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Engineering Technology Division

**HIGH-TEMPERATURE FLAW
ASSESSMENT PROCEDURE**

A STATE-OF-THE-ART SURVEY

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HIGH-TEMPERATURE FLAW ASSESSMENT PROCEDURE: A STATE-OF-THE-ART SURVEY

**M. B. Ruggles
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ABSTRACT

High-temperature crack growth under cyclic, static, and combined loading is reviewed with an emphasis on fracture mechanics aspects. Experimental studies of the effects of loading history, microstructure, temperature, and environment on crack growth behavior are described and interpreted. The experimental evidence is used to examine crack growth parameters and theoretical models for fatigue, creep, and creep-fatigue crack propagation at elevated temperatures. The limitations of both elastic and elastic-plastic fracture mechanics for high-temperature subcritical crack growth are assessed. Existing techniques for modeling critical crack growth/ligament instability failure are also presented. Related topics of defect modeling and engineering flaw assessment procedures, non-destructive evaluation methods, and probabilistic failure analysis are briefly discussed.

1. INTRODUCTION

The ever increasing demand for safety requires that very stringent and conservative design rules be used for liquid-metal reactor (LMR) components. Present design methodology is geared toward prevention of the initial existence and subsequent development of cracks. Most structures, however, contain cracks or crack-initiating defects, which are either inherent in the material or created during manufacturing and installation. To predict correctly the structural response to the in-service loading, it is necessary to account for these existing cracks.¹ The high-temperature flaw growth is schematically shown in Fig. 1. If the initial flaw is of less than the critical size, subcritical crack growth caused by fatigue, creep, or a combination of the two will ensue. With time the crack growth may self-arrest; the crack will then remain subcritical, and the component will continue to function. Conversely, the crack may, in a stable mode, grow through to the opposite face and cause a leak-before-break in the component. Should the initial flaw be of critical size (or reach the critical size with time), materials with the low fracture toughness will in most cases exhibit unstable rapid crack growth and a fast cleavage fracture. For the highly ductile materials (like those used in the LMR components), the critical crack growth does not have to be unstable. In fact, stable ductile crack growth (ductile tearing) may take place before the leak-before-break situation develops or before the remaining ligament fails by plastic collapse mechanism.

Fig. 1. Schematic of high-temperature flaw growth.

Most cracks or cracklike defects can be detected by visual inspection and/or indirect nondestructive evaluation (NDE) techniques. The initial examination establishes a limit on the largest defect existing in the component at the beginning of the service life. Subsequent modeling of the crack or defect should determine whether the crack is critical, namely whether a single service cycle will cause crack growth on the order of magnitude of the structure dimensions. Clearly such a crack could lead to a rapid catastrophic failure and therefore make the component unfit for service. In case of a noncritical crack, it is important to determine whether continued service loading will result in stable crack growth. If stable crack growth is possible, a noncritical crack may eventually grow to a critical size. It is then necessary to describe the rate and direction of the stable crack growth to establish the limits that it will impose on the service life of the structure. Complex thermomechanical cyclic loading, which results in significant inelastic behavior in LMR components, and the use of highly ductile materials make such an assessment a challenging task.

The purpose of this effort is to survey the current state of the elevated-temperature nonlinear fracture technology suitable for LMR components. First, several crack growth parameters are discussed, and examples of laboratory test data correlated with these controlling parameters are documented. Then experimental investigations of the effects of temperature, microstructural changes, and environment on the elevated-temperature crack growth behavior are considered. The experimental evidence is used to examine the characteristic features of some of the existing crack growth models. The capability of these theories to predict slow stable crack growth is assessed for the cases of fatigue, creep, and the interaction of the two. Existing techniques for modeling critical crack growth/ligament instability

failure are also considered. Finally, excursions are made into related fields of defect modeling, NDE procedures, and probabilistic failure analysis. Some of the existing engineering flaw assessment procedures are surveyed as well.

The large number of papers published on the subject of nonlinear fracture mechanics makes it impossible to find and incorporate all of them. Apologies are extended to any authors whose important contributions may have been overlooked.

2. HIGH-TEMPERATURE CRACK GROWTH

The complex phenomenon of high-temperature crack growth requires an interdisciplinary approach.² From the engineering fracture mechanics standpoint, the variation of the flow stress with temperature and presence of creep significantly limits the applicability of linear-elastic fracture mechanics. From a materials standpoint, temperature, environment, and loading history considerably affect the micromechanics of crack growth. In addition, microstructural instabilities dependent on time and temperature can occur. The micro- and macromechanics aspects are strongly interrelated.

2.1 CRACK GROWTH PARAMETERS

Several crack growth controlling parameters have been proposed. These parameters include stress-intensity factor K and stress-intensity factor range ΔK , contour integral J , and energy rate line integral C^* .

2.1.1 Stress Intensity Factor K and Stress Intensity Factor Range ΔK

Under conditions of small-scale yielding at the crack tip (i.e., the size of the plastic zone at the crack tip is small compared to the crack length and specimen dimensions) crack growth may be controlled by the stress-intensity factor K . K defines the magnitude of the crack tip stress field singularity, and for mode I crack-ing in an infinite plate it is given by

$$K = K_I = \sigma(\pi a)^{1/2}, \quad (1)$$

where σ is the applied tensile stress and a is the half crack length. In the brittle fracture mode, crack propagation becomes unstable once K reaches a critical value (called fracture toughness), which generally is a function of specimen thickness and geometry. However, under plane strain conditions (maximum crack tip constraint), the critical value (called plane strain fracture toughness) is independent of the specimen thickness. Figure 2 shows several fracture mechanics specimens and the corresponding relations for determination of K (compliance relations). Plane strain conditions at the crack tip are assumed in the use of these relations.

In case of time-dependent crack growth, K or K_{max} may be used as a controlling parameter if nonlinear effects are contained in the immediate vicinity of the crack tip. In case of cycle-dependent crack growth, where the growth is determined largely by the cyclic stress amplitude, stress-intensity factor range ΔK is more suitable. ΔK and K_{max} are related via

$$\Delta K = K_{max}(1 - R), \quad (2)$$

where R is the ratio of minimum to maximum loads.²

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Fig. 2. Commonly used specimens and compliance relations. Source: Ref. 64.

Siverns and Price³ attempted to correlate creep crack growth rate with K for 2 1/4 Cr-1 Mo steel. A single specimen geometry was used. Results given in Fig. 3 show a wide scatterband for the growth rate at a given value of K . The studies by Harrison and Sandor⁴ likewise demonstrate that K is not a successful correlating parameter for creep crack growth. Conversely, Sadananda and Shahinian⁵ report satisfactory correlation between da/dt and K for Udimet 700 at 843°C. Although only one specimen geometry was used in Ref. 5, tests conducted at different loads appeared to correlate well. Saxena⁶ explains these observations by introducing plastic, creep, and K zones at the crack tip, shown in Fig. 4. According to Saxena, a K zone of a finite size exists at the crack tip, in which the near crack tip elastic stress and deformation fields apply. A plastic zone is formed at the crack tip upon loading. As the plastic zone becomes comparable in size to the K zone, K ceases to be a controlling parameter for crack growth; then the J integral applies. Because of creep deformation, stresses in the vicinity of the crack relax with time. The creep zone therefore increases. When the creep zone becomes comparable in size to the K zone, both K and J become unsuitable for characterizing the crack growth rate (as reported in Ref. 3). In case of the Udimet 700 studied in Ref. 5, creep deformation

Fig. 3. Crack-growth rate vs stress-intensity factor. Source: Reprinted by permission of Kluwer Academic Publishers.³

Fig. 4. Schematic representation of deformation zones ahead of crack tip at elevated temperatures. Source: Copyright ASTM. Reprinted with permission.⁶

was limited, and the creep zone was smaller than the K zone. Under such conditions K adequately characterized the crack-growth rate behavior.

2.1.2 Elastic-Plastic Parameter

The J integral concept was introduced by Rice,⁷ and later proposed as a geometry-independent static fracture toughness criterion for elastic-plastic conditions.^{8,9} Values of J may be determined experimentally from load-deflection curves (see Fig. 5) using the following equation:

$$J = -\frac{1}{B} \left(\frac{dU}{da} \right). \quad (3)$$

where B is the specimen thickness. Thus for a given displacement δ_0 , J integral is interpreted as the strain energy change dU caused by a small increase in the crack length da (Ref. 10). If the specimen behaves in the linear-elastic fashion, the load-deflection curves in Fig. 5 become straight lines and J reduces to the linear-elastic strain energy release rate G , which is related to the stress intensity K as follows:

$$J = G = \frac{K^2}{K}. \quad (4)$$

For elastic-plastic materials, quantity U may also be interpreted as the elastic-plastic work necessary to deflect the specimen. In this case, J can be viewed as a measure of the intensity of the crack tip strain field.¹⁰

Fig. 5. Determination of J from load vs deflection curves. (a) Potential energy difference. (b) Rice et al. approximation. Source: Copyright ASTM. Reprinted with permission.¹⁰

Although, in a rigorous sense, the J integral concept is valid only for monotonic loading, Dowling and Begley¹⁰ successfully applied the J integral to fatigue crack growth under the conditions of large-scale yielding where the stress intensity factor K becomes ineffective. Ohtani¹¹ demonstrated that, for a sufficiently high loading rate, high-temperature fatigue crack growth rate in various metallic materials correlates well with the fatigue J integral range ΔJ (see Fig. 6). Sadananda and Shahinian² outlined an experimental procedure for determination of the J integral for displacement-controlled fatigue, which is schematically illustrated in Fig. 7. Here the rising parts of the load-displacement curves [Fig. 7(a)] for crack lengths a_1 and a_2 are displaced to a common origin [Fig. 7(b)]. The J integral is then given by the shaded area. However, Sadananda and Shahinian² also demonstrated that it is impossible to determine J_{max} for time-dependent crack growth under a static load without loading and unloading for each crack increment and thus introducing cyclic load effects.

While the fatigue J integral range ΔJ can be used as a controlling parameter for the high-temperature cycle-dependent crack growth under fast loading conditions, it does not give good results when creep deformation at elevated temperatures becomes more pronounced, that is, when low loading rates or long hold times are introduced. For correlating crack-growth rate under cyclic loading with hold times,

Fig. 6. Correlation between crack propagation rate and fatigue J integral range for cycle-dependent crack-growth conditions. Source: Ref. 11.

Fig. 7. Determination of J integral for displacement-controlled fatigue. (a) Hysteresis loops for crack lengths a_1 and a_2 , and (b) translation of loading parts of loops to common reference for integration. Source: Ref. 2.

Taira et al.¹² proposed the creep J integral range ΔJ_C . The values of ΔJ_C are experimentally determined from the displacement increment during hold time using an equation similar to that used to evaluate the line integral C^* , which is discussed below. As seen in Fig. 8, Ohtani¹¹ obtained good correlations between ΔJ_C and crack growth rate during time-dependent crack propagation tests for a number of engineering alloys.

Combined time-independent and time-dependent crack growth was studied by Ohji et al.¹³ Based on the detailed evaluation of the test results the following relation was proposed for the crack growth rate:

$$\frac{da}{dN} = a_1 \Delta J_c + a_2 \Delta J_f . \quad (5)$$

Fig. 8. Correlation between crack propagation rate and creep J integral range for time-dependent crack-growth-rate conditions. Source: Ref. 11.

As seen in Fig. 9, crack growth rates calculated with this equation agreed well with those obtained experimentally for type 304 stainless steel at 650°C.

Fig. 9. Correlation between crack growth rate (da/dN) and $[(\Delta J + 0.015\Delta J_I)]$ for various loading conditions. Source: Ref. 13.

2.1.3 Line Integral C^*

Landes and Begley¹⁴ introduced an energy rate line integral C^* , which uniquely characterizes the crack tip stress and strain rate field for materials following a nonlinear steady-state creep law. C^* was defined as the power difference between two identically loaded bodies having incrementally differing crack lengths (see Fig. 10).

$$C^* = -\frac{1}{B} \left(\frac{dU^*}{da} \right), \quad (6)$$

where B is the specimen thickness, and U^* is the power or energy rate defined for a load P and displacement rate $\dot{u}$. Experiments on Fe-Ni-Cr superalloy (Discaloy), using center-cracked panel specimens and wedge-opening loading specimens, were performed¹⁴ at 649°C. Good correlation between C^* and creep crack growth rate was obtained.

A number of investigators attempted to represent high-temperature creep crack growth data for type 304 stainless steel in terms of C^* for various specimen geometries. Saxena⁶ used center crack tension (CCT) specimens and compact

Fig. 10. (a) Crack-tip coordinate system and arbitrary line integral contour, and (b) schematic illustration of the energy rate interpretation of C^* . Source: Copyright ASTM. Reprinted with permission.⁶

tension (CT) specimens at 540, 594, and 700°C. At 650°C Koterazawa and Mori^{15*} used single-edge notched (SEN), double-edge notched (DEN), and center-notched (CN) specimens of variable dimensions; Ohji^{16*} used SEN, DEN, CT, notched round bar (NRB), and CN specimens; Taira^{17*} and Ohtani^{18*} used center notched cylindrical specimens, CN, and NRB specimens. Results of these investigations are shown in Fig. 11, where the crack growth rate is plotted vs C^* . All the sets of data fall in a relatively narrow band; crack growth rates for a given value of C^* vary by a factor of 10. In contrast, crack growth rates in terms of K vary by more than 2 orders of magnitude for a given value of K for one set of data.² Data obtained by Taira^{17*} for type 316 stainless steel at 600 and 650°C are also reported in Fig. 11. The extent of these results, however, is insufficient to make a valid comparison with the results reported on type 304 stainless steel. Saxena⁶ also reported that crack growth rates in 304 stainless steel as a function of C^* were not significantly affected by temperature in the range of 540 to 700°C.

*As cited in Ref. 2.

Fig. 11. Creep crack-growth rate in austenitic stainless steel in terms of C^* integral. Source: Ref. 2.

2.1.4 Modifications of Line Integral C^*

The energy rate integral C^* was originally proposed by Landes and Begley for the steady-state (secondary) creep regime. A considerable body of experimental evidence discussed previously demonstrates that C^* correlates creep crack-growth data for the steady-state conditions reasonably well. Attempts have been made to extend the C^* concept to the nonsteady-state creep range. Riedel¹⁹ proposed a path-independent integral C^*_A to correlate the near crack tip field to the applied load for primary (strain-hardening) or tertiary (strain-softening) creep conditions. The concept of the creep zone (with clearly distinguished primary, secondary, and tertiary creep zones) is introduced, and for a sufficiently pronounced primary creep regime, the integral C^*_A determines the near crack tip field amplitude and the growth of a secondary creep zone. Once the secondary creep starts to displace primary creep, the C^* integral becomes dominant. This theory, however, is restricted to stationary cracks and to special material laws and cannot therefore be immediately applicable to crack growth tests in a quantitative sense.

Saxena²⁰ focused on characterizing subcritical cracking in the presence of creep deformation in the crack tip region, considering both nonsteady-state (including small-scale creep and the transition regime) and steady-state creep conditions. The levels of creep deformation under which creep crack growth can occur are schematically shown in Fig. 12. Small-scale creep, transition creep, and the steady-state regions are identified. A crack tip parameter C_t for describing creep crack-growth behavior under non-steady-state creep is defined as

$$C_t = -\frac{1}{B} \left(\frac{dU^*}{da} \right) . \quad (7)$$

where B is the specimen thickness, and ΔU_t^* represents the difference in the energy rates supplied to the two cracked bodies with identical creep deformation histories

Fig. 12. Schematic representation of levels of creep deformation under which creep crack growth can occur, t = time, t_1 = transition time. Source: Copyright ASTM. Reprinted with permission.²⁰

as they are loaded to different load or deflection-rate levels (see Fig. 13). As the ratio of fixed time t to transition time t_1 goes to infinity, namely when the creep regime becomes steady state, C_t reduces to the C^* integral. Experimental evaluation of C_t was conducted using the A470 Class 8 steel under a variety of creep conditions at 538°C and at 482°C. The correlation was found to be satisfactory for a wide range of da/dt and C_t values and independent of specimen geometry. The da/dt vs C_t behavior was also independent of temperature in the temperature range considered.

For standard test specimens, the experimental formulae developed for evaluating C_t give results close to those obtained using C^* experimental estimation formulae, even when steady-state creep conditions have not been reached.

Fig. 13. (a) Load-line deflection, V_c , as function of time for bodies of crack lengths a and $a + \Delta a$ at various load levels, and (b) definition of C_t parameter. Source: Copyright ASTM. Reprinted with permission.²⁰

2.2 EFFECTS OF LOADING HISTORY, MICRO-STRUCTURE, TEMPERATURE, AND ENVIRONMENT ON CRACK GROWTH

2.2.1 Effect of Loading History

A number of experimental studies addressed the effect of load history on high-temperature crack growth. James²¹ studied the fatigue crack growth behavior of annealed types 304 and 316 stainless steels at several frequencies in an air environment over the temperature range 427 to 649°C. Increase in the crack growth rates with decreasing cyclic frequency was observed. A similar result regarding the effect of frequency on the rate of crack propagation was reported by Plumtree and Yu²² for AISI type 304 stainless steel at 570°C. Plumtree and Yu²² also reported significant influence of the waveform on the fatigue crack growth rate. A slow-fast waveform was found to be the most damaging, and a fast-slow waveform — the least damaging. In Ref. 23 James studied the effect of waveform and hold time on the fatigue crack propagation behavior of type 304 stainless steel at 537°C. At higher frequencies, tests with the sawtooth waveform and those with the square waveform produced essentially the same crack-growth response. At lower frequencies (0.083 cycle/min), however, a somewhat lower crack-growth rate was observed for cyclic loading with hold times.

The effect of hold time on fatigue crack propagation rate was considered for a 20% cold-worked type 316 stainless steel at 593°C by Shahinian²⁴ in the unirradiated condition and by Michel et al.²⁵ after radiation to 2.2×10^{22} neutrons/cm² at 427°C. Tests were conducted with a waveform of the type shown in Fig. 14 with

Fig. 14. Cycles used in elevated temperature fatigue testing. (a) Fully plastic low-cycle fatigue, and (b) predominantly elastic fatigue crack growth. Source: Ref. 26.

hold periods ranging from 0.1 to 10 min. For both unirradiated and irradiated material, application of hold periods resulted in drastic increases in crack-growth rates compared to those observed in continuous cycling tests. For an unirradiated material a 3 orders-of-magnitude growth rate increase was reported for 10-min hold period tests, while a 2 orders-of-magnitude growth increase was reported for 1-min hold period tests in irradiated material.²⁶ Michel and Smith²⁷ observed an increase in fatigue crack propagation rate due to hold time for type 316 stainless steel at 593°C in both solution-annealed and 20% cold-worked conditions. Sooley et al.²⁸ reported an increase in crack-growth rates with decreasing frequency and/or increasing hold time at maximum load for 2 1/4 Cr-1 Mo steel.

Raj and Min²⁹ investigated the effects of cycle shape on creep-fatigue interaction at 625°C for type 316 stainless steel that had been solution-annealed and aged at 800°C. The loading sequence included a single cycle with a 5-min hold period at maximum compression prior to a 60-min hold period at maximum tension, followed by a block of cycles without hold periods. Repeated application of this loading sequence demonstrated that all the creep damage occurred in the first creep cycle and subsequent hold cycles had no further effect on crack growth. Data presented in Fig. 15 show that the crack growth rates were initially similar for test histories with and without the single hold period cycle. However, for a crack length ~1 mm, an increase in crack growth rate was noted for the hold period test. A 6-h compressive hold, applied after a certain period of crack growth and prior to further no-hold cycling, caused the growth rate to be reduced to that observed in no-hold cycling tests. The prior creep damage was eliminated by the compression hold.

Fig. 15. Relationship between crack-growth rate and crack length for high strain tests on type 316 stainless steel at 625°C. Source: Ref. 29.

Shahinian³⁰ studied the effect of cycling on subsequent creep and that of creep on subsequent cycling in type 304 stainless steel at 593°C. The crack growth was reported to virtually cease with the change of loading from cyclic to static. However, the change of loading from static to cyclic caused no apparent retardation. Similar observations were reported by James,³¹ who considered the effect of prior creep damage on the fatigue crack propagation of annealed type 304 stainless steel at 538°C. For both intergranular and transgranular cracking mechanisms, fatigue crack growth behavior of the material subjected to prior creep damage (of 36 to 50% of the creep strain to rupture) did not differ significantly from that of the material with no prior creep damage history.

The effect of cyclic stress ratio $R = \sigma_{min}/\sigma_{max}$ on fatigue crack growth behavior was examined by James³² for type 304 stainless steel at 537°C. James demonstrated that at a given value of stress intensity factor range, fatigue crack growth rate increased with increasing stress ratio. Effective stress intensity factor, defined as $K_{max}(1-R)^{0.5}$, was found to correlate fatigue crack growth rate better than the stress intensity factor range ΔK (see Fig. 16).

2.2.2 Effect of Microstructure

One of the factors that can significantly affect high-temperature crack growth is the microstructure. Sadananda and Shahinian³³ demonstrated that modifying the heat treatment for Alloy 718 improved its crack growth resistance by nearly 2 orders of magnitude (see Fig. 17). Gooch et al.³⁴ showed that microstructure has considerable influence on the deformation-controlled crack growth in Cr-Mo-V steel. Given in Fig. 18 are crack-growth rates vs K for various percentages of bainite (percentages of bainite and ferrite control ductility and strength) in the material. As seen in Fig. 18, high strength and low ductility increase the crack growth rates.

The influence of thermal aging on creep-fatigue behavior of both solution-annealed and 20% cold-worked type 316 stainless steel at 593°C has been investigated by Michel and Smith.²⁷ Before testing, specimens were aged for 5000 h at 593°C. This treatment ensures complete precipitation of the $M_{23}C_6$ carbides. Results are presented in Figs. 19 and 20. For the case of continuous cycling, aging causes reduced growth rates in the solution-annealed material. Thermal aging results in bulk hardening of solution-annealed material, and stable low-stress fatigue crack growth rate depends inversely on material yield strength. Conversely, thermal exposure produces bulk softening of cold-worked material, which in turn results in increased crack growth rates. In case of the hold time tests, prior aging totally removes the increase in the crack growth rates observed when hold periods are introduced for solution-annealed material. For 20% cold-worked steel, thermal aging considerably suppresses the dramatic increase in growth rate observed upon the introduction of 1-min hold times.

Fig. 16. Effect of cyclic stress ratio on fatigue-crack growth of type 304 stainless steel at 538°C. (a) Crack-growth rate vs stress-intensity factor range, and (b) crack-growth rate vs effective stress-intensity range. Source: Ref. 32.

Fig. 17. Changes in time-dependent crack growth behavior as a result of modification in microstructure for Alloy 718 in air and vacuum. *Source:* Ref. 33.

James³⁵ conducted extensive studies of the effect of thermal aging on the fatigue crack growth behavior of type 304, 304L, and 316 stainless steels in both annealed and 20% cold-worked conditions. Aging temperatures of 538 and 649°C were used. Exposure times were 1500, 3000, and 6000 h. Material subjected to prior aging generally exhibited lower crack growth rates at elevated temperatures than the unaged material. At room temperature, however, slightly higher crack growth rates were observed for the aged specimens.

Fig. 18. Changes in time-dependent crack growth behavior as a result of modification in microstructure for Cr-Mo-V steel. Source: Ref. 34.

Brinkman et al.^{36,37} used push-pull strain cycling tests with and without tensile hold periods to investigate effects of thermal aging on type 316 stainless steel at 593°C. Results are given in Fig. 21. An increase in tensile hold period causes dramatic reduction in the fatigue life of solution-annealed material. Prior aging for 8760 h at 400°C produces negligible improvement in the hold time life; a preaging treatment of 3096 h at 600°C somewhat enhances endurance. However, an aging treatment of 5040 h at 593°C produces a tenfold increase in fatigue life in the 30-min hold period tests. Similar tests were conducted by Jaske et al.³⁸ at 566 and 649°C on type 316 stainless steel that was solution-annealed and preaged for 1000 h at test temperature. Results indicated that exposure to certain combinations of time and temperature prior to testing can considerably improve the creep-fatigue resistance of type 316 stainless steel at both 593 and 649°C.

Fig. 19. Effect of hold time on fatigue crack propagation rate, da/dN , in air at 593°C for solution annealed, unaged, and thermally aged type 316 stainless steel. Source: Ref. 27.

Fig. 20. Effect of hold time on fatigue crack propagation rate, da/dN , in air at 593°C for 20% cold worked, unaged, and thermally aged type 316 stainless steel. Source: Ref. 27.

Fig. 21. Influence of thermal aging on creep-fatigue endurance of type 316 stainless steel at 593°C. Source: Refs. 36 and 37.

2.2.3 Effects of Temperature and Environment

Surface crack initiation and growth are open to environmental influence; hence, a number of investigators have examined environmental effects. James and Knecht³⁹ conducted fatigue crack-growth tests for type 304 stainless steel at 427 and 537°C in a liquid-sodium environment, as well as in air and in vacuum. In air environment crack growth rates increased with temperature. On the other hand, crack growth rates in vacuum and liquid-sodium environments appeared to be unaffected by the temperature change. Therefore the thermally activated component of crack propagation observed in air was concluded to be an environmental effect. Wood et al.⁴⁰ performed continuous cycling and tension hold period cycling tests on type 316 stainless steel at 625°C in air (aggressive) and helium (inert) environment. Data in Fig. 22 show that the continuous cycling life in air is reduced compared to that in helium. Conversely, the creep-fatigue life under 2-h tensile hold period cycles is virtually independent of environment. Sadananda and Shahinian⁴¹ reported slightly higher crack growth rates in air than in vacuum in type 316 annealed stainless steel at 593°C. For 20% cold-worked material at 593°C, crack growth rates were found to be nearly the same in air and in vacuum under fatigue, creep, as well as combined creep-fatigue conditions. An increase in crack growth rate with temperature was observed in both air and vacuum and was attributed to temperature dependence of the elastic modulus. Zamrik and Zahiri⁴² concluded that temperature had a strong effect on the failure modes of the waspaloy material at 649°C — precipitation of carbide particles caused formation of embrittled regions. Huthmann and Gossmann,⁴³ however, studied fatigue and creep crack growth behavior of the X6CrNi18 steel (type 304 stainless steel) in air and flowing sodium at 823°K and reported no significant influence of the environment.

Fig. 22. Effect of environment on creep-fatigue behavior of type 316 stainless steel at 625°C in air and helium. Source: Ref. 40.

3. THEORETICAL MODELS

3.1 FATIGUE CRACK GROWTH

Micromechanistic interpretation of the fatigue (cycle-dependent) crack growth has been a subject of extensive research. A simple mechanism for crack advance was proposed by Cottrell.⁴⁴ Under plane strain conditions, continuous stable crack growth is produced by the injection of edge dislocations with the accommodating shear displacements at $\pm 45^\circ$ to the crack plane ahead of the crack. While conducting high-strain-fatigue experiments on aluminum, Tomkins and Biggs⁴⁵ observed this process directly on the specimen surface. Tomkins and Biggs also noted the ensuing striation formation. Striations are the regular markings on the fatigue fracture surface, which are generally parallel to the crack front (see Fig. 23). The overall size of striations is considerably less than the grain size. Striations can be viewed as a manifestation of localized crack tip plastic deformation.⁴⁶

Assuming a one-to-one correspondence between the number of cycles and the number of striations, Laird⁴⁷ proposed a plastic blunting model to account for the formation of striations. The mechanism of crack growth by the plastic blunting process is schematically shown in Fig. 24. Tomkins⁴⁸ analyzed the cycle-dependent crack growth using (1) the qualitative features of the plastic blunting model and (2) the Dugdale model for determining the plastic zone size or the crack tip displacement due to plastic flow in each cycle. The following constitutive relation was used

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Fig. 23. Fatigue striations in type 316 stainless steel after low-temperature crack propagation at 1 Hz. Source: Ref. 46.

Fig. 24. The plastic blunting process of fatigue crack growth. (a) Zero load, (b) small tensile load, (c) maximum tensile load, (d) small compressive load, (e) maximum compressive load, and (f) small tensile load. Source: Copyright ASTM. Reprinted with permission.⁴⁷

to describe the cyclic stress-strain response of the material ahead of the crack tip:

$$\Delta\sigma = B (\Delta\epsilon_p)^\beta . \quad (8)$$

The crack growth rate was then expressed in terms of the plastic strain amplitude $\Delta\epsilon_p$ as

$$\frac{da}{dN} = \left(\frac{\pi^2}{B}\right) \left(\frac{B}{2T}\right)^2 \left[1 + \left(\frac{\pi^2}{B}\right) \left(\frac{B}{2T}\right)^2 (\Delta\epsilon_p)^{2\beta} \right] a (\Delta\epsilon_p)^{2\beta+1} , \quad (9)$$

where T is the average tensile stress in the plastic zone ahead of the crack tip and B and β are constants.

Based on the power law and a dislocation stress-velocity equation $V = V_0 \tau^m$ Yokabari et al.⁴⁹ developed the following expression for the crack growth rate:

$$\frac{da}{dN} \propto \left(\frac{\Delta K}{\sigma_{yc}}\right)^\delta \left(\frac{\sigma_{yc}}{E}\right)^\eta , \quad (10)$$

where $\delta = (1+2\beta\eta)(1+\beta)$, $\eta = (m+1)^2/(m+2)$, is the cyclic yield stress of the material.

According to the plastic blunting crack growth mechanism, during the application of a tensile load the crack opens by a certain amount δ (crack tip opening displacement) and advances by an amount da/dN . The difficulty lies in calculating

δ and isolating the amount of crack growth for a given applied stress-strain field. Tomkins⁵⁰ considered a material with the stress-strain relationship given by

$$\epsilon = \epsilon_e + \epsilon_p = \left(\frac{\sigma}{E}\right) + \left(\frac{\sigma}{K}\right)^{\frac{1}{n}}. \quad (11)$$

He developed an approximate general solution for δ in an elastic-plastic strain field:

$$\delta = \left(\frac{\Lambda a}{\beta E}\right) n [\sec(\beta \sigma)] + \Lambda a \epsilon_p \left[\frac{\beta \sigma}{1+n} + \frac{(\beta \sigma)^3}{3(1+3n)} + \dots \right], \quad (12)$$

where $\Lambda = 2$ for plastic relaxation on two 45° planes at the crack tip, $\beta = \pi/2T$, T is the material flow stress equal to the ultimate tensile stress, n is the work hardening exponent, and k is a material constant.

3.2 CREEP CRACK GROWTH

3.2.1 Diffusion Models

Several models have been advanced to predict crack growth rate and life of a component under creep conditions. Diffusion models analyze creep crack growth assuming that it is controlled purely by diffusion of deformation. According to the microstructural studies, crack growth can occur either by volume or by grain boundary diffusion of vacancies to the crack tip.² This results in the deposition of matter coming from the crack tip, which relaxes the stress field of the crack. Increase in the stress field due to the crack length increment compensates for this phenomenon to some extent. Puls et al. originally analyzed crack growth by bulk diffusion,⁵¹ later extending this model to include the growth by grain boundary diffusion.⁵² In Ref. 52 the crack growth rate is assessed by considering the source of vacancies at a distance from the crack tip and computing the total free energy change for the process.

Sadananda⁵³ also analyzed kinetics of crack growth by diffusion, but considered the grain boundary diffusion and both the size effect and inhomogeneity interactions of vacancies with the crack tip stress field. In addition, this analysis also considered the effect of plastic flow that provides additional sinks for vacancies via bulk diffusion. Crack growth is shown to occur if $D_g t > 2D_b r_p$, where D_g and D_b are, respectively, the grain boundary and bulk diffusion coefficients, and t and r_p are the grain boundary thickness and plastic zone size. The diffusional growth mode was shown to occur only in certain temperature ranges.

Vitek⁵⁴ considered the effect of stress relaxation due to the deposition of matter from the crack tip. The following expression for the crack growth was proposed:

$$\frac{da}{dt} = \left(\frac{0.516 D_b t \Omega \mu}{k T \delta^4} \right) \left(\frac{K}{\sigma} \right)^4, \quad (13)$$

where

Ω = atomic volume,
 μ = shear modulus,
 k = Boltzmann's constant,
 t = grain boundary thickness,
 δ = crack tip displacement.

The threshold value K_{min} was defined as

$$K_{min} = 3.07 \sqrt{\frac{\beta_s}{\mu}}. \quad (14)$$

where β_s is the surface energy. If $K > K_{min}$ the steady state conditions dominate. For $K < K_{min}$, the growth rate decreases, reaches a minimum, and increases as soon as steady-state conditions are established. This model is limited by the assumption of constant δ . Moreover, if δ is identified with the crack opening displacement (COD), the crack growth rate becomes unrealistic. Vitek therefore identified the diffusional growth δ with the growth of the microcracks ahead of the main crack, rather than the growth of the main crack itself. Vitek and Wilkinson⁵¹ extended the model given in Ref. 54 by considering the diffusional growth of cavities of cylindrical cracklike shapes in the stress field of the main crack. The results, however, again predict the fourth-power dependence on K , which is rarely found in practice.

3.2.2 Deformation Models

Because the time-dependent crack growth at elevated temperatures is mainly due to creep, attempts have been made to relate crack growth to creep deformation rates. Most of the deformation models assume a certain criterion for crack growth (e.g., critical strain or displacement, or time for fracture of a critical size ligament ahead of the crack tip). Validity and usefulness of such a model then necessarily rest on the validity of the critical criterion selected.²

Barnby⁵⁶ assumed that under steady-state conditions, creep rupture life is inversely proportional to the minimum creep rate. It was also postulated that crack growth rates are proportional to creep rates. An expression for crack growth rate in terms of stress intensity and net section stress was proposed. However, because high creep rates do not always correspond to high crack growth rates (and vice versa), the validity of this model is questionable. Vitek⁵⁷ extended the Dugdale model to include creep by replacing yield stress with the fictitious time-dependent friction stress and assuming a power law for creep. Crack growth occurs, provided a critical value of COD is reached and a constant width of the plastic zone h is maintained. The critical value of the COD for crack extension and the width of the plastic zone h are to be determined experimentally. Ewing⁵⁸ assumed the critical COD and related it to critical creep strain by expressing steady-state creep strain as COD/ h . Primary creep was included by modifying the creep exponent. Thus the

correspondence between the incubation/early crack growth and primary creep, and between steady-state crack growth and secondary creep was established. Experimental data, however, do not support such a direct correspondence between different stages of crack growth and different creep processes.

Floreen and Kane⁵⁹ modified the McClintock and Irwin fracture model by assuming a critical strain at a distance D from the crack tip. Distance D was identified with the grain size. Crack growth rate was assumed to be proportional to strain rate, while strain rate was postulated as an exponential (with the exponent P) function of strain. The crack growth rate was then expressed as

$$\frac{da}{dt} = FG \left[\frac{K^2}{HYED} \right]^P. \quad (15)$$

where F , G , H , and P are material constants, Y is the yield stress, D is the grain size, and E is the Young's modulus. Thus defined, growth rates are inversely related to yield stress. Experimental observations, however, indicate that high yield strength materials are generally more susceptible to crack growth.

Several models postulated a functional form for damage in terms of stress or strain rate and related the fracture process to critical damage accumulation. Janson^{60*} assumed a power law relation for damage rate and stress and developed a crack growth expression, which needs to be solved numerically by an iteration process. Ohji,¹⁶ having assumed a similar functional form for damage rate, proposed a crack growth rate expression in terms of the C^* integral. In this case the crack growth rate is proportional to $(C^*)^{b/(n+1)}$, where b is the exponent relating damage rate to stress, and n is the exponent in the creep power law.

3.2.3 Combined Models

Deformation models recognized creep as a thermally activated process involving either grain boundary or bulk diffusion. Thus, even though the crack growth is not directly attributed to diffusion, diffusion terms are implicit in the crack growth laws. Combined models, in contrast consider both processes as explicitly related to the crack growth process. Beere and Speight^{61*} analyzed the problem of diffusive crack growth in a plastically deforming solid. Drift of vacancies to the crack tip is counterbalanced by the drift of atoms from the tip, which causes displacement of the grain ahead of the crack. Away from the crack tip this displacement is accommodated by dislocation creep. Raj and Baik⁶² postulated crack propagation by coalescence of voids that nucleate and grow ahead of the crack tip by diffusion. Void growth is accommodated by the relaxation of stresses ahead of the crack tip. This model, however, does not consider elastic distortion of the medium ahead of the crack tip, which becomes significant as the small cavities grow into a cracklike defect.

*As cited in Ref. 2.

3.3 CREEP-FATIGUE CRACK GROWTH

Understanding of the creep-fatigue interaction in terms of fundamental processes at the crack tip is very limited.² Separation of the cycle- and time-dependent components responsible for crack growth under creep-fatigue conditions is a difficult task. Several phenomenological models have been proposed to predict creep-fatigue crack growth.

Carden⁶³ used the data obtained by James⁶⁴ on type 304 stainless steel to develop an empirical relation for crack growth:

$$\frac{da}{dN} = A(f,T)(K_e^2 - K_{th}^2)^m, \quad (16)$$

where

$$K_e = K_{max}(1 - R)^n, \quad (17)$$

$$K_{th} = \Delta K_{th}(1 - R)^n. \quad (18)$$

Powers m and n are material constants, and A is an experimentally determined function of temperature and frequency. Döner⁶⁵ modified Carden's model by expressing $A(f,T)$ as:

$$A(f,T) = C_0(T) + C_1(T)\left(\frac{1}{f}\right)^k + C_2(T)\left(\frac{1}{f}\right), \quad (19)$$

where the three terms represent, respectively, cycle-dependent crack growth, environmental effects, and creep crack growth. This model was used to predict frequency dependence of crack growth in HS-188 cobalt alloy. Comparison with the experimental results of Ohmura et al.⁶⁶ indicated that prediction was satisfactory as long as the environmental contribution could be neglected.

Jaske and Begley⁶⁷ expressed the cycle- and time-dependent components in terms of ΔJ and C^* integrals, respectively, as

$$\frac{da}{dN} = C_1(\Delta J)^p, \quad (20)$$

$$\frac{da}{dt} = C_2(C^*)^q. \quad (21)$$

The creep and fatigue load conditions were then combined by employing both Eqs. (20) and (21) to compute the crack growth increment in each individual cycle. The analysis, however, lacked substantial experimental support.

Riedel and Rice⁶⁸ presented the time-dependent stress analysis of a stationary crack in mode I tension. Material behavior was modeled as elastic-nonlinear

viscous, where the nonlinear term described power law creep. Definition of the small-scale yielding represented an important result of this analysis. Under creep conditions small-scale yielding implies that a small "creep zone" is formed around the crack tip, while elastic strains dominate the rest of the material. If the crack growth occurs while the creep zone is still small compared with the crack length and the specimen size, the stress intensity factor K governs crack growth behavior. The following approximate description of the small-scale yielding stress and strain fields was derived:

$$\sigma \propto \left(\frac{K^2}{Er} \right)^{\frac{1}{k+1}}, \quad (22)$$

$$\epsilon \propto \left(\frac{K^2}{Er} \right)^{\frac{1}{k+1}} t^{\frac{1}{k+1}}, \quad (23)$$

where k , r , and t represent the creep exponent, distance from the crack tip, and time, respectively. The small size of the creep zone indicates that the secondary creep deformation is localized. When the secondary creep deformation becomes widespread (the calculated size of the creep zone becomes large compared to that of the specimen), the strain rate at the crack tip is governed by the energy rate line integral C^* :

$$\sigma \propto \left(\frac{C}{r} \right)^{\frac{k}{k+1}}, \quad (24)$$

$$\epsilon \propto \left(\frac{C}{r} \right)^{\frac{k}{k+1}}. \quad (25)$$

Riedel⁶⁰ analyzed the time-dependent crack-tip stress fields in an elastic-nonlinear viscous body for the case of variable load. Analysis was of a continuum mechanics nature; therefore, the micromechanisms of crack growth were ignored. For slow cycling the C^* integral was found to determine the crack tip fields and therefore the crack growth rate. In case of the rapid load variations within otherwise slow cycles, the stress intensity factor ΔK_I was found to determine the severity of the stress peaks that follow rapid load changes. Stress intensity factor ΔK_I was also found to control the time-dependent stressing and straining near the crack tip for the case of rapid cycling. For the case of short periods under constant load following the rapid load application, parameter $C(t)$ was advanced, which represents the amplitude of the near-tip stress field as

$$C(t) = \frac{K_I^2 (1 - \nu^2)}{(k + 1)Er}. \quad (26)$$

To obtain the steady-state condition, $C(t)$ becomes equal to C^* after a sufficiently long time. The time required for the transition from the K_I -controlled to the C^* -

controlled state is given by

$$t = \frac{K_1^2 (1 - \nu^2)}{(k + 1)EC^*} . \quad (27)$$

Based on the stress analysis of Riedel and Rice,⁶⁸ Saxena et al.⁷⁰ advanced a model to represent mathematically as well as to predict the effect of hold time on the fatigue crack growth behavior at elevated temperature. The K^2/t parameter is selected to characterize crack growth rate during the hold time according to

$$\frac{da}{dt} = b \left(\frac{K^2}{t} \right)^r . \quad (28)$$

Then the total crack growth rate per cycle is given by superposition:

$$\left(\frac{da}{dN} \right)_A = \left(\frac{da}{dN} \right)_0 + \int \left(\frac{da}{dt} \right) dt , \quad (29)$$

where the integral is taken between zero and hold time t_h , and $(da/dN)_0$ - the crack growth rate for the same loading/unloading rates with no hold time - is given by

$$\left(\frac{da}{dN} \right)_0 = C(\Delta K)^m . \quad (30)$$

To verify the model, fatigue crack growth rate data with hold times of 0, 5, and 50 s were generated for A470 class 8 steel at 538°C. Additional data obtained by Shahinian and Sadananda⁷¹ on Inconel alloy 718 at 538°C were used to evaluate the model further. It was concluded that the model was capable of accurately representing and predicting time-dependent crack growth caused by creep deformation at the crack tip.

The model presented in Ref. 70 applies to situations where the influence of creep on crack growth rate dominates over the influence of environment. To complement this work, Saxena⁷² proposed a semiempirical model for predicting the influence of frequency and hold time on the high-temperature fatigue crack growth rate behavior of metals in the regime where the influence of environment is more pronounced than the influence of creep deformation. A three-region relationship between crack growth da/dN , at a constant ΔK , and the time period of a cycle, $1/\nu$ (where ν is the frequency), for a given wave shape, is defined (see Fig. 25). Within the cycle-dependent region I, crack growth is approximately independent of the frequency, and the amount of crack extension for constant ΔK depends primarily on the number of cycles. Crack growth in region II consists of a cycle-dependent contribution and a time-dependent one. Region III is characterized by a constant time rate of crack growth; here the amount of crack extension at a constant ΔK depends solely on time. Figure 26 gives a schematic representation of the various zones in the region of a stationary crack loaded in mode I. The model is confined to region II and to the situations where the kinetics of the growth of the environment-

Fig. 25. Schematic representation of high-temperature fatigue crack-growth behavior with frequency. Source: Ref. 72.

affected zone are considerably faster than those of the creep zone. The following expressions are developed for crack growth under triangular and trapezoidal wave forms, respectively:

$$\frac{da}{dN} = C_0(\Delta K)^{m_0} + C_1(\Delta K)^m(v^{-0.5}), \quad (31)$$

and

$$\frac{da}{dN} = \left(\frac{da}{dN}\right)_f + C_2(\Delta K)^m \sqrt{t_h}, \quad (32)$$

where C_0 , C_1 , C_2 , m , and m_0 are constants to be determined empirically; $(da/dN)_f$ is

Fig. 26. Schematic of the various zones in the region of a stationary crack loaded in mode I. D is the diffusion coefficient, σ_{ys} is the yield strength, B and n are constants representing the steady-state creep deformation behavior in the material, $\epsilon = B\sigma^n$. Source: Ref. 72.

the crack growth rate corresponding to the loading and unloading portion of the fatigue cycle; and t_h is the hold time. Fatigue crack growth data on Cr-Mo-V steel at 427°C (Ref. 73), Astroloy at 655°C (Ref. 74), and Inconel 718 at 650°C (Ref. 75) were used to validate the model. Results were satisfactory. The model was also shown to predict accurately the fatigue crack growth behavior at slow frequencies from the data obtained at considerably faster frequencies in each of these materials.

4. CRITICAL CRACK GROWTH

Following the subcritical crack growth discussed above, cracks may reach the size at which time-independent "critical" crack growth will commence. Depending on the material and temperature, critical crack growth may be brittle or ductile. Brittle failure mode is typical for many ferritic steels at relatively low temperatures. Cleavage is the principal crack growth mechanism in this case. Plastic deformation accompanying such failure is usually negligibly small, therefore linear-elastic fracture mechanics can be used to evaluate the crack growth. On the other hand, such highly ductile materials as austenitic stainless steels do not normally exhibit brittle fracture. The failure mode in this case is ductile, with the crack growth mechanism being a successive coalition of voids. Within the last two decades, design of light-water reactor (LWR) pressure vessels and piping motivated development of analytical methods for evaluating ductile fracture conditions. Most of such methods are based on fracture mechanics and/or simplified limit load analysis.

4.1 FRACTURE MECHANICS APPROACH

Extensive plastic deformation preceding the initiation of ductile crack growth makes the application of linear-elastic fracture mechanics impractical. As an alternative, time-independent nonlinear fracture mechanics has been developed. Among various parameters the J integral, defined by Rice⁷ for nonlinear-elastic (deformation theory plastic) materials, has received the broadest attention.

Under increasing applied load the initiation of ductile crack growth is generally followed by stable ductile tearing. At that stage a number of materials, including austenitic stainless steels, exhibit an increase in resistance. Because this fact can be of considerable importance in the assessment of the final failure load, analytical methods for representing stable ductile crack growth have been studied with some effort. The use of J integral in modeling the initiation of ductile tearing was investigated by Begley and Landes.⁸ A technique for obtaining J_{IC} , the critical value of the J integral, which is considered a material property at a given temperature, is documented in the ASTM standards.⁷⁶ The value of J integral is established from the load-displacement curve, where specimens are either of the CT or the three-point bend type. Size restriction is specified in the ASTM standard in to ensure plane strain conditions and the dominance of the Hutchinson-Rice-Rosengren singular stress-strain field (intensity of which is given by J) at the crack tip. The use of the J integral resistance curve (the graph of J vs crack extension Δa) for the assessment of crack growth behavior was proposed by Paris et al.⁷⁷ Like the critical value J_{IC} , the J integral resistance curve is viewed as a temperature-dependent material property.

After a certain amount of stable crack growth the load carrying capacity of the structure starts to decrease. At this point the dead-load type loading will produce unstable fracture. Under displacement-controlled loading, in contrast, stable crack growth may continue beyond the maximum load point. Instability

under the displacement-controlled loading will occur, when the elastic shrinkage of the structure due to the load drop surpasses the elongation due to crack growth. This observation served as a basis for the tearing instability criterion presented by Paris et al.⁷⁷ once the applied tearing modulus T_{app} is greater than the material tearing modulus T_{mat} , the tearing instability sets in. The material and applied tearing moduli are defined as

$$T_{mat} = \left(\frac{E}{\sigma_f} \right) \left(\frac{dJ_R}{da} \right), \quad (33)$$

$$T_{app} = \frac{E}{\sigma_f} \cdot \frac{dJ_A}{da}, \quad (34)$$

respectively. Here E is the Young's modulus; σ_f is the flow stress; J_R is the value of J corresponding to the ductile crack length Δa on the resistance curve; and J_A is the value of J due to the external loading.

The main conceptual problem of this approach is the applicability of J integral to the stable crack growth. In a strict sense, J integral loses meaning due to the inconsistency between the actual material behavior and assumptions made in the definition of the J integral. Crack growth in the plastically deformed material causes local elastic unloading as well as significantly nonproportional loading near the crack tip. Material behavior under these conditions cannot be described by the simple nonlinear-elastic theory. Hutchinson and Paris⁷⁸ studied conditions under which application of J integral to crack growth in elastic-plastic materials is justified. The following restriction was derived:

$$\omega = \frac{b}{J_R} \cdot \frac{dJ_R}{da} \gg 1, \quad (35)$$

$$\Delta a \ll b,$$

where b is the ligament length of the specimen. The ASTM E24.08 subcommittee is currently developing a standard procedure for establishing J integral resistance curve that takes these restrictions into account. However, these conditions can hardly be met in a number of practical problems, including the large crack growth. A modification of J integral, called J_M , was proposed by Ernst⁷⁹ for cases where these conditions are violated.

Another difficulty lies in the complexity of the J integral evaluation procedures. Because considerations of nonlinear material behavior are essential in J integral applications, the evaluation method necessarily becomes much more complicated than in the case of the elastic stress intensity factor. The η -factor method, based on the potential energy release rate interpretation of the J integral, was discussed in Refs. 80 and 81. This method is helpful in determining J from the experimental load-displacement curve and generating the J integral resistance

curve. But it cannot be used in predictive analysis. Fracture behavior can be predicted with the detailed nonlinear finite element analysis. This, however, requires a fair amount of effort, expertise, and computation time. As an alternative, use of normalized nonlinear finite element solutions for power law hardening materials was proposed by Kumar et al.⁸² In this case, J integral solutions and representative deformation parameters are tabulated for various geometries and types of loading in a handbook format.^{82,83} Wilkowski et al.⁸⁴ pointed out the difficulty of approximating the uniaxial stress-strain relation with the power law. A modification of this approach was suggested by Zahoor.⁸⁵ Expressions based on the limit load analysis were developed for the pipe with the circumferential through-wall crack by Wilkowski et al.⁸⁶ Within the framework of an extensive research program on fracture of flawed piping in nuclear applications, Wilkowski et al.⁸⁶ presented comparison of these methods. This study demonstrated a considerable dependence of the predicted results on the analytical method used (see Fig. 27).

To overcome the difficulty of approximating the uniaxial stress-strain relation by a power law, Ainsworth⁸⁷ developed approximate reference stress

Fig. 27. Predicted load vs load-line displacement record (using J_M - R curve with fit of entire σ - ϵ curve for determining Ramberg-Osgood coefficients) compared with experimental data. Source: Ref. 86.

J -estimation techniques that allow any stress-strain relation to be used. The reference stress approximation has been compared to the handbook solutions for power-law materials and demonstrated to be accurate for the wide range of geometries and loadings in the handbook.⁸⁸ The approximation may be calculated provided stress intensity factor and limit load solutions are available. It has been incorporated into the Central Electricity Generating Board's (CEGB's) R6 approach,⁸⁹ which has been the subject of experimental validation⁹⁰ and for which a compendium of limit load solutions for cracked geometries has been produced.⁹¹

A number of researchers addressed the applicability of the J integral to surface crack problems. Attempts were made to construct the fully plastic solutions for surface cracks in plates and cylinders.⁹² Several experimental and numerical studies focused on evaluating the applicability of the J integral approach to surface crack extension in the ductile mode.^{93,94}

In an attempt to overcome the fundamental problem with applying J to crack growth analysis, other parameters were proposed. Kishimoto et al.⁹⁵ advanced the $\hat{J}$ integral, which consists of both path and area integral terms. In the event of crack growth in elastic-plastic materials, this parameter remains path-independent. Within this theory thermal and plastic strains can be treated in the same manner. Atluri⁹⁶ proposed $(\Delta T)c$, a path-independent parameter for elastic-plastic materials. In some ways $(\Delta T)c$ is similar to $\hat{J}$. Crack tip opening angle (CTOA) has also been considered as a possible controlling parameter for the ductile crack growth.⁸⁶ The $\hat{J}$, $(\Delta T)c$, and the CTOA share one advantageous characteristic: rather than increase with the crack growth like the J integral, they remain virtually constant. However, a detailed analysis is required to determine these parameters, and this is somewhat of a shortcoming.

For structures operating at high temperatures, effects on the ductile fracture resistance of prior creep damage, creep-fatigue damage, and prior loading in general, may be important and should be considered. It is known, that thermal aging may cause a considerable decrease in tearing resistance⁹⁷ in cast stainless steel. Hence it would be beneficial to assess the effect of thermal aging on the fracture resistance, as is done for cast stainless steel.⁹⁸

4.2 NET SECTION COLLAPSE CRITERION

Net section collapse criterion is based on the limit load concept for elastic-perfectly-plastic material. The failure loads are derived from the stress distribution in the crack plane, which in turn is obtained from the limit load analysis and equilibrium equations. To account for strain hardening, the "equivalent" flow stress, defined as an average of the yield or proof stress and the ultimate strength, is introduced.

The effectiveness of this criterion was thoroughly investigated for austenitic stainless and carbon steel pipes with circumferential cracks at the operating temperatures of the LWR plants. Experimentally obtained failure loads were compared to those predicted for pipes containing through-wall or surface cracks and subjected to bending loads. Predictions were found to be satisfactory for pipes of relatively small diameter and large fracture toughness, where the failure occurred after the general yielding of the cracked section. In cases where the failure took place before general yielding, failure loads were significantly overpredicted. Wilkowski et al.⁸⁶ introduced a dimensionless plastic zone parameter based on the concept of Irwin's plasticity correction. A reasonably good correlation between this parameter and the ratio of experimental to predicted failure loads was obtained (see Figs. 28 and 29).

While this approach predicts the collapse load of the structure, it does not address the local plastic ligament instabilities, which may develop in the presence of deep surface cracks. Such local instability is likely to occur before the global limit load is reached and would normally cause a limited crack penetration through the wall rather than total failure. In the creep temperature range large deformations and consequently a ligament instability may develop due to creep. Modeling of local instabilities is particularly important in the leak-before-break assessment.

Fig. 28. Ratio of experimental load to predicted net-section-collapse load as function of dimensionless plastic-zone parameter. The flow stress definition used in development of this figure is average of yield and ultimate strengths of material. (Data points for crack initiation only.) Source: Ref. 86.

Fig. 29. Ratio of maximum experimental stress for surface cracks subjected to axial membrane loading to net-section-collapse stress as a function of a dimensionless parameter relating the plastic-zone size to the remaining legament to neutral axis. Source: Ref. 86.

5. RELATED TOPICS

5.1 DEFECT MODELING AND ENGINEERING FLAW ASSESSMENT PROCEDURES

Guidance on elevated-temperature modeling of defects and the interactions between neighboring defects or between defects and structural boundaries is currently given in Sec. XI of the *ASME Boiler and Pressure Vessel Code*. The procedures were originally developed specifically for LWR systems and were therefore applicable to ferritic steels in aqueous environments at temperatures below the creep regime. A modified flaw size evaluation procedure and acceptance criteria were recently developed for stainless steel piping and its application in boiling water reactors. Maximum allowable flaw sizes were first specified for wrought and cast stainless steel piping in the "Winter 1983 Addendum" of the *ASME Code*.⁹⁹ Because research demonstrated that flawed piping failed by plastic collapse, the allowable flaw sizes were developed based on limit load analysis.¹⁰⁰ Crack geometry was related to loads and material properties by simple tension and bending equilibrium equations. The allowable flaw depth was specified as a function of applied load and measured flaw length. Subsequent experimental data indicated that pipes with flaws in welds could have lower load-carrying capacity than that predicted for the wrought material. Such reduced load-carrying capacity was estimated by determining the ratio of the load at crack instability (obtained from elastic-plastic fracture mechanics analysis) to collapse load (obtained from limit load analysis) as a function of crack length and pipe diameter. As a result, two new flaw size tables for evaluating flux weldments for (1) normal and upset conditions and (2) emergency and faulted conditions, were added to the *ASME Code* in 1985.

The flaw assessment methodology given in the *ASME Code* is not intended for creep-regime temperatures and does not address such phenomena as thermal creep.¹⁰¹ However, elements of an assessment procedure of defects in structures operating in the creep regime are fairly similar to those required below the creep range.¹⁰² The first step is to characterize the shape of an initial defect, notch acuity, and the spatial distribution of interacting flaws. The second step is to assess the time needed to initiate a propagating crack from the initial defect and the subsequent crack growth rates. The final step is to determine the point at which structural failure will occur due either to creep rupture mechanisms or to fast fracture.

Ainsworth and Goodall¹⁰³ discussed predictive and statistical approaches to the determination of crack initiation and growth in structures subjected to a steady load. These authors considered creep ductile response as one extreme of behavior. In this case the life of the component is governed by continuum damage mechanics. An approximate life assessment can also be based on a reference stress determined from limit analysis. The presence of the initial defect is accounted for by considering its effect on the limit load. The limit analysis then provides the load-bearing capacity of a component. For a particular defect category, it is possible to establish the maximum allowable defect size, which would not grow beyond an allowable fraction of the critical size during service or between inspection periods. Thus, according to

Ref. 103, an allowable defect size may be determined during the fabrication stage, and repair criteria can be established before in-service inspection. The approach described in Ref. 103 has recently been produced as a more formalized assessment procedure for cracked components under steady load at elevated temperature.¹⁰⁴ An associated worked example illustrates application of the procedure to a circumferential crack in a pressure vessel.¹⁰⁵ A feature of the approach is that reference stress methods, similar to those used for J -estimation in R6, are provided for calculating C^* for defects in structural geometries. However, when the structure is subjected to significant cyclic loading, the problem becomes complicated and the solution uncertain. A satisfactory approach requires the development of a suitable correlating parameter for the initiation and growth of cracks under cyclic loading. A restricted approach dealing with crack growth, calculated as in Eqs. (20) and (21) by addition of cycle- and time-dependent components, has recently been produced¹⁰⁶ and compared with experimental data on austenitic specimens.^{107,108}

In 1976 Harrison et al.¹⁰⁹ developed the defect assessment procedure commonly referred to as R6 and introduced the concept of the failure assessment diagram. In the R6 method the severity of loading on a structure with a flaw of a certain size is represented in terms of the material yield and material toughness after some stable crack growth. Two criteria limit the load-carrying capacity of a flawed structure: (1) the linear-elastic stress intensity factor must not exceed the fracture toughness and (2) the applied load must not exceed the plastic collapse load of the flawed structure.¹¹⁰ Graphic representation of these criteria constitutes the failure assessment diagram (see Fig. 30). A second revision of the R6 procedure¹¹¹

Fig. 30. Typical R6 failure assessment diagram. Source: Ref. 89.

presented a detailed method for including stable tearing in the CEGB's defect assessment procedure. Recently Ainsworth et al.⁸⁹ described a third revision of the R6 defect assessment procedure. In this revision a major change has been made in the shape of the failure assessment diagram to account for strain hardening effects. Revision 3 also gives a simplified procedure to facilitate the inclusion of stable growth in defect assessments.

The Engineering Critical Assessment (ECA) method was proposed by the British Standards Institution¹¹² for evaluating the fitness for use of reactor components containing real or hypothetical flaws. The ECA procedure provides guidance for modeling defects, analyzing slow crack growth due to cyclic fatigue loading, and defining a single crack-induced failure mode. Both linear-elastic fracture mechanics and elastic-plastic fracture mechanics serve as bases for the ECA. Within this method a concept of a tolerable effective flaw size a_m is introduced. In general, the effective size of a real defect depends on its size, shape, location in the structure, and orientation with respect to both the structure and applied loading.¹ These factors are incorporated in the ECA concept of effective flaw size, which is defined as the length of a straight, through-thickness crack in an infinitely wide plate, and is expressed as

$$a_m = C \delta_c \frac{E}{\sigma_f} . \quad (36)$$

Here C is a loading parameter, E is elastic modulus, σ_f is material yield strength, and δ_c is the critical COD corresponding to the onset of crack extension in a monotonically loaded precracked specimen. The toughness measure δ_c can be related to the J -integral toughness J_{IC} by

$$\delta_c = \frac{J_{IC}}{\alpha \sigma_f} , \quad (37)$$

where α is a geometric constraint factor of order unity, and σ_f is a flow stress characteristic of the material. The ECA method is generally suitable for elevated-temperature nuclear applications; however, its use is prohibited when thermal creep is significant.

An analytical technique for estimating the influence of defects on structural integrity of reactor coolant piping was developed by Maxey, Kiefner et al.^{113,114} Reuter and Place¹¹⁵ presented an experimental program, designed to study behavior of type 304 stainless steel in the presence of defects using surface-flawed plate specimens and pressurized pipe specimens with internal and external surface defects. Results of these experiments were used to evaluate models proposed in Refs. 114, and 115. Kitagawa et al.¹¹⁶ discussed the fracture mechanics approach to the assessment of structural integrity of the loop type fast breeder reactor (FBR) when the existence of a certain defect was conservatively postulated. Fatigue crack growth from the initial semielliptical surface flaw was predicted numerically and was found to be insignificant.

Konish¹¹⁷ presented a set of guidelines for evaluation of flawed components in elevated-temperature nuclear service. Konish defined such evaluation as the process consisting of three major phases: modeling, evaluation of incremental crack growth, and definition of the failure modes of interest and the associated factors of safety. The modeling phase addresses structure and defect geometry, component loading and resulting local stresses, and pertinent aspects of material behavior. Defects are modelled as one- or two-dimensional, perfectly sharp, planar cracks having continuous geometrically regular boundaries (crack fronts) with no vertices. Elliptical crack front models, including elliptical embedded cracks, semielliptical surface cracks, and quarter-elliptical corner cracks, are chosen. A minimum depth dimension and a minimum surface length dimension are prescribed.

Bros et al.¹¹⁸ presented the methodology for determination of the initial critical crack size developed by Tecnatom in Spain. This approach implies that the appearance of the crack or defect is to be anticipated. Thus, the critical zones in the component must be selected, and a parametric study carried out to determine different critical crack sizes corresponding to the most representative sections of these critical zones. For each of these representative sections, surface and subsurface flaws as well as circumferentially and longitudinally oriented defects are assumed. Once the critical crack sizes are established, dimensions of a crack that could grow to a critical size in a given period of time can be calculated. Such dimensions are then defined as the initial critical crack size. Determination of the initial critical sizes is based on the smallest of the critical sizes and a safety coefficient (0.1 for normal conditions and 0.5 for emergency and/or accident conditions). Critical zones in a reactor vessel include the vessel beltline region surrounding the core, where neutron bombardment has a degrading effect on the mechanical properties of the material, and areas with structural discontinuities due to changing thickness.

5.2 PROBABILISTIC APPROACH

Recently the probabilistic approach began to be widely utilized in assessment of structural integrity of various components. In the nuclear industry, possibilities of guillotine-type break and leakage failure were estimated for the piping systems of LWR plants.¹¹⁹⁻¹²¹ Large-scale computer programs based on the Monte Carlo simulation method have been developed for that purpose. The probability of unstable break of the pressure vessel due to pressurized thermal shock was estimated using a similar program.¹²² Some applications to LMR technology also began to appear in the literature.^{123,124}

As a typical example, the flow diagram of the computer program PRAISE (Piping Reliability Assessment Including Seismic Event) is shown in Fig. 31. Here the input data necessary for fracture mechanics evaluation are treated as distributed variables obeying some probability functions. Application of this program or an equivalent procedure requires information on the various kinds of input data, such as initial crack size distribution, nondetection probability in preservice and inservice inspections, stress conditions, crack growth characteristics, failure conditions, and leak detection probability.

Fig. 31. Flow diagram of PRAISE. Source: Ref. 120.

Among the input data, determination of the initial crack size distribution presents the largest difficulty. This can be factored into the density of the cracks, distribution of crack depth, and distribution of the aspect ratio (ratio of width to depth of the crack). Several functional forms, including exponential, normal, and lognormal functions based on limited available data, have been used for these distributions in the past studies. Results of the assessment show that the choice of function has a fairly significant influence on the final values of failure probability (see Fig. 32). This is largely due to the uncertainty at the tails of the distributions used in analysis (i.e., cracks of large depth and/or large aspect ratio). Because

Fig. 32. (a) Probability distribution vs aspect ratio, and (b) break probability vs service time evaluated for pipe elbow of a German Fast Breeder Reactor. Source: Ref. 124.

amounts of data samplings needed for the justification of distributions are prohibitively large, this problem will continue to complicate quantitative evaluation of the failure probability. However, such an approach can be useful in quantitative assessment of the influence of various factors on the final failure. For example, the effects of in-service inspection, improvement of stress conditions, and difference in material properties on the failure probability can be assessed in a quantitative way (see Fig. 33). The influence of the choice of fracture mechanics models (e.g., inclusion or exclusion of a creep crack growth mechanism) can be also assessed quantitatively.

Fig. 33. Probabilities of leak and break for different inspection schemes evaluated for PWR primary coolant piping. (a) Leak probability, and (b) break probability. Reprinted from: Wittman, Folker, Eds., *Structural Mechanics in Reactor Technology - Transactions of the 9th International Conference on Structural Mechanics in Reactor Technology, Lausanne, 17-21 August 1987*. Volume M. 1987. (All 14 volumes, 6800 pp., Hfl.895/US\$450.00/£268.50. Separate volumes, Hfl.125/US\$63.00/£37.50 each). A. A. Balkema, P.O. Box 1675, Rotterdam, Netherlands/ A. A. Balkema, Old Post Road, Brookfield, Vermont 05036.¹⁴²

As regards the fracture mechanics models that address both subcritical crack growth and final failure conditions, a probability assessment procedure requires useful simplified methods, detailed calculation being difficult to implement. Due to lack of more sophisticated methods, past studies employed relatively crude procedures. For example, in many cases effective stress intensity factor range was used to estimate the crack growth due to cyclic loading excluding the nonlinear effect. A simple plastic collapse criterion was used as a condition for final failure. However, recent work of Matsumoto et al.¹²³ on the LMR elevated-temperature piping uses parameter C_I to treat the subcritical creep crack growth and the tearing instability criterion to establish the final failure condition. Because the choice of the fracture mechanics model considerably affects the results,¹²³ the best available method should be continuously incorporated into the assessment procedure.

5.3 NDE TECHNIQUES

To evaluate the fitness of a flawed component for service, the defect must be detected and its size, shape, location, and orientation defined.¹ Should a component be found suitable for continued use, monitoring of material degradation becomes an important aspect of remaining life assessment and must be performed either continuously during service or intermittently during outages.

The most widely used approach of directly measuring material degradation is in-service nondestructive inspection for the presence of cracks and flaws.¹²⁵ NDE methods give rise to a somewhat different approach to evaluation of structural integrity of a component than a direct analysis of crack behavior. At the beginning of the service life, the component is examined with one or more NDE techniques. Even if no cracks are detected, a limit on the largest possible defect can be established based on the sensitivity of the NDE methods. It is then possible to estimate conservatively the stable crack growth of a hypothetical defect during service life. After some period of service, NDE can be conducted again. At this time the crack may have stably grown to a detectable size. Then the fitness of the component for use can be explicitly evaluated. If no cracks are detected, a limit on a possible defect is again established, and the next in-service inspection scheduled.

Some cracks can be detected, defined, and monitored by direct visual observation. Penetrant and magnetic particle inspection methods provide enhanced visual indications of cracks and flaws. However, it is difficult to estimate the size of defects through visual inspection. Moreover, these methods are not sensitive to subsurface flaws. Many shortcomings of the visual examination can be overcome with the indirect techniques, like ultrasonic, radiographic, and eddy-current methods. These techniques provide volumetric indications of defects. However, information that they provide on size and distribution of defects is often inadequate for fracture mechanics analysis. Sensitivity and accuracy of the NDE techniques also have a significant influence on the fitness-for-service or remaining life assessment.

Among the present procedures for nondestructive testing during fabrication and operation, some need significant improvements for safety reasons. However, in many more areas improvements are needed for economic benefit and technical feasibility.¹²⁶ For instance:

- (1) Radiography is not sufficiently developed for detection and sizing of cracklike defects, especially in heavy structures.
- (2) Ultrasonic testing of austenitic weldments, forgings, and castings is complicated by microstructure.
- (3) Evaluation from the outside of cracks located close to the inner surface is a difficult NDE problem, especially for austenitic components and bimetallic weldments.

Crutzen¹²⁷ summarized the results of comprehensive round-robin ultrasonic testing recently conducted in Europe. Large pressure vessel steel preflawed plates were subjected to ultrasonic examinations by 50 groups of researchers in 15 different countries. While the results of this program, called PISC-II, provide a positive validation of several defect detection procedures, they also demonstrate the need for improvement of the defect sizing capability. Based on the test results, modification of existing design codes, such as the *ASME Boiler and Pressure Vessel Code*, Sec. XI, is suggested. Recommendations for future work included development of expert systems that could replace qualified technicians currently needed for signal analysis. Several remaining questions will be addressed in the PISC-III program.

Schmitz and Höller¹²⁸ commented on the following areas of special emphasis in NDE research and development in West Germany: ultrasonic testing of austenitic steels, detectability and interpretation in ultrasonic testing, electromagnetic ultrasonic transducers, interpretation in acoustic emission testing, and validation of NDT systems. Detectability is generally defined with respect to signals. A signal is detectable when its amplitude or intensity is significantly above the noise level caused by instruments, material, etc. In ultrasonics, detectability is defined by echo amplitudes, which do not provide information on the geometry of defects. A few of the signals related to geometry are time-of-flight, phase, and spectra in the time and frequency domain. Evaluation of these signals considerably improves detectability. Automated systems recently developed in West Germany for flaw detection and interpretation by ultrasonic techniques include: Phased Arrays, Interactive Reconstruction for Tandem and Pulse-Echo (ALOK), Holography, and Synthetic Aperture Focusing Techniques (SAFT).¹²⁸⁻¹³⁰

Okamoto et al.¹³¹ discussed the ongoing development in Japan of advanced NDE methods involving supersonic wave holography, electromagnetic supersonic wave, as well as the software needed for processing the NDE results. The expert system called Defect Identification Program (DIP) is among the results of the current effort. Designed to replace engineering experts during the in-service inspection, DIP is a software package that has the following features:

1. automatic interface for NDE information;
2. relational data base for storing design, manufacturing and operational data ready for access;
3. knowledge base for storing the experts' knowledge;
4. coded logic for inferring the cause and kind of defect and for assessing the structural integrity for future operation.

The system will receive NDE information from conventional ultrasonic testing and from more advanced NDE techniques that are currently being developed. In addition to locating the defect, NDE should provide a fairly distinct profile and, possibly, a visual image of it. DIP is designed to handle primarily the stress corrosion cracks and fatigue cracks growing from defects of other kinds. Once the cause and type of defect have been established, the system will assess it by referring to Sec. XI of the *ASME Boiler and Pressure Vessel Code* for allowable flaw sizes. In the future,

capabilities of the DIP will be extended to predicting behavior of the flawed structure based on the fracture mechanics analysis.

Inspection Level Evaluation Program, presented by Hayashi et al.,¹³² is another method for predicting defect initiation probability, recently developed in Japan. This method focuses on stress corrosion cracking (SCC), corrosion fatigue and thermal fatigue. A database of SCC, corrosion and thermal fatigue, and other factors affecting crack initiation is prepared. The magnitude of crack initiation probability is approximated by several types of functions. A database of the production history (including design conditions, material properties, manufacturing and heat treatments) and operational history of the structure is also established in the assessment program. Based on the approximate functions and operational history, the probability of crack initiation can be predicted, which in turn determines inspection levels. Such an approach helps to rationalize and improve inspection. The in-service inspection level can be determined more accurately by combining the Inspection Level Evaluation Program with the DIP expert system. The combined program is called the Assessment Program for In-Service Inspection Levels.¹³² Experimental data confirm the accuracy of this program. Application of the program to the actual operational plants is believed to result in a reduction of inspection time and improvement of inspection reliability.

Among the current NDE techniques are the conventional low-frequency eddy current and the microwave eddy current used with the ferromagnetic resonance (FMR) probe. A detailed description of the theory and operation of the FMR probe was given by Auld et al. in Refs. 133-135. Because of its high operating frequency, the probe induces eddy currents in a skin depth δ , which is on the order of several microns in aluminum. For high ratios of flaw depth a to δ , the probe is quite sensitive to crack mouth opening displacement (CMOD). Conventional, lower frequency eddy-current probes, with larger δ and thus lower a/δ , do not exhibit such sensitivity. Resch et al.¹³⁶ demonstrated the ability of the FMR probe to detect fatigue microcracks. Nelson and Auld¹³⁷ summarized preliminary experimental results on the use of the FMR probe for crack detection and for monitoring COD. Monitoring COD behavior of flaws was shown to provide important information for estimating the depth of surface flaws and the crack opening stresses.

Carre and Vérité¹³⁸ reported on the application of a three-dimensional computer code to optimization of the eddy-current probe used for the NDE of steam generator piping. The probe is constructed from two magnetic cores, a discoid, six poles on each core, and a coil on each pole. The poles create a radial induction field. The probe, which moves along the axis of the pipe without rotating, should detect longitudinal as well as circumferential flaws. A numerical procedure is introduced to simulate control signals created by the movement of the probe, and thus to optimize the probe design parameters (namely, frequency, number of coils on a core, distance between the two cores, the width of a core, the distance between two successive poles). It is then demonstrated that an eddy-current probe design can be optimized with the help of a three-dimensional computer code.

The electric potential drop method has been considered as one of the possible NDE techniques for detecting crack location and size. Ohtani et al.⁹⁸ investigated applicability of this method to the measurement of crack growth as well as crack initiation at elevated temperatures. Ohtani demonstrated that creep crack growth in low-ductility materials could be measured fairly accurately with the potential drop method. In the case of highly ductile materials, such as stainless steels, creep crack growth is accompanied by large creep deformations, which cause changes in microstructure and macroscopic reduction in the cross-sectional area of the specimen. This leads to an increase in the electric resistance of the specimen. However, in elevated-temperature low-cycle fatigue tests on type 304 stainless steel, both crack length and crack initiation at the notch root could be measured with reasonable accuracy. Ohji and his coworker^{139,140} developed the so-called electric-potential CT method for quantitatively detecting location and size of cracks. To identify crack location and size, measured distribution of electric potential over the surface of a cracked body is used as an input for an inverse numerical analysis. Boundary element analyses are carried out for various assumed crack locations, the most probable location is determined by the least residual method. Effectiveness of this approach was demonstrated for various types of two-dimensional cracks.

An application of NDE in safe-life evaluation of nuclear pumps was presented by Raymond.¹⁴¹ In this case, a postfabrication NDE revealed the existence of linear surface indications on the shroud section of an intermediate cover of a pump. Further NDE revealed the cause to be hot tears, which were overlooked during manufacturing and were not due to any time delay cracking phenomenon. A study was performed to characterize a CA-6NM casting alloy. Test specimens were machined from an actual intermediate pump cover, and the material properties were measured within the framework of fracture mechanics. Test data were used to perform a safe-life cycle analysis. Conservative analysis that took into account the worst-case operating conditions was also carried out. Results indicated that the safety of the pump was not in jeopardy, primarily because the low operating stresses were below the threshold operating fatigue stress intensity range.

6. CONCLUDING REMARKS

A brief review of the high-temperature crack growth behavior under cyclic, static, and combined loading was conducted. Cycle- and time-dependent crack growth was assessed from both materials and engineering fracture mechanics standpoints. Effects of several variables, such as loading frequency, wave shape, hold time, environment, temperature, and microstructure, on crack growth at elevated temperatures were discussed.

A number of fracture mechanics parameters that can uniquely describe crack growth under given loading conditions were considered. The linear-elastic parameter K appears to be useful for correlating purely cycle-dependent crack growth under small scale yielding conditions. In case of time-dependent crack growth, K may be used as a controlling parameter only if nonlinearity is limited to the immediate vicinity of the crack tip. An elastic-plastic parameter J , though originally developed for monotonic loading, is shown to correlate fatigue crack growth for elastic-plastic conditions as well. A growing body of experimental evidence demonstrates that the energy rate integral C^* can be considered a controlling parameter for creep crack growth. As pointed out in Ref. 2, experimental da/dt vs C^* data for a large number of engineering alloys that show crack growth by creep fall on a relatively narrow band. It is possible that a unique da/dt vs C^* relation exists for creep crack growth independent of material and temperature. However, the line integral C^* , which is an adaptation of the J integral involving substitution of strain rates and displacement rates for strains and displacements, is limited to time-dependent crack growth in the steady-state regime. For the case of non-steady-state conditions, modifications C^*_h and C_t were discussed. Several theoretical models for predicting high-temperature crack growth were surveyed. Analytical methods advanced by Riedel and models proposed by Saxena appear to be most promising.

While the effectiveness of the fracture mechanics parameters has been demonstrated in laboratory tests, analytical techniques for evaluating such parameters need to be further developed. This task should include development of both rigorous and simplified methods, and is likely to require considerable effort. In addition to that, validation of the fracture mechanics parameters for more realistic loading conditions, such as long-time hold periods and/or temperature variations, is important for practical application.

Existing methods for modeling ductile tearing phenomenon were briefly surveyed. While the J integral is widely used, several problems regarding its applicability to ductile crack growth analysis still remain. The net section collapse criterion, a conceptually simpler method, predicts global failure loads reasonably well, but does not address local instabilities. Analytical techniques for treating surface cracks, based either on the J integral approach or the ligament instability concept, need to be further developed.

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