CREEP CRACK GROWTH IN 2219-T851 ALUMINUM ALLOY: APPLICABILITY OF FRACTURE MECHANICS CONCEPTS

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ABSTRACT

Creep crack growth rate data measured in 2219-T851 aluminum alloy at 175°C in air are reported. 2219-T851 is found to behave as a typical creep brittle material in agreement with results derived from concepts of fracture mechanics of creeping solids. A unique correlation is shown to exist between the creep crack growth rates (da/dt) and the stress intensity factor for simple K-histories in a regime of quasi-steady or steady state creep crack growth under highly constrained conditions, i.e., ideally, under plane strain conditions. There is no correlation between da/dt and the net section stress, the nominal stress, the reference stress nor the C*-integral.

KEYWORDS

Creep Crack Growth; Fracture Mechanics; Stress Intensity Factor; J-Integral; C*-Integral; Net Section Stress; Nominal Stress; Reference Stress; Crack Growth Rate Correlations; Aluminum Alloys; 2219 Aluminum Alloy.

INTRODUCTION

Single macroscopic cracks can initiate and propagate in metallic parts at high temperature under the influence of creep damage, fatigue damage and/or environment induced damage. The need to develop reliable methods to estimate the remaining life of high temperature components containing cracks has motivated extensive research programs on creep crack propagation. Creep crack growth (CCG) i.e., the propagation of a single macroscopic crack under a sustained load at temperatures well within the creep regime (i.e., $T/T_m > 0.4$) has been studied in several structural alloys. Most materials susceptible to creep crack growth can be said to be either creep brittle or creep ductile. Creep brittle materials fail by creep crack growth with almost no general creep deformation, while creep ductile materials fail by creep crack growth with extensive general creep deformation, even after initial small scale yielding loading conditions. The choice of a correlating parameter for creep crack growth is often guided by the creep brittle or creep ductile characteristic of a given material. The elastic stress intensity factor K and the C*-integral, which describe the time - dependent near tip stress - strain fields around the tip of a sharp crack for short times and long times, are often used for creep brittle and creep ductile materials respectively. For extremely creep ductile materials in which the fracture process zone spreads across the unbroken ligament ahead of the crack average, stresses and displacement rates which describe the global specimen behavior have also been proposed.

The applicability of fracture mechanics concepts to creep crack growth in 2219-T851 aluminum alloy at 175°C in air is discussed in the present paper.

I) FRACTURE MECHANICS OF CREEPING SOLIDS

The stresses around the tip of a Mode I stationary sharp crack in a creeping solid can be calculated in an approximate way as a function of time and distance from the crack tip (1-4). In such a solid, the total strains can be assumed to be the sum of time-independent elastic and plastic strains which develop instantaneously upon loading, and of time-dependent creep strains which accumulate as time increases.

Under small scale yielding loading conditions the stresses in the elastic region where the elastic strains are dominant are well approximated for $r \to 0$ by the usual elastic singularities (e.g., 5):

$$\sigma_{ij} = \frac{K}{\sqrt{2 \pi r}} f_{ij}(\theta) \qquad (1)$$

where K = the stress intensity factor, $f_{ij}(\theta)$ = geometric factors (see e.g. (5)), (r, θ) = the standard polar coordinates centered at the crack tip.

Upon loading, a plastic zone where the plastic strains are dominant develops around the crack tip. The strain rate dependence of the high temperature plastic behavior can be neglected if the high strain rate limit is considered. This assumption is certainly reasonable at the crack tip. For power law hardening materials, the stresses in the crack tip plastic zone are accurately approximated for $r \to 0$ by the Hutchinson-Rice-Rosengren singularities (6-8):

$$\sigma_{ij} = \left(\frac{J}{B_p I_{n_p} r}\right)^{1/(n_p+1)} \tilde{\sigma}_{ij}(\theta, n_p) \qquad (2)$$

where J = the J-integral (9), n_p and B_p = the constants of the plastic law $(\varepsilon_{pl} = B_p \sigma^{n_p})$, $I_{n_p} \approx \pi$, and $\tilde{\sigma}_{ij}(\theta, n_p)$ = geometric factors (see (6-8)).

Under small scale yielding loading conditions (5):

$$J \sim K^2 / E \qquad (3)$$

As time increases, a creep zone, where the creep strains are dominant, develops and grows around the crack tip. For a material which creeps by secondary power law creep only, the stresses in the creep zone are given

for $r \to 0$ by the time-dependent Riedel and Rice (RR) singularities (10):

$$\sigma_{ij}(t) = \left(\frac{C(t)}{B_c I_{n_c} r}\right)^{1/(n_c+1)} \tilde{\sigma}_{ij}(\theta, n_c) \qquad (4)$$

where n_c and B_c = the constants of the creep law $(\dot{\varepsilon}_{cr} = B_c \sigma^{n_c})$, and:

$$t < t_{tr} \qquad C(t) = \frac{J}{(n_c+1)t} \qquad (5a)$$

$$t > t_{tr} \qquad C(t) = C^* \qquad (5b)$$

where:

$$t_{tr} = \frac{J}{(n_c+1)\, C^*} \qquad (5c)$$

and where C* = the C* - integral (10,11).

The approximate time dependent spatial distribution ahead of the crack tip of the tensile stress across the crack plane is shown in figure 1 for 2219-T851 at 175°C for conditions typical of the crack growth tests reported in reference (1). The calculations were carried out using equations 1 through 5 on the basis of data listed in table I (1). K and C* were computed according to the formulas listed in table II (see section IV).

Although this analysis assumes a stationary crack, it can be applied to rationalize to a certain extent the differences creep crack growth between creep brittle and creep ductile behaviors. If the transition time t_{tr} calculated for typical conditions of creep crack growth tests is much larger than test characteristic times such as the time to failure, creep deformation will certainly remain localized at the crack tip during crack growth. Such a material is expected to be creep brittle. Since the near tip stresses are fully characterized by K for $t < t_{tr}$ (see equations (1-5)) under small scale yielding loading conditions, K is the logical load-geometry parameter to correlate with da/dt in creep brittle materials. In the cases where t_{tr} is short compared to the time to failure, or compared to the time for the crack to grow over a given microstructural distance, extensive creep deformation may accumulate during crack growth. Such a material may thus be creep ductile. Since the near tip stresses are fully characterized by C* for $t > t_{tr}$ (see equations (4-5)), C* is the logical load-geometry parameter to correlate with da/dt in creep ductile materials. The analysis presented above assumes that the crack remains sharp, even for $t >> t_{tr}$. For extensively ductile materials, crack tip blunting cannot be neglected, especially in specimens with low deformation constraints. Under such conditions, the stresses around the crack tip and in the fracture process zone are not accurately approximated by singularities characterized by K or C*. Parameters which describe the overall deformation of the specimen such as the net section stress, the nominal stress or the reference stress may then be better suited to correlate with da/dt than C* (12-14).

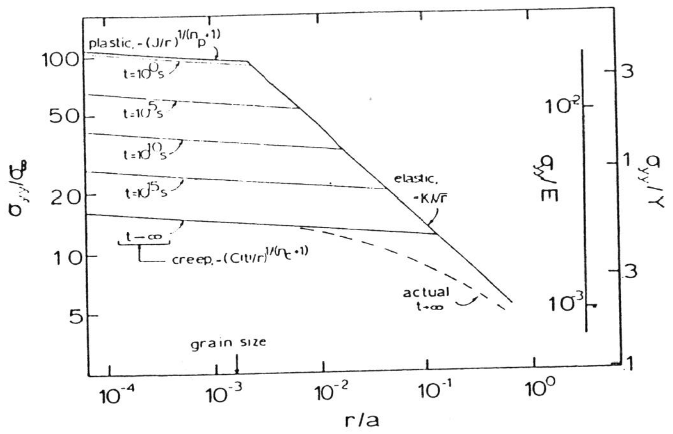

Fig. 1. Approximate spatial distribution ahead of a Mode I stationary sharp crack of the tensile stress across the crack plane for 2219-T851 at 175°C (log scales). The variations of the singular terms only are shown (see Table I for a list of the data used in the calculations).

TABLE I

Material Constants and Typical Testing Conditions Used to Calculate the Stresses in Figure 1.

MATERIAL:	2219-T851
TEMPERATURE:	175°C
YOUNG'S MODULUS:	$E = 7.1 \times 10^{4}$ MPa
YIELD STRESS:	$Y = 272$ MPa
PLASTIC LAW:	$n_p = 23$, $B_p = 7.0 \times 10^{-60}$ $(MPa)^{-23}$
CREEP LAW:	$n_c = 24$, $B_c = 1.2 \times 10^{-63}$ $(MPa)^{-24}$ $\times\ s^{-1}$
SPECIMEN:	C.T.
	$w = 6.35$ cm, $b = 1.27$ cm, $b_{net} = .76$ cm
LOAD:	$P = 5344N$
CRACK LENGTH:	$a = 3.18$ cm $(a/w = 0.5)$
STRESS INTENSITY FACTOR:	$K = 20.8\ MPa\sqrt{m}$
C*-INTEGRAL:	$C^* = 6.0 \times 10^{-22}$ $MPa.m.s^{-1}$
TRANSITION TIME:	$t_{tr} = 4 \times 10^{17}$ s
FAILURE TIME:	$t_f = 10^{5} - 10^{6}$ s

Table II

Expressions of Different Loading Parameters for CT Specimens

- Stress intensity factor:

$$K = \frac{P}{(bb_{net}w)^{1/2}} \frac{(2 + a/w)}{(1 - a/w)^{3/2}} f(a/w) \qquad (15\text{-}17)$$

with $f(a/w) = .886 + 4.64\,(a/w) - 13.32\,(a/w)^2 + 14.72\,(a/w)^3 - 5.6(a/w)^4$

- Net section stress:

$$\sigma_{net} = \frac{P}{b_{net}(w - a)}$$

- Nominal stress:

$$\sigma_{nom} = \frac{P}{b_{net}(w - a)} \left(1 + 3\,\frac{w + a}{w - a}\right) \qquad (11,18,19)$$

- Reference stress:

$$\sigma_{ref} = 2.02\,\frac{P}{b_{net}w} \frac{(1 + a/w)}{(1 - a/w)^2} \qquad (20\text{-}22)$$

- C* integral:

$$C^* = \frac{b}{b_{net}} B_c(w-a)\, h_1(a/w,n_c)\, [P/(\alpha b(w-a)\eta(a/w))]^{n_c + 1} \qquad (23,24)$$

with $\alpha = 1.455$ in plane strain

$\alpha = 1.072$ in plane stress

$$\eta(a/w) = \left[\left(\frac{2a}{w - a}\right)^2 + 2\left(\frac{2a}{w - a}\right) + 2\right]^{1/2} - \left[\left(\frac{2a}{w - a}\right) + 1\right]$$

$$h_1(a/w,\ 20)^{\dagger} = 27.33 - 228.8(a/w) + 745.7(a/w)^2 - 1181(a/w)^3$$

$$+ 912.7\,(a/w)^4 - 275.0\,(a/w)^5 \qquad \text{for } 0.25 \leq a/w \leq 1.0$$

† The error which arises from the fact that $h_1\ (a/w,\ n_c)$ is assumed to be equal to $h_1(a/w,\ 20)$ is estimated as being neglible.

According to the data listed in table I, 2219-T851 is expected to be creep brittle at 175°C. Whether K correlates with the creep crack growth rates in this alloy is discussed below.

II) EXPERIMENTAL PROCEDURES

Creep crack growth rates were measured in 2219-T851 aluminum alloy at 175°C in air in the (T-L) orientation. The experimental procedures were described in detail elsewhere (1) and they are only summarized here.

The crack growth tests were performed on 6.35 cm wide and 1.27 cm thick side grooved CT specimens. Side grooves were machined up to a depth of 0.254 cm on most of the specimens with a 0.025 cm maximum tip radius.

All the specimens were fatigue precracked at room temperature, the load range being continuously adjusted in order to maintain the maximum stress intensity factor at a value lower than the initial stress intensity factor of the creep crack growth tests.

The specimens were cyclically loaded on a servohydraulic testing machine under computer control, maintained at maximum load for a given hold time at each cycle, unloaded and then reloaded. Crack lengths were obtained from unloading compliance measurements recorded during each cycle.

Compliance and stress intensity factor calibrations for CT specimens with side grooves were reported in reference (1). Both constant load and constant stress intensity factor tests were performed. For hold times longer than 10 seconds, no cyclic effect on the crack growth rates could be detected and it was shown that purely time-dependent crack growth where measured (1).

III) EXPERIMENTAL RESULTS

2219-T851 aluminum alloy was found to behave in a typical creep brittle manner at 175°C in air. The creep crack growth rates were thus primarily reported versus the stress intensity factor. Eventual correlations with the C*-integral, the net section stress, the nominal stress or the reference stress were also considered.

The creep crack growth rates measured in constant-K tests performed on 40% side grooved specimens were found to remain essentially constant after a short transient where the crack growth rates were observed to increase rapidly (see fig. 2). A regime of steady state crack growth where a balance is established between creep relaxation at the crack tip and crack tip blunting on the one hand, and damage accumulation ahead of the crack on the other hand can be reached under the experimental conditions corresponding to these tests.

The creep crack growth rates for constant load tests performed on 40% side grooved specimens show a typical three stage behavior when they are plotted versus K (see fig. 3 and 4). Stage I corresponds to an initial transient which depends on the initial conditions (see fig. 3). The stage II creep crack growth rates were found to be independent of the initial stress intensity factor, and to depend only on the instantaneous stress

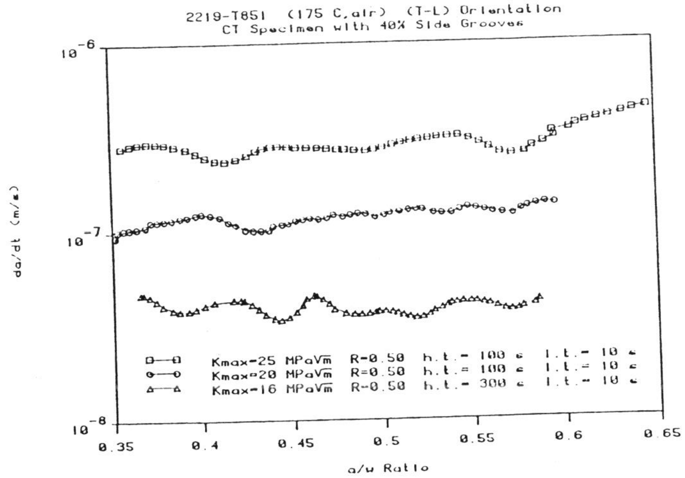

Fig. 2. Variations of da/dt as a function of a/w for constant K tests.

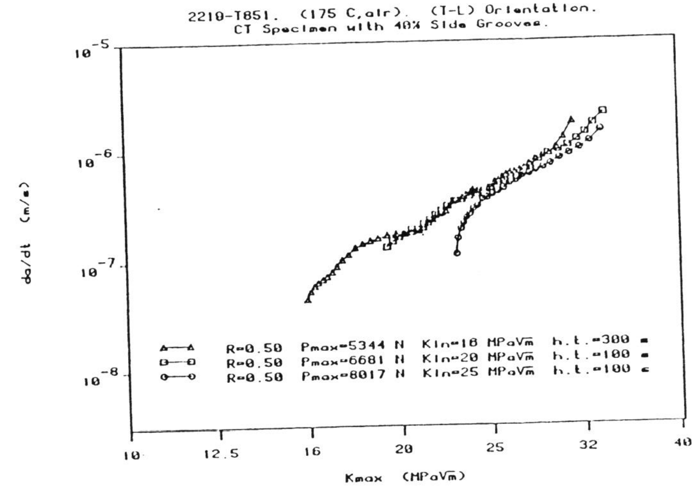

Fig. 3. Variations of da/dt as a function of K for constant load tests.

intensity factor (see fig. 3). The creep crack growth rates measured in the stage II regime of constant load tests were found to be equal, within the experimental scatter, to those measured in the steady state regime of the constant stress intensity factor tests at the same K level. A regime of quasi-steady-state crack growth is thus established in stage II, where the kinetic balance described above for steady-state crack growth is displaced very slowly as K increases with crack length. The creep crack growth rates in this regime of steady or quasi-steady state crack growth can be expressed as a function of the stress intensity factor as:

$$\frac{da}{dt} = A\ K^{n} \tag{6}$$

with n = 3.8. This apparent correlation between the stress intensity factor and the stage II creep crack growth rates is thus independent of the initial stress intensity factor in constant load tests. The creep crack growth rates in stage II were shown to increase with the specimen side groove depth, an upper bound of the crack growth rates being reached for more than 20% side grooved specimens (1). The constraints in the latter specimens were found to be sufficient to guarantee approximate plane strain conditions over most of the net thickness of the specimens. As a matter of fact, flat fracture surfaces with no shear lips and showing evidence of straight crack front markings were observed in more than 20% side grooved specimens(1). Wide shear lips and extensive crack tip tunnelling were observed in 1.27 cm thick specimens with no side grooves (1). The correlation between the stage II creep crack growth rates and the stress intensity factor given by equation (6) is thus valid only for simple K histories such as those followed during constant load or constant stress intensity factors tests, in a regime of quasi-steady or steady state crack growth, and under highly constrained conditions. Finally, the stage III creep crack growth rates for constant load tests correspond to the onset of critical fast fracture.

IV) DISCUSSION

The net section stress, the nominal stress (i.e., the tensile stress at the crack tip which would be obtained by the application of the linear elastic beam theory) (11, 18, 19), the reference stress (i.e., the stress which would give rise to the same load point displacement rate if applied to an uncracked specimen of same geometry and same dimensions, and which can be obtained by plastic limit analysis) (12,14,20,25,26), and the C* integral were also estimated for the constant load tests and the constant-K tests whose results are shown in figs. 2 and 3. The equations which were used to calculate these loading parameters and the stress intensity factor are listed in table II along with references where they were originally given. Kumar and Shih's semi-analytical method to calculate C* under plane strain conditions (23,24) was preferred to the multi-specimen method developed by Landes and Begley (11) because of the complexity of the latter method. It was also preferred to Harper and Ellison's methods ($C^* \propto P\,\Delta$) (27-31) since the creep component of the displacement rate was shown to be negligible compared to the crack growth component over most of the range of crack growth rates studied (see (1)). Harper and Ellison's methods are thus invalid for the creep crack growth tests discussed here (1,32,33).

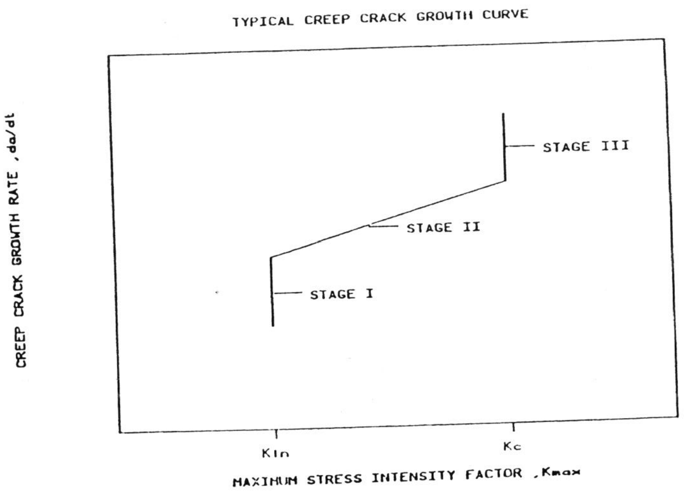

Fig. 4. Typical log - log creep crack growth rate versus stress intensity factor curve.

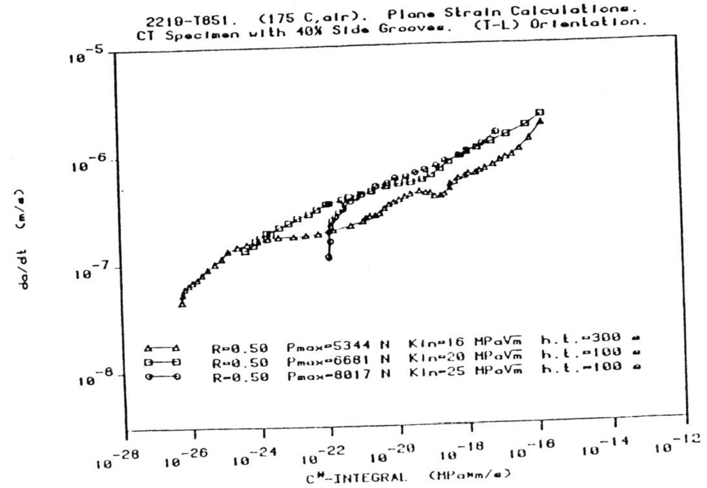

Fig. 5. Variations of da/dt as a function of C* for constant load tests (same data as in figure 3).

The creep crack growth rates for the constant load tests fall on totally separate curves when they are plotted versus the net section stress. That the net section stress does not correlate with the creep crack growth rates in CT specimens is not surprising since the bending stresses at the crack tip are dominant as can be verified from the equation giving the nominal stress. Both the nominal stress and the reference stress are found to correlate satisfactorily with the creep crack growth rates in the stage II regime of crack growth independently of the initial conditions for constant load tests. An exponent close to 4 was found for the correlation of the stage II creep crack growth rates with either the nominal stress or the reference stress. The similitude between the correlations with the nominal stress, the reference stress and the stress intensity factor can be explained by the fact that the variations of both stress parameters are dominated by the variations of $((1 + a/w) / (1 - a/w)^2)$ while the variations of the stress intensity factor are dominated by the similar variations of $((2 + a/w) / (1 - a/w)^{3/2})$. The C*-integral appears to correlate also with the stage II creep crack growth rates with an exponent of 0.1 and not 1 as would have been predicted if Harper and Ellison's method had been used (see fig. 5). Since the stress intensity factor, the C*-integral, the nominal stress and the reference stress all seem to correlate with the stage II creep crack growth rates for constant load tests, it appears very difficult to determine unambiguously which loading parameters correlate best with the creep crack growth rates.

Although the creep crack growth rates were found to remain essentially constant during constant stress intensity factor tests (see fig. 2) the nominal stress, the reference stress and the C*-integral were shown to increase steadily with crack length during the same tests (see fig. 6). In spite of the results of the constant load tests, neither the nominal stress, the reference stress nor the C*-integral can thus correlate with the stage II creep crack growth rates in 2219-T851 aluminum alloy at 175°C under constrained conditions. It is only with the stress intensity factor that such a correlation can be attained. In order to decide upon the best correlating parameter for creep crack growth, it is thus essential that tests where the creep crack growth rates can be maintained constant be performed. This brings therefore into doubt the conclusions of studies where the choice of such a loading parameter was merely based on the extent of scatter in the stage II creep crack growth curves for constant load tests under different initial conditions.

CONCLUSIONS

1) In agreement with concepts of fracture mechanics of creeping solids, 2219-T851 was found to behave as a typical creep brittle material at 175°C.

2) A unique correlation was shown to exist between K and da/dt in the stage II regime of creep crack growth in 2219-T851 for simple K histories under constrained conditions, i.e., ideally, under plane strain conditions. There is no correlation between da/dt and the net section stress, the nominal stress, the reference stress nor the C*-integral.

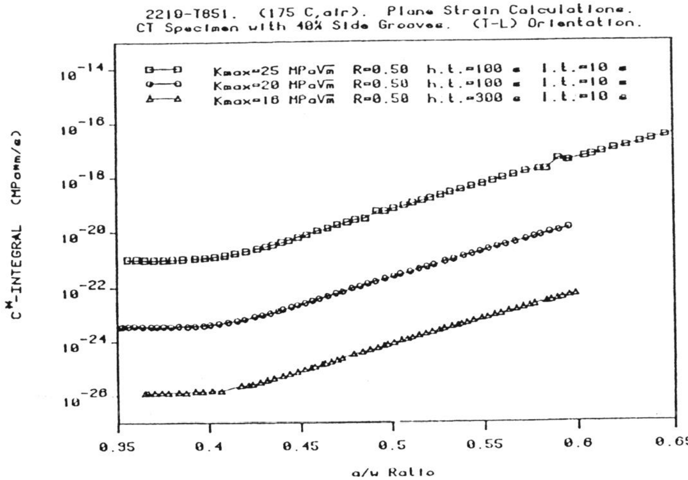

Fig. 6. Variations of C* as a function of a/w for constant K tests (same data as in figure 2).

3) This study also demonstrates that:
(i) The methods to calculate load-geometry parameters should be selected with great care since invalid methods would yield misleading results; and

(ii) it is essential in order to determine unambiguously which parameter can correlate with da/dt to perform, in addition to constant-load tests, tests where, for example, the creep crack growth rates remain constant.

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REFERENCES

1. P.L. Bensussan, D.A. Jablonski and R.M. Pelloux: Metall. Trans. A, 1984, Vol. 15A, pp. 107-120.
2. H. Riedel: J. Mech. Phys. Solids, 1981, vol. 29, pp. 35-49.
3. F.A. Leckie and R.M. McMecking: Int. J. Fracture, 1981, vol. 17, pp. 467-476.
4. F.A. McClintock and J.L. Bassani: in Three Dimensional Constitutive Relations and Ductile Fracture, J. Zarka and S. Nemat-Nasser, eds., North Holland Publishing Co., Amsterdam, Holland, 1981, pp. 119-141.
5. D. Broek: Elementary Engineering Fracture Mechanics, 2nd ed., Sijthoff and Noordhoff, Alphen aan den Rijn, The Netherlands, 1978.
6. J.R. Rice and G.F. Rosengren: J. Mech. Phys. Solids, 1968, vol. 16, pp. 1-12.
7. J.W. Hutchinson: J. Mech. Phys. Solids, 1968, vol. 16, pp. 13-31.
8. J.W. Hutchinson: J. Mech. Phys. Solids, 1968, vol. 16, pp. 337-347.
9. J.R. Rice: J. Appl. Mech., Trans. ASME, 1968, vol. 35, pp. 379-386.
10. H. Riedel and J.R. Rice: in Fracture Mechanics: Twelfth Conference, ASTM STP 700, P.C. Paris, ed., American Society for Testing and Materials, Philadelphia, PA, 1980, pp. 112-130.
11. J.D. Landes and J.A. Begley: in Mechanics of Crack Growth, ASTM STP 590, J.R. Rice and P.C. Paris, eds., American Society for Testing and Materials, Philadelphia, PA, 1976, pp. 128-148.
12. G.J. Neate: Mater. Sci. Eng., 1978, vol. 33, pp. 165-173.
13. S. Taira and R. Ohtani: in Proceedings of the International Conference on Creep and Fatigue at Elevated Temperature Applications, I. Mech. E. Conf. Publ., vol 13, Edmunds, England, 1973, Paper No. C213/73.
14. B.L. Freeman: Int. J. Fracture, 1979, vol. 15, pp. 179-190.
15. ASTM E399-81, Annual Book of ASTM Standards, Part 10, ASTM, Philadelphia, PA, 1982, p. 611.
16. C.N. Freed and J.M. Krafft: J. Mater., 1966, vol. 1, pp. 770-790.
17. P. LeFort and D.F. Mowbray: J. Test. Eval., Trans. ASME, 1978, vol. 6, pp. 114-119.
18. K. Sadananda and P. Shahinian: Metall. Trans. A, 1977, vol. 8A, pp. 439-449
19. K. Sadananda and P. Shahinian: Metall. Trans. A, 1978, vol. 9A, pp. 79-84.
20. J.A. Williams and A.T. Price: J. Eng. Mat. Technol., Trans. ASME, 1975, vol. 97, pp. 214-222.
21. K. Sadananda and P. Shahinian: Eng. Fract. Mech., 1981, vol. 15, pp. 327-342.
22. K. Sadananda and P. Shahinian: in Cavities and Cracks in Creep and Fatigue, J. Gittus, ed., Applied Science Publishers, London, England, 1982, pp. 109-195.
23. V. Kumar and C.F. Shih: in Fracture Mechanics: Twelfth Conference, ASTM STP 700, P.C. Paris, ed., American Society for Testing and Materials, Philadelphia, PA, 1980, pp. 406-438.
24. V. Kumar and C.F. Shih: in Proceedings of the International Conference on Engineering Aspects of Creep, I. Mech. E. Conf. Publ., vol. 2, Edmunds, England, 1980, Paper No. C228/80, pp. 211-214.

25. J.R. Haigh and C.E. Richards: in Proceedings of the International Conference on Creep and Fatigue at Elevated Temperature Applications, I. Mech. E. Conf. Publ., vol. 13, Edmunds, England, 1973, Paper No. C 159/73.
26. J.R. Haigh: Mater. Sci. Eng., 1975, vol. 20, pp. 225-235.
27. M.P. Harper and E.G. Ellison: J. Strain Analysis, 1977, vol. 12, pp. 167-179.
28. R. Koterazawa and T. Mori: J. Eng. Mater. Technol., Trans. ASME, 1977, vol. 99, pp. 298-305.
29. K.M. Nibkin, G.A. Webster and C.E. Turner: in Cracks and Fracture, ASTM STP 601, J.L. Swedlow and M.L. Williams, eds., American Society for Testing and Materials, Philadelphia, PA, 1976, pp. 47-62.
30. K.M. Nibkin, G.A. Webster and C.E. Turner: in Fracture 1977 - Advances in Research on the Strength and Fracture of Materials, D.M.R. Taplin, ed., vol. 2, University of Waterloo Press, Waterloo, Canada, 1977, pp. 627-634.
31. G.A. Webster: in Mechanics and Physics of Fracture, The Institute of Physics and The Metals Society, Cambridge, England, 1975, Paper No. 18.
32. E. Maas and A. Pineau: to be published in the Proceedings of the Fourth International Conference on the Mechanical Behavior of Materials (ICM4), Stockholm, Sweden, August 1983.
33. P.L. Bensussan: An Experimental and Theoretical Study of Creep Crack Growth in 2219 Aluminum Alloy, Ph.D. Thesis, Dept. of Materials Science and Engineering, Massachusetts Institute of Technology, Cambridge, MA, February 1984.