

AN OVERVIEW OF FAILURE MECHANISMS IN CREEP

G. W. Greenwood*

An assessment of the rate of build up of "damage" resulting from the formation and growth of cavities and cracks is necessary for the prediction of creep life. Quite small changes in chemical composition and heat treatment and their influence on microstructure and segregation can induce large changes in behaviour. Fracture processes are closely linked with mechanisms of deformation and with the form of the applied stresses. Internal stress redistribution is an important feature and depends on the distinctive contributions of vacancy fluxes and dislocation motion.

INTRODUCTION

Failure at elevated temperatures is always influenced by the presence or generation and development of defects (1). These may be: macroscopic, in the form of cracks, notches or large holes; microscopic, in terms of microstructural features, such as grain size, shape, inclusion type and distribution; or crystallographic, with regard to dislocations, vacancies, segregated elements and the structure of grain boundaries.. If metallic materials contained none of these defects, except for thermal vacancies that are always present, (and materials approaching such quality can now be made, albeit on a small scale), temperature would have only a minor effect on their mechanical properties. Their elastic range would then be great and only decreased by a factor of about two from temperatures near zero to the melting point. This factor follows directly from the values of Young's and Shear Moduli which typically change by this amount. But for virtually all engineering materials required to operate at elevated temperatures the situation is totally different from one that is dominated by elastic behaviour.

*Department of Engineering Materials, University of Sheffield, SHEFFIELD S1 3JD UK

At elevated temperatures elastic limits are almost meaningless since deformation is time dependent and anelastic recovery presents further problems. Threshold stresses, below which no permanent deformation can occur, have frequently been recorded but their validity is subject to the sensitivity of measurement. Processes leading to fracture are inextricably linked with the mechanism of deformation. For fracture to occur, two conditions must be met: firstly the thermodynamic requirement of an overall decrease in free energy and secondly an atomic process whereby interfacial separation can take place. There are no further conditions and so mechanisms range from those applicable at low temperatures to entirely different modes that become operable through the changing form of deformation when the temperature is raised. An extreme example of lies in the enormous achievable ductilities in tension of superplastic alloys. Here, interfacial separation is difficult, but flow in a nearly viscous manner is easy, when a very small equiaxed grain size is maintained. The strain rate then has a low stress sensitivity which retards neck formation through which failure ultimately occurs after extensive elongation. In complete contrast, some materials in service have failed under low stresses at very low strains, illustrating the diversity of behaviour at elevated temperatures.

Direct applications of fracture mechanics that have become well established at low temperatures do not cover all the features of high temperature behaviour. For example, numerous failures at low strains, often $< 1\%$, have occurred in low alloy steel pipelines in steam raising plant, particularly at welds, after many years' service under relatively small stresses (2),(3). These same alloys display tensile ductilities more than an order of magnitude greater under much higher stresses at normal temperatures. This is one example of the extreme difference in the energy input sufficient to cause fracture, brought about entirely by a difference in temperature.

Since increasing temperature generally leads to easier flow through greater variety of atomic movement, it is not immediately clear why a decrease in strain to fracture should arise to give some ductility minimum, which occurs typically just above half the absolute melting temperature for many materials.. Two additional deformation processes, however, are significant under these conditions: one is the increasing significance of vacancy fluxes, driven by the applied stress: the other is opportunity for grain boundaries to slide. Either or both of these processes can contribute to an atomic mechanism for fracture when the thermodynamic condition is also met. The analysis of fracture at high temperatures of alloys used in long term service and the prediction of their safe and economic service life require a critical appreciation of interactive effects. Of particular significance is the form of redistribution of internal stress.

STRESS REDISTRIBUTION AT ELEVATED TEMPERATURES

At normal temperatures, the prediction of fracture relies heavily on an evaluation of the stress distribution in relation to the geometry of macroscopic defects and the mechanical properties of the material. Within the elastic range, the stress at the crack tip is described by the stress intensity factor $K=Y\sigma\sqrt{a}$ where Y is a non dimensional function of the form of loading and the crack and specimen geometry, a is the crack length and σ is the applied stress. At elevated temperatures, the situation is more complex because the mobility of crystal defects allows a progressive redistribution of the stress as illustrated

in Fig.1. Many problems are presented in attempting to evaluate the time dependence of this redistribution. Deformation by creep is not linear with time, except in the steady state, nor is the stress dependence, and distinctive mechanisms control its occurrence through the behaviour of crystal defects

Quite different conditions arise depending upon whether the deformation is governed principally by lattice dislocation movement or by stress induced vacancy fluxes (4). These conditions are further influenced by the form of any macroscopic defects that may be present or by microscopic cavities or cracks that may be generated during the creep life. Cavities are associated with the creation of volume and thus have especial sensitivity to the hydrostatic component of stress (5). Since cavities and cracks at elevated temperatures are almost invariably situated on grain boundaries, the microstructural characteristics and any anisotropy in the material are also significant (6) with respect to the principal stress directions.

The ability of grain boundaries to provide a source of vacancies, together with a path for their enhanced diffusion, may determine the conditions for cavity growth. This feature can be analysed in detail and contrasts the extreme difference in mechanisms that can characterise failures at elevated temperatures from those that occur at low temperature. Moreover, the local stress distributions that can be established concurrently with cavity growth are entirely different from that illustrated in Fig.1 and this aspect will next be considered.

CAVITATION

Cavity Growth

One sign of impending failure, through microstructural observations of cavitation, has long been recognised. As in Fig. 2, cavities are situated along grain boundaries oriented nearly normal to the principal tensile stress. They sometimes assume crystallographic facets only showing a crack-like morphology (7) when they subsequently coalesce. It is instructive first to present an outline of the theory of growth by vacancy condensation (8) because it illustrates how the stress patterns can develop that are quite different from those on which low temperature fracture toughness is evaluated. The theory has received experimental support but special conditions are necessary for its operation. Its extension, range of applicability and limitations will be discussed later.

The chemical potential ϕ of a vacancy in a grain boundary perpendicular to a tensile stress σ is given by $\sigma\Omega$, where Ω is the atomic volume. At the surface of a hole of cavity of radius r , the potential is $2\gamma\Omega/r$. (An effective value of r can still be defined when cavities are faceted and when shape is altered by grain boundary intersections). The potential gradient $\nabla\phi$ leads to a flux $j = (D_g w / kT\Omega) \nabla\phi$ where D_g is the grain boundary self diffusion coefficient and w is the grain boundary width. The equation of continuity requires that $\nabla_j + \beta = 0$, where β is the vacancy creation rate per unit area of grain boundary. Combining these equations, we have :-

$$\nabla^2\phi + (\beta kT\Omega / D_g w) = 0 \quad (1)$$

Solving this equation for radial symmetry for $r < x < \lambda$ to obtain the variation of ϕ with the distance x from the cavity centre, where adjacent cavities in the grain boundary are a distance 2λ apart, and noting that $d\phi/dx = 0$ at $x=\lambda$, we obtain :

$$\phi = (2\gamma/r)\Omega - (\beta kT\Omega/4D_g w)[x^2 - r^2 - 2\lambda^2 \ln(x/\lambda)] \quad (2)$$

Now the local stress σ_x at any point on the boundary is equal to ϕ / Ω and so σ_x at a position x is given by equation (2) such that the stress distribution is as shown in Fig 3(a). We note the striking contrast of this distribution of internal stress from that normally envisaged in a fracture mechanics approach and illustrated in Fig 1. The total forces across the boundary must balance and so the absolute values of σ_x can be determined by integration. The flux of vacancies, of course, is balanced by a counterflow of atoms leaving the cavity. Thus the rate of increase of each cavity volume $dv/dt = \pi\lambda^2\beta\Omega$ under applied stress σ and evaluating β from equation (2), we obtain :-

$$dv/dt = 2\pi D_g w [\sigma - (2\gamma/r)\Omega] / kT f(r/\lambda) \quad (3)$$

$$\text{where } f(r/\lambda) = \ln(\lambda/r) - [1 - (r/\lambda)^2][3 - (r/\lambda)^2] / 4$$

The acquisition of vacancies from the grain boundary is accompanied by a deposition of atoms that effectively plate out on the boundary by their attachment to the grains on either side. This gives rise to a strain rate $\dot{\epsilon} = \beta\Omega/d$ where d is the grain size. Thus, substituting for β , we obtain :-

$$\dot{\epsilon} = \beta\Omega/d = (dv/dt) / \pi\lambda^2 d \quad (4)$$

If there are N cavities per unit volume, then $N \approx 1/\pi\lambda^2 d$ and the rate of density change of the material $dp/dt = d/dt (V_c / V_s)$ where V_c is the total volume of all the cavities and V_s is the volume of the material in which they are situated. Thus, it follows from this that $V_c / V_s = Nv \approx v/\pi\lambda^2 d$ and so :-

$$dp/dt \approx \dot{\epsilon} \quad (5)$$

This equation has been validated for copper (9) under low stresses with cavities that have been pre-nucleated, initially in the form of small gas bubbles introduced by hydrogen reaction with small cuprous oxide particles. The fracture condition can be derived from the increasing area of grain boundary that the growing cavities remove.

This approach, however, cannot be widely used and in many cases it may be inadequate for one or more of four principal reasons; (1) the nucleation stage is not included in the analysis, (2) lattice dislocation movement can alter the form of the internal stress redistribution, (3) grain boundaries are not always able to operate as perfect vacancy sources and (4) it does not take into account pre-existing or developing crack-like defects. These aspects will be individually considered.

Cavity Nucleation

The precise mechanisms of cavity nucleation remain one of the most difficult areas to unravel. Any pre-existing nuclei must exist at a substantial size (1) equivalent to an effective radius $r \geq 2\gamma / \sigma$ where γ is the surface energy of the material per unit area and σ is the stress. For many materials, γ is typically of the order of 1 J/m^2 and so for $\sigma = 50 \text{ MPa}$, r is about 40 nm . In transmission electron microscopy, for convincing observations, nuclei must be entirely contained within films sufficiently thin for electron penetration and problems are compounded by wide nucleation spacing. Small angle neutron scattering techniques are appropriate in this range and have provided some useful but limited information. Scanning electron microscopy is useful, especially when it is possible to fracture the material along grain boundaries on which cavities are situated. Particles have often been seen associated with cavities, thus revealing the origin of nucleation.

The results of different workers have not always led to similar conclusions (2). There is some general indication, however, that in a variety of steels sulphides often form the easiest sites for nucleation, oxides are the next most potent with carbides and carbonitrides less effective. Gas bubbles have also been considered as nucleants because interfaces are then already separated. It is important to appreciate nevertheless that a critical minimum bubble size is necessary for a bubble to act in this way. The reason is that, as the bubble expands, the internal gas pressure drops to a level where it is of little assistance to cavity growth. Detailed analysis (10) shows that a bubble size $> 4\gamma / 3\sqrt{3}\sigma$ is necessary, so that a gas filled bubble is still required to be nearly half the size of an unbonded precipitate to be equally effective in nucleation.

In the arguments so far, the stress level required for nucleation has been related to the overall applied stress but stress concentrations may occur that substantially alter the requirements. This remains a difficult area to explore, because the rate of stress relaxation is also a factor and the lifetime of a nucleus is relevant before sintering forces cause re-bonding and render the nucleus ineffective.

For effective nucleation it is also apparent that some overall strain must be incurred to provide the increase in volume that cavity formation demands. The extent to which strain plays a part in the nucleation process is not always clear but much evidence has accumulated to indicate that, in many cases, the number of cavities can be directly related to the overall strain (1). Such observations open up the possibility that second phase particles may not necessarily be required as nucleants since the form of the intrinsic deformation may be such that space is provided for cavity growth to continue. The extent of initial interfacial separation must be large enough for a cavity growth mechanism to become operative before thermal sintering processes cause significant re-bonding.

For many years, discussion has ranged around the likelihood of ledges in grain boundaries, though there is now common agreement that such ledges would have to be at least of the order of 10 nm before they could act effectively as cavity nucleants. Such dimensions are not feasible as an equilibrium condition but speculation has continued about the possibility of their formation by grain boundary sliding or by interactive effects

such as the intersection of slip within grains with sliding grain boundaries. It is difficult to obtain definitive information on this point, but in several materials cavities have been observed to form without the identification of inclusions or second phase particles as nucleating agents.

The Coupling of Dislocation Motion and Vacancy Fluxes

It has been noted earlier that under small stresses, there is a smooth redistribution of internal stress as in curve (a) in Fig. (3) with a maximum value mid-way between adjacent cavities on the grain boundary. When the stress is larger, however, a level is reached whereby dislocation motion by glide and climb is activated. This level is attained in the central region between the cavities and creates a relatively uniform stress there, in which the creep rate is maintained by dislocation movement (11). Where grain boundaries continue to act as a source of vacancies, then the potential gradient driving their flux becomes steeper whilst being confined to a small region immediately surrounding the cavities. This has the effect of increasing the rate of cavity growth with respect to time but decreasing it with respect to strain, thereby becoming less effective in decreasing ductility. It is informative to analyse this situation in more detail.

As indicated in Fig. (3)(b), each cavity now only draws in vacancies from a region of the grain boundary up to a distance m from its centre and so its rate of growth dv/dt is given by equation (3) where the half distance between the cavities λ is replaced by m . To pursue this analysis, the stress levels are calculated by equating the total stress across the grain boundary to the applied stress and the ensuing creep rate in the dislocation creep region $\dot{\epsilon} = C(\sigma_m/G)^n$ where G is shear modulus and C and n are constants. We have the further relationship that creep rate $\dot{\epsilon} = \beta\Omega/m$ caused by vacancy flux must equal that by dislocation motion to maintain continuity. When the stress level is sufficiently high for dislocation motion to occur over the whole region, such that $m \Rightarrow r$, then we find that :-

$$dv/dt = 2\pi r^3 C [\sigma \lambda^2 / G(\lambda^2 - r^2)]^n \tag{6}$$

It follows from this that for widely spaced cavities, where $\lambda \gg r$, then $dv/dt = 2\pi r^3 \dot{\epsilon}$ and so, approximately :-

$$\Delta V_c / V_c \approx \epsilon \tag{7}$$

This implies that the ratio of the increase in cavity volume to the initial cavity volume is roughly equal to strain. Thus dislocation creep gives a density change with strain much less than that which applies through equation (5) when vacancy fluxes are important

This result is of widespread application, for it is derived solely from the effect of a uniform deformation on the rate of opening up of cavities with strain. Thus it is essentially independent of temperature when vacancy fluxes are not introduced, though time to failure is then dominated by strain rate effects when the stress is high.

Of future significance is the distinction that must be made between the effects of different forms of stresses that are applied. When vacancy fluxes make the major

contribution to cavity growth such that equation (5) applies, the controlling stress is the tensile component that is approximately perpendicular to the grain boundaries on which cavity growth occurs. Several features follow from this. It implies that anisotropy of grain shape and the distribution of potential nuclei are important. Because a significant volume increase results from cavity growth, there is a large influence of hydrostatic pressure (6).

These features contrast strongly with the different behaviour when cavity growth is governed by deformation by dislocation movement. The stress under which equation (7) holds is effectively the Von Mises stress and so the fracture criteria are much less influenced by the orientation of individual stress components or by hydrostatic pressure.

Grain Boundaries as Vacancy Sources

The models that rely on cavity formation and growth by the diffusion and condensation of vacancies assume that vacancies can be produced at perfect grain boundary sources. The definition of such a source is the establishment of a vacancy concentration enhanced by a factor $\exp(\sigma_x \Omega / kT)$ at a point on a grain boundary acted on by a normal tensile stress σ_x . The assumption is linked with the ability of a material to deform in the manner first proposed by Nabarro (12). Evidence over several years has indicated that many materials, especially of high purity and without second phases, can behave in this way, (Copper and many of its alloys are examples), but such behaviour is by no means universal and Aluminium and its alloys are notable exceptions. For steels, the position is complex (2), in that their behaviour is intermediate and is highly sensitive to microstructure. The ability of grain boundaries to emit vacancies is to some extent inhibited in these materials. The rapidly increasing knowledge of grain boundary structure may help to resolve these problems. The concept of the climb of grain boundary dislocations has already been helpful in considering these situations.

Analyses have been presented of possible forms of inhibition of the effectiveness of grain boundary vacancy sources where it is the rate of vacancy production and not the diffusion control of their flux that may determine the rate of cavity growth. This is an inadequately understood area but may prove to be important in predicting creep life.

CRACKING AT TRIPLE POINTS

The occurrence of grain boundary sliding can lead directly to internal crack formation at triple points where three grain boundaries intersect (13). It can be a sole cause of failure or it may occur concurrently with cavity formation. The geometry of crack development under these conditions is easily understood but a fully quantitative description of the kinetics of crack growth is proving more difficult to achieve.

The most promising approaches have been based on analogies with dislocation pile-ups in slip processes for which the stress patterns are well established. Factors of importance are the length available for sliding and the applied stress. At high temperatures, the sliding length is closely related to the grain size and a time factor can be introduced through the rate of grain boundary sliding. Detailed analysis (13), making a number of approximations, leads to the estimation of a time to fracture given by :-

$$t_f \approx 2\gamma / \sigma \dot{\epsilon}_s d \quad (8)$$

where $\dot{\epsilon}_s$ is the rate of grain boundary sliding under applied stress σ , d is grain size and γ is the energy to separate unit area of interface. Though this formulation is clearly oversimplified, it does contain the correct influences of the relevant parameters. It illustrates the importance of reducing grain size to increase service life and this aspect has had major technological benefits resulting from improvements in welding processes. Since, typically, $\dot{\epsilon}_s$ is proportional to σ raised to some high power, it follows that the product $t_f \dot{\epsilon}_s$ is roughly constant, in support of the Monkman-Grant relationship when $\dot{\epsilon}_s$ is proportional to the creep rate.

The estimation of the value of γ remains one of the most difficult problems. In practice, it may vary substantially, being lowered by segregated elements even at very small overall concentration but being raised by the additional energy that is required to drive deformation processes as well as to separate the interface. Such problems are similar to those arising in low temperature fracture and for material with pre-existing cracks.

MATERIALS WITH NOTCHES OR SHARP CRACKS

The Effect of Notches

The sensitivity of the modes of cavity growth to the form of applied stress system are related to the effects caused by notches (14). The elastic stress distribution at a notch is well characterised and in test pieces of circular cross section a hydrostatic dilatation is created in the neck through the radial tensile stresses that result from an axial load. Creep deformation gradually leads to a more uniform stress distribution and the concept of a skeletal stress that remains constant around a specific annulus within the notch throat has proved valuable. In addition to these features, the growth of cavities in the neck region depends on the relative contributions made by vacancy fluxes or dislocation motion.

In the simplest situation where growth follows exclusively from the stress induced flux of vacancies, only the stress nearly perpendicular grain boundaries on which they are situated is significant. Thus, in a cylindrical notched material, only the axial stress is then effective.

For the other extreme behaviour, where vacancy fluxes play no part and cavity growth results entirely from dislocation motion, the radial stress component causes hydrostatic dilatation in the notch and can reduce the value of the Von Mises stress that is rate controlling. This is typical of many aluminium alloys and these are regarded as notch ductile.

For most steels and many other materials, behaviour in the presence of notches is less extreme in so far as both vacancy fluxes and dislocation motion appear to operate concurrently in varying proportions. These proportions depend sensitively on the composition and microstructural state of the material that govern the ability of its grain

boundaries to act as efficient vacancy sources. They also depend on the magnitude of the applied stress where a low value leads to a greater role played by vacancy fluxes until cavities are enlarged to an extent such that the effect of dislocation motion becomes more substantial.

Sharp Cracks

When a sharp crack is present perpendicular to a tensile stress, the useful service life of the material is dependent on the rate of growth of the crack. When the creep rate is very small, such that there is negligible stress relaxation, the stress at the crack tip is characterised by the stress intensity K . In these circumstances, the crack growth rate has been found to be approximately proportional to K raised to a power that is close to the stress exponent of the creep rate.

When the creep rate is not negligible, the situation is substantially altered, since K cannot then be appropriately defined because of the stress relaxation that takes place. Instead, it has been found useful to adopt an approach similar to that employed at ordinary temperatures through an extension of the concept of the contour integral J . At elevated temperatures, a parallel concept has been the introduction of the parameter C^* which is representative of the stress state ahead of the crack tip when creep is taking place. By analogy with J , C^* must be evaluated for a contour surrounding the crack. For a crack on the plane $y=0$ in a body of unit thickness in the z direction subjected to a stress perpendicular to the crack, C^* is given by

$$C^* = \int W^* dy - T (\partial \dot{u} / \partial x) ds \quad (9)$$

here W^* is the creep strain energy release rate per unit volume, T and \dot{u} are respectively the components of the traction and creep rate vectors and s is the length of the contour surrounding the crack tip.

For specific specimen and crack geometry and loading conditions numerical solutions are available for the calculation of C^* . Otherwise or alternatively, this parameter can be determined in a way that corresponds to the evaluation of J in fracture mechanics. Under creep conditions this leads to the a value of C^* equal to a dimensionless geometrical constant multiplied by the area under the curve of load versus displacement rate divided by the specimen thickness and uncracked ligament length. With these evaluations, it has been shown (15) that the crack growth rate is proportional to C^* raised to a power slightly below unity, with the constant of proportionality dependent upon the material.

Considerations of this kind have assisted in the development of useful practical procedures for the estimation of remaining service life in materials in which maximum flaw dimensions are assumed.

CONCLUSIONS

The prediction of the service life of components operating at elevated temperatures depends upon a knowledge of the progressive build up of creep 'damage'. Simple

concepts of this 'damage', such as the fractional area of grain boundaries separated, have continued to form a useful basis for modelling fracture processes. Where it is possible to monitor this parameter in practical situations, such models can prove helpful in assessing service life.

The importance of detecting deterioration during creep through observation of the development of cracks and cavities is now well recognised. Where this can be achieved non-destructively it is of particular value. When coupled with appropriate analysis, it forms the most reliable indication of the extent of the safe lifetime in service.

The complexity of the mechanisms underlying the creation and progression of this 'damage' should not be underestimated. Quite small changes in chemical composition and heat treatment can lead to major changes in performance. In this brief overview emphasis has been placed on mechanical aspects. Here, detailed consideration of atomic mechanisms of deformation are shown to be important and illustrate the limitations of analyses based only on continuum models. Factors such as stress redistribution on a local scale can be significant in circumstances where grain boundaries are able to provide a source of vacancies. This also leads to 'damage' in terms of volume change as well as interfacial separation. The understanding of these features provides a basis for interpretation of the effects of multiaxial stresses, of different geometrical conditions and of anisotropic microstructures.

REFERENCES

- (1) Cadek, J., "Creep in Metallic Materials", Academia Press, Prague 1988.
- (2) Strang, A., Editor "Rupture Ductility of Creep Resisting Steels", The Institute of Metals, London. Book Number 522, 1991.
- (3) Earthman, J. C. & Farghali, A. H., Editors "Creep & Fracture of Engineering Materials & Structures", TMS Publications, Warrendale, Pa. 1997.
- (4) Edward, E. H. & Ashby, M. F. *Acta Metallurgica* Vol. 27, 1979, p1505.
- (5) Ratcliffe, R. T. & Greenwood, G. W. *Philosophical Magazine* Vol. 12, 1965, p59.
- (6) Greenwood, G. W. *Philosophical Magazine A* Vol. 43, 1981, p281.
- (7) Wei, R.P., Liu. H. & Gao, M., *Acta Materialia* Vol 6, 1997, p313.
- (8) Hull, D. & Rimmer, D. E., *Philosophical Magazine*, Vol. 4, 1959, p673.
- (9) Hanna, M. D. & Greenwood, G. W., *Acta Metallurgica*, Vol. 30. 1982, p719.
- (10) Hyam, E. D. & Sumner, G., "Radiation Damage in Solids" IAEA Vienna p323.
- (11) Beere, W. B. & Speight, M. V. *Metal Science*, Vol. 12. 1978 p172.

ECF 12 - FRACTURE FROM DEFECTS

- (12) Nabarro, F. R. N., "The Strength of Solids" The Physical Society, London, p 75.
- (13) Williams, J. A., Philosophical Magazine, Vol. 15, 1967 p1289.
- (14) Leckie, F.A. & Hayhurst, D. A., Acta Metallurgica, Vol.25, 1977 p1059.
- (15) Webster, G. A. Jour. of Strain Analysis, Vol.29, 1994 p215.

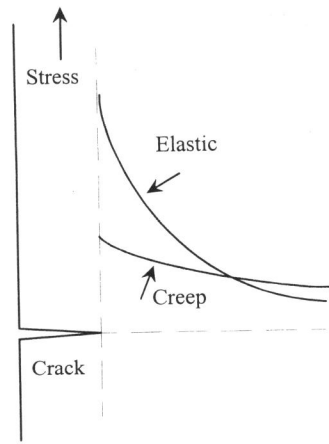


Fig. 1. Stress distribution near a crack

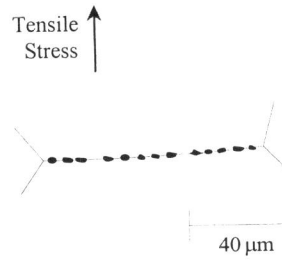


Fig. 2. Cavities on a grain boundary

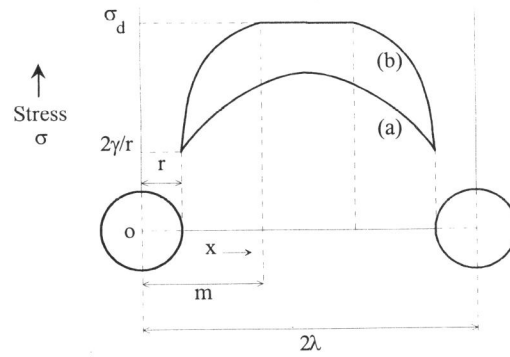


Fig. 3. Local stress redistribution between cavities