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Fracture during Creep [and Discussion]

G. W. Greenwood, P. E. Evans, H. H. Schloessin and R. W. Cahn

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Fracture during creep

By G. W. Greenwood
Department of Metallurgy, University of Sheffield, St George's Square, Sheffield S1 3JD

Although materials generally become more ductile with increase in temperature and decrease in strain rate, many metals have a local minimum in ductility at specific temperatures and strain rates. There is thus a region where a decrease in stress level results in a decrease of elongation before fracture. Failure is observed to occur at interfaces and particularly at grain boundaries nearly perpendicular to the maximum principal tensile stress. The nucleation of cavities and cracks, however, appears to arise initially from shear processes at these interfaces at temperatures and stresses where grain boundary sliding is significant but transverse boundary movement is negligible. Thus, tensile and shear stresses play different rôles. Since cavities have a finite volume and link together to cause failure, the hydrostatic component of stress is important and so, in evaluating criteria for fracture, a precise definition of the stress system is necessary. Further, chemical composition, segregation and microstructural features play a vital rôle and materials that are only slightly different may show quite dissimilar behaviour.

INTRODUCTION

Brittleness is often associated with rapid impact and high stresses and with sharp cracks in hard materials. The effects of temperature rises to levels, say, of half the absolute melting temperature, are normally regarded as softening materials significantly to permit stresses to be relieved, cracks to heal and voids to sinter. The overall situation, however, is now widely recognized to be much more complex (Garofalo 1965). At relatively low stresses, at temperatures when materials are soft and where they are capable of slow deformation, it is not uncommon to find that they fracture suddenly, particularly when tensile forces are applied.

Despite the apparent suddenness of fracture at elevated temperatures, it is now clear (Leckie 1978, this volume) that it originates from the slow deformation, or creep, progressively causing 'damage' in the material that can be identified by metallographic techniques. The most general observations have indicated that the separation that ultimately causes fracture occurs at interfaces and particularly at grain boundaries that are nearly normal to the principal tensile stress. Because these interfaces are involved, it appears that the crystallography of the material is relatively unimportant, and even metals of face centred cubic structure and close packed hexagonal structure have been observed to fail by this mode (Greenwood 1973). Nevertheless, such fracture is not observed universally and aluminium, titanium and lead have only fractured in this manner when specific alloying elements or impurities were present.

The time scale of laboratory experiments is such that fracture during creep takes place in a temperature range roughly in the range from 0.4 to 0.7 of the absolute melting temperature (Taplin 1973). Where the time scale is enlarged, for example, by a factor of about 10^5, it is expected that this temperature range would be greatly exceeded, and that creep fracture could occur at quite low fractions of the melting temperature.

It is often observed that the ductility is a minimum at a given temperature and at a certain rate of creep. This minimum, however, is usually a shallow one and the strain before fracture does not vary greatly over a rather wide range of temperature and creep rate (Perry 1974).

The strain before failure, nevertheless, is strongly dependent on the stress system. Under compressive forces there is often a very large ductility and some increase in the strain to fracture is also observed if a tensile component of stress is partly offset by hydrostatic compression. It seems necessary to distinguish between the rôles of the stresses that change shape and those that affect the volume.

The broadest interpretation of the influence of strain rate and of temperature is that, for fracture to occur, there should be significant amount of sliding on interfaces but these interfaces should not move transversely. It is for this latter reason that ductility almost invariably appears to be restored at low strain rates when the temperature rises above about 0.7 of the absolute melting temperature.

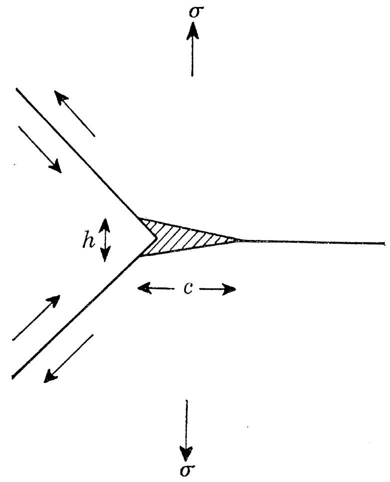

FIGURE 1. The sliding at grain boundaries to open up a crack of length c and maximum thickness h perpendicular to an applied stress σ.

In this paper the creation of cracks by sliding at grain boundary junctions will first be explored and the condition evaluated whereby these may lead to a complete separation.

It is often observed, however, that fracture does not occur purely by progressive crack growth at an interface but by the formation of cavities at specific points along a boundary. Proposals are next considered that attempt to account for cavity nucleation. The rôle of particles at the boundaries is often found to be of particular significance in this respect. The segregation of impurities can also have a large influence. Interfacial separation can be made easier when gases are precipitated as bubbles along a grain boundary, but their rôle as nucleants for fracture is less potent than has sometimes been concluded. Geometrical factors make it possible to envisage ways in which deformation processes can directly lead to fracture but these cannot provide a complete approach.

It is now widely recognized that much of the deformation in creep resistant materials over long periods of time, and possibly many materials in the Earth's crust, depends more on the flow of vacancies than on the effects of dislocation mobility. It is pointed out in this paper that the vacancy flow effects can be even more important in assisting the growth of cracks and cavities because the diffusion paths for flow of vacancies along grain boundaries are somewhat easier than through the lattice; these paths are relatively short and the vacancies can be produced within the boundary itself. Such vacancy flows cannot, however, produce the initial step of interfacial separation. The mode of nucleation thus requires another mechanism to operate until a critical size of cavity nucleus has been reached. The continued growth by vacancy accumulation can be analysed by well established procedures. This also permits some account to be taken of the type of stress system as well as of its magnitude.

FAILURE BY SLIDING AT INTERFACES

It is easy to visualize geometrically how sliding at interfaces can cause cracks to open and to propagate from points where three interfaces meet. This situation has been explored in depth by considering the effects of the coalescence of dislocations slipping on different planes to form a crack on a third plane. This approach has led to interesting developments of the theories of fracture arising from the concepts of Zener (1948), Stroh (1954, 1957) and Cottrell (1958) in particular.

It is more recently that a similar approach has been adopted to interpret one of the possible ways in which cracks can grow during creep.

Following the analysis of Williams (1967*a*, *b*), where grain boundaries intersect as shown in figure 1, a wedge-type crack can be formed of length c and of maximum thickness h under a tensile stress σ with the equilibrium condition given by

$$c = \frac{4\mu\gamma}{\pi(1-\nu)\,\sigma^2}\left\{\left(1-\frac{\sigma h}{4\gamma}\right)-\left(1-\frac{\sigma h}{2\gamma}\right)^{\frac{1}{2}}\right\},$$

where μ is the shear modulus, ν is the Poisson ratio and γ is the energy per unit area to create the surfaces. Now continued sliding at the grain boundaries increases the value of h and the corresponding rate of increase in crack length c is given by

$$\frac{\mathrm{d}c}{\mathrm{d}h} = \frac{\mu}{\pi(1-\nu)\,\sigma}\left\{\left(1-\frac{\sigma h}{2\gamma}\right)^{-\frac{1}{2}}-1\right\}.$$

It is immediately clear from this that the crack becomes unstable and can propagate rapidly when $\sigma = 2\gamma/h$. Now $\mathrm{d}h/\mathrm{d}t$ is related to the rate of grain boundary sliding $\dot{\epsilon}_{\mathrm{s}}$ and to the grain size x by the expression $\mathrm{d}h/\mathrm{d}t = x\dot{\epsilon}_{\mathrm{s}}$.

Where t_{f} is the time to fracture, the maximum value of h is then given by

$$h = x\dot{\epsilon}_{\mathrm{s}}\,t_{\mathrm{f}}.$$

Now h cannot exceed $2\gamma/\sigma$ and so it follows that

$$2\gamma/\sigma \approx x\dot{\epsilon}_{\mathrm{s}}\,t_{\mathrm{f}}.$$

This provides a simple evaluation of the life-time before failure occurs under a tensile stress σ, with

$$t_{\mathrm{f}} \approx 2\gamma/x\dot{\epsilon}_{\mathrm{s}}\,\sigma.$$

The above situation can also be developed to evaluate the case where the crack may traverse a complete grain boundary before becoming unstable. It may now be assumed, purely on a statistical basis, that when grain boundaries are completely cracked this is a criterion of a complete failure of the material. This situation is not always true, for in superplastic materials it is well known that complete interfaces can be fractured whilst the material continues to elongate up to 20-fold (Ritchie 1970). This case, however, only appears true for small grains of only a few micrometres in diameter.

To evaluate the time necessary for a crack to propagate to a length equal to a grain diameter we may proceed as follows: the rate of crack growth $\mathrm{d}c/\mathrm{d}t = (\mathrm{d}c/\mathrm{d}h)\,(\mathrm{d}h/\mathrm{d}t) = x\dot{\epsilon}_{\mathrm{s}}\,\mathrm{d}c/\mathrm{d}h$. Hence

$$\frac{\mathrm{d}c}{\mathrm{d}t} = x\dot{\epsilon}_{\mathrm{s}}\frac{\mathrm{d}c}{\mathrm{d}h} = \frac{x\dot{\epsilon}_{\mathrm{s}}\,\mu}{\pi(1-\nu)\,\sigma}\left\{\left(1-\frac{x\dot{\epsilon}_{\mathrm{s}}\,t\sigma}{2\gamma}\right)^{-\frac{1}{2}}-1\right\}.$$

Since in this instance $x\epsilon_{\mathrm{s}}\,t\sigma$ is substantially less than 2γ,

$$\mathrm{d}c/\mathrm{d}t \approx x^2\dot{\epsilon}_{\mathrm{s}}^2\,\mu t/4\pi(1-\nu)\,\gamma.$$

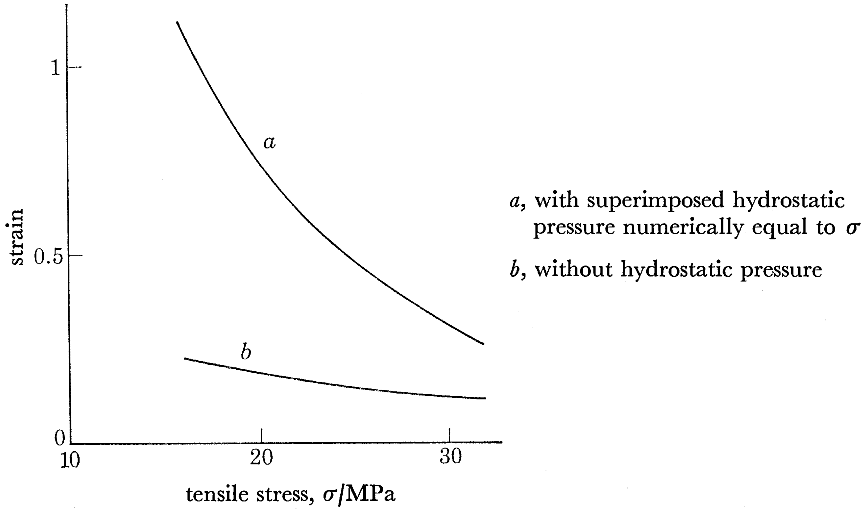

FIGURE 2. The effect of superimposed hydrostatic pressure on strain at fracture at 260 °C as observed by Williams (1967*c*) for an Al (20 mass %)–Zn alloy heat-treated to give precipitate free zones about 4 μm wide adjacent to the grain boundaries. The applied tensile stress was numerically equal to the independently superimposed hydrostatic pressure. Failure at the higher tensile stress levels was mainly by triple point cracking.

It follows that $4\pi(1-\nu)\,\gamma\int_0^x \mathrm{d}c = \int_0^{t_{\mathrm{f}}} x^2\dot{\epsilon}_{\mathrm{s}}^2\,\mu t\,\mathrm{d}t$ and so the time to fracture is given on integration by

$$t_{\mathrm{f}}^2 \approx 8\pi(1-\nu)\,\gamma/\mu x\dot{\epsilon}_{\mathrm{s}}^2;$$

thus we may write

$$t_{\mathrm{f}} \approx (4/\dot{\epsilon}_{\mathrm{s}})\,(\gamma/\mu x)^{\frac{1}{2}}.$$

The above approaches are clearly very much simplified and leave out many factors that can influence the failure process. Nevertheless, they do indicate some important features that have been observed experimentally in studies of the failure of metals and alloys at elevated temperatures.

First, it is noted that $\dot{\epsilon}_{\mathrm{s}}\,t_{\mathrm{f}} \approx \epsilon_{\mathrm{f}}$ and this is a measure of the strain to fracture although it does not take into account the primary and tertiary creep stages (Feltham & Meakin 1959). It was mentioned earlier that although ductility minima occur, these minima are shallow. It is only when grain boundary sliding is absent or when grain boundaries may move transversely that substantial ductility can be restored and the latter situation is not included in the analyses.

A further factor is that for a given boundary sliding rate both the time and strain at fracture are decreased by a decrease in the interfacial energy. There is a good deal of evidence that indicates that the segregation of impurities at interfaces can reduce γ and consequently reduce the fracture strain. Further, both the above expressions suggest that brittleness is enhanced by an increase in grain size and this again is observed experimentally (Taplin 1973).

It would be wrong to regard these equations as a means of evaluating accurately the conditions under which complete interfacial separation occurs, even under the simplest case of a tensile stress. The principal reason is that they ignore other modes of deformation that occur concurrently and that may have a substantial influence on the strain energy distribution created by the crack and the interfacial sliding. Phenomenologically, it may be adequate to group these effects together in the term relating to the energy required to separate the interfaces but this would then give a somewhat artificial meaning to the term γ. Comparison of experimental results with the equations, however, does suggest that the value of γ is somewhat higher than the true surface energy and this may be related to the reasons indicated.

A further feature in the above analysis is that the applied tensile stress does work which includes a term due to the increase in crack volume. It is then clear that with superimposed hydrostatic pressure the energy required for the crack opening is consequently increased. This is indicated in figure 2.

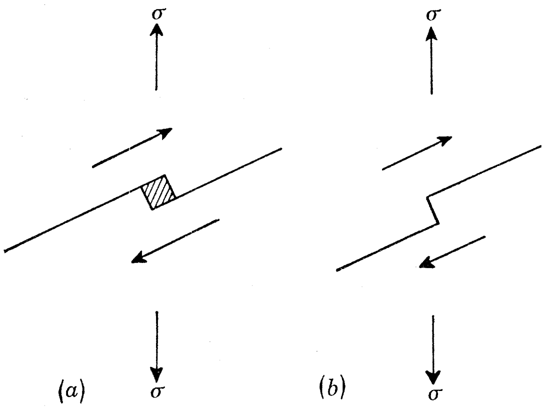

FIGURE 3. If ledges are present or can be created in grain boundaries, then they may have a different reaction to a shear stress depending on whether they are of type (*a*) where sliding may possibly cause cavity formation or type (*b*) where the ledge makes boundary movement more difficult.

THE CREATION OF CAVITIES BY INTERFACIAL SLIDING

Many observations relating to fracture during creep show that the growth of a crack from triple point junction, as previously indicated, is only one of the possible failure mechanisms (Gifkins 1959). Instead, discrete cavities may be formed on grain boundaries (Perry 1974), usually nearly perpendicular to the applied stress, and these cavities often preserve their geometrical characteristics as they grow. It is the nucleation of such cavities that has proved difficult to elucidate. This is partly because the critical nucleus size is believed, theoretically, to lie generally between 0.1 and 1 μm which is a particularly difficult region to explore experimentally, for nuclei are too small for light microscopes and too large to be contained in thin films that can be studied easily by transmission electron microscopes operating at 100 kV. Transmission electron microscopes operating at 1 MV hold more promise in this area (Fleck, Taplin & Beevers 1975) and a much used technique in cavity observation is to study the surface by scanning electron microscopy on fracture surfaces or on prepared sections that have in some instances been bombarded by ions to preserve the cavity shape as far as possible.

The character of these cavities has indicated that they have some features similar to those of grain boundary precipitates (Greenwood, Miller & Suiter 1954) and it was natural to assume that their formation and growth arose from precipitation processes; in this case by the agglomeration of vacancies. From considerations of generally accepted nucleation theory, however, it was soon clear that the vacancy supersaturation would not be sufficiently high to result in such precipitation unless nuclei were present (Cottrell 1961). Thus the problem of identifying such nuclei was recognized.

Geometrically, it is clear from figure 3 that ledges in grain boundaries could play a rôle in causing cavity nucleation. The problem then becomes one of deciding whether such ledges can exist or whether they can be caused by the deformation processes. It is conceivable that deformation within the grains may cause the impingement of slip bands at grain boundaries while sliding is taking place and the resultant step formation could result in cavity formation (Gifkins 1956). It is to be noted, however, that ledges are essentially of two types (Perry 1974). As illustrated in figure 3 one type of ledge is capable of opening, and the other simply locks the boundary.

An analysis was proposed (McLean 1963) whereby quite small ledges in grain boundaries that could exist in thermal equilibrium would be of sufficient size to cause cavity nucleation. On this basis, the rate of grain boundary sliding was calculated that would cause the cavity to continue to open and to resist sintering forces. Unfortunately, in the numerical calculations, the product was taken of the vacancy concentration and self-diffusion coefficient in the grain boundary and not simply the self-diffusion coefficient of atoms within the grain boundary and so the sintering rates were underestimated by many orders of magnitude (Harris 1965). Thus, the analysis served to show that there is considerable difficulty in developing a satisfactory quantitative theory to describe the circumstances whereby cavity nucleation may be caused by grain boundary ledges although much discussion remains in this field.

It is much easier to envisage the formation of cavities at particles within the grain boundaries (Cottrell 1961) although metallographic observations have not always been successful in positively identifying the presence of such nuclei where cavities have been observed.

Cavity nucleation at particles must clearly depend on the strength of the interfacial bond between the particles and matrix and the criterion for interfacial separation at particles has been approached in different ways.

The criterion to nucleate a crack is that the shear stress σ_s should exceed a value given by $\sigma_s = \{2\gamma\mu/\pi(1-\nu)\,x\}^{\frac{1}{2}}$. This expression can be written alternatively by considering the number of dislocations n of Burgers vector b piled up within the grain boundary such that fracture occurs when $\sigma_s = 2\gamma/nb$ (Smith & Barnby 1967). It is worth noting that this value is only one sixth of the value often quoted from earlier work.

An essential feature of an analysis of this type is that the high stresses generated by the sliding interface are supported by an infinite amount of material. It has been pointed out that this assumption may not be valid (Eborall 1961). It follows that nucleation may be possible at much lower stresses than those predicted by the above relation. A detailed calculation (Smith & Barnby 1967) indicates that the stress for nucleation decreases with the width x_b of the barrier to sliding provided that the width of such discontinuities is less than a critical value. The formula above then becomes modified to the form $\sigma_s = \{\pi\mu\gamma x_b/(1-\nu)\}^{\frac{1}{2}}/x$. It has been pointed out that this results in the possibility that the stress for nucleation may be only one twenty-fifth of that previously expected.

Another way of approaching the nucleation problem is simply to consider the influence of the interfacial sliding rate. In such analyses it is necessary to calculate the rate at which a potential nucleus would disappear by sintering forces and to compare this with the rate at which the interfacial sliding would cause opening to occur. Where the latter process was most rapid a nucleus would be formed. As previously pointed out, arguments such as these (Harris 1965) have been used to cast some doubt on ledge mechanisms of cavity formation and they also show that, even for particles, a rather fast rate of sliding is required. The velocity v of grain boundary sliding for cavity opening must be greater than a value given by

$$v > \{D_g w/r^2 \ln (b/r)\} \exp \{(2\gamma \Omega/kTr) - 1\}$$

in order for the cavity of radius r to open, where D_g is the grain boundary self-diffusion coefficient, w the grain boundary width, Ω the atomic volume, k Boltzmann's constant and T absolute temperature. For many metals and alloys this rate of grain boundary sliding is calculated to be somewhat faster than that observed in practice and this is true in situations where cavities have been observed. To overcome this problem, the inhomogeneous nature of sliding processes has been invoked (Chen & Machlin 1959) whereby sliding occurs discontinuously and so, at certain times, the sliding velocity can very considerably exceed that given by the above equation. Analyses of this kind clearly depend on the value of the interfacial energy but it is not true that where this is very low a minute sliding rate would be sufficient to cause permanent separation. Otherwise, any segregation of elements and particularly the formation of gas bubbles would be seen immediately to be an effective nucleus but this is known not to be the case.

Since no interfacial separation is required in the case of gas bubbles it may be considered that these would form ready nuclei for unlimited cavity growth. This, however, is not the situation, because, as the cavity enlarges, its surface area and consequently the surface energy must be increased. The gas pressure soon becomes of little assistance to cavity growth for as a bubble enlarges, the pressure, at constant temperature, falls inversely with the increase in bubble volume (Sykes & Greenwood 1965). Thus the cavity becomes subject to sintering forces unless it exceeds a given size. The evaluation of this size may be calculated by following the treatment of Hyam & Sumner (1962).

If an applied tensile stress σ acts perpendicularly to a grain boundary on which a bubble of radius r is situated, then the equilibrium condition can be written $\sigma = 2\gamma/r - p$ where p is the gas pressure within the bubble. If $r = r_0$ and $p = p_0$ when $\sigma = 0$, then $pr^3 = p_0 r_0^3$ and $p_0 = 2\gamma/r_0$. Hence

$$\sigma = 2\gamma/r - 2\gamma r_0^2/r^3.$$

By differentiation, $\mathrm{d}\sigma/\mathrm{d}r = 2\gamma(-1/r^2 + 3r_0^2/r^4)$. When $\mathrm{d}\sigma/\mathrm{d}r = 0$, the bubble will continue to expand, and this condition is met when $r = \sqrt{3}\, r_0$. Thus the tensile stress to cause the gas bubble to grow must be $\geqslant 4\gamma/3\sqrt{3}\, r_0$.

THE GROWTH OF CAVITIES

Where cavities are of an effective radius r greater than the critical size then convincing thermodynamic arguments can be presented to show that vacancy fluxes should permit continued growth principally of the cavities on those grain boundaries that are approximately perpendicular to the applied stress (Balluffi & Seigle 1957; Hopkin 1957). It may be anticipated that such a mechanism would be entirely adequate to account for cavity growth in a majority of cases. The true situation, however, does not seem to be so simple. There is a possibility that deformation mechanisms may enhance or may detract from cavity growth (Dyson 1976) and there are the further possiblities that grain boundaries may not be able to act as perfect sources of vacancies (Ashby 1969). It now seems that these factors are partly responsible for the variety of results and interpretations that have been reported. The implications of these will be considered later but, first, it is appropriate to consider the basis of the theory of cavity growth by vacancy condensation and the extent of the experimental support that it has so far received.

It is some 25 years since it was pointed out that cavities often have the character of precipitates and this led to the proposal that vacancy condensation was required for their growth (Greenwood *et al.* 1954). Support for such a model also arose independently from studies of inter-diffusion processes between two metals (Barnes & Mazey 1958), where the growth of voids could be correlated with the rate of condensation of vacancies as they flowed to compensate for the unequal diffusion of different atoms across the interface.

Some further implications of the vacancy growth model were subsequently considered and led to the prediction that a hydrostatic pressure numerically equivalent to the applied tensile stress would effectively prevent cavity growth by this process (Hull & Rimmer 1959). The basis of this approach was that near a grain boundary perpendicular to an applied stress σ the average vacancy concentration is increased by a factor $\exp(\sigma\Omega/kT)$ and near a cavity of radius r the vacancy concentration is increased by a factor $\exp(2\gamma\Omega/kTr)$. These equations form the basis of the criterion previously mentioned for cavity growth, namely that $\sigma > 2\gamma/r$. When a hydrostatic pressure P is applied, however, then the vacancy concentration at a grain boundary perpendicular to the principal tensile stress is modified to the form

$$\exp\{(\sigma-P)\,\Omega/kT\}.$$

This implies that if $P = \sigma$ there can be no cavity growth, at least by a vacancy condensation mechanism. There is considerable experimental support (Hull & Rimmer 1959; Ratcliffe & Greenwood 1965).

Some confusion has occasionally arisen between the rôle of vacancy flow in cavity growth and that in Nabarro–Herring or in Coble creep (Coble 1963), where vacancy fluxes are responsible for the deformation. The situation here is to evaluate the effective concentrations of vacancies at the respective grain boundaries in relation to the stress system and also the vacancy concentration at voids, and from these the relative importance of the various fluxes may be determined (Greenwood 1976). It is not difficult to show that most of the vacancy flux can take place simply by vacancy creation in grain boundaries in which the voids are situated. Conversely, this may be visualized as a process whereby atoms leave the cavities to cause their enlargement and subsequently plate on those grain boundaries that are nearly perpendicular to the tensile stress (Harris, Tucker & Greenwood 1974). This process can in itself cause a component of creep strain that can be significant. The entire situation, however, may be difficult to analyse in complete detail because of the overall adjustments that may be required to accommodate the increase in cavity volume (Dyson 1976).

When cavities are growing in this prescribed manner, then the stress distribution is modified. The stress is relaxed particularly at places on the grain boundary close to voids and is a maximum at points midway between the voids on grain boundaries that are nearly perpendicular to the principal tensile stress. It is this stress concentration gradient that gives rise to the vacancy concentration gradient and so to the vacancy flux. This aspect may also be related to the fact that cavities often remain quite distinct until a time near to the final fracture.

Although theories have generally been based on spherical cavities, the theory is not modified if cavities have crystallographic form or where the ends of the cavities are pointed to preserve an equilibrium configuration. In such an instance, effectively, it can be considered that surfaces of the cavity form sections of spheres and it is the radius of these spheres that corresponds to the effective radii of cavities in growth models. It may be highly significant, however, that such cavities need fewer vacancies per unit area of grain boundary that they occupy than do spherical cavities and so they may cause a more severe form of creep damage that leads to a reduced creep life. Segregation of those elements that cause a decrease in surface energy may be especially relevant here.

Several assessments have been made of the rate of cavity growth and the theory is now widely regarded as well established, although detailed treatments lead to slightly different results (Hull & Rimmer 1959; Weertman 1973; Raj & Ashby 1975; Speight & Beere 1975). Where x_c is half the cavity spacing, the equation which now appears to be most widely accepted is

$$\frac{\mathrm{d}v}{\mathrm{d}t}=\frac{2\pi D_g w\{\sigma-P-(2\gamma/r_0)\}\,\Omega}{kT}\bigg/[\ln\{x_c/r_0\}-\tfrac{1}{4}\{1-(r_0/x_c)^2\}\{3-(r_0/x_c)^2\}];$$

when cavities are relatively widely spaced and $v \gg \frac{4}{3}\pi r_0^3$, this formula reduces to the expression

$$\mathrm{d}v/\mathrm{d}t \approx 10D_g w\sigma\Omega/kT \quad \text{when} \quad P=0.$$

Although this equation clearly cannot hold throughout creep life it has the important characteristic that it predicts that the volume of an individual cavity would be expected to grow approximately proportionally with time over the greater part of creep life.

If it is further assumed that all cavities are nucleated at the start of creep and that creep failure occurs when a certain fraction of the grain boundary is covered by cavities, then it follows from this approach that the time to fracture is given approximately by $t_f \propto 1/\sigma$.

In practice, this relation has generally been found not to hold and the time to facture has been noted to be a much higher power of the applied tensile stress (Feltham & Meakin 1959). A further discrepancy between this theory and experiment has been that quite often (Woodford 1969; Needham, Wheatley & Greenwood 1975) the volume increase due to cavitation in creep is given by a formula of the type $\Delta V/V \propto \epsilon t\sigma^n$. This contrasts with the equation $\Delta V/V \propto \sigma t$ that would be expected to hold on the theory so far stated.

The simplest explanation of such discrepancies is that the cavities are not all nucleated at the start of the test (Greenwood 1963) and that a finite amount of deformation is necessary for their creation as well as a given magnitude of the applied stress.

When cavity nucleation is taken to be some function of strain then it is possible to obtain a much closer fit between theory and experiment. Most exhaustive studies in the field of high-temperature fracture have been carried out on copper. In some respects this may be considered an ideal material because it is well established that grain boundaries readily act as a source of vacancies which emerge with a negligible threshold stress (Greenwood 1976). There is good evidence for the operation of Nabarro–Herring and Coble creep processes in the régimes anticipated (Burton & Greenwood 1970) and results on α-particle injection into copper (Barnes 1960) and sintering experiments all lead to the view that vacancies may be readily emitted or absorbed at grain boundaries under conditions previously mentioned.

Although a number of aspects of the cavity nucleation mechanism remain obscure, an increasing amount of evidence (Kelly 1975; Morris 1977) is pointing to an approximate relation for a number of cavities given by $N \propto \sigma^2\epsilon$. With such an expression then it becomes clear that the volume increase, given by $\Delta V/V \propto \epsilon t \sigma^n$, is obtained from the cavity growth relation $\mathrm{d}v/\mathrm{d}t \propto \sigma$ and is in accord with experimental results when $n \approx 3$ (Needham & Greenwood 1975). This approach further permits an assessment to be obtained for variation of the time to then fracture with tensile stress and this is deduced as follows.

A given fraction of the area of grain boundaries is occupied when $Nv^{\frac{2}{3}}$ reaches a given value which, at fracture, is $N_\mathrm{f} v_\mathrm{f}^{\frac{2}{3}}$. This value is proportional to $(\sigma^2\epsilon_\mathrm{f})\,(\sigma t_\mathrm{f})^{\frac{2}{3}}$ and so, if $\epsilon_\mathrm{f} \propto \sigma^5 t_\mathrm{f}$, then $t_\mathrm{f} \propto 1/\sigma^{4.6}$.

One of the biggest difficulties that has been faced by cavity growth models by diffusion fluxes is in relation to the observation that the product $\dot{\epsilon} t_\mathrm{f}$, an approximate measure of the strain at fracture, is roughly constant (Feltham & Meakin 1959). This expression seems to imply that the cavitation mechanism should have an activation energy similar to that for the creep process which is often identical with the self-diffusion activation energy. Such a correlation would imply that vacancies grow by lattice diffusion rather than by grain boundary diffusion and, of the many proposals that have been put forward to reconcile the situation, some suggest that vacancies can only be produced in a grain boundary at the same rate as that permitted by deformation of the grains (Ishida & McLean 1967).

Although this explanation seems an entirely plausible one, it does not always appear necessary because steady pursuit (Morris 1977) of the separate factors controlling nucleation and growth of cavities in copper has revealed that the growth of individual cavities, taking into account temperature, follow a relation of the form $v \propto \sigma t \exp(-Q_\mathrm{g}/kT)$. Additionally, the formula for nucleation, $N \propto \sigma^2\epsilon$, has a negligible temperature dependence, at least over a range where the rate of grain boundary sliding does not differ substantially from the rate of overall deformation.

Taking again the criterion that failure takes place when a given area of grain boundary is occupied by cavities, this condition may be written

$$(\sigma^2\epsilon_\mathrm{f})\,\{\sigma t_\mathrm{f} \exp(-Q_\mathrm{g}/kT)\}^{\frac{2}{3}} = \text{a constant.}$$

Where creep life is related to the secondary creep stage, so that $\epsilon_\mathrm{f} \approx \dot{\epsilon} t_\mathrm{f}$, and the creep rate $\dot{\epsilon} \propto \sigma^5 \exp(-Q/kT)$, where Q is the activation energy for creep, then the fracture criterion becomes

$$\sigma^{\frac{8}{3}} \dot{\epsilon} t_\mathrm{f}^{\frac{5}{3}} \exp(-2Q_\mathrm{g}/3kT) = \text{a constant.}$$

Substituting for σ, we obtain the condition

$$\dot{\epsilon}^{\frac{23}{15}} t_\mathrm{f}^{\frac{5}{3}} \exp(8Q/15kT - 2Q_\mathrm{g}/3kT) = \text{a constant.}$$

Now the activation energy for creep, Q, is similar to that for self-diffusion under the conditions considered (Perry 1974) and the activation energy for grain boundary self-diffusion $Q_\mathrm{g} \approx 0.6Q$ (Gibbs & Harris 1969); thus the exponential term is approximately $\exp(2Q/15kT)$ which shows that the fracture condition is relatively insensitive to temperature. This is in agreement with most observations, except where the temperature is too low for grain boundary sliding or sufficiently high for grain boundary migration since these aspects are not incorporated in the analysis. It also follows that the approximate measure of the strain to fracture, ϵ_f, is roughly constant. It was noted that $\epsilon_\mathrm{f} \approx \dot{\epsilon} t_\mathrm{f}$ and the above analysis indicates approximately that, at fracture, $\dot{\epsilon} t_\mathrm{f}^{\frac{25}{23}}$ is constant.

It follows from this that the elongation to fracture is not strongly sensitive either to temperature or to the level of tensile stress. This is in accord with many experimental results and follows from the theory of cavity growth by vacancy diffusion where strain is necessary for cavity nucleation.

The question next arises of the possible extension of the analysis to interpret the influence of more complex stress systems on creep fracture. It is noted that the rate of vacancy flow to a cavity is dominated by the principal tensile stress and, where a hydrostatic pressure P is superimposed, this stress is $(\sigma - P)$. More precisely, the term $(\sigma - P - 2\gamma/r)$ must be used since $2\gamma/r$ may not be negligible when compared with $(\sigma - P)$. It is also assumed that the same stress term governs nucleation. In contrast, the secondary creep rate is governed primarily by the shear stress components σ_s which is thus not strongly dependent on P and may be taken to be some fraction of σ. Thus, the fracture criterion now becomes

$$(\sigma - P - 2\gamma/r)^2 \epsilon_f \{(\sigma - P - 2\gamma/r)\, t_f \exp(-Q_g/kT)\}^{\frac{2}{3}} = \text{a constant.}$$

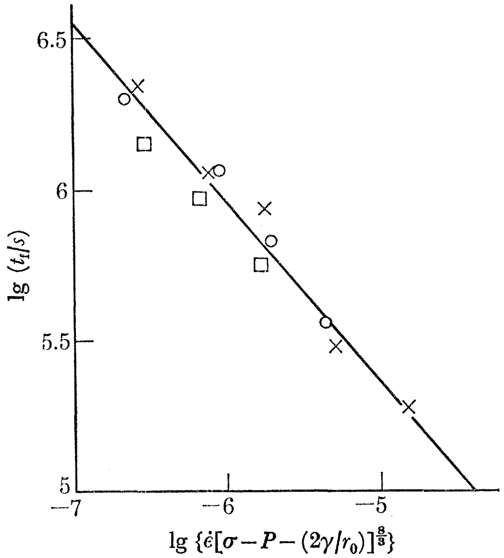

FIGURE 4. The relation between the time to fracture t_f in seconds and the parameter $\dot{\epsilon}[\sigma - P - (2\gamma/r_0)]^{\frac{8}{3}}$ for copper with grain size 230 μm at 500 °C, where σ is the applied tensile stress, P is the superimposed hydrostatic pressure and $(2\gamma/r_0)$ is the stress required for cavity nucleation, taken to be 5 MPa. $\dot{\epsilon}/\mathrm{s}^{-1}$ is the creep rate. The circles represent points for $\sigma = 20.7$ MPa; the squares represent points for $\sigma = 24.2$ MPa; the crosses represent points for $\sigma = 27.6$ MPa; P has values between 0.1 and 13.6 MPa. The graph has a gradient equal to -0.60 (Needham & Greenwood, to be published.)

At present, studies have only been made at constant temperature and so, neglecting the temperature term when substituting for ϵ_f, the criterion can be written:

$$(\sigma - P - 2\gamma/r)^{\frac{8}{3}} \dot{\epsilon} t_f^{\frac{5}{3}} = \text{a constant.}$$

Since $\dot{\epsilon}$ has been found to depend more strongly on P than originally suspected (Needham & Greenwood 1975) it is not possible at this stage to make further substitutions, but some experimental evidence supporting the above equation is presented in figure 4.

It is tempting to pursue these arguments to explain all situations of fracture at high temperatures. They have the merit that they can distinguish between the effects of different stress systems and particularly between the deviatoric and non-deviatoric stresses. There appears to be quite strong evidence, however, that the theory cannot be applied in all or even in the majority of practical situations. The reason is probably that even in the case of pure metals there is a small but finite threshold stress that is necessary for vacancies to be produced at the grain boundaries, but in solid solutions, and particularly in materials with precipitates (Harris 1976), the threshold stress may become quite substantial. At this point the theory of cavity growth by vacancy condensation must be linked closely with deviations from the theories of Nabarro–Herring and Coble creep. A number of theories have recently emerged to suggest why the action of grain boundaries as vacancy sources can be inhibited and these eventually must be expected to relate closely to the prediction of creep life.

Thus the situation concerning cavity growth is not fully resolved but there are some clear indications of the directions in which further work may be most profitably pursued.

Conclusions

Although the proposed interpretations of modes of failure in creep are far from complete, a number of features now seem reasonably well established. The areas of greatest uncertainty appear to lie in the initial stages of interfacial separation and in the ability of grain boundaries in alloys to produce vacancies when subjected to low stresses.

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Discussion

P. E. Evans (*Joint University and U.M.I.S.T., Metallurgy Building, Grosvenor Street, Manchester M*1 7*HS*). I feel that Professor Greenwood's hope that creep in metals may throw some light on creep in the mantle may not be *directly* realized. However, there is no doubt that the study of creep in metals has provided useful ideas on creep in ceramics. In turn I see ceramics as intermediate, both in terms of complexity and in terms of our understanding of their mechanical properties, between metals and rocks. I personally tend to consider ceramics as model rocks and I am sure that the study of creep in ceramics will increase our understanding of creep processes in the mantle.

H. H. Schloessin (*Department of Geophysics, University of Western Ontario, London, Ontario, Canada*). Is not the situation of stress concentration for faceted bubbles with sharp edges as described here similar to that of crack propagation by internal pressure in silicon–iron observed by Tetelman & Robertson (*Acta metall.* **11**, 1963)? There the internal pressure was generated by preferential accumulation of hydrogen in cracks. Stable crack propagation can be obtained if the accumulation of hydrogen by exsolution from the surrounding matrix keeps pace with the pressure drop due to the extension of the crack volume.

There is evidence from laboratory studies and actual rocks and minerals in the field that bubble nucleation and growth can occur under considerable hydrostatic pressure in the presence of partial melting. For example, basalts drilled from several hundred metres below the ocean floor show high concentrations of bubbles, usually thermal segregation vesicles or decorated vesicles. Their sizes vary from several millimetres to below 1 μm. As shown by scanning electron microscopy the walls of some decorated vesicles exhibit glassy linings which are riddled with multitudes of spherical bubbles all equal in size. Because of the higher solubility of gases in liquids compared with solids and, for most gases, decreasing solubility with decompression both liquid and solid phases become supersaturated with gases and it is thus possible for bubbles to nucleate during crystallization of partial melts under considerable hydrostatic pressures. The mechanical effects of gas bubbles may play a significant rôle in the ascent and emplacement of magmatic liquids. The large volume of gases indicated by high concentrations of vesicles in oceanic basalts gives evidence of very strong effervescence.

G. W. Greenwood. Dr Schloessin enquires about the possible similarity of the stress concentration for faceted bubbles with sharp edges and that which has been shown to arise during crack propagation by internal pressure in silicon–iron.

I should feel that there are some similarities and some significant differences. The similarities may arise because in both instances there may be a continued accumulation of gas in the bubbles or the cracks and their growth can be influenced if the gas pressure is maintained at a sufficiently high level. Nevertheless, there appear to be some significant differences. In creep processes, failure almost invariably occurs at grain or interface boundaries rather than along cleavage planes and so the shape of bubbles can be further modified by grain boundary tension. A still more important difference between the work mentioned on silicon–iron and observations on creep fracture is that diffusional movements appear to play a major rôle during creep and the tendency to form equilibrium shapes of bubbles can become predominant. Where bubble growth is controlled by diffusional processes, the tensile stress component is redistributed in a way that can be calculated and can become a maximum across grain boundaries in regions midway between adjacent bubbles.

The information that Dr Schloessin provides of bubble nucleation and growth in actual rocks and minerals could have close analogy with some forms of creep fracture in metals. It would be interesting to apply some of the theories that have been developed for the latter to the laboratory studies and observations that have been made on rocks and minerals to allow some quantitative predictions to be made of macroscopic behaviour.

The effects of pressure can be readily incorporated into theoretical treatments and the effects caused by crystallization and other phase changes should be amenable to interpretation to some extent.

I was interested to learn of the importance of gas precipitation in materials of the Earth and it seems that, as is now well appreciated in metallic systems, the influence of the gases may be all-important in some situations.

R. W. Cahn (*School of Applied Science, University of Sussex*). The tendency for pores to form at a grain boundary under uniaxial stress is a function of the inclination of the boundary to the stress vector. Is it feasible to enhance fracture resistance under creep conditions by controlling grain shape in such a way as to reduce or exclude the boundaries which are orientated favourably for pore formation?

I believe that creep ductility can also be enhanced by resorting to a very fine grain size. This might need to be stabilized by means of a disperse second phase. Would the presence of such a dispersion be likely to generate so many extra pores that this strategy would prove counterproductive?

G. W. Greenwood. It is quite feasible to enhance the resistance to creep fracture by controlling grain shape and size. It is generally true that cavities form most readily on grain boundaries perpendicular to the maximum principal tensile stress and that they have the most deleterious effect when they occur on such boundaries. Further, when cavities are situated on grain boundaries of large area, then they present the most serious limitation to creep ductility. Thus there is the immediate possibility of using materials with a microstructure consisting of elongated grains where only grain boundaries of relatively small area are orientated nearly perpendicularly to the tensile stress.

These comments also link up with Professor Cahn's second question. Creep ductility can be enhanced simply by decreasing the grain size of a material and it may be thought that a material with fine equiaxed grains might be entirely adequate. There is, however, the allied problem that fine equiaxed grains often result in material of reduced creep strength and so a careful balance must be made in alloy design to reach a suitable compromise between creep strength and ductility. From this point of view, grains elongated in the direction of the tensile stress, but with a relatively small area of cross-section perpendicular to this stress, may have the best combination of properties. Such structures, however, may still require second phase particles for stabilization. The presence of such particles can cause a nucleation of pores in the material and so reduce the creep ductility, but a good deal remains to be learnt in this area since not all second phase particles seem capable of nucleating pores.