Grain Boundary Sliding and Creep Fracture of Metals under Multiaxial Stresses

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ABSTRACT

The mechanisms of high temperature creep fracture are known to involve the nucleation, growth and coalesence of intergranular cavities. Yet these processes are rarely the controlling factors in creep fracture. The redistributions of stress that occur in a creeping solid are usually more important. Grain boundary sliding can lead to stress concentrations on grain boundaries that are perpendicular to the maximum principal stress and this can cause those grain boundaries to cavitate and fail more quickly than others. This observation has important implications for creep fracture under multiaxial stresses and can be used to develop a parameter for predicting creep fracture of metals and alloys under different loading conditions. The parameter is expected to be valid for alloys that exhibit grain boundary sliding and fail by cavitation of grain boundaries perpendicular to the maximum principal stress. An analysis given by Anderson and Rice is used to show that, prior to cavitation, the average tensile stress on grain boundary facets perpendicular to the maximum principal stress is approximately

$$\sigma_F = 2.24\,\sigma_1 - 0.62\,(\sigma_2 + \sigma_3)$$

where $\sigma_1 > \sigma_2 > \sigma_3$ are the principal stresses. This quantity, called the principal facet stress, can be used to predict multiaxial creep fracture from uniaxial rupture data for several materials.

KEYWORDS

Creep fracture; grain boundary sliding; facet stresses; multiaxial fracture.

INTRODUCTION

The mechanisms of creep fracture have been studied for more than 30 years. As a result of these studies, the various physical processes involved in this failure mode are now reasonably well understood. Because this subject has been reviewed recently (Argon, Chen and Lau, 1980; Nix, 1983; Cocks and Ashby, 1982; Nix and Gibeling, 1985; Riedel, 1987; Nix,1988) and because most of the fundamentals are now quite well established, it is not necessary for us to review these mechanisms in detail. For the purposes of this paper it is sufficient to note that creep fracture occurs by the formation, growth and coalesence of intergranular cavities. These cavities nucleate at stress concentrations created by localized deformation, grow by the diffusional transport of matter away from the cavities, usually along the adjoining grain boundaries, and finally coalesce to produce fracture. If the diffusion distance is small compared to the spacing between cavities, the displacements associated with cavity growth must be accommodated by creep of the surrounding material. Thus the mechanisms of creep fracture involve both the nucleation of cavities at stress concentrations and their growth by both diffusional and creep processes.

This discussion of mechanisms usually relates to a hypothetical situation in which a single grain boundary is subjected to a constant tensile stress that promotes nucleation and growth of cavities on the boundary. The effects of stress redistribution associated with grain boundary sliding and inhomogeneous cavitation, which can be more important, are often ignored. In this paper we focus our attention on the stress redistribution associated with grain boundary sliding and the effects of that stress redistribution on creep fracture. We show that significant stress concentrations can be produced by grain boundary sliding. These stress concentrations help to explain why cavitation is typically concentrated on those grain boundaries that are perpendicular to the maximum principal stress. We also show that the stress redistribution associated with grain boundary sliding can be used to develop a parameter for describing creep fracture under multiaxiaal stresses. The ideas presented here are described in more detail in other papers (Nix,1988; Nix, Earthman, Eggeler and Ilschner, 1988).

GRAIN BOUNDARY SLIDING

As noted above, theoretical studies of creep fracture commonly assume a single grain boundary subjected to a tensile stress that does not change in the course of creep. Such an assumption is valid for certain experiments involving bicrystals and for polycrystalline solids in which cavities are present on all of the boundaries. In these cases, cavitation can occur in an unconstrained manner. But for most cases, a major redistribution of the stress occurs during creep. In such circumstances, the laws describing cavity growth have to be used in conjunction with a description of the stress redistribution. In this section of the paper, we consider the stress redistribution that occurs because of grain boundary sliding.

Grain boundaries slide at high temperatures because atomic diffusion in the boundaries quickly accommodates any incompatiblities that develop at ledges in the sliding boundaries. As a result, the shear stresses that are supported by grain boundaries at low temperatures are relieved at high temperatures. In the discussion that follows, we assume that the shear stresses on the sliding boundaries are fully relaxed and that the imposed stress is supported entirely by the normal stresses on the boundaries. Of course this causes a major redistribution of the stress to occur within the solid.

First, we consider a two dimensional arrangement of hexagonal grains of the kind shown in Fig.1. The imposed tensile stress σ is assumed to be greater than the transverse stress σ_T. For $\sigma > \sigma_T$ the inclined grain boundaries slide in the direction shown and relieve the shear stresses. The state of stress in the solid after grain boundary sliding is complex, especially if power law creep occurs in the grains. Although we are not able to give a complete description of the stresses in the solid, it is possible to determine the average normal stresses on the grain boundaries using the principles of statics. For the kind of loading shown in Fig.1, the average normal stresses on the grain boundaries can be expressed as

$$\sigma_F = \frac{3}{2}\sigma - \frac{1}{2}\sigma_T \quad \text{and} \quad \sigma_I = \sigma_T \qquad (1)$$

Here σ_F is the average normal stress on the transverse boundary facets and σ_I is the average normal stress on the inclined boundaries. We note that for purely uniaxial tension (σ_T=0) the inclined boundaries support no stress at all and the applied stress must be supported entirely by the transverse facets (σ_F=1.5σ). This simple model shows why cavitation is strongly concentrated on the grain boundaries that are perpendicular to the tensile axis (Hayhurst,

Dimmer and Morrison, 1984; Stanzl, Argon and Tschegg, 1983; Davies, Williams and Wilshire, 1968; B.J. Cane, 1978; Yu and Nix, 1984). Indeed, for this model no cavitation is expected on the inclined boundaries. Any cavity that might be nucleated on an inclined boundary would eventually shrink because the average normal stress on the boundary is zero.

It is of interest to consider the consequences of cavitation on the transverse facets using this simple two dimensional model. If the cavitation occurs at the same rate on all of the transverse facets, then the facet stress rather than the applied stress should be used to calculate the rate of cavity growth and the fracture time. When the cavities interlink simultaneously on all of the facets, fracture occurs immediately because the applied stress can no longer be supported. The separation of the grains is accommodated completely by grain boundary sliding. Of course it is unrealistic to expect the cavitation to occur at the same rate on all of the transverse facets. The kinetics of cavitation depend strongly on both the structure (Watanabe, 1983) and composition (George, Li and Pope, 1987) of the grain boundaries. Thus one can expect the cavitation processes to occur non-uniformly in the solid. In this case normal stresses will develop on the inclined boundaries even for uniaxial tension. If a few transverse facets fail at some point in the solid, the subsequent separation of the grains will be constrained by the surrounding material. Because the creep rate in this region is bounded by the creep rate of the surroundings and the shear stresses on the inclined boundaries are zero, it follows that an equal biaxial stress state will develop in the region. Thus the failure of some of the transverse facets causes tensile stresses to develop on the inclined boundaries. One can expect that they too would begin to cavitate and that damage would spread to the entire sample, eventually resulting in fracture.

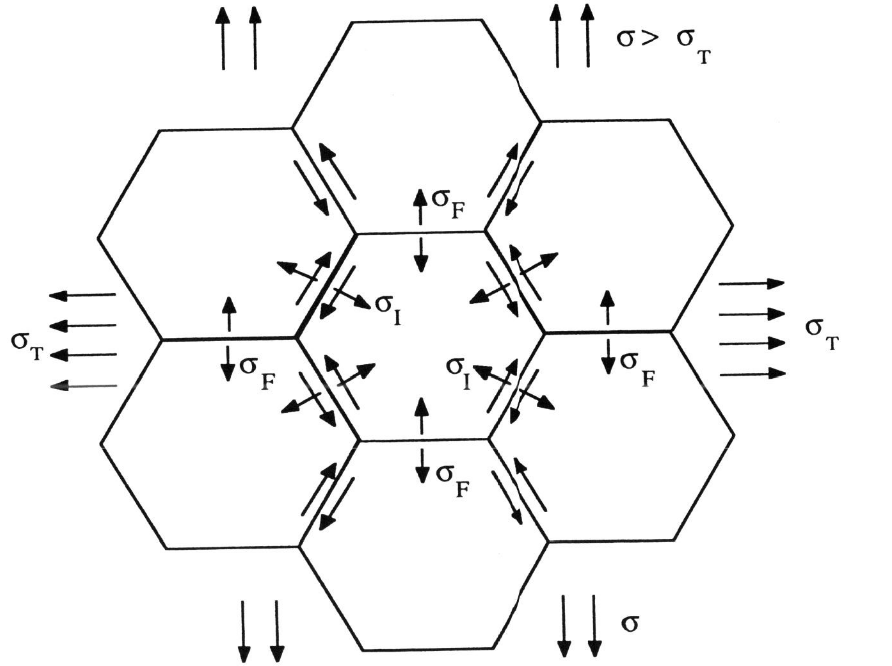

Figure 1. Grain boundary sliding and stress redistribution.

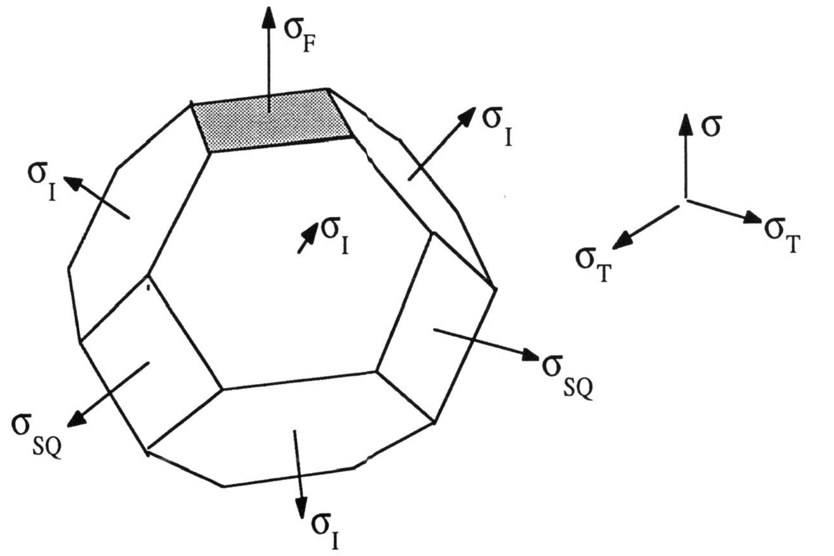

Figure 2. Three dimensional grain structure of Anderson and Rice (1985).

The simple estimate of the average normal stress on the transverse grain boundaries given above for a two dimensional hexagonal grain structure is not adequate for our purposes. It does not include multiaxial stresses and it is not based on a three dimensional grain structure. A better estimate for a three dimensional grain structure can be made using the results of an analysis given by Anderson and Rice (1985). They have recently analyzed grain boundary sliding and cavitation in a polycrystalline solid composed of a periodic array of identical grains shaped like the Wigner-Seitz cells. Figure 2 shows the geometry of the grains they considered. Each grain has 6 square facets and 8 hexagonal facets and all of the edges have the same length. Two orientations were considered: one with the tensile axis perpendicular to one set of square facets and one with the tensile axis perpendicular to one set of hexagonal facets. Anderson and Rice analyzed the average normal stress acting on the transverse grain boundaries for each of these two orientations. Before cavitation has started to occur but after grain boundary sliding has relaxed the shear stresses on the boundaries, they show that the average stress on the transverse facets is

$$\sigma_F^{sq} = 3\sigma_1 - 2\sigma_T \tag{2}$$

for the tensile axis perpendicular to one set of square facets and

$$\sigma_F^{hex} = 1.67\sigma_1 - 0.67\sigma_T \tag{3}$$

for the tensile axis perpendicular to one set of hexagonal facets. Here σ_1 is the maximum principal stress and $\sigma_T = \sigma_2 = \sigma_3$ is the axisymmetric transverse stress. To obtain the average stress on the transverse grain boundary for a polycrystalline structure we take a weighted average of the facet stresses given by Eqns.(2) and (3). Since there are 14 facets, of which 6 are square and 8 are hexagonal, we define the average stress on the transverse facets as

$$\sigma_F = \frac{6}{14}\sigma_F^{sq} + \frac{8}{14}\sigma_F^{hex} \tag{4}$$

Using Eqns.(2) and (3) this becomes

$$\sigma_F = 2.24\sigma_1 - 1.24\sigma_T \tag{5}$$

Although the analysis of Anderson and Rice (1985) was for an axisymmetric stress state ($\sigma_2 = \sigma_3 = \sigma_T$), we generalize Eqn.(5) by replacing the transverse stress with the mean value of the two transverse stresses σ_2 and σ_3: $\sigma_T = (\sigma_2 + \sigma_3)/2$. With this modification, the average stress on the transverse boundaries is given by

$$\sigma_F = 2.24\sigma_1 - 0.62(\sigma_2 + \sigma_3) \tag{6}$$

We note that for a hydrostatic stress state ($\sigma_1 = \sigma_2 = \sigma_3 = \sigma_H$) the facet stress is simply equal to σ_H. This corresponds to the limit of no grain boundary sliding and no stress redistribution. Because the grain boundaries of interest are those perpendicular to the maximum principal stress, we call the average facet stress on these boundaries the principal facet stress. Below we show that the principal facet stress, as calculated using Eqn.(6), can be used as a parameter to predict creep fracture under multiaxial stresses.

It should be emphasized that our estimate of the principal facet stress is expected to be valid after grain boundary sliding has caused a redistribution of the stresses to occur but before the transverse grain boundaries have started to cavitate. During the course of creep, cavitation on the transverse boundaries will lead to additional changes in the facet stresses. Eventually the transverse facets will fail completely and the stresses will then be supported entirely by the other boundary facets. Presumably the inclined facets would then begin to cavitate rapidly leading to final rupture. These complexities, although important to a prediction of the absolute rupture time, need not be included in the multiaxial creep rupture parameter.

CREEP FRACTURE UNDER MULTIAXIAL STRESSES

Creep fracture has been studied extensively, most commonly under uniaxial stress conditions. Although uniaxial stress experiments have led to a good understanding of the physical processes involved, they do not provide sufficient information to predict failure under multiaxial stress conditions. Bending and torsion are examples of loading that can cause multiaxial stresses to develop in smooth components. Multiaxial stress states are also produced at notches and other geometric irregularities, even when the remote loading is purely uniaxial. Thus, the prediction of creep rupture in engineering structures requires that the multiaxial stresses be taken into account. The most widely accepted approach to the prediction of creep fracture under multiaxiaal stresses is that based on continuum damage mechanics. Below we give a brief review of that approach. This is followed by an alternative approach based on the principal facet stress described above.

Continuum Mechanics Approach

Hayhurst (1972) and Hayhurst and Leckie (1973,1984) and their colleagues (Hayhurst, Leckie and Morrison (1978)) have studied the effects of multiaxiality on creep rupture and have used the principles of damage mechanics (Kachanov, 1958; Robotnov, 1969) to describe these effects. Here we give a brief review of their results. They show that for a smooth round bar subjected to uniaxial tension, the rupture lifetime at a given temperature can be expressed as

$$t_f = M\sigma^{-\chi} \tag{7}$$

where σ is the uniaxial stress and M and χ are parameters that characterize the evolution of damage at the temperature in question. The need to consider multiaxial stresses is shown by considering the creep rupture properties of notched round bars. Hayhurst and Leckie (1973) pointed out that Eqn.(7) does not correctly predict creep rupture of notched bars even if the nominal stress in the notch is used in the expression. Instead they have argued that the equation must be generalized to allow other components of the multiaxial stress state to enter the equation. They retain the form of Eqn.(7) and write, for a general state of multiaxial stress,

$$t_f = M(\alpha\sigma_1 + \beta J_1 + \gamma J_2)^{-\chi} \tag{8}$$

for the rupture time. In this expression and in others in this paper we describe the multiaxial stress state in terms of the principal stresses, $\sigma_1 > \sigma_2 > \sigma_3$. Here σ_1 is the maximum principal stress, J_1 is the first stress invariant defined by

$$J_1 = \sigma_1 + \sigma_2 + \sigma_3 = 3\sigma_H \tag{9}$$

and J_2 is the second stress invariant defined by

$$J_2 = \frac{1}{\sqrt{2}}\left[(\sigma_1 - \sigma_2)^2 + (\sigma_2 - \sigma_3)^2 + (\sigma_3 - \sigma_1)^2\right]^{\frac{1}{2}} = \sigma_e \tag{10}$$

The first stress invariant is related to the hydrostatic tension stress through $J_1 = 3\sigma_H$ and the second invariant is the von Mises equivalent shear stress or the effective stress, $J_2 = \sigma_e$. The coefficients α, β and γ describe the relative contributions of the different multiaxial stress components to the rupture life. The terms M and χ again characterize the evolution of damage; their values are assumed to be independent of stress state. In order for Eqn.(8) to be consistent with Eqn.(7) for the case of uniaxial tension the coefficients α, β and γ must be chosen such that $\alpha + \beta + \gamma = 1$. Then for the case of uniaxial tension ($J_1 = \sigma_1$, $J_2 = \sigma_1$), Eqn.(8) reduces to Eqn.(7) ($\sigma_1 = \sigma$).

Hayhurst and Leckie (1984) have reviewed the multiaxial creep rupture data for several metals and have found that the maximum principal stress, σ_1, and the effective stress σ_e are much more important than the hydrostatic stress σ_H in determining creep rupture. Thus they let β = 0 in Eqn.(8) as an approximation and rewrite the equation as

$$t_f = M[\alpha\sigma_1 + (1-\alpha)\sigma_e]^{-\chi} \tag{11}$$

where α is a single parameter that describes the relative importance of σ_1 and σ_e in determining rupture. It is evident that the parameter α cannot be determined from uniaxial creep rupture tests alone. Multiaxial tests in which σ_1 and σ_e can be varied independently are needed to determine α. Materials in which the maximum principal stress is most important are called maximum principal stress materials and are characterized by α=1. Effective stress materials are characterized by the effective stress, α=0. Most materials fall in between these limits. In such cases it is necessary to measure the creep fracture properties under various multiaxial stress states to determine the parameter α. In some cases, α is not constant and much more multiaxial creep rupture data is needed to fully characterize the rupture process. Below we offer an alternative procedure. We suggest that for many materials the principal facet stress can be used as a parameter for predicting creep rupture under multiaxial stresses.

The Principal Facet Stress

In this section of the paper we show that the principal facet stress, defined as

$$\sigma_F = 2.24\sigma_1 - 0.62(\sigma_2 + \sigma_3) \tag{6}$$

can be used to describe creep fracture under multiaxial stress conditions for several materials. Figure 3 shows the creep rupture data of Cane (1982) for a $2^1/_4$ Cr - 1 Mo steel. In this figure, and the ones to follow, three different stress parameters: the effective stress, σ_e; the maximum principal stress, σ_1 and the principal facet stress, σ_F are plotted against the logarithm of the rupture time. In this way the correlation between the different stress parameters and the rupture time can be assessed visually. In Fig.3 the data for uniaxial tension are distinguished from those for the torsion and notched bar experiments by different symbols. Evidently, neither the effective stress nor the maximum principal stress correlate well with the time to rupture for $2^1/_4$ Cr - 1 Mo steel. We consider especially the failure of the maximum principal stress to describe the rupture data. At the same maximum principal stress, the samples tested in torsion fail sooner than those tested in tension while the notched samples fail later. By contrast, the principal facet stress, as defined by Eqn.(6), brings all of the creep rupture data onto a single curve.

The success of the principal facet stress in bringing the data together in Fig.3 suggests that the data points for torsion and the notched bar experiments are separated from those for uniaxial tension (at a fixed maximum principal stress) because of different degrees of grain boundary sliding and stress redistribution. High principal facet stresses are produced in torsion and they cause the torsion samples to fail sooner than samples subjected to uniaxial tension. Conversely, small principal facet stresses are developed in the notch, with the consequence that the notched samples fail later than those subjected to uniaxial tension.

The creep rupture data of Hayhurst, Dimmer and Morrison (1984) for 316 stainless steel are shown in Figure 4. Again the different stress parameters are plotted against the logarithm of the rupture time. Here the data for uniaxial tension tests are distinguished from those for the notched bar tests. The data for two different notch geometries are shown: The British Standard notch and the Hayhurst circular notch. The rupture times for the notched samples are greater than those for uniaxial tension at the same maximum principal stress. This represents a notch strengthening effect. We note that the Hayhurst circular notch is more severe than the British Standard notch and it causes a greater notch strengthening effect. Here again we find that the principal facet stress is a correlating parameter that brings all of the data onto a single curve. Evidently the transverse stresses acting in the notch makes the principal facet stress smaller than for uniaxial tension This causes the rupture life of the notched bar to be greater than that for uniaxial tension under the same maximum principal stress.

Stanzl, Argon and Tschegg (1983) have measured the creep rupture life of uncavitated copper at 500°C in both tension and torsion. Their results are shown in Figure 5. We observe that samples tested in torsion fail sooner than those tested in tension at the same maximum principal stress just, as for the $2^1/_4$ Cr - 1 Mo steel data discussed above. We find again that the principal facet stress brings the tension and torsion data together.

Figure 6 shows the creep rupture data of Dyson and McLean (1977) for Nimonic 80A. We see that neither the effective stress nor the maximum principal stress alone can characterize the creep rupture data. Again we find that the principal facet stress correlates well with the time to rupture.

In the four examples cited here the principal facet stress brings the multiaxial creep rupture data into coincidence with the uniaxial data. This correlation should prove to be quite useful because it allows a prediction of the rupture time under multiaxial stresses from uniaxial creep rupture data. Of course, this prediction requires a knowledge of the multiaxial stress state imposed.

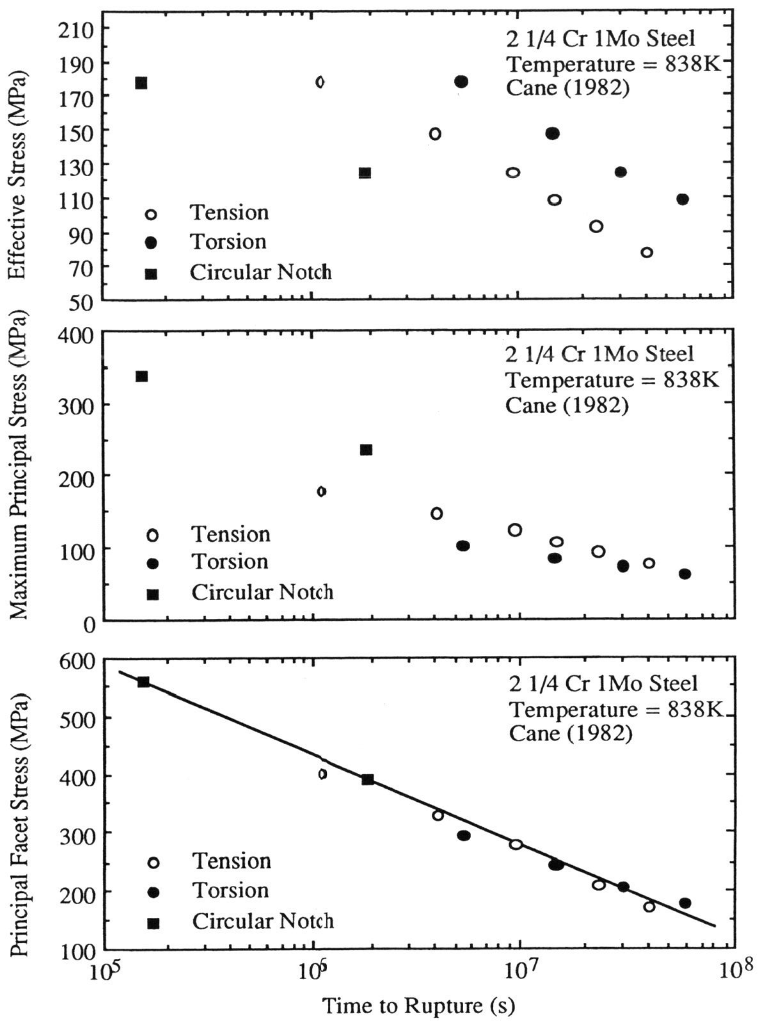

Figure 3. Creep rupture data for a 21/4 Cr - 1 Mo steel plotted as the effective stress (top), the maximum principal stress (middle) and the principal facet stress (bottom) vs the logarithm of the rupture time.

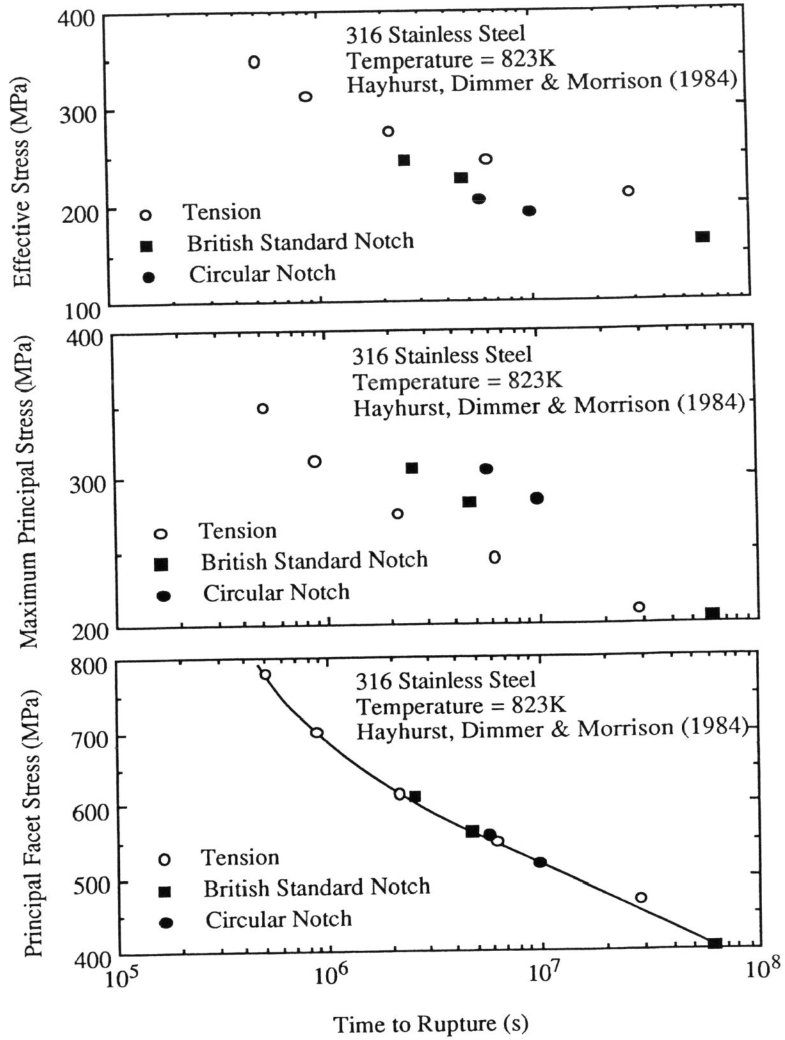

Figure 4. Creep rupture data for 316 stainless steel plotted as the effective stress (top), the maximum principal stress (middle) and the principal facet stress (bottom) vs the logarithm of the rupture time.

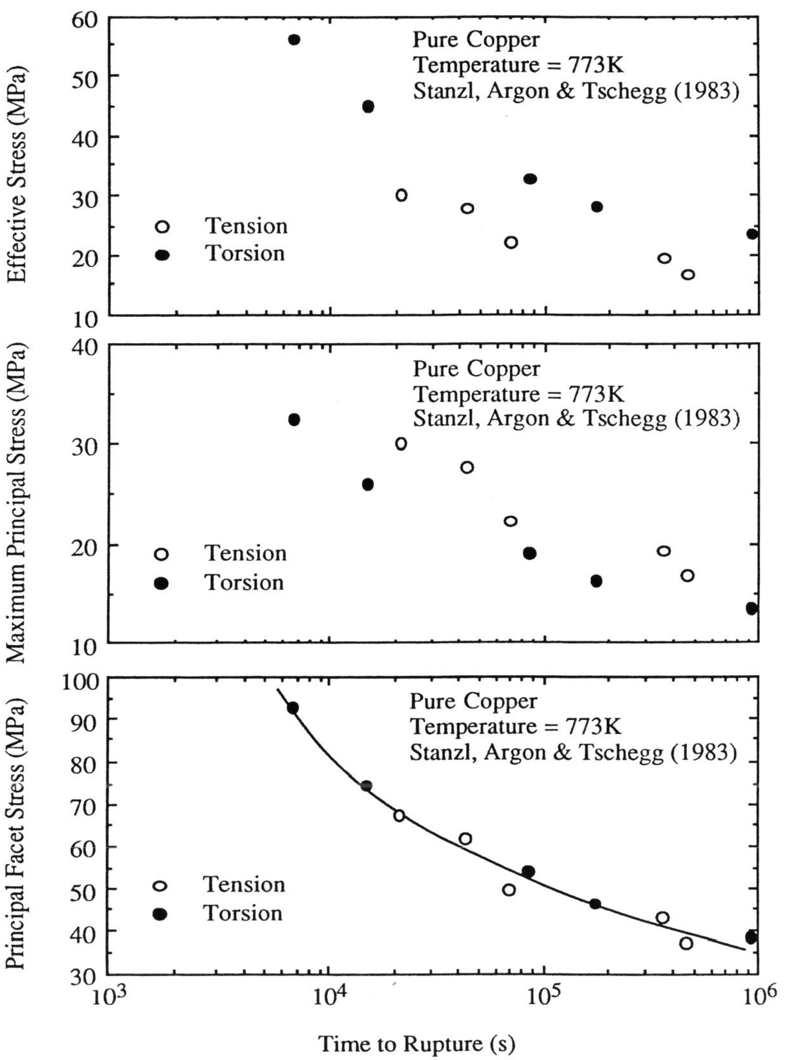

Figure 5. Creep rupture data for pure copper plotted as the effective stress (top), the maximum principal stress (middle) and the principal facet stress (bottom) vs the logarithm of the rupture time.

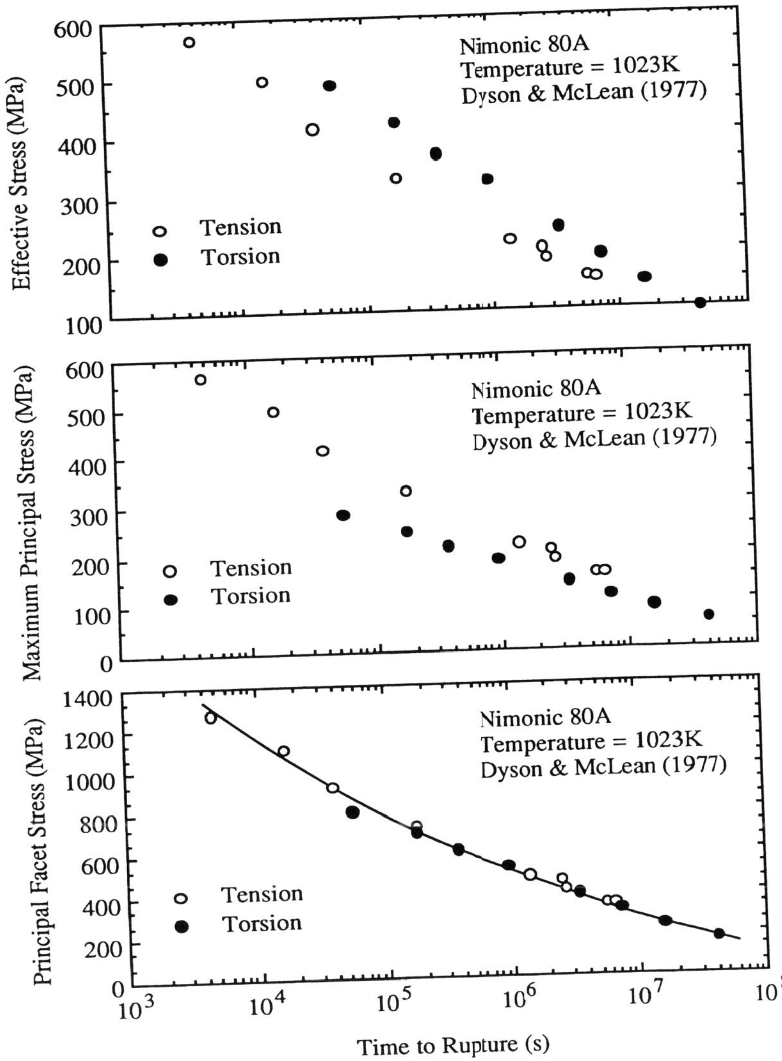

Figure 6. Creep rupture data for Nimonic 80A plotted as the effective stress (top), the maximum principal stress (middle) and the principal facet stress (bottom) vs the logarithm of the rupture time.

Although the principal facet stress is expected to describe multiaxial creep fracture for many materials, there are some materials for which it fails to bring the uniaxial and multiaxial data into coincidence. One example is found in the work of Hayhurst, Dimmer and Morrison (1984) who showed that creep fracture of an aluminum alloy is well characterized by the effective stress (α=0). Although we do not fully understand the failure of the principal facet stress in this case, we believe it may be related to the failure of this material to exhibit grain boundary sliding and cavitation of transverse boundaries, as envisioned by the present approach. It is known that pure aluminum does not cavitate easily. If the aluminum alloy cited here behaves like pure aluminum, then the rupture time would not be expected to correlate with the principal facet stress.

Stanzl, Argon and Tschegg (1983) have also performed tension and torsion experiments for precavitated copper specimens. They found that, for a given maximum principal stress, the time to rupture for precavitated copper under tension and torsion is the same. It follows that the principal facet stress theory is not valid for this case. Precavitated copper is so brittle that rupture may occur before the stresses are adequately redistributed by grain boundary sliding. If so, we would not expect the principal facet stress to be applicable.

Three separate studies of creep fracture of copper under multiaxial stresses have indicated that failure of this material is characterized completely by the maximum principal stress (Hayhurst, Dimmer and Morrison, 1984; Henderson, 1979; Needham and Greenwood, 1975). The possible reasons for the failure of the principal facet stress to predict fracture in these have been discussed recently by Nix et al. (1988).

In spite of the exceptions mentioned here, the principal facet stress appears to be a useful parameter for predicting multiaxial creep rupture for many materials. Further experimental work along these lines will determine how widely the principal facet stress concept can be applied.

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