile dwell 760°C cross section.

analysis did not reveal any significant damage at grain boundaries. The retained strength of the boundaries did not allow cracks to propagate along boundaries as they did in the tensile dwell case. Cracks observed in compressive dwell specimens were transgranular in nature. The lack of accumulated creep damage



a.) Fracture Surface



b.) Cross section

FIG. 11—*Compression dwell 760°C post test inspections.*

likely leads to the longer lifetimes for compressive dwell LCF testing compared to tensile dwell.

In LCF testing, a specimen that develops more cracking at grain boundaries will likely fail before one that does not experience the same level of grain

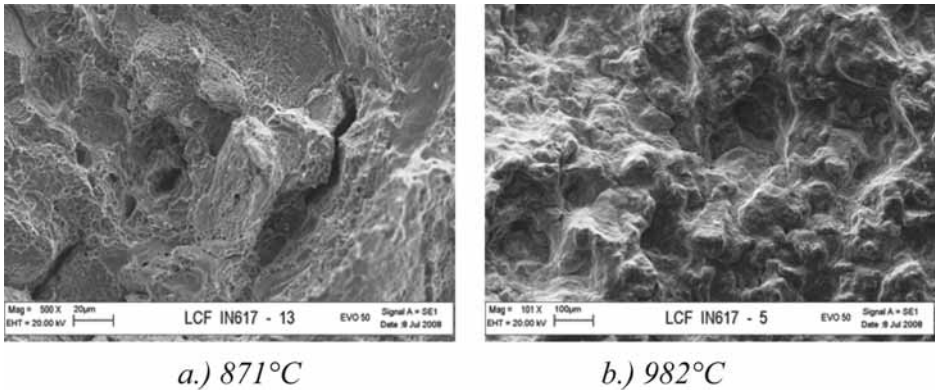


FIG. 12—Transgranular fracture in high temperature dwell fatigue tests.

boundary creep damage. It is a strong indication that lifetime reductions due to tensile dwells are associated with creep grain boundary damage that leads to intergranular failure. Void formation at grain boundaries and subsequent intergranular fracture is characteristic of the common creep failure mode in IN617 [7]. Grain boundary cracks in the interior of the specimen can link up with one another, or with surface nucleated cracks, to rapidly form a multiple cracks that lead to rapid failure. Even if a crack does not form at a boundary, cavitation can weaken the boundary to the point that any crack contacting the boundary can easily propagate. This provides a crack propagation path through the sample that can lead to drastically reduced cyclic lifetimes.

Failure Mechanism (High Temperatures)

Fractography was performed on tensile dwell LCF specimens at 871°C and 982°C, with the same total strain ranges. At the high temperature tensile dwell tests; however, failure occurred by transgranular or mixed mode fracture, see

Figure 12 Failure also appeared to be more ductile in nature at these higher temperatures. This indicates that creep damage to grain boundaries is not as dominant at 871°C and 982°C to the extent that it is at the moderate temperatures.

Conclusions

At moderate temperatures, 649°C and 760°C, the effect of dwell fatigue is quite pronounced when measured on strain range versus cycles. There appears to be a 60 % reduction in life due to a compressive dwell, and nearly 90 % reduction in life due to a tensile dwell. As measured by strain range, the fatigue strength reduction is 30 % on compressive strain and 60 % on tensile strain. Reduced cyclic lifetimes at moderate temperature tensile dwell are due to creep damage accumulation at grain boundaries that leads to intergranular cracking and

failure. Tensile dwell conditions at high stresses are more conducive to creep damage at grain boundaries.

At high temperatures, 871°C and 982°C, the effect of dwell fatigue is much less pronounced when measured on strain range versus cycles. There appears to be a 45 % reduction in life due to a compressive dwell, and 65 % reduction in life due to a tensile dwell. The test results at 982°C indicate there is no influence on the life from either the compressive or tensile dwell. At 871°C and 982°C, the stresses that develop are lower and the tensile dwell sensitivity of IN617 is significantly reduced (871°C) or almost entirely eliminated (982°C). Higher temperature cyclic plasticity is more dominant than creep. At the higher temperature tensile dwell tests failure occurred by transgranular or mixed mode fracture.

Decreased LCF lifetimes observed during tensile dwell testing of IN617 cannot be attributed to serrated yielding. Significant serrated yielding only occurs during the first cycles and seems to occur in similar levels in both tensile and compressive dwell testing.

The mean stress behavior does not follow the general rule that development of positive mean stresses reduces LCF lifetimes compared to negative mean stresses. The observed mean stress trend is counter to the expected effect on life at moderate temperatures.

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L. J. Carroll,¹ C. Cabet,² R. Madland,³ and R. N. Wright¹

Creep and Environmental Effects on the High Temperature Creep-Fatigue Behavior of Alloy 617

ABSTRACT: Alloy 617 is the leading candidate material for an intermediate heat exchanger (IHX) of the very high temperature reactor (VHTR), expected to have an outlet temperature as high as 950 °C. Acceptance of Alloy 617 in Section III of the ASME Code for nuclear construction requires a detailed understanding of the creep-fatigue behavior. Strain-controlled low cycle fatigue (LCF) tests including hold times up to 1800 s at maximum tensile strain were conducted at 950 °C and a total strain range of 0.3 % in air and in a simulated VHTR impure helium. For similar test conditions, the creep-fatigue behavior was globally similar in both environments. The fatigue resistance decreased when a hold time was added at peak tensile strain, consistent with the observed change in fracture mode from transgranular to intergranular. Further increases in the tensile hold time, beyond 180 s, were not detrimental to the creep-fatigue resistance. The influence of oxidation on the fatigue crack initiation and propagation is discussed. Bulk damage in the form of grain boundary cracking was present in the interior of the creep-fatigue specimens. This bulk cracking was quantified and found to be similar for hold times of up to 1800 s consistent with the saturation in failure lives.

KEYWORDS: creep-fatigue, Alloy 617, environment, helium, creep-fatigue interaction, nickel alloy

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¹ Idaho National Laboratory, 1955 Fremont, P.O. Box 1625, Idaho Falls, ID 83415-2218.

² CEA, DEN, DPC, SCCME, Laboratoire d'Etude de la Corrosion, Non Aqueuse 91191, Gif-sur-Yvette, France.

³ Colorado School of Mines, 1500 Illinois Street, Golden, CO 80401.

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Introduction

The very high temperature reactor (VHTR) is being developed as a nuclear system with a gas as the primary coolant for cogeneration of hydrogen and electricity. Conceptual design requires an outlet temperature of greater than 850 °C to efficiently generate hydrogen, with a maximum expected temperature of 950 °C [1]. A critical component in the VHTR system is the intermediate heat exchanger (IHX), which will operate at the reactor outlet temperature of up to 950 °C. The combination of very high temperature operation and long duration of service requires structural materials with good thermal stability as well as high temperature creep and oxidation resistance. Based on these material requirements, the nickel base alloy UNS N06617, Alloy 617, is the leading IHX candidate alloy [1]. Alloy 617 is strengthened by solid solution strengthening provided by the alloying elements chromium, cobalt, and molybdenum as well as by intra and intergranular carbide precipitates. The high temperature oxidation resistance is derived from the high chromium content.

Creep-fatigue deformation is expected to be an important damage mode for the IHX as a result of transients during start up and shut down producing cyclic loadings, while the stress relaxation during steady power operation induces creep damage. To address the needs identified by the Alloy 617 Draft ASME Code Case [2] and potential licensing concerns [3], a comprehensive program was launched to assess the creep-fatigue resistance of Alloy 617. To approximate the expected deformation in a laboratory setting, creep-fatigue testing introduces a hold time in a strain-controlled fatigue cycle. Previous work on Alloy 617 has suggested that a hold at peak tensile strain is most damaging [4] and therefore this type of loading was investigated.

In addition, licensing of a high temperature reactor will likely require an understanding of the environmental effects of VHTR-type impure helium on the creep-fatigue behavior [3]. A number of helium cooled reactors have been built and operated for extended periods. The helium coolant in the primary circuit has been found to contain low levels of impurities at steady state operation [5] that can corrode high temperature alloys used for internals and heat exchangers. Depending on the impurity nature and on the temperature, nickel alloys can undergo oxidation, carburization, or decarburization at high temperatures [6]. These two later processes were observed to be deleterious to the microstructure and strength [7,8]. The optimum coolant chemistry is slightly oxidizing to enable the formation of a continuous and dense chromia scale on the alloy surface without significantly impacting the lifetime of the graphite components. It is expected that the surface oxidation would protect the bulk microstructure and have little influence on the mechanical properties [6]. The results of previous studies suggest that adequately controlled helium fostering alloy oxidation may be slightly beneficial for the continuous cycle fatigue life of nickel alloys [9–11]. Less work is available for the creep-fatigue behavior of Alloy 617 in a helium environment [9,12], and little is known about the relationship between the environment, the deformation mechanisms, and the changing microstructure.

To enhance the understanding of creep-fatigue deformation mechanisms, the fatigue and creep-fatigue behavior of Alloy 617 was investigated at 950 °C

with systematic characterization of the evolving microstructure, crack initiation, and crack propagation. A previous paper [13] on the creep-fatigue life of Alloy 617 in air discussed the effect of oxidation on the crack initiation and propagation. The goal of the current work is to provide insight into the microstructure development during creep-fatigue deformation and to discuss its influence on the crack propagation. Experiments in air as well as in controlled VHTR-simulated helium are presented to account for any possible environmental effect.

Experimental Procedure

Cylindrical creep-fatigue specimens, 7.5 mm diameter in the reduced section and a gage length of 12 mm, were machined from an Alloy 617 annealed plate; the long axis of the specimen aligned with the rolling direction. The composition of the investigated heat of Alloy 617 is given in Table 1, and representative images of the solution annealed microstructure are shown in the reference [13].

Cyclic testing was conducted utilizing servo-hydraulic test machines in axial strain-control mode. Radio-frequency induction heating was used to heat the specimens. Temperature control was achieved using a combination of spot-welded thermocouples located near the shoulder of the specimen and a thermocouple loop contacting the center of the gage section. The temperature gradient was measured with a specimen instrumented with spot-welded thermocouples along the gage length and was found to deviate by less than 1 % within the gage section.

For the cyclic tests in controlled chemistry helium, a servo-hydraulic test frame equipped with a gas-tight chamber was used. A gas system was developed to feed VHTR simulated coolant into the chamber as is shown in Fig. 1. It is comprised of a manifold containing bottles of helium with controlled partial pressures of CO, CH₄, and H₂, mass flow controllers to regulate the helium flow, and pressure controllers. Small quantities of H₂O are added to control the partial pressure of water vapor in the helium flow (as low as 5 ppm). A gas chromatograph and two solid-state hygrometers record the gas chemistry at the inlet and outlet of the chamber. Feedback is employed at the inlet hygrometry.

Fully reversed strain-controlled low cycle fatigue and creep-fatigue testing was completed on Alloy 617 at a total strain range of 0.3 % in air and in helium at 950 °C. Selected creep-fatigue tests were also conducted in air at 950 °C and a 0.6 % total strain. A triangular waveform and a ramp rate of 10⁻³/s were used for low cycle fatigue testing. Creep-fatigue testing followed a trapezoidal waveform with a tensile hold imposed at the maximum tensile strain. Tensile holds of 180, 600, and 1800 s were investigated. The number of cycles to failure, N_f, was defined as a 20 % decrease in the ratio of the peak tensile stress to the peak

TABLE 1—Alloy 617 composition in wt. %.

Alloy	Ni	C	Cr	Co	Mo	Fe	Al	Ti	Si	Cu	Mn
617	Bal.	0.05	22.2	11.6	8.6	1.6	1.1	0.4	0.1	0.04	0.1

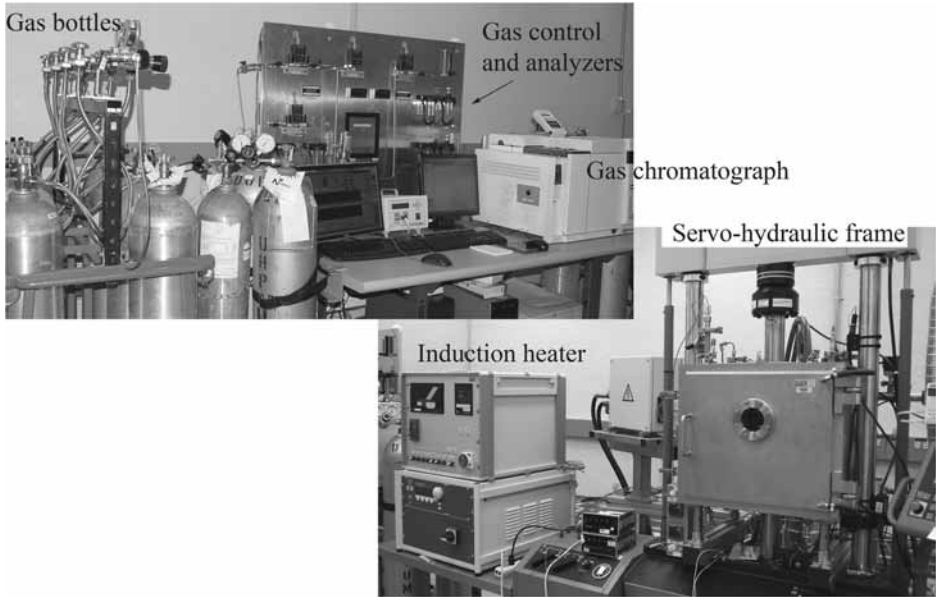


FIG. 1—Creep-fatigue bench with the gas system capable of controlling the chemistry of the helium.

compressive stress from the point at which it initially deviated from its level slope [14]. Test completion was prior to actual specimen separation and was based on a predetermined drop in load. Upon achieving the load drop, the test was switched to zero load, and power to the induction coil was shut off allowing the specimen to rapidly cool.

For testing in helium, a gas mixture was selected to roughly simulate the controlled VHTR coolant. The gas mixture contained 50 ppm CO and 350 ppm H₂ in addition to approximately 8 ppm H₂O at 1.1 bar total helium pressure and was oxidizing with respect to Alloy 617; i.e., a Cr-rich oxide scale formed on the surface of the static (no load) alloy [6]. The total gas flow rate was approximately 0.5 L/min. It was demonstrated that the gas system was capable of stable control to the target chemistry, including the partial pressure of the water vapor. Furthermore, the inlet gas chemistry was similar to the chemistry at the creep-fatigue chamber outlet. Figures 2(a) and 2(b) show the P(CO) and P(H₂O) analyzed at the chamber inlet and outlet during a creep-fatigue test of about 500 h. Static (no load) coupons were exposed at 950 °C to air and to controlled VHTR-simulated helium for comparison.

After creep-fatigue deformation, the gage sections of the deformed specimens were cut along the stress axis through the largest surface crack. The specimen was then coated with a thin layer of gold followed by nickel plating to ensure retention of the surface oxide. The coated specimens were mounted in phenolic, polished, and then etched with a 2 % bromine and 98 % methanol solution. Metallurgical evaluation was conducted using optical microscopy and scanning electron microscopy (SEM).

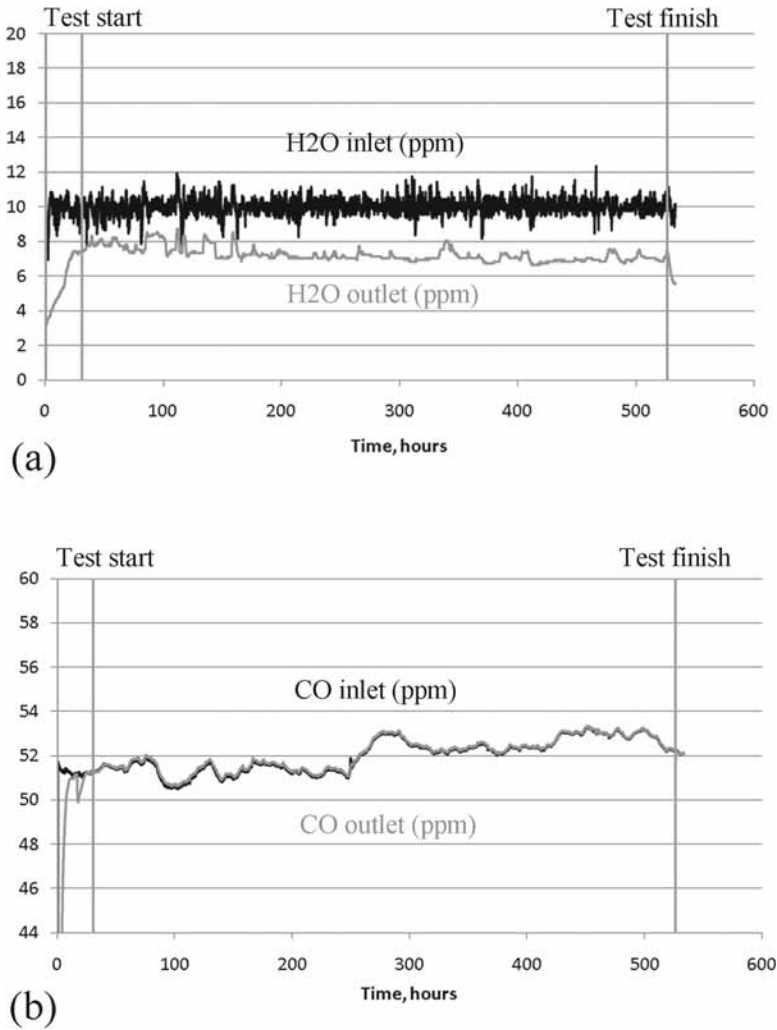


FIG. 2—Recorded level of CO (a) and H₂O (b) in part per million (ppm) at the chamber inlet and outlet for a creep-fatigue test in helium under 1.1 bar total pressure.

Results

Fatigue and Creep-Fatigue

Continuous cycle fatigue and creep-fatigue testing was completed in air and in helium at a 0.3 % total strain range at 950°C, and failure lifetimes were found to be similar for the two environments, as shown in Fig. 3(a). Table 2 gives the cycles to failure and time to failure for fatigue and creep-fatigue specimens tested in air and in helium. The continuous cycle fatigue lives were significantly longer than the corresponding creep-fatigue lifetimes.

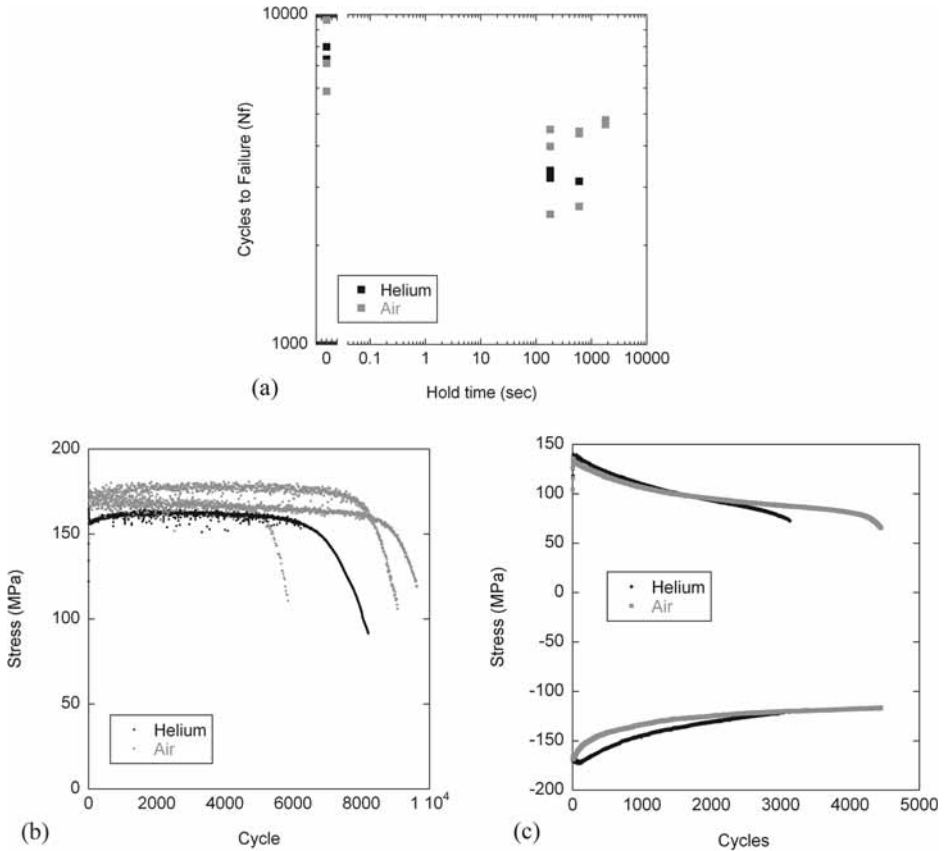


FIG. 3—Fatigue and creep-fatigue in air and in impure helium at 950°C and a total strain range of 0.3 %. Cycles to failure as a function of hold time (a), peak tensile stresses as a function of cycle for continuous cycling (b), and peak tensile and compressive stresses for creep-fatigue with a 600 s hold (c).

However, the increasing duration of the tensile hold did not further decrease the creep-fatigue life, i.e., the 180 s hold time cycle life was comparable to the 1800 s hold time cycle life. For continuous cycle fatigue, for which the peak tensile stresses are shown in Fig. 3(b), the peak stresses were relatively symmetric, and a steady state stress was achieved prior to macro-crack initiation. For creep-fatigue, the magnitudes of the peak stresses in compression were slightly greater than in tension and cyclic softening was observed, as shown in Fig. 3(c). The peak stresses as a function of cycle were also similar for both environments. Typical creep-fatigue hysteresis loops and stress relaxation curves are shown in Fig. 4. The stresses relaxed rapidly during the hold and were almost completely relaxed in 200 s.

TABLE 2—Continuous cycle fatigue and creep-fatigue cycle life and time to failure for Alloy 617 tested at 0.3 % total strain and 950 °C in air and in helium (containing 50 ppm CO, 350 ppm H₂ and about 8 ppm H₂O).

Hold time (s)	Environment	N _f (cycles)	Time (h)
0	Air	9641	16
0	Air	7133	12
0	Air	5867	10
0	Helium	8000	13
0	Helium	7333	12
180	Air	3989	210
180	Air	2485	130
180	Air	4486	230
180	Helium	3004	155
180	Helium	3373	174
600	Air	4096	690
600	Air	2623	440
600	Air	4361	735
600	Helium	3130	530
1800	Air	4805	2410
1800	Air	4650	2330

Cracking Mode and Surface Microstructure Evolution

Metallurgical analysis of the deformed specimens parallel to the stress axis revealed the surface microstructure and the primary and secondary cracking. The microstructures and cracking in the specimens cycled in air were observed

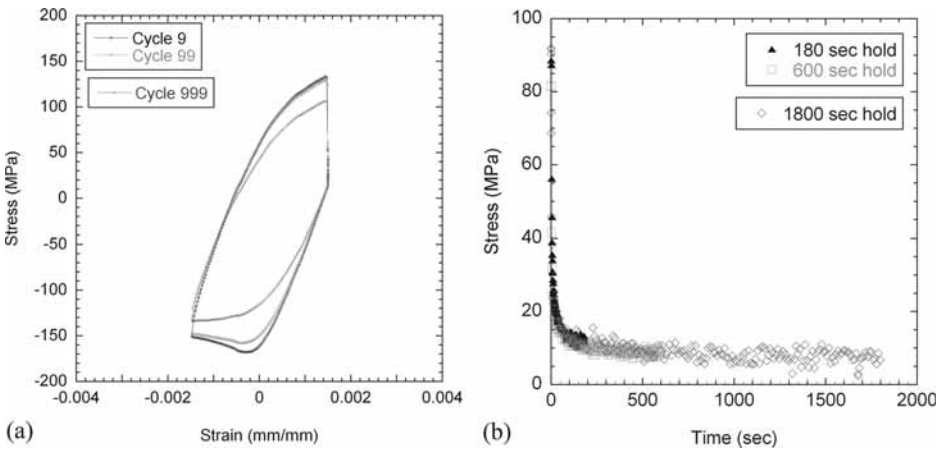


FIG. 4—Hysteresis loops from cycle 9, 99, and 999 from a creep-fatigue specimen tested at 950 °C, 0.3 % total strain, and a 600 s hold in air (a). Stress relaxation behavior during the tensile hold for creep-fatigue specimens tested at 180, 600, and 1800 s in air (b).

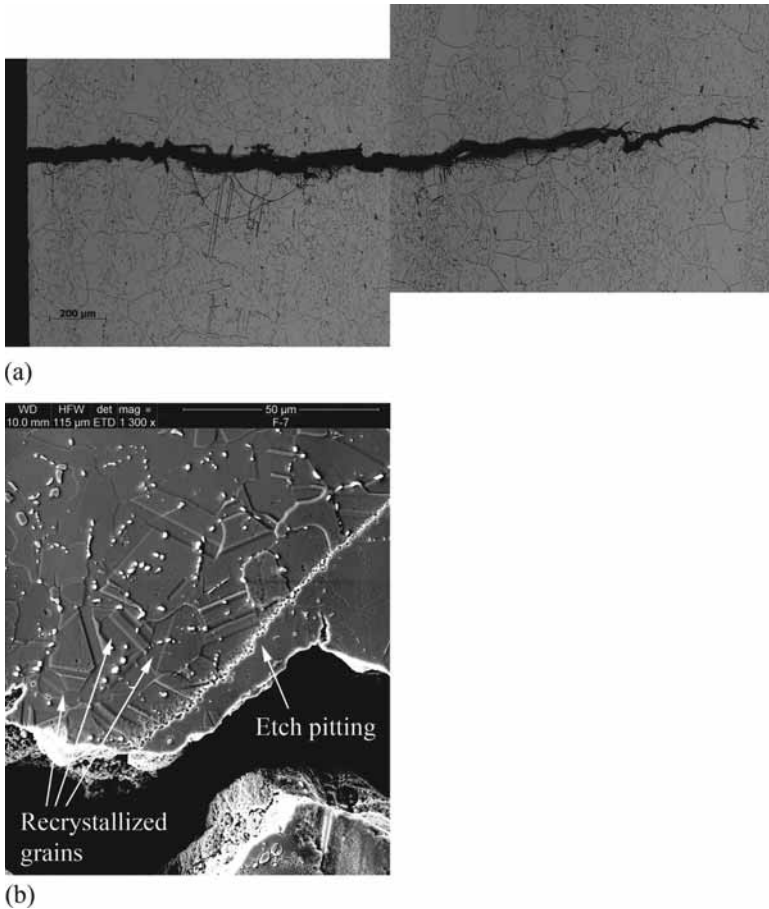


FIG. 5—Cracking in a continuous cycle fatigue specimen deformed at 950°C and a 0.3 % total strain in air (a) and impure helium (b). The stress axis is vertical and in the plane of the page.

to be qualitatively similar to those cycled in impure helium. Therefore, the results reported in this section do not differentiate between testing environments.

In the continuous cycle fatigue specimens a single primary crack is initiated and propagated in a transgranular manner perpendicular to the stress axis, as shown in Fig. 5(a). Oxide protrusions were observed along the length of the crack, and a few of the surface connected grain boundaries perpendicular to the stress axis were also oxidized. Secondary intergranular cracks were typically not observed. Interestingly, recrystallization was observed at several locations along the fatigue crack. These new grains were smaller, approximately 10 μm in diameter, and free of carbides, as shown in Fig. 5(b).

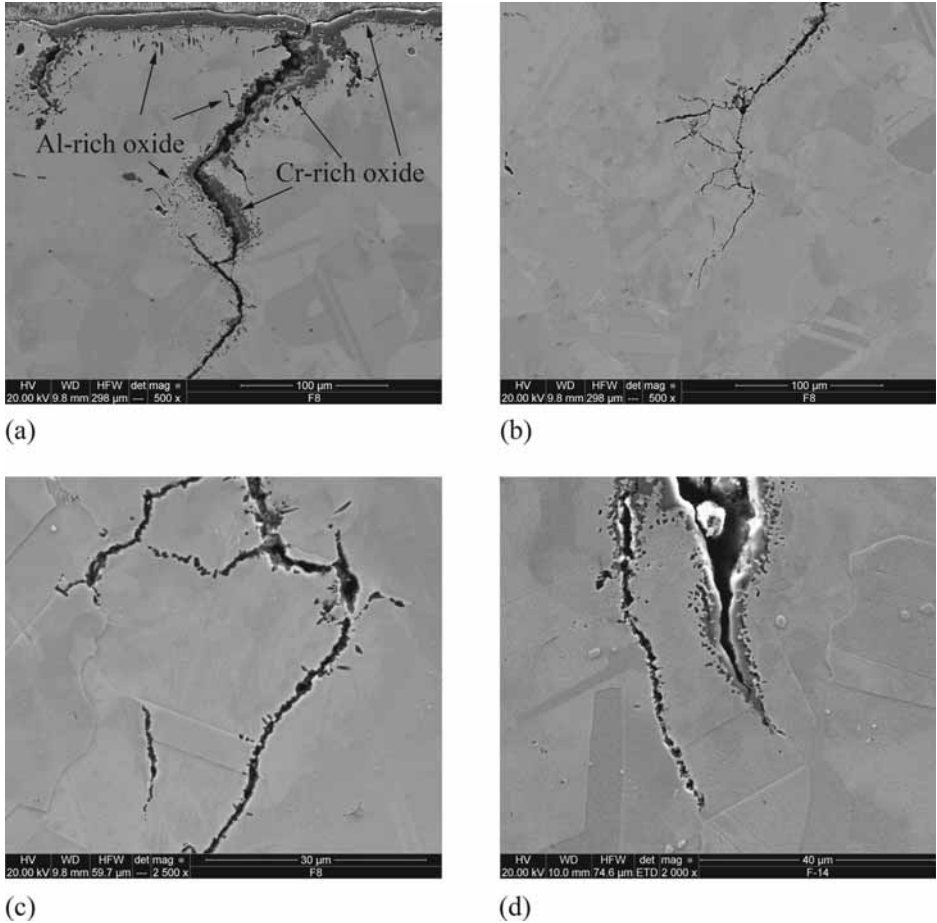


FIG. 6—SEM images of creep-fatigue specimens deformed at 950°C and a 0.3 % total strain with a 600 s hold in impure helium (a)–(c) and a 180 s hold in air (d). The stress axis is horizontal and in the plane of the page.

The addition of a hold time resulted in multiple larger or primary cracks originating from the specimen surface as well as in an evolving microstructure, particularly at the specimen surface. In contrast to continuous cycle fatigue, secondary intergranular cracking was observed at some of the grain boundaries perpendicular to the stress axis, as shown in Fig. 6(a). It is difficult to determine whether the primary crack(s) initiation occurred intergranularly or transgranularly because of the evolving microstructure at the surface and surrounding the cracks. However, based on the many shorter secondary cracks which started at oxidized grain boundaries, it is likely that the primary cracking also initiated intergranularly. Propagation was in an intergranular manner. An oxide, rich in Cr, formed along the crack flanks and Al-rich oxides formed at grain boundaries

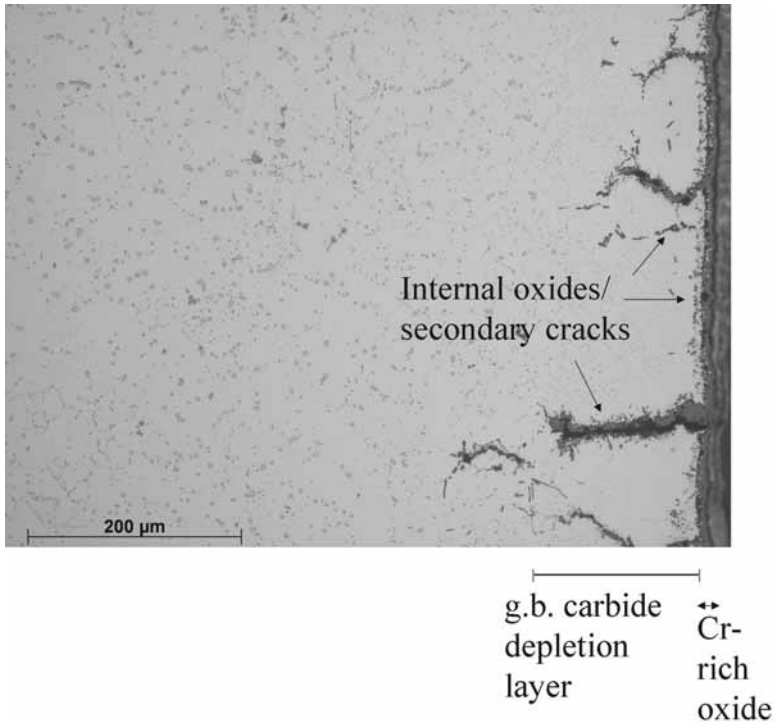


FIG. 7—Optical image of the surface of a creep-fatigue specimen deformed in air at 950 °C and a 0.3 % total strain with a 600 s hold. The stress axis is horizontal and in the plane of the page.

close to the crack and ahead of the crack tip. Figures 6(b)6(d) illustrate oxidized grain boundaries at the crack tip. The region surrounding the primary or larger creep-fatigue cracks was also depleted of grain boundary carbides. An 1800 s hold creep-fatigue test interrupted at approximately 15 %–25 % of life exhibited substantial secondary cracking of oxidized near surface grain boundaries. The depth of these secondary cracks was less than a grain diameter.

The surface microstructure of the deformed specimens was consistent with the statically exposed specimens with the exception of the presence of primary and secondary cracking. The surface of a creep-fatigue specimen tested at a 0.3 % total strain and a 600 s hold is shown in Fig. 7. The specimen was at temperature for approximately 700 h at 950 °C. The surface of the oxidized material consisted of:

- An external Cr-rich oxide scale. The scale shows a fairly uniform thickness, about 6 μm, but is wavy. Variation observed in the scale thickness for different specimens correlates well with the amount of time exposed at 950 °C.
- Internal Al-rich oxides, formed beneath the external scale and at grain boundaries close to the surface. In the statically exposed specimens, the

internal oxides were thin fingerlike, intergranular, and predominately comprised of aluminum oxide; their depth increased with the amount of time spent at 950 °C. In the case of creep-fatigue specimens tested for a similar duration at 950 °C, the depth of internal oxides was greater and grain boundaries close to the surface have initiated larger veins filled with Cr-rich oxide, as shown in Fig. 7.

- Grain boundary secondary surface cracks (see above) to a depth of 250 μm , as shown in Fig. 7.

As commonly observed for oxidized Alloy 617 [15], a region depleted of grain boundary carbides has formed close to the surface. The depth of this region is related to the formation of the surface chromia scale which consumes chromium and destabilizes the Cr-rich carbides. The depletion depth of the creep-fatigue specimens was approximately 50 μm for the shorter hold times and as much as 200 μm for the longest hold times. After long times, such as in the 600 and 1800 s creep-fatigue tests, this grain boundary carbide depleted surface region consisted of a fine distribution of re-precipitated intragranular carbides and/or nitrides.

Post-Cyclical Deformation Microstructure

Numerous cracks were observed in the interior or bulk of the creep-fatigue tested specimens, examples of which are shown in Fig. 8. The bulk is defined as the material that is a sufficient distance from the specimen surface to not be influenced by the surface microstructure evolution or surface cracking. The bulk cracks were relatively thin and typically followed grain boundaries aligned perpendicular to the stress axis. In some cases, intergranular carbides were observed along the interior cracks. The width of the cracking varied as did the length of the cracks, from several micrometers to several grain diameters in length. In some cases, there were boundaries with multiple small cracks or cavities. Some regions contained a significant amount of this type of damage, whereas most of the grain boundaries were completely free of cracking. Grain boundaries not associated with the bulk cracking did not exhibit grain boundary cavitation and instead microscopically appeared free of grain boundary damage (at magnifications of up to 100 000 times). The bulk cracks were absent of any oxide and thus it can be concluded that they were not exposed to the environment. It should be mentioned that there was no grain boundary cracking observed in the pure fatigue specimens (or in the static oxidized coupons) indicating that the cracking was a result of the hold time. Furthermore, fine bulk cracking was observed in an 1800 s hold creep-fatigue specimen interrupted at 15 %–25 % of life; the fine cracks were less than 10 μm in length and were significantly smaller than those observed after failure.

The amount of bulk material cracking in the failed specimens was quantified to determine if a relationship existed between the duration of the hold time and the amount of grain boundary damage. The total number of cracks, total length of cracks, and the average length of the cracks were determined for a 0.2 mm² area at five locations, as shown schematically in Fig. 9(a). The areas analyzed were 1 mm in each direction (in the plane of the page) from the primary crack tip and 3 mm in each direction along the stress axis (in the plane of the

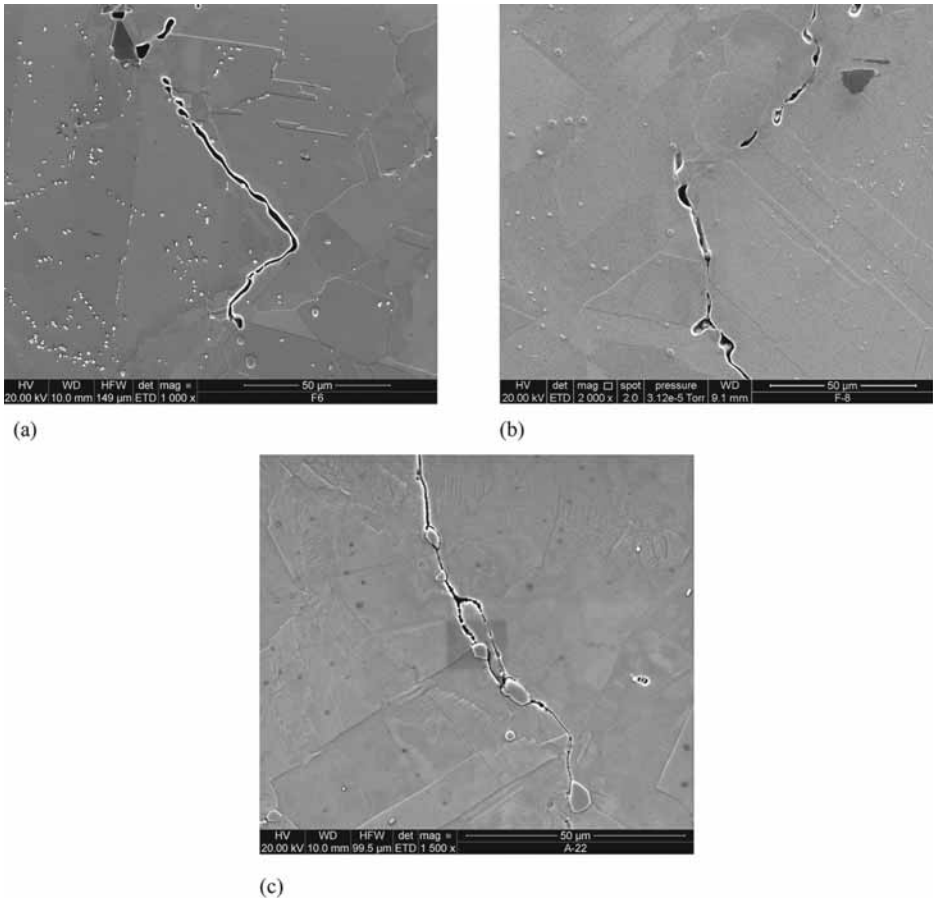


FIG. 8—SEM images of grain boundary cracking in the interior of specimens deformed at 950°C and 0.3% total strain range with a 180 s hold in impure helium (a), a 600 s hold in impure helium (b), and a 600 s hold in air (c). The stress axis is horizontal and in the plane of the page.

page) from the primary crack tip. For all specimens, the cracking was statistically similar among the five areas and no general trends were observed. In particular, the amount or length of the bulk cracking did not appear to be related to location relative to the primary crack. Hence, the cracking features (number, average length, and cumulated length of the cracks) were calculated for the all of the areas combined. Figures 9(b)–9(d) show the results graphically as a function of hold time. As mentioned previously, there was no cracking observed in the continuous cycle fatigue specimens, whereas increasing hold times resulted in relatively similar amounts of grain boundary cracking. Additionally, regardless of the environment, air or impure helium, the material exhibited a similar degree of bulk grain boundary cracking.

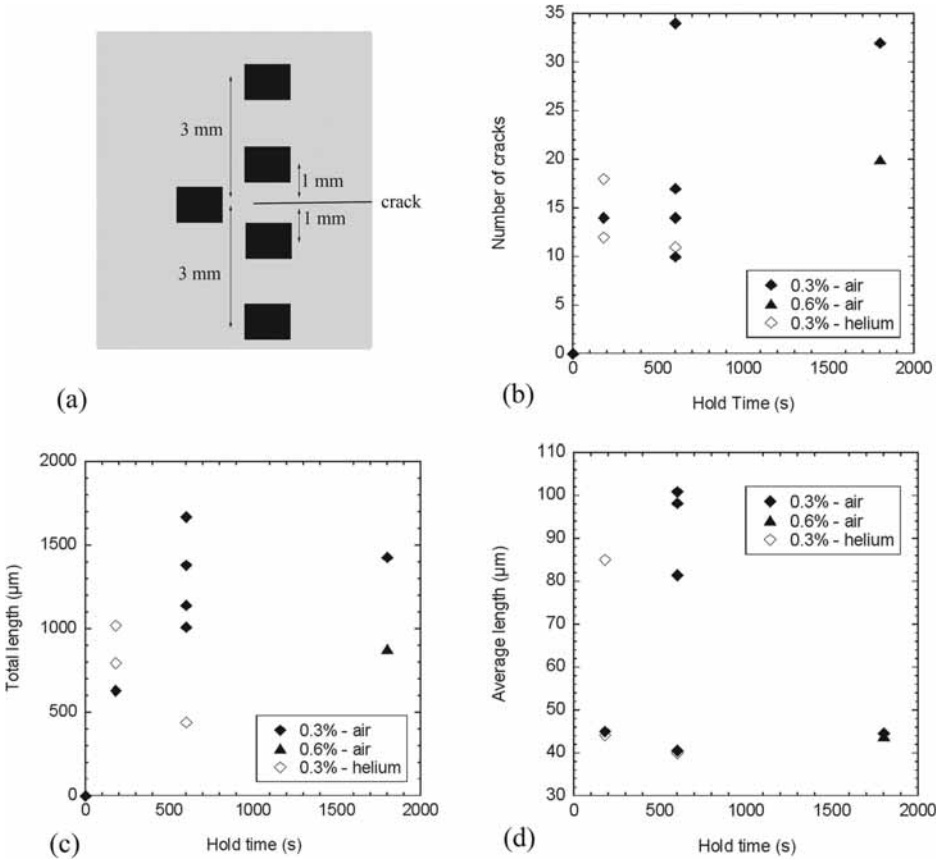


FIG. 9—Schematic of the method used to quantify interior grain boundary damage in the deformed fatigue and creep-fatigue specimens (a). The stress axis is vertical and in the plane of the page. Results of the grain boundary cracking analysis for the total number (b), total length (c), and average length (d). The specimens represented by open diamonds were tested in impure helium, the filled diamonds were tested in air, and the triangles were tested at a 0.6% total strain in air.

Discussion

One faces a challenge in discussing the role of environment at elevated temperature. In an ideal case, comparable creep-fatigue testing performed in an inert atmosphere would be used as a baseline, exemplifying the effect of oxidation on fatigue life. However, creep-fatigue in inert conditions is not practically achievable at high temperature. Exposure at 950°C in vacuum and in (never completely) pure helium or argon is accompanied by carbon loss due to reaction with the residual oxidizing impurities (oxygen, water vapor...) [6]. The continuous carbon removal from the alloy surface induces instability of strengthening carbides to an increasing depth [8]. The dissolution of the carbides that pin the

grain boundaries enables grain boundary sliding and the formation of wedge-cracks in the carbide-free regions [4,16]. These metallurgical instabilities may contribute to the fatigue crack initiation and/or growth, and the fatigue life may be significantly lower than observed in air [12]. In the absence of a comparable test under a strictly inert atmosphere, the most relevant data are collected in air or in any other environment that promotes formation of a protective oxide scale such that the alloy microstructure is preserved.

The failure lives of Alloy 617 specimens cycled in air and in VHTR simulated helium were similar at 950 °C, as shown in Fig. 3(a). In addition, the tensile/compressive peak stresses versus cycle, the hysteresis loops, and the stress relaxation during the peak tensile hold were also similar, as shown in Figs 3(b), 3(c), 4(a), and 4(b). No differences in the crack initiation behavior, propagation mode, or oxidation of the cracks were evidenced between the air and impure helium cycled specimens. Finally, metallurgical characterization of the specimens cyclically deformed in air and in helium failed to reveal any significant differences in the oxidation products or surface morphology. The observed evolution of the surface morphology is consistent with general understanding of oxidation of Alloy 617 in air and in a controlled simulated VHTR cooling helium. In an oxidizing environment (as is air or impure helium with significant water vapor and carbon monoxide partial pressures at 950 °C), Alloy 617 develops a chromia surface scale and aluminum oxidizes internally [6]. These similarities in the overall behavior of the specimens cycled in air and in simulated impure VHTR helium allow for the joint discussion of the results for both environments without specific reference to the testing atmosphere.

The continuously cycled fatigue lifetimes were significantly longer than the creep-fatigue lifetimes, as shown in Table 2. The introduction of a 180 s hold at peak tensile strain decreased the life by approximately a factor of 2. This decrease in cycle life is likely associated with the change in the cracking mode. In the continuously cycled specimens, the primary crack initiated and propagated in a transgranular manner, as shown in Fig. 5. When a hold time is added at peak tensile strain, the fracture mode changed from transgranular to intergranular. For each of the hold times, often several longer cracks propagated intergranularly, as shown in Fig. 6. It is believed that the development of the intergranular fracture results in higher crack-growth rates [17]. The surface of reference static coupons (no load) and continuously cycled specimens showed intergranular-aluminum oxides. In the creep-fatigue specimens, many of the surface grain boundaries were cracked and filled with a Cr-rich oxide, as shown in Fig. 7. Accordingly, it was proposed [13] that surface oxidized grain boundaries perpendicular to the applied stress may act as crack initiation sites. Thus crack initiation was likely in an intergranular manner as well. An interrupted 1800 s hold time creep-fatigue test at a 15 %–25 % of life exhibited a large number of near surface grain boundary oxides and secondary cracks indicating that initiation occurred relatively early in life and the majority of cycles were spent propagating a crack. The early crack initiation in Alloy 617 and the faster crack propagation rates typically observed for intergranular cracking at least partially explain the creep-fatigue decreased cycle life.

Both the creep portion of the cycle (stress relaxation) and the environment may influence the driving force for crack propagation. Separating the influence of these two time-dependent processes is challenging. In the following, creep deformation is discussed first and then the role of environment.

Stainless steels exhibit creep damage in the form of voids at grain boundary precipitates [18]. These voids accelerate the propagation of intergranular cracks thus leading to the well known creep-fatigue interaction [4,19]. The Alloy 617 creep-fatigue specimens evaluated in this study exhibited grain boundary cracking. Although a couple of the specimens had isolated regions of cracking with evidence of multiple voids, as shown in Fig. 6(b), the strong majority of the grain boundaries were absent of any voiding (not visible at magnifications of up to 100 000 times). Therefore, the mechanism by which these grain boundary cracks form is not clear. The fact that interior grain boundary cracks are not present in the continuously cycled specimens indicates that they are a result of the peak tensile hold. The interrupted creep-fatigue specimen observed after 15 %–25 % of life (1800 s hold) exhibited some fine interior grain boundary cracks suggesting that these cracks also form early in the fatigue life. Further work involving the metallurgical analysis of interrupted creep-fatigue tested specimens as well as transmission electron microscopy analysis (TEM) of the dislocation substructure in grains adjacent to the grain boundary cracking is necessary to investigate the development of the intergranular cracks in the specimen interior.

Either way, grain boundary cracking, absent from continuously cycled specimens, is typically associated with creep deformation occurring during the tensile hold and likely interacts with the fatigue crack resulting in accelerated crack growth. As shown in Fig. 9, the increasing duration of the peak tensile hold time in creep-fatigue did not further increase the amount of grain boundary cracking in the interior of the specimens. The lack of further evidence of creep damage as quantified by the bulk cracking is consistent with the rapid stress relaxation during the tensile hold, during which the stresses were almost completely relaxed within 200 s, as shown in Fig. 4(b). The driving force for crack propagation during the tensile hold would have been similar in all of the hold time conditions in line with the absence of further degradation in cycle life with increasing hold times.

A previous paper described the interaction between oxidation and cracking during creep-fatigue of Alloy 617 in air at 950 °C [13]. When a fatigue crack propagates, unoxidized alloy is exposed to the oxidizing atmosphere (air or impure helium), and oxidation occurs along the crack flanks and at the crack tip. As shown in Fig. 6, aluminum-rich oxides were observed in the vicinity of the main crack: at grain boundaries along the oxidized crack and ahead of the crack tip. Oxygen penetration at the crack tip and precipitation of brittle alumina can cause high stresses to buildup [20]. The localized stress field would cause rupture at grain boundary intersection, especially in the global direction of the crack growth, and accelerate crack propagation.

A large percentage of creep-fatigue life of Alloy 617 at 950 °C is likely spent propagating a crack. Creep deformation occurring during the tensile hold and oxidation likely influence crack propagation. Grain boundary cracking and

oxidation may play a role in determining the crack propagation mode, tending the creep-fatigue crack propagation to intergranular as opposed to transgranular as observed in the fatigue case. The bulk grain boundary cracking, quantifying the creep damage, may have accelerated the crack propagation. Interestingly, longer peak tensile hold times did not result a further degree of grain boundary cracking in the alloy interior consistent with the rapid stress relaxation and saturation in the number of cycles to failure with increasing hold times.

Conclusion

To support Alloy 617 acceptance in the nuclear section (Section III) of the ASME code for application at high temperature as well as licensing of VHTR systems with an IHX, creep-fatigue testing of Alloy 617 was performed in air and in a VHTR simulated impure helium at 950 °C and a total strain of 0.3 %. For specimens tested at similar conditions, fatigue and creep-fatigue behavior was globally similar in air and in helium, and tests carried out in both environments were discussed together. The crack propagation shifted from transgranular in pure fatigue to intergranular in creep-fatigue. In addition, the introduction of a hold time at the peak tensile strain decreased the number of cycles to failure. However, increasing the duration of the hold time from 180 s to as long as 1800 s did not further reduce the cycle life. Both the environment and creep deformation occurring during the tensile hold may contribute to the premature failure observed in the creep-fatigue specimens relative to the continuously cycled ones. Oxidation may have played a role in the fatigue crack initiation (surface aluminum-oxides at grain boundaries acting as crack initiation sites) and potentially in the crack propagation by precipitation of alumina ahead of the crack tip. Cracking in the interior of the specimen, most likely due to creep deformation, was observed in the cycled creep-fatigue specimens and likely accelerated the fatigue crack propagation. The interior intergranular cracking was quantitatively similar for all of the creep-fatigue hold times consistent with the cycles to failure saturating with increasing hold duration.

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Mauro Filippini¹

Creep-Fatigue at High Temperature of Notched Single Crystal Superalloys

ABSTRACT: The effect of hold time on the high temperature notched fatigue strength of the CMSX-4 single crystal superalloy has been experimentally investigated. Notched specimens with orientation $\langle 001 \rangle$ have been tested at 950 °C in load control in the low-cycle fatigue (LCF) life regime both without and with dwell. For the same applied stress range (net-section stress), the effect of hold time results in an appreciable reduction of fatigue life. Several scanning electron microscope (SEM) observations of the failed specimens have been made in order to associate the actual failure (crack initiation) location to the correct crystallographic orientation and to the critical location in the notch as given by suitable multiaxial fatigue parameters. Moreover, SEM observations reveal the change of the failure mode from pure fatigue to fatigue/creep conditions. In order to model the material behavior in the notch, stress-strain analysis of the notched specimens was performed by an elastic-plastic finite-element method (FEM) procedure; the parameters describing the material behavior of the single crystal superalloy have been identified by comparison with experimentally determined stress-strain fatigue behavior and creep data of smooth specimens. A suitable model for the assessment of the reduction of fatigue life in creep-fatigue conditions is proposed.

KEYWORDS: creep-fatigue, single crystals, superalloys, notch behavior, high temperature testing, elasto-visco-plastic model

Introduction

One of the basic tools that may be employed to increase the energy efficiency and to reduce the consumption of fossil fuels, thus reducing the output of CO₂ in the already endangered atmosphere, consists of an increase of gas

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¹ Politecnico di Milano, Dipartimento di Meccanica, Via La Masa 1, 20156 Milano, Italy, e-mail: mauro.filippini@polimi.it

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temperatures both in power plants and in the turbine engines for civil airplanes [1,2]. However, the need of increasing the service temperatures cannot be done at the cost of accelerated damage and reduced durability of critical components, and eventually affect the engine service life.

Turbine inlet temperatures have now reached 1650 °C (3000 °F) for the latest large commercial turbofan engines, resulting in higher fuel efficiency and thrust levels. High reliability and durability must be achieved by designing these turbine engines to meet operating economic targets and Extended-range Twin-engine Operational Performance Standard certification requirements. This level of performance has been obtained by a combination of advances in film cooling for turbine blades and vanes, novel design technology for reducing stresses and increasing aerodynamic efficiency, and the development and application of nickel-base single crystal superalloys. At high gas temperatures, several issues are critical to turbine engine performance, blade life, and structural integrity.

Gas turbine rotor blades are subjected to very high levels of stress and temperature during each engine operating cycle. Rotor blade fatigue failures account for up to 40% of the total number of issues that arise during an engine development program and a similar proportion of in-service problems. Thus, the gas turbine industry is being forced to minimize the resources required to bring a new design to market due to global competitive demands and increasing customer expectations. Accurate design and prediction tools are key to success in this process.

In gas turbine engines for aerospace propulsion, the assessment of the fatigue damage mechanisms due to the interaction of low cycle fatigue (due to ground-air-ground major engine cycling) and creep damage (due to cruise) are of paramount importance for the assessment of the structural integrity of the critical components [1,3,4]. For this purpose, within the frame of the experimental activities of the European Research Project named PREMECCY (Predictive Methods for the Combined Cycle Fatigue in Gas Turbine Blades) aimed at developing new and improved fatigue life prediction methods for the use in the design process, the effect of hold time on the high temperature notched fatigue strength of the CMSX-4 single crystal superalloy has been experimentally investigated. Notched specimens with theoretical stress concentration factor $K_t = 2.15$ and with crystal orientation $\langle 001 \rangle$ have been tested at 950 °C in load control in the low-cycle fatigue (LCF) life regime both without and with dwell.

The aim of this paper is to show how to associate the actual failure (crack initiation) location to the correct crystallographic orientation and to the critical location in the notch by employing suitable multiaxial fatigue damage parameters. By integrating the modeling of the stress-strain behavior of the material in the notch, such parameters should be able to correlate the fatigue lives of notched specimens with those of smooth specimens, and at the same time, predicting the likely position of failure initiation in agreement with the analysis of the scanning electron microscope (SEM) observations of the failed specimens. As SEM observations reveal the change of the failure mode from pure fatigue to fatigue/creep conditions, in order to model the material behavior in the notch, stress-strain analysis of the notched specimens was performed by a non-linear finite-element method (FEM) procedure. The parameters describing the material behavior of the single crystal superalloy CMSX-4 at 950 °C have been

identified by comparison with experimentally determined stress-strain fatigue behavior and creep data of smooth specimens. The aim of this work is to assess the validity of simplified damage models for the prediction of fatigue life in the creep-fatigue regime suitable for the purpose of the design of turbine engine components.

Material and Methods

Material

Single crystal Ni-base superalloys are commonly employed for use as high temperature creep and oxidation resistant blade alloys in the high pressure stages of modern gas turbine aero-engines [4]. The superior high temperature creep properties of the CMSX-4 superalloy are due to its microstructure, consisting of 70% volume fraction of coherent, cuboidal γ' -precipitates in a matrix of solid solution strengthened γ -phase [5–8]. The chemical composition of the CMSX-4 material is given in Table 1. Material for the production of the test specimens has been produced in the form of bars by investment casting under vacuum, followed by full heat treatment. The typical microstructure of the CMSX-4 superalloy is shown in Fig. 1. The axes of all rods were shown to lie within 12° of $\langle 001 \rangle$.

Specimens

The notched specimens for the creep-fatigue tests have been designed with a (nominal) stress concentration factor $K_t = 2.15$ (i.e., evaluated as in the case of isotropic material). The shape and dimensions of the notched specimens are shown in Fig. 2. Prior to machining bars have been inspected by Non-Destructive Testing (NDT) (x-ray, visual, and penetrant inspection); specimens have been produced by conventional machining and their exact dimensions recorded.

High Temperature Fatigue Testing

Fatigue tests with notched specimens have been conducted at the Laboratories of the Dipartimento di Meccanica, Politecnico di Milano, by employing a servo-hydraulic MTS 810 ± 100 kN load capacity test machine, equipped with a Hüttinger type induction heating furnace. Prior to fatigue tests, a thermal set-up activity has been conducted with dummy specimens: five type K thermocouples, as according to ASTM E230-03 [9], have been welded by discharge welding at the root of the notch and on the parallel sections of the specimen at a distance of ± 2.5 and ± 10 mm from the center of the specimen, following the recommendations described in Ref. [10]. In this way, the thermal gradient along the

TABLE 1—Chemical composition of the CMSX-4 superalloy (nominal values, weight Percentage).

Element	Co	Ta	Cr	W	Al	Re	Ti	Mo	Hf	Ni
wt. %	9.6	6.5	6.5	6.4	5.6	3.0	1.0	0.6	0.1	bal.

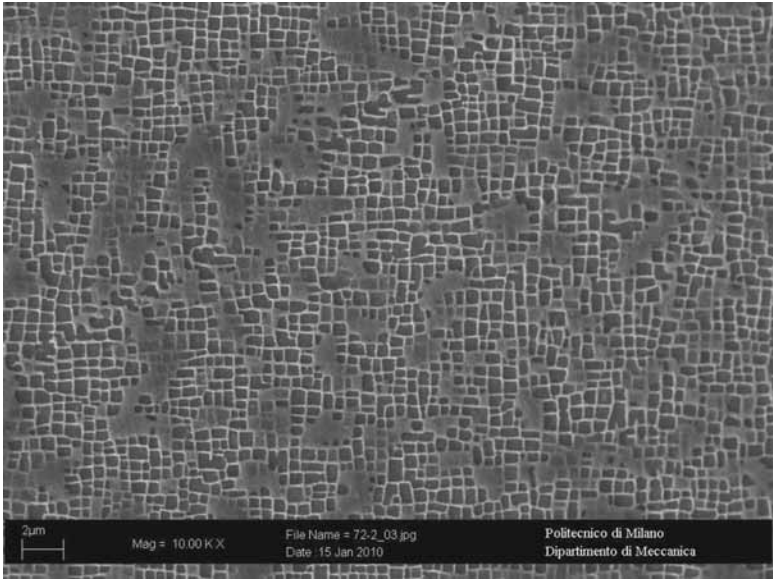


FIG. 1—SEM image of the microstructure of the CMSX-4 alloy.

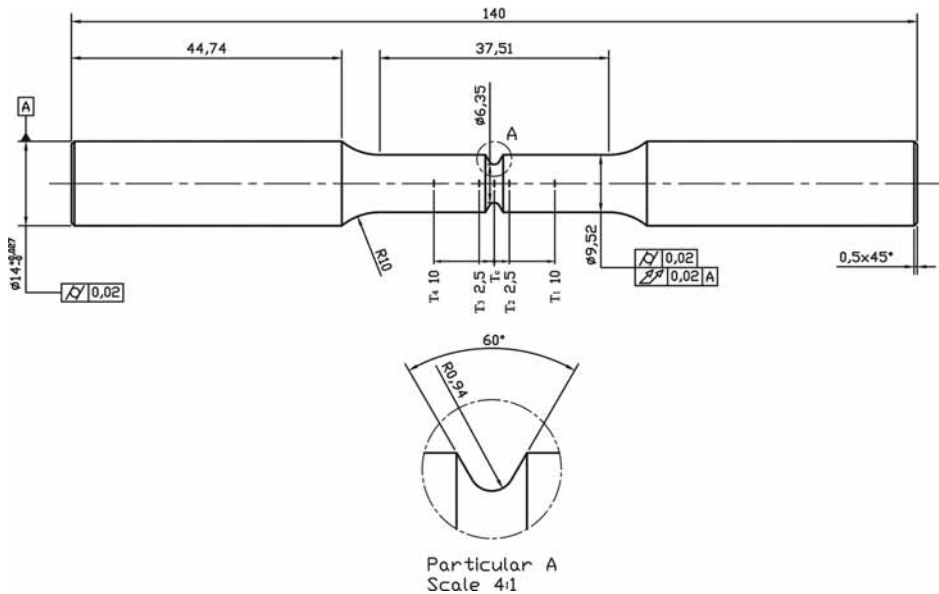


FIG. 2—Shape and dimensions of the notched specimens employed for the LCF and the creep-fatigue tests (dimensions in mm).

specimen could be recorded and a correlation between the temperature measured outside the notch and that achieved at the notch root could be established. In particular, all the tested specimens have been instrumented with two type K thermocouples positioned at ± 2.5 mm across the notch: the first one feeding the control channel of the induction heating furnace and the second one has been continuously measured with an external signal meter in order to cross-check the symmetry of the temperature distribution across the notch and giving the possibility to adjust the induction heater coil prior to testing if needed. Before each test, while keeping the load at zero, specimens have been gradually heated to the specified temperature, and they have been maintained at that temperature for at least 30 min before loading.

Pure LCF fatigue tests have been carried out at the temperature of 950°C in load control with a triangular wave shape with a frequency of 0.5 s^{-1} , i.e., with 1 s loading followed by 1 s unloading, with a loading ratio $R = F_{\min}/F_{\max} = 0$.

Creep-fatigue tests with hold time of 15 s (tensile dwell) have been performed at the temperature of 950°C in load control at $R = 0$ with a trapezoidal wave shape, with 1 s loading followed by 15 s tensile dwell (load control) and 1 s unloading, by adopting the same net stress levels employed for the continuous cycling (pure LCF) tests.

Test Results

For both the load controlled pure LCF and creep-fatigue conditions, nine tests have been carried out in each condition (for a total of 18 tests). The different nominal (net) stress ranges applied in the pure LCF tests have been chosen, so that the range of fatigue lives between 10^4 and 10^5 cycles could be explored. The same values of net stress ranges have been applied in the fatigue test with hold time, with the aim of studying the effect of the presence of dwell on the fatigue life. The results of the high temperature fatigue tests with notched specimens without and with dwell are shown in Fig. 3. However, due to confidentiality issues set by the project consortium agreement, experimental results are presented here in normalized form or, when needed, by omitting the indication of the unit labels. By comparing these two set of test, for the same applied net stress, when a hold time of 15 s is introduced in the fatigue cycle a reduction of fatigue lives by about a factor of 10 has been observed, Fig. 3. In order to gain a more in-depth understanding of the differences observed between the pure fatigue tests and the creep-fatigue tests, all the specimens have been analyzed in the SEM from two different points of view. First, it is of interest to identify the position of the failure initiation site with respect to the notch root. Second, fracture surfaces have been analyzed with the aim to identify the main differences between the pure fatigue and the creep-fatigue tests and to identify the position of failure initiation sites with respect to the secondary crystallographic orientation of the material.

Identification of Failure Initiation Sites with Reference to Notch Geometry

All the specimens have been analyzed in the SEM by a lateral view, so that the exact positions of failure initiation sites could be measured with a high degree

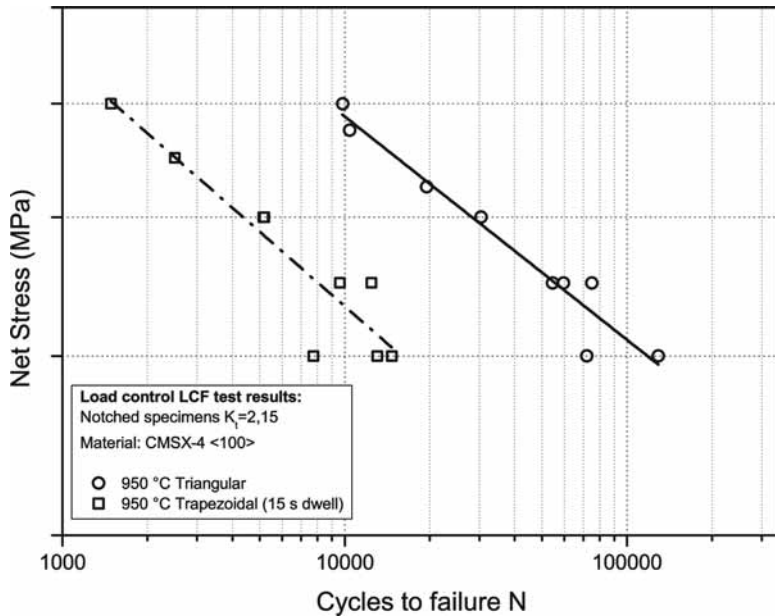


FIG. 3—Pure fatigue and creep-fatigue test results with $\langle 001 \rangle$ notched specimens at 950 °C.

of precision. It has been observed that failure initiation sites are not located at the notch root in correspondence of the minimum net-section area but in a position away from the specimen's notch root. For every specimen, several measurements have been made, and maximum and minimum values for each of the failure initiation positions have been recorded. The mean values of the measurements of the distance a are shown in Fig. 4 in the form of frequency count histograms: for specimens tested with triangular and with trapezoidal waveform cycles at 950 °C they are distributed around a mean value of $a = 1.132$ mm and $a = 1.129$ mm, respectively. This observation, in agreement with the work of other researchers [11,12], needs to be validated by the application of damage models based on the local concepts, that may demonstrate that the most critical conditions in the notch region are found in positions different from the notch minimum cross-section [13].

Identification of Failure Initiation Sites with Reference to Crystallographic Orientation

The failure surface of all the tested specimens has been also analyzed in the SEM with the aim of examining the specific microstructural features and of highlighting the differences between the pure LCF and the creep-fatigue tests.

In the specimens tested with triangular waveform cycles (pure LCF), it has been clearly observed that the single failure initiation site is located almost perpendicular to the crack propagation front and one dominant crack brings the

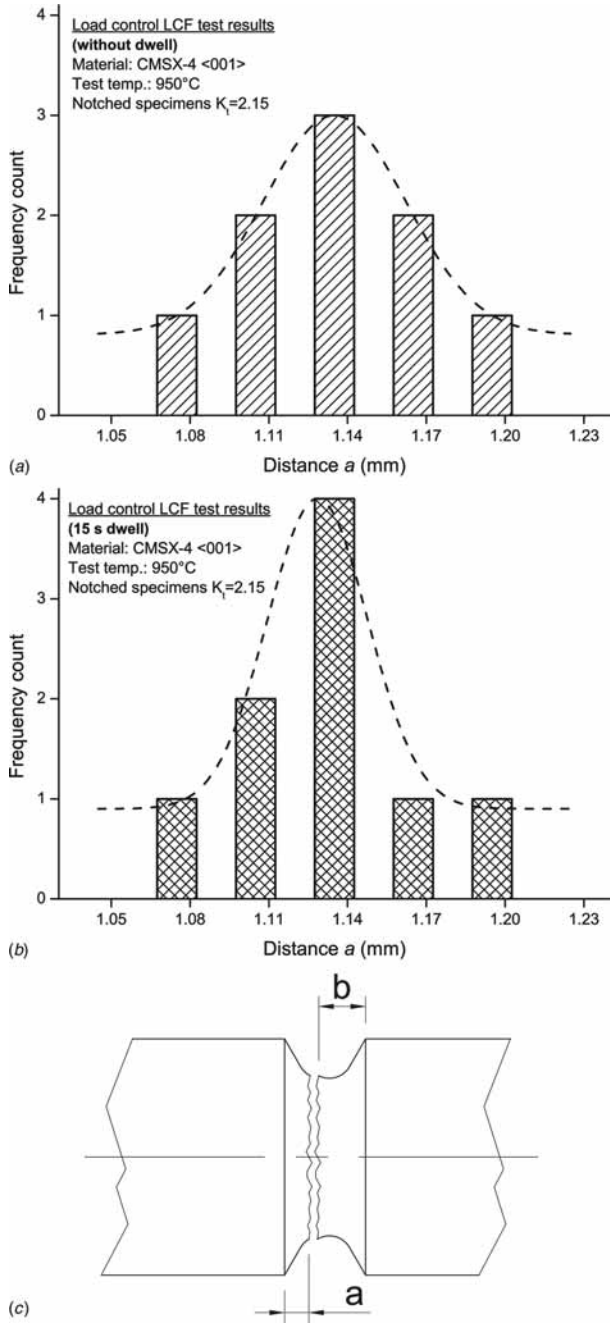


FIG. 4—Measured positions of the fatigue failure initiation sites in the notched specimens respect to the notch geometry (above) and frequency count distributions of distance a measurements in the case of load controlled LCF tests without (a) and with 15 s dwell (b).

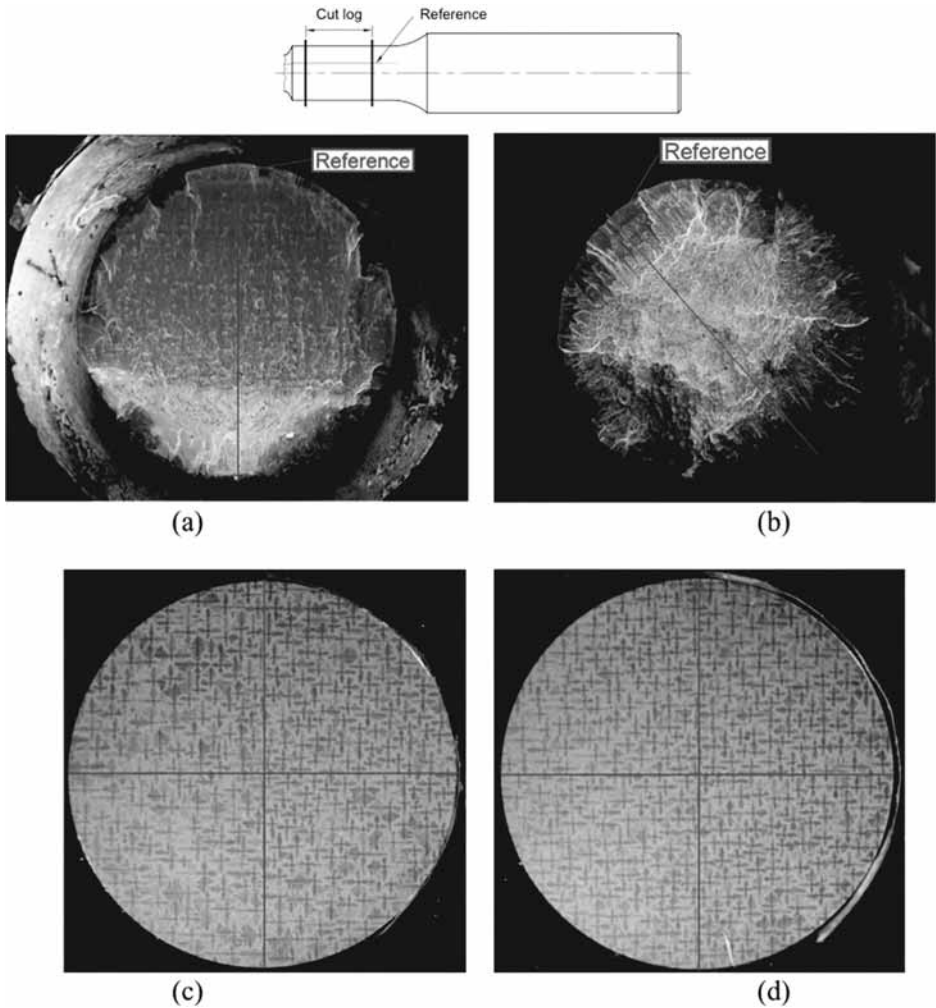


FIG. 5—SEM images of the fracture surfaces in: (a) pure LCF tests; (b) creep-fatigue test with tensile dwell of 15 s. The metallographic analysis reveals that both in case of pure fatigue tests (c) and in that of creep-fatigue tests (d), failure initiation sites are located at the 0° secondary crystallographic direction.

specimen to failure, Fig. 5(a). On the contrary, in the specimens tested with trapezoidal waveform cycles (15 s dwell at maximum load) it has been observed that the failure initiation and propagation region are distributed at the periphery of the failure surface in four angular positions at 90° angular distance from one another, with a distinct final failure core section with an irregular square shape, Fig. 5(b). These SEM observations reveal a distinctive change of the failure mode from pure fatigue to fatigue/creep conditions. Also in this case, it has been clearly observed that the multiple failure initiation sites are located almost

perpendicular to the crack propagation front. By marking the angular position of the failure initiation sites with the help of a thin strip made of copper bonded to the lateral surface of the specimens, the single crystal material microstructure has been analyzed by polishing and etching a surface cut few millimeters away from the failure surface. This analysis has shown very clearly that, in both cases of pure LCF and LCF + dwell, the failure initiation sites are located in exact correspondence and closely oriented at 0° with reference to the secondary crystal orientation of the single crystal cubic FCC (face-centered cubic) structure, Fig. 5(c) and 5(d).

The detailed analysis of the failed specimens has revealed that different mechanisms come into play when notches are introduced in single crystal materials. First of all, the most critical location from the point of view of fatigue damage is the result of a combination of the local crystal orientation and the state of stress and strain that it is found in the notch region, and it is not governed only by the notch geometry as in the case of isotropic material. Second, it has been observed a marked difference in the characteristics of the fracture surface between the pure LCF tests and the creep-fatigue tests. In the latter case, the tight link between the failure initiation site and the crystallographic directions is maintained, but the peculiar extension of the damage process zone (in four separate zones at 90° angular distance) seems to justify, at least partially, the reduction in the fatigue lives observed in the fatigue tests with tensile dwell. For these reasons, it is essential to model the material behavior in the notch region for the purpose of discriminating the damage parameters suitable for the prediction of fatigue lives.

Modeling Material Behavior

One of the most successful models to simulate the anisotropic behavior of single crystal materials is the Cailletaud model [14]. This model uses a crystallographic approach, involving the viscoplastic constitutive equations and isotropic and kinematic hardening variables of the material [15]. This type of model provides a detailed and quantified assessment of material behavior at a microscopic level when subjected to specific conditions of loading and environment.

Single Crystal Material Behavior

In most of the single crystal materials, the prevalent deformation mechanism is governed by a slip along crystallographic directions [4,16]. From the point of view of the mechanical behavior, that means that the plastic strain rate of the material is the result of a sum of plastic strain rate coming from each active slip system [17]. In single crystal materials, the occurrence of visco-plastic conditions is strongly dependent on the temperature, but in any case the governing variable is the resolved shear stress. Slip in metal crystals often occurs on planes of high atomic density in closely packed directions. The four octahedral planes corresponding to the high-density planes in the FCC crystal are shown in Fig. 6(a). The four octahedral slip planes have three

primary slip directions resulting in 12 independent primary $\langle 110 \rangle \{111\}$ slip systems. The four octahedral slip planes also have three secondary slip directions resulting in 12 independent secondary $\langle 112 \rangle \{111\}$ slip systems. Thus, there are 12 primary and 12 secondary slip systems associated with the four octahedral planes [11]. In addition, the three cube slip planes have two slip directions resulting in six independent $\langle 110 \rangle \{100\}$ cube slip systems, as shown in Fig. 6(b).

Non-Linear Behavior of Single Crystals

The Caillietaud model [14,15] assumes that the plastic strain is caused by plastic slip on slip systems, defined by a slip plane of normal $n^{(s)}$, and a slip direction $l^{(s)}$. Different slip systems can be active in the material: in the case of FCC single crystal materials such as the CMSX-4 superalloy, the octahedral slip and the cubic slip systems are of interest. Each system will have its own material parameters.

The arrangement of the slip systems is used to build the orientation tensor, $\mathbf{m}^{(s)}$, that allows the calculation of the resolved shear stress for the slip system (s), $\tau^{(s)}$. Assuming that the slip behavior is described by Schmid's law, the resolved shear stress can then be used as a critical variable to evaluate the inelastic flow and a critical resolved shear stress, τ_0 , is introduced to characterize the initial yield on each slip system. This threshold value is introduced both in positive and negative direction on each slip system: twelve octahedral slip systems (primary octahedral slip systems) and six cubic slip systems are used.

Two variables are defined for each slip system (s), $r^{(s)}$, and $x^{(s)}$, corresponding, respectively, to isotropic hardening (expansion of the elastic domain), and kinematic hardening (translation of the elastic domain). A slip system is activated when its resolved shear stress $\tau^{(s)}$ is greater than $(x^{(s)} + r^{(s)})$ or less than $(x^{(s)} - r^{(s)})$, and the slip rate will be known as long as stress and the hardening variables are known. The internal state variables used to define the evolution of $r^{(s)}$ and $x^{(s)}$ are the accumulated slip $v^{(s)}$ for isotropic hardening and the variable $\alpha^{(s)}$ for kinematic hardening [18].

If the stress tensor applied to a given point in the single crystal, σ , is known, the resolved shear stress for system $\tau^{(s)}$ can be derived according to

$$\tau^{(s)} = \sigma : \mathbf{m}^{(s)} = \frac{1}{2} \sigma : (n^{(s)} \otimes l^{(s)} + l^{(s)} \otimes n^{(s)}) \quad (1)$$

where:

$n^{(s)}$ = the normal to the slip plane and

$l^{(s)}$ = a slip direction in that plane.

The kinematic hardening behavior of the material is described by adopting the kinematic variable $x^{(s)}$, representing the equivalent of a back stress for the given slip system (s), and expressed as a function of the internal variable $\alpha^{(s)}$

$$x^{(s)} = c\alpha^{(s)} \quad (2)$$

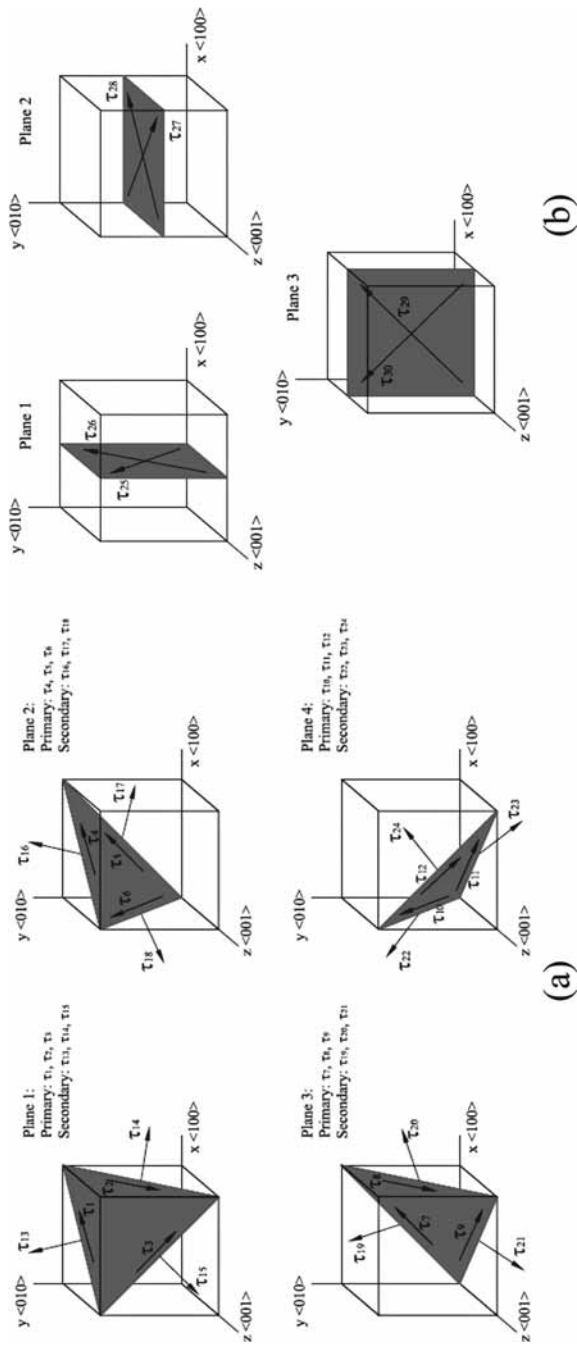


FIG. 6—Octahedral slip system (a) and cubic slip system (b) of an FCC single crystal material (cubic symmetry).

while the viscoplastic strain rate tensor $\dot{\epsilon}^p$ is computed from the shear strain rate on every slip system (s) as

$$\dot{\epsilon}^p = \sum_{(s)} \mathbf{m}^{(s)} \dot{\gamma}^{(s)} \quad (3)$$

where

$$\dot{\gamma}^{(s)} = \dot{\nu}(s) \text{sign}(\tau^{(s)} - x^{(s)}) = \left\langle \frac{|\tau^{(s)} - x^{(s)}| - r^{(s)}}{K} \right\rangle^n \text{sign}(\tau^{(s)} - x^{(s)}) \quad (4)$$

where:

$\langle z \rangle = \max(z, 0)$ are the Macauley brackets.

The isotropic hardening of the material is described by the following relationship:

$$r^{(s)} = R_0 + Q \left(1 - \exp(-b\nu^{(s)}) \right) \quad (5)$$

while the non-linear kinematic hardening rule is given by the equation

$$\dot{\alpha}^{(s)} = \dot{\gamma}^{(s)} - d\alpha^{(s)}\dot{\nu}^{(s)} \quad (6)$$

The non-linear material model summarized here needs, for every slip system family and temperature, seven constants. The parameters K and n describe the evolution of the inelastic flow with a “Norton like” evolution law, Eq 4, while R_0 , Q , and b express the change of the extension of the elastic domain with the accumulated inelastic strain, Eq 5. The two constants c , and d completely describe the non-linear kinematic hardening behavior, as given in Eqs 2 and 6.

Identification of Model Parameters and Comparison with Experiments

The elasto-visco-plastic material model was employed using the multi potential constitutive equations to represent CMSX-4 behavior at 950 °C, as implemented in the Zmat/Zset material library [19–21]. This model considers the characteristic FCC cubic symmetry slip systems, in particular both octahedral, Fig. 6(a), and cubic slip systems, Fig. 6(b).

The general elasto-visco-plastic material model was divided in three blocks: (i) elastic behavior: cubic symmetry material (three constants) [22]; (ii) visco-plastic behavior on octahedral slip systems: inelastic flow, isotropic hardening, kinematic hardening, inter-system hardening of the isotropic variable; (iii) visco-plastic behavior on cubic slip systems: inelastic flow, isotropic hardening, kinematic hardening, inter-system hardening of the isotropic variable [23].

Material parameters were determined by comparisons between numerical simulation results and real experimental data of tensile tests, creep tests, strain controlled LCF tests and stress relaxation tests (SRT) at 950 °C on smooth specimens $\langle 001 \rangle$. The first three sets of test data have been generated by other

partners of the project [24,25], while stress relaxation tests have been performed by the present author in accordance with ASTM E328-02 [26]. In order to obtain a complete set of parameters, additional data for test carried out with plain specimens in the $\langle 111 \rangle$ crystallographic direction are needed for the identification of material model parameters for the cubic slip systems. In fact, by testing FCC single crystal materials along principal crystallographic directions, i.e., $\langle 100 \rangle$, $\langle 010 \rangle$ or $\langle 001 \rangle$ directions, respectively, cubic slip systems are not activated. Thus for the identification of the visco-plastic parameters related to the cubic slip systems, experimental tests carried out with specimens oriented in $\langle 111 \rangle$ direction, where the generated slip is predominately cubic, must be available. Tensile, creep, and stabilized stress-strain hysteresis loops obtained by means of strain controlled LCF tests with plain specimens along $\langle 111 \rangle$ crystallographic direction for CMSX-4 superalloy at 950 °C have been taken from the literature [27,28].

The material parameters have been obtained by employing the Z-mat/Z-set model simulation/optimization module, which can be used to simulate material behavior for a volume element, based on the material constitutive equations [29]. Final parameters that have been obtained by the optimization procedure represent a reasonably good compromise to cover all the experimental observations. As an example, in Fig. 7 it is shown the comparison between the experimental tests and the behavior predicted by the material model in the case of tests in the $\langle 001 \rangle$ crystallographic direction. For example, LCF tests conducted in total strain control at 950 °C with smooth specimens in the $\langle 001 \rangle$ direction with a tensile dwell of 15 s could be simulated and compared with the experimental results, Fig. 8. It is worth observing that in the simulation of these LCF tests, the same stabilized cycle can be obtained after a complete simulation of the single cycles with dwell or by simulating a single “long” dwell time with a total hold time equal to the sum of the dwell times up to half life. This means that the material model is able to correctly capture the stress relaxation behavior of the material, either in pure SRT tests or in the strain controlled LCF tests with hold times. The final set of parameters obtained by the optimization procedure represents a solution of compromise covering material behavior under different loading conditions. It must be emphasized that if, for example, only the creep behavior would be of interest, the LCF data should be left out from feeding the optimization process. Herein the main emphasis is given to the stress relaxation behavior more than to the creep strain accumulation. If the latter would be the primary interest of the analysis, and the precise description of the creep strain versus time curves with primary and tertiary creep phases would be exactly captured, different models, with an even more elaborate mathematical form, should be adopted [30].

Modeling the Behavior of Notched Specimens

Once a complete set of parameters for the anisotropic (cubic) elasticity have been identified, and the visco-plastic model parameters for both octahedral and cubic slip systems have been obtained for the temperature of 950 °C, the

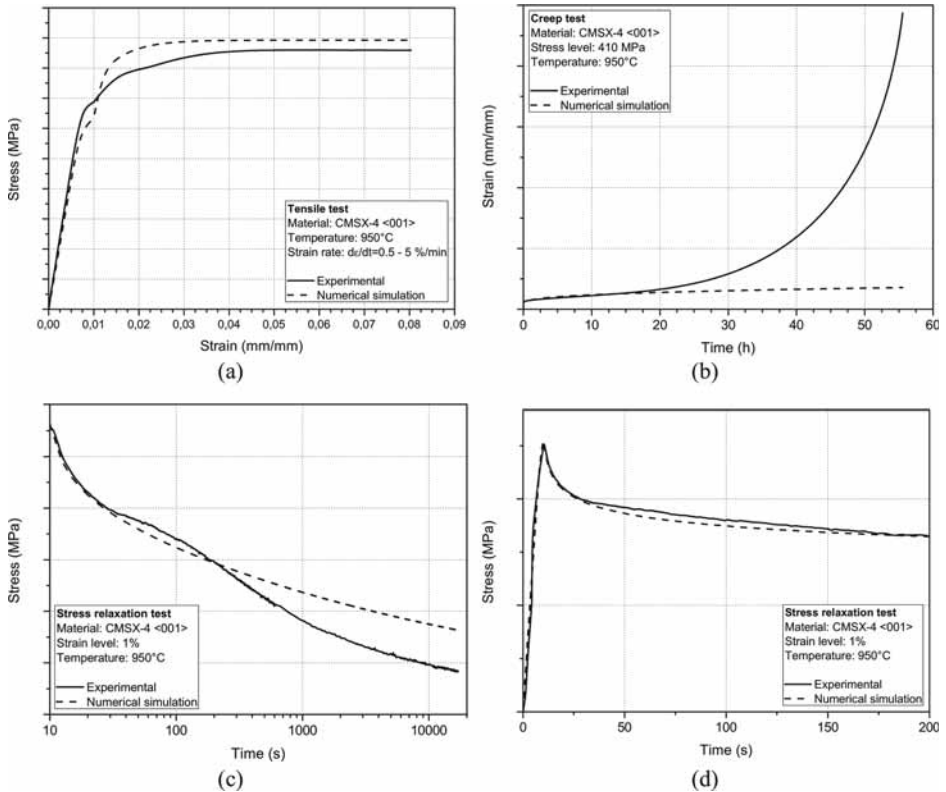


FIG. 7—Comparison between experimental test results on CMSX-4 at 950°C and simulation material behavior: tensile test (a), creep test (b), stress relaxation test (c, d). The graph (d) is a magnified view, with time in linear scale, of the graph (c), with time in log scale.

material behavior has been implemented in the Z-mat interface for the ABAQUS finite element analysis code [31]. The Z-mat code is mainly a library of material behavior routines (constitutive equations), which can be used in combination with non-linear FEA general purpose codes [20,29].

Modeling the Fatigue Tests

Applying the material model to an ABAQUS finite element model of the notched specimen, the response to triangular waveform cycles of LCF tests was simulated to evaluate stress-strain behavior inside the notch. The FE model has been built with a 3D geometry to cover all the possible directions due to the anisotropy of the single crystal material but, taking into account both the specimen and the material symmetry, it is composed of the minimum portion of the specimen necessary to completely represent the FCC cubic symmetry. In particular, a 45° segment of half of the specimen was modeled, Fig. 9. Simulations of the

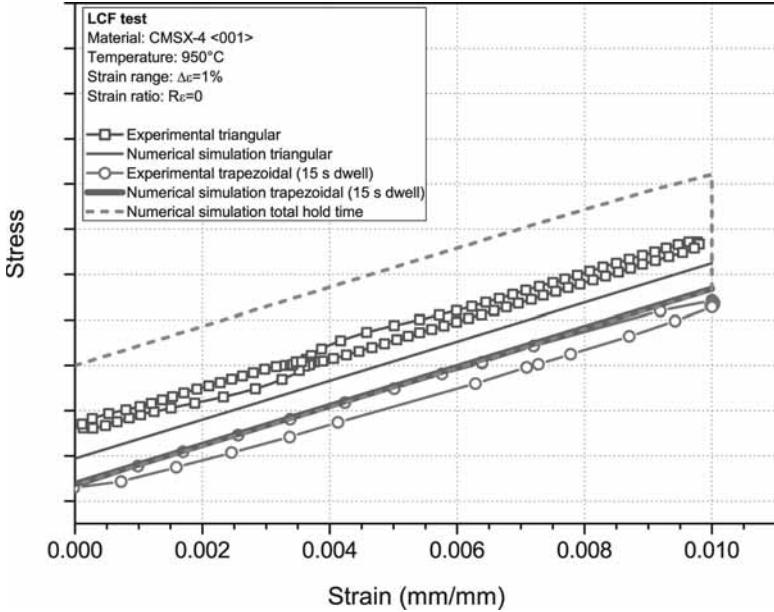


FIG. 8—Comparison between experimental test results on CMSX-4 at 950°C and simulation material behavior: strain controlled LCF tests without and with tensile dwell.

stress-strain behavior of the notched specimens have shown that an elastic shakedown occurs after few cycles also when the highest stress range applied in the actual fatigue tests was used in the simulation. This result must not be seen as reductive, but it is a consequence of the application of the single crystal material model, whose parameters have been identified by fitting a fairly large and consistent set of simple test results (tensile, creep, stress relaxation, and LCF with plain specimens). Thus, the solution obtained under the described hypothesis must be regarded as representative of the loading conditions of the CMSX-4 notched specimens under pure fatigue loading. However, it must be observed that the model does not exclude the possibility that for higher net stress or for temperatures higher than 950°C for which new values of the model parameters should be identified by comparison with tests conducted at the temperatures of interest, the CMSX-4 material could experience extensive plastic deformation.

From the analysis of the stress state in the notch, it can be seen that the maximum value of von Mises equivalent stress is located in a position away from the centre of the specimen and it is oriented at 0° respect to one of the reference axis of the cubic material. Even if the von Mises equivalent stress is not apt to represent the condition of yielding in single crystal materials [16,18], here it can be viewed simply as a measure of the intensity of the shear stresses acting on all the active slip systems. It can be proved that, at least in the case of a polycrystalline material, the arithmetic mean of the squares of the resolved shear stresses acting on all material planes at a given point can always be interpreted as the von Mises equivalent stress. As a consequence, the von Mises

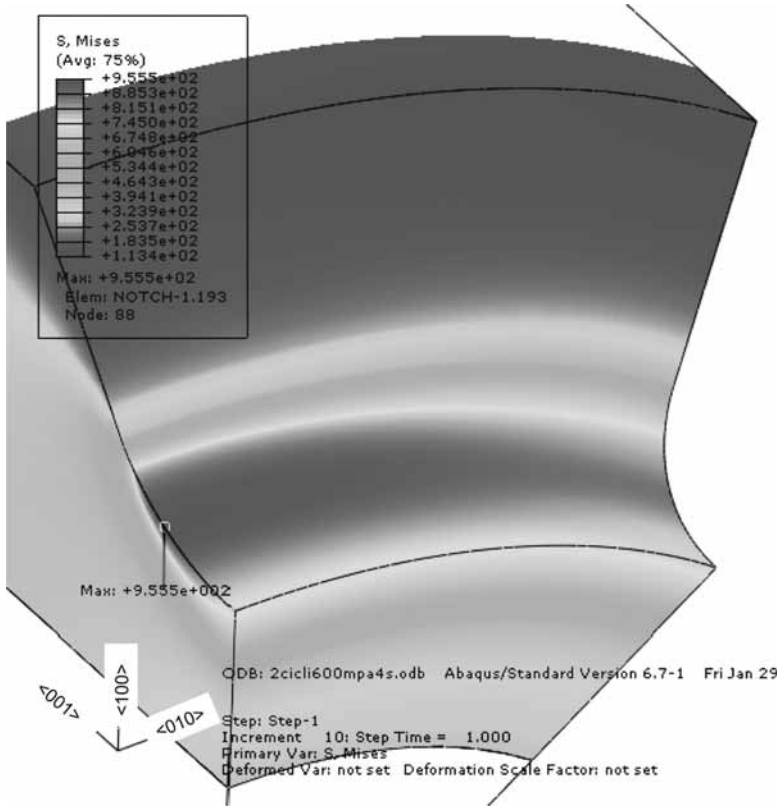


FIG. 9—The most critical point according to the von Mises effective stress distribution in the notch region exactly corresponds to the location of failure initiation site as observed in the SEM analysis.

stress can be employed as an indicator for selecting the most critically stressed point in the notch root. It must be observed that its calculated position is coincident with the position of the failure initiation site as observed in the SEM analysis and accurately measured as shown in section titled “Test Results.”

Numerical results of simulations performed imposing the same net stress ranges of the LCF experimental tests at 950 °C were used to identify a fatigue parameter able to represent the fatigue behavior of the material regardless of component’s geometry. As shown from the SEM analysis, the fatigue crack initiation is a very localized phenomenon: for this reason, a local parameter was considered for the calculation of fatigue life. At the same time, one of the main objectives in the development of a suitable fatigue damage parameter is to ensure the transferability of fatigue test results obtained with simple plain specimens to more complicated stress-strain conditions, either due to the presence of notches or introduced in a mechanical component by multiaxial loading. The damage parameters considered were the range (in a cycle) of von Mises equivalent stress $\Delta\sigma_{VM}^*$ as given by

$$\Delta\sigma_{\text{VM}}^* = [\Delta\sigma_{11}^2 + \Delta\sigma_{22}^2 + \Delta\sigma_{33}^2 - \Delta\sigma_{11}\Delta\sigma_{22} - \Delta\sigma_{11}\Delta\sigma_{33} - \Delta\sigma_{22}\Delta\sigma_{33} + 3(\Delta\sigma_{12}^2 + \Delta\sigma_{13}^2 + \Delta\sigma_{23}^2)] \quad (7)$$

and the range of the maximum value of shear stress $\Delta\tau_{\text{max}}$ between those identified on the 30 slip systems of an FCC cubic symmetry material

$$\Delta\tau_{\text{max}} = \max_{(s)} \Delta\tau^{(s)} = \max_{(s)} (\max_t \tau^{(s)}(t) - \min_t \tau^{(s)}(t)) \quad (8)$$

These quantities were calculated on whole notch surface as shown in Fig. 10.

So, in agreement with experimental observations of fracture surfaces, for each of the experimental data point, the values of parameter $\Delta\tau_{\text{max}}$ calculated in the point, where the maximum value of von Mises stress variation was located were taken into account. In Fig. 10(b), it can be observed that in this point $\Delta\tau_{\text{max}}$ does not assume the maximum value on the notch surface.

Then, these fatigue damage parameters were used to compare the results of load controlled LCF tests at $R=0$ on notched specimens and strain controlled LCF tests at $R_\epsilon = \epsilon_{\text{min}}/\epsilon_{\text{max}} = 0$ on smooth specimens [25]. From the analysis of experimental stabilized hysteresis loops of the strain controlled LCF tests, it can be observed that a nearly elastic behavior is achieved, so that is quite simple to calculate the range of the von Mises stress or the maximum range of resolved shear stress in the plain specimens. In this case, the same result is obtained by using the actual experimental turning points from the acquired stabilized cycles.

The equivalent von Mises stress $\Delta\sigma_{\text{VM}}^*$ calculated for these two set of experimental results is given in Fig. 11(a). It can be observed very clearly that this parameter does not adequately represent the experimental results of LCF tests on notched and smooth specimens and the two fitting curves follow different trends. This is an additional proof of the fact that fatigue test results are tightly linked to both the geometry and to the crystallographic direction of the material of the notched component.

Vice versa, by calculating the maximum range of the resolved shear stress $\Delta\tau_{\text{max}}$ at the point, where the maximum value of von Mises stress range ($\Delta\sigma_{\text{VM}}^*$) is achieved, experimental results of LCF tests on notched and smooth specimens fall on the same curve, as shown in Fig. 11(b). This apparently surprising result of the remarkably good agreement between the strain controlled LCF tests and the load controlled notched LCF tests, once they are interpreted with the correct fatigue damage parameter, is a consequence of the fact that the fatigue damage in single crystal materials is mainly governed by slip and thus the range of resolved shear stress must be considered as a valid parameter for predicting the fatigue life of notched components by using tests data obtained from strain controlled LCF tests with plain specimens. The most critical position has been associated with the location in the notch where the maximum range of von Mises stress is achieved. Due to the periodic symmetry of the material structure, this position is coincident with the 0° secondary crystallographic direction, $\langle 010 \rangle$ or $\langle 100 \rangle$. In order to discriminate between the two hypothesis, whether the single crystal material microstructure or the combination of shear stress or, again, an equivalent stress determines the position where the damage parameters needs to be evaluated, fatigue tests

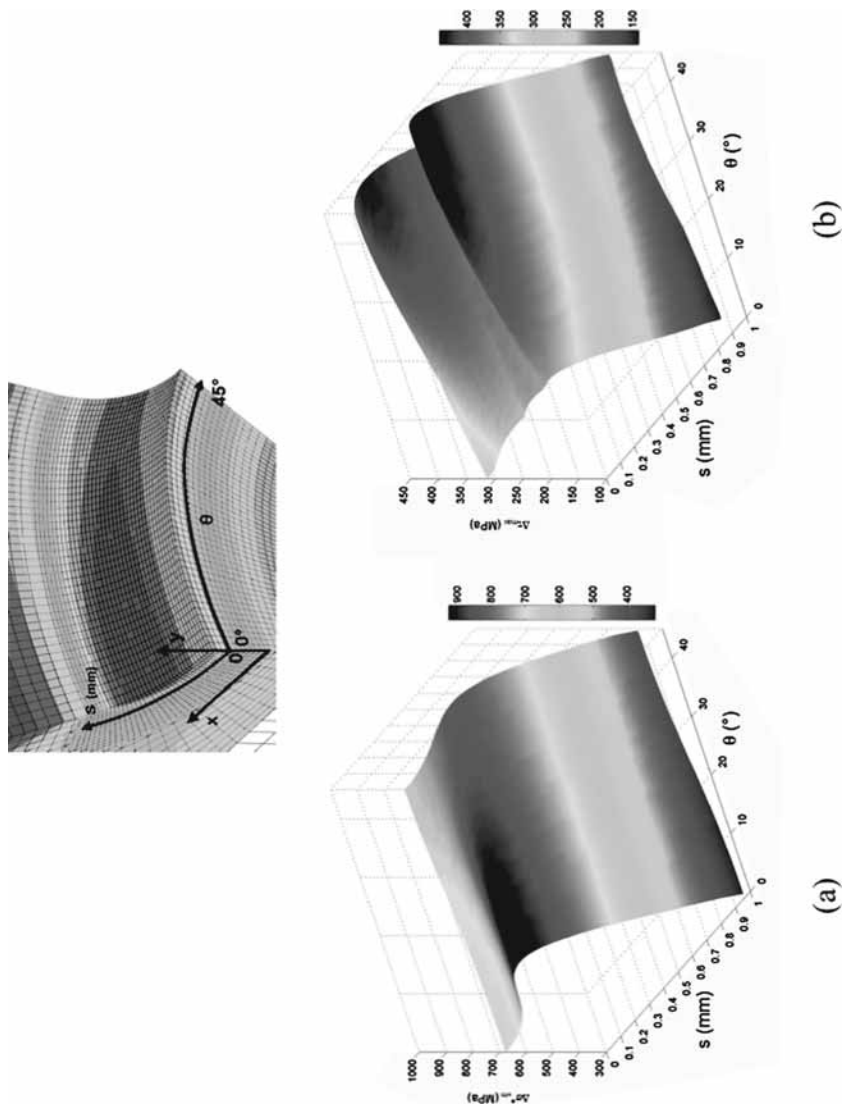


FIG. 10—Distribution of the damage parameters at the notch surface (coordinate reference above): range of von Mises equivalent stress $\Delta\sigma_{vM}$ (a); maximum range of resolved shear stress on crystallographic planes $\Delta\tau_{max}$ (b).

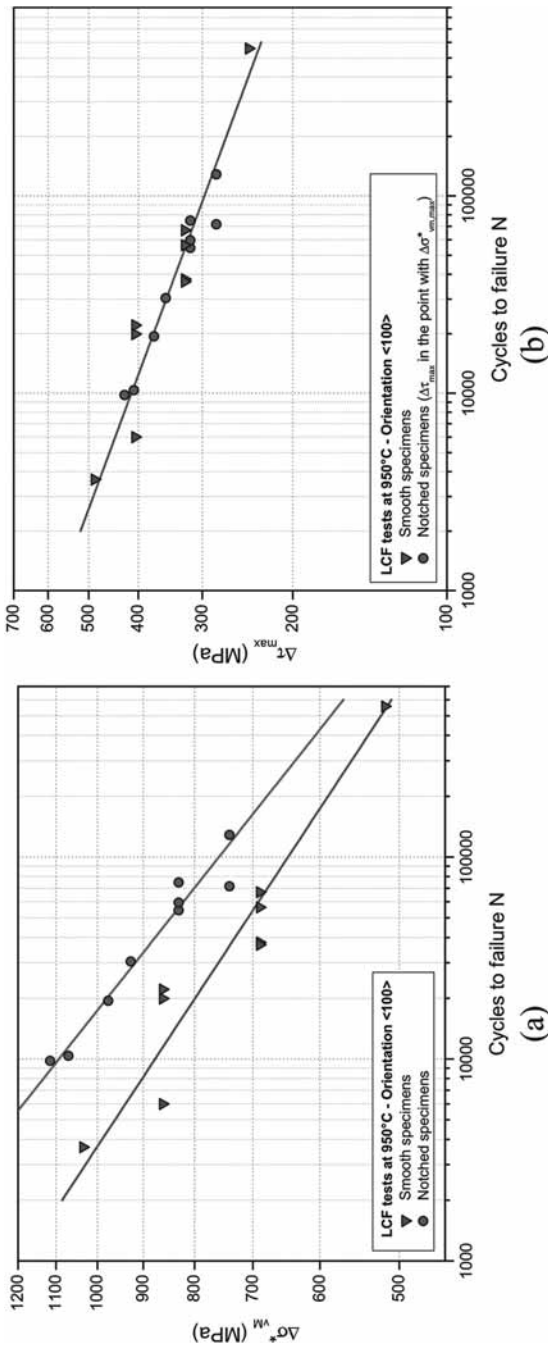


FIG. 11—Comparison between strain controlled LCF tests with plain specimens and load controlled LCF tests with notched specimens of CMSX-4 at 950°C; damage parameter is calculated as the maximum range of von Mises equivalent stress $\Delta\sigma^*$ (a) or as the maximum range of resolved shear stress on crystallographic planes $\Delta\tau_{max}$ in the position of maximum von Mises stress (b).

with a non-symmetrical notches may be carried out. However, this issue is beyond the scope of the present paper and will not be discussed further.

Modeling Creep Tests With Notched Specimens

One of the most widely used parameters for the prediction of time to rupture of metals under creep is the Larson–Miller parameter:

$$P_{LM} = T[\log t_R + C] \quad (9)$$

In the case of CMSX-4 single crystal superalloy, creep rupture stresses σ plotted versus Larson–Miller parameter P_{LM} with a material constant $C = 20$ show a limited scatter 0. If creep rupture tests performed with notched specimens are plotted in the same diagram of creep stress rupture tests of plain specimens as a function of nominal net stress, it becomes evident that the mechanisms of creep failure in the two sets are quite different, as they data points fall on very distinct curves. In fact, the material in the notch region is loaded in completely different way because, due to the non-uniform stress distribution, stresses tend to relax and to be progressively redistributed during time [32,33].

By employing the elasto-visco-plastic material model already described above, the behavior of the material in the notched region was simulated by modeling the complete geometry of the specimen and by performing a FEA simulation of creep tests at 950 °C at different applied net stresses with the aim of deriving a suitable stress parameter able to interpret the creep stress-rupture tests both with plain and with notched specimens in a unified way. So, instead of considering the local maximum of an equivalent stress, the volumetric average of the equivalent von Mises stress in the notch region

$$\sigma_{vM}^{\wedge} = \frac{1}{V} \iiint_V \sigma_{vM} dV$$

can be taken into account, where the volume V corresponds to the highly stressed region in the notch region, as illustrated in Fig. 12. Such calculation is made after the redistribution of the stresses in the notch due to creep stress relaxation has reached a stabilized state, which is observed in the simulations at different time scales depending on the level of applied net stress.

By employing the volumetric average of the equivalent von Mises stress, the data points corresponding to the creep stress rupture tests with notched specimens at 950 °C fall on the same curve as those derived by creep tests with plain specimens [24,28,30], as shown in Fig. 13.

The possibility to collapse all the data on the same curve permits the application of simple models for the calculation of creep life, as it is the case of the Larson–Miller parameter P_{LM} , also in the case of notched specimens. This simplified engineering approach, based however on the application of a sophisticated model of the visco-plastic behavior of the single crystal material, gives the chance to indicate a suitable method to assess the fatigue/creep life of notched specimens with the aim of developing

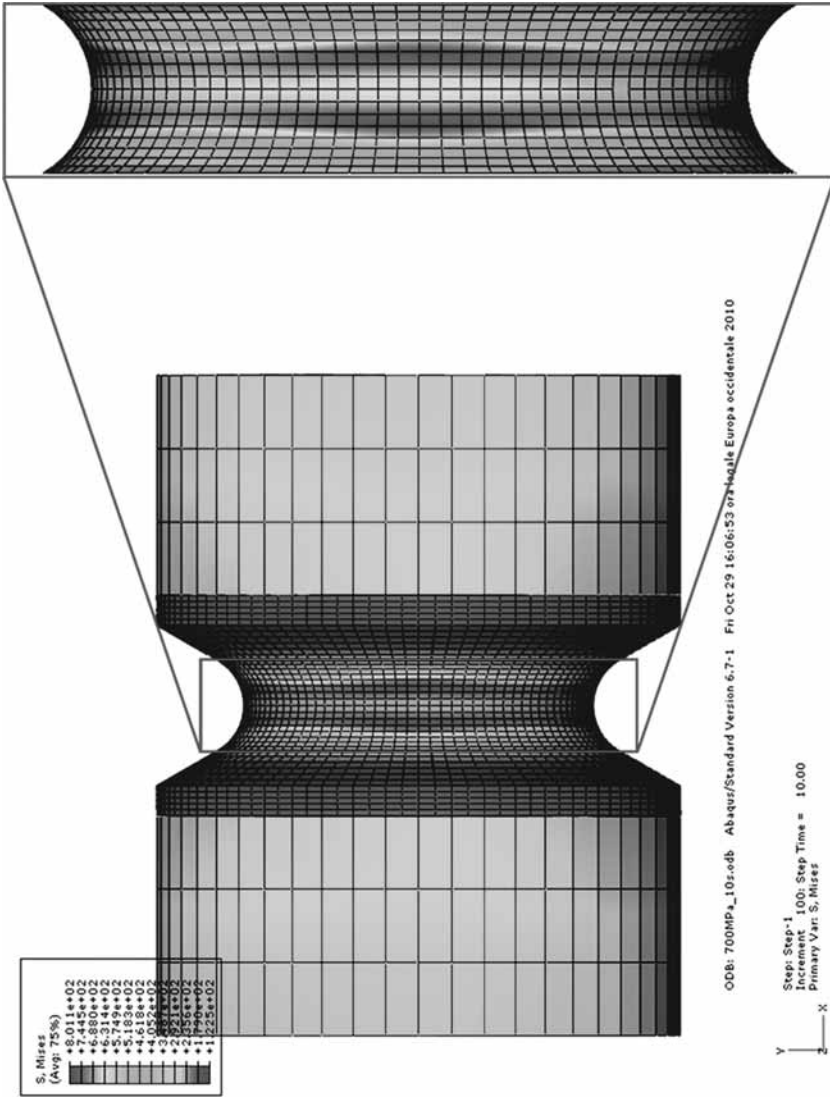


FIG. 12—FEM model employed for the calculation of the volumetric average of the Mises equivalent stress, in the simulation of creep tests with notched specimens.

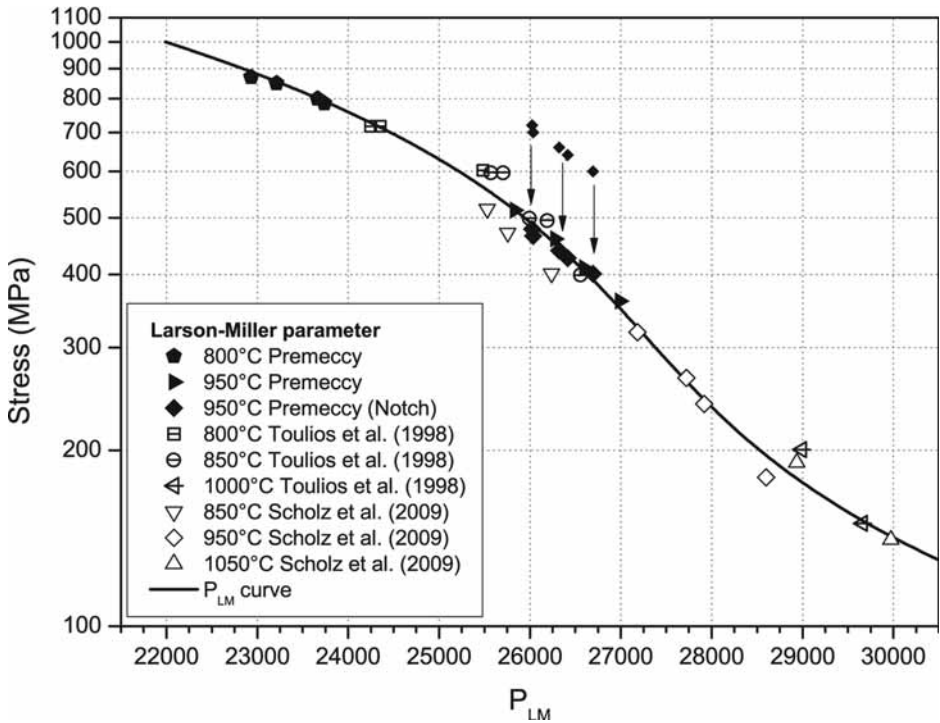


FIG. 13—Larson–Miller parameter plot for creep rupture behavior (001): notched specimen data points fall on the plain specimens P_{LM} curve when the volumetric average of the von Mises effective stress is used.

engineering methods for the assessment of the structural integrity of components.

Creep-Fatigue Tests

In the case of creep-fatigue conditions, one of the most widely used approaches is that of damage fraction summation, also known as Miner–Robinson rule, which combines the Miner rule for fatigue damage and the life fraction rule for creep damage:

$$\phi_C + \phi_F = \sum_1^{N_f} \left(\frac{t_h}{t_R} \right)_{\text{creep}} + \sum_1^{N_f} \left(\frac{n}{N_f} \right)_{\text{fatigue}} = D_{CF}$$

In the case of constant amplitude tests, the damage fraction summation is

$$\phi_C + \phi_F = \left(\frac{\sum t_h}{t_R} \right)_{\text{creep}} + \left(\frac{n}{N_f} \right) = D_{CF}$$

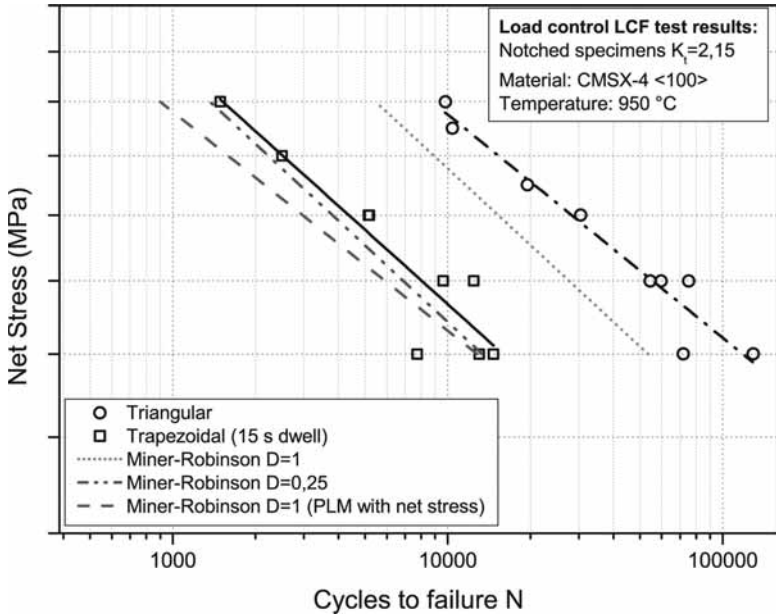


FIG. 14—Fatigue life prediction curves of creep-fatigue of CMSX-4 notched specimens according to the modified damage summation rules.

Where:

- t_h = the total dwell time at the given maximum stress,
- t_R = the time to rupture at a given stress,
- n = the number of cycles at failure in the case of fatigue-creep interaction,
- and
- N_f = the number of cycles at failure in the case of pure fatigue.

The damage fraction relative to the fatigue tests with dwell has been calculated by employing the experimental curves for fatigue according to the $\Delta\tau_{\max}$ parameter, Fig. 11(b), and the Larson–Miller parameter curve. For the latter, as it as been shown in Fig. 13, the curve derived from creep stress rupture tests with smooth specimens can be adopted also in the case of notched specimens, provided that the volumetric average in the notch region of the von Mises equivalent stress is used to calculate the P_{LM} needed to estimate the time to creep rupture. In the case of the fatigue tests with dwell carried out in the present work, a damage sum of about 0.25 has been calculated, witnessing a strong interaction of creep and fatigue for the temperature and the stress amplitudes applied in the tests described in this paper.

The simplified approach of calculating the total life by means of the damage fraction rule has been applied with some differences, Fig. 14. First of all, a ultra-simplified approach can be adopted if the creep damage fraction ϕ_C is calculated by assessing the time to rupture from the P_{LM} curve by using the “nominal” net stress. By considering a damage sum $D_{CF}=1$, an

over-conservative result has been obtained, predicting a reduction of fatigue life due to tensile dwell higher than that observed in the experiments. This approach, for which it would not be necessary to implement a visco-plastic material model, corresponds to indicate the mean stress over the net-section as the fitting creep parameter. However, as it has been demonstrated by both the SEM analysis, for the identification of failure initiation sites, and the numerical simulations of the stress-strain behavior in the notch region, due to anisotropic behavior of CMSX-4 single crystal material, in the presence of notches, the stress distribution in the net minimum section is not significant for the correct assessment of the creep damage. As a consequence, the evolution of the stress distribution in the three-dimensional notched region must be taken into account by employing a suitable stress parameter. Thus, the damage fraction summation rule has been applied by considering a damage sum $D_{CF}=1$ and calculating the life fraction for creep damage by means of the volumetric average of the equivalent von Mises stress. Again, it can be observed that the predicted reduction of fatigue life due to the superposition of a 15 s tensile dwell is not correctly captured when this calculation route is followed, predicting only a small reduction of fatigue life, i.e., a non-conservative result. Vice versa, when a damage sum $D_{CF}=0.25$ is considered, the reduction of fatigue life in the presence of a fatigue-creep damage interaction is correctly described.

Conclusions

Fatigue life behavior of CMSX-4 single crystal superalloy in presence of notches has been explored from both the experimental and the numerical simulation point of view. In the fatigue tests with notched specimens, for the same applied stress range (net-section stress), the effect of hold time results in an appreciable reduction of fatigue life.

Fatigue failure is controlled by both the geometry of the notch and by the microstructure of FCC single crystal materials: microstructure (crystallographic directions) play a major role in defining the failure initiation site. Testing with hold times change the behavior of the notched specimens, by changing the extension of the damage process zone. Again, fatigue/creep interactions are inextricably linked with the microstructure of single crystal materials.

As a consequence, modeling the material elasto-visco-plastic behavior is a necessary tool for analyzing the stress-strain conditions in the notch. This requires consistent and well documented data sets of experiments on different crystallographic directions: it's a cost that must be paid for building a numerical model apt to reproduce the anisotropic material behavior of CMSX-4 single crystal at the temperatures of interest.

In the case of this model, emphasis is not on "perfect copy" between the test results and the numerical simulations but on covering multiple and often conflicting material behaviors (tensile, LCF, SRT, and creep). This result can be summarized as following: when the behavior of the single crystal materials can be modeled by employing a complete elasto-visco-plastic model, the actual stress-strain conditions in a given geometry can be correctly simulated.

On the basis of the knowledge of the stress-strain behavior of the material in the notch, a suitable fatigue damage parameter has been clearly identified: maximum range of resolved shear stress on the crystallographic planes allows the transferability of tests results from plain specimen data to notched specimen data. At the same time, indication of likely position of crack initiation is given by a “companion” fatigue parameter. However, discriminating experiments are needed to rank the causes (crystallographic orientation versus critical stress parameter).

To evaluate creep damage, the volumetric average of a “classical” equivalent stress in the notch volume can be employed, provided that redistribution of stresses due to creep stress relaxation can be simulated. Fatigue/creep interaction has been numerically assessed by the application of simple approach (damage fraction summation), and the selection of a suitable method to evaluate the creep damage fraction has been suggested.

Even though new experimental results with different hold times or with different notch geometries may be needed to refine and enhance the model, the proposed model may be considered for the implementation in a design environment for the assessment of structural integrity of components.

Acknowledgments

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