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Creep cavitation in metals

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Abstract

This review concisely describes the state-of-the-art of the understanding of cavity, or r-type void, formation during stages I and II (primary and secondary) creep in polycrystalline metals and alloys, particularly at elevated temperatures. These cavities can directly lead to Stage III, or tertiary, creep and the eventual failure of metals. There have been, in the past, a variety of creep fracture reviews that omitted important developments relevant to creep cavitation or are less than balanced in their discussions of conflicting ideas or theories regarding various aspects of cavity nucleation and growth. This concise, comprehensive, review discusses all of the important developments over the past several decades relating to both the nucleation and growth of cavities. The nucleation section discusses the details and limitations of the approaches based on “classic” nucleation theory, slip-induced nucleation as well as grain boundary sliding effects. Growth is discussed starting from the Hull–Rimmer diffusion controlled cavity growth (DCCG) model. This will be followed by refinements to DCCG by others. Next, there will be a discussion of plastic cavity growth and diffusion-plasticity coupling theories. This will be followed by the particularly important development of constrained cavity growth, initially proposed by Dyson, and probably under-appreciated. Other growth effects by grain boundary sliding will also be discussed. All of these mechanisms will be compared with their predictions in terms of creep fracture phenomenology such as the Monkman–Grant relationship. Finally, there will be a discussion of creep crack propagation by cavitation ahead of the crack tip.

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1. Introduction to creep plasticity

Creep of materials is classically associated with time-dependent plasticity under a fixed stress at an elevated temperature, often greater than roughly $0.5 T_m$, where T_m is the absolute melting temperature. The plasticity under these conditions is described in Fig. 1 for constant stress (a) and constant strain-rate (b) conditions. Several aspects of the curve in Fig. 1 require explanation. First, three regions are delineated: Stage I, or primary creep, which denotes that portion where [in (a)] the creep-rate (plastic strain-rate), $\dot{\epsilon} = d\epsilon/dt$ is changing with increasing plastic strain or time. In Fig. 1(a) the primary-creep-rate decreases with increasing strain, but with some

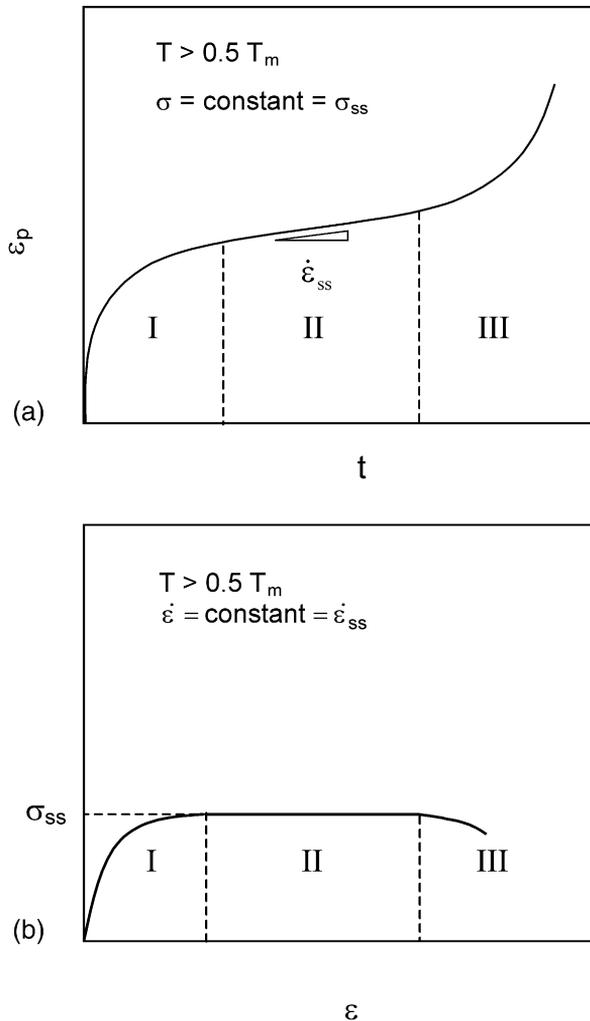


Fig. 1. Constant true stress and constant strain rate creep behavior in pure and Class M (or Class I) metals.

types of creep, such as solute drag with “three-power creep,” an “inverted” primary occurs where the strain-rate increases with strain. Analogously, in Fig. 1(b), under constant strain-rate conditions, the metal hardens, resulting in increasing flow stresses. Often, in pure metals, the strain rate decreases or the stress increases to a value that is constant over a range of strain. This phenomenon is termed Stage II, secondary, or steady-state creep. Eventually, an increase in the apparent strain-rate or decrease the flow stress is observed. This regime is termed Stage III, or tertiary creep, and leads to fracture. Sometimes, under constant stress, Stage I leads directly to Stage III and an “inflection” is observed. Thus, care must be exercised in concluding a mechanical steady-state (ss).

The term “creep” as applied to plasticity of materials likely arose from the observation that at modest and constant stress, at or even below the macroscopic yield stress of the material (at a “conventional” strain-rate), plastic deformation occurs over time as described in Fig. 1(a). This is in contrast with the *general* observation that a material deformed at ambient temperature (for example, $0.1\text{--}0.3 T_m$) shows very little plasticity under constant stress at or below the yield stress, again, at “conventional” or typical tensile testing strain-rates (e.g., $10^{-4}\text{--}10^{-3} \text{ s}^{-1}$). [The latter observation is not always true as it has been observed that some primary creep is observed (e.g., a few percent strain, or so) over relatively short periods of time at stresses less than the yield stress in some “rate-sensitive” and relatively low strain-hardening alloys such as titanium (Suri et al., 1999) and steels (Kassner et al., 1990).]

2. Background

Creep plasticity can lead to tertiary or Stage III creep and failure. It has been suggested that creep fracture can occur by w or wedge-type cracking, illustrated in Fig. 2a, at grain boundary triple points. Some have suggested that w -type cracks

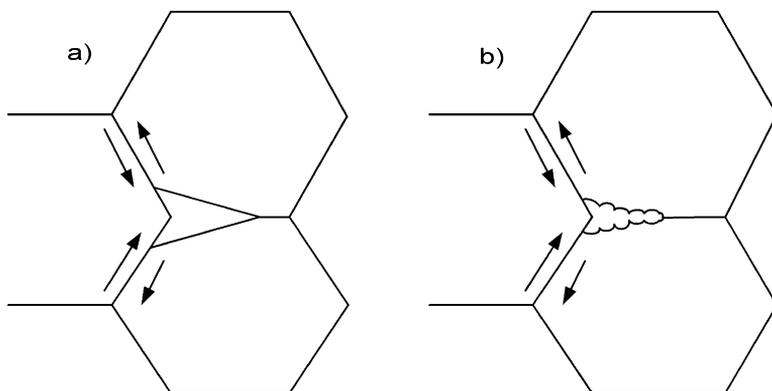


Fig. 2. (a) Wedge (or w -type) crack formed at the triple junctions in association with grain boundary sliding; (b) illustrates a wedge crack as an accumulation of spherical cavities.

form most easily at higher stresses (lower temperatures) and larger grain sizes (Waddington and Lofthouse, 1967) when grain boundary sliding is not accommodated. Some have suggested that the wedge type cracks nucleate as a consequence of grain boundary sliding. Another mode of fracture has been associated with r -type irregularities or cavities illustrated in Fig. 3. The wedges may be brittle in origin or simply an accumulation of r -types voids [Fig. 2(b)] (Stiegler et al., 1967). These wedge cracks may propagate only by r -type void formation (Chen and Argon, 1981a; Courtney, 1990). Inasmuch as w -type cracks are related to r -type voids, it is sensible to devote this short summary of creep fracture to cavitation.

There has been, in the past, a variety of reviews of creep fracture by Cocks and Ashby (1982a,b), Nix (1988), and Needleman and Rice (1980) and a series of articles in a single issue of a journal (Argon, 1983; Beere, 1983; Chen, 1983a; Goods and Nieh, 1983; Dyson, 1983; Nix, 1983), chapter by Cadek (1988) and particularly books by Riedel (1987) and Evans (1984), although most of these were published 15–20 years ago. This chapter will review these and, in particular, other more recent works. Some of these works, compiled in recent bibliographies (Mackerle, 2000), are quite extensive, of course, and this chapter is intended as a balanced and brief summary. The above two books are considered particularly good references for further reading. This review will particularly emphasize those works published subsequent to these reviews.

Creep fracture in uniaxial tension under constant stress has been described by the Monkman–Grant relationship (Monkman and Grant, 1956) which states that the

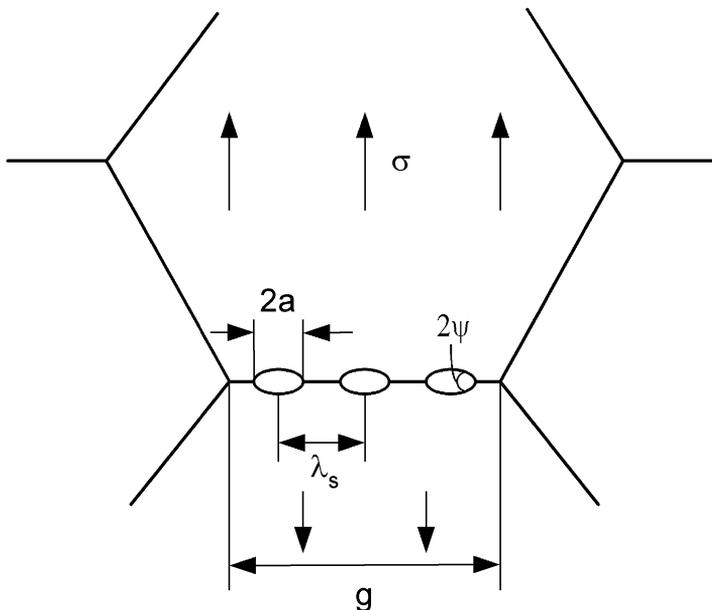


Fig. 3. Cavitation (r -type) or voids at a transverse grain boundary. Often, ψ is assumed approximately 70° .

fracture of creep deforming materials is controlled by the steady-state creep rate, ($\dot{\epsilon}_{ss}$)

$$\dot{\epsilon}_{ss}^m t_r = k_1 \quad (1)$$

where k_1 is sometimes referred to as the Monkman–Grant constant and m is a constant typically about 1.0. Some data that illustrates the basis for this phenomenological relationship is in Fig. 4, based on Feltham and Meakin (1959) and Evans (1984). Although not extensively validated over the past 20 years, it has been shown recently to be valid for creep of dispersion strengthened cast aluminum (Dunand et al., 1999) (where, interestingly, cavities nucleate at particles not located at grain boundaries). Modifications have been suggested to this relationship based on fracture strain (Dobes and Milicka, 1976). Although other more recent work on Cr–Mo steel suggests that Eq. (1) is valid (Molinie et al., 1991), the same data has been interpreted to suggest a modified vision (Dobes and Milicka, 1976). The Monkman–Grant phenomenological relationship(s), as will be discussed subsequently, places constraints on creep cavitation theories.

Another relationship to predict rupture time utilizes the Larson–Miller parameter (Larson and Miller, 1952) described by

$$LM = T[\log t_r + C_{LM}] \quad (2)$$

where T is the absolute temperature. This equation is not derivable from the Monkman–Grant or any other relationships presented. The constant C_{LM} is phenomenologically determined as that value that permits LM to be uniquely described in terms of the logarithm of the applied stress, usually illustrated by a $\ln \sigma$ vs LM graph. The technique appears to be currently used (Murty et al., 2002) for zirconium alloys. C_{LM} is suggested to be about 20, independent of the material.

One difficulty with these equations is that the constants determined in a creep regime, with a given rate-controlling mechanism, may not be reliable for extrapolation to the rupture times within another creep regime where the constants may actually change (Murty et al., 2002). The Monkman–Grant relationship appears to be more popular.

The fracture mechanisms that will be discussed are those resulting from the nucleation of cavities followed by growth and interlinkage, leading to catastrophic failure. Fig. 5 illustrates such creep cavitation in Cu, already apparent during steady-state (i.e., prior to Stage III or tertiary creep). It will be initially convenient to discuss fracture by cavitation as consisting of two steps, nucleation and subsequent growth.

3. Cavity nucleation

It is still not well established by what mechanism cavities nucleate. It has generally been observed that cavities frequently nucleate on grain boundaries, particularly on those transverse to a tensile stress (e.g., Chen and Argon, 1981a; Chen and Weertman,

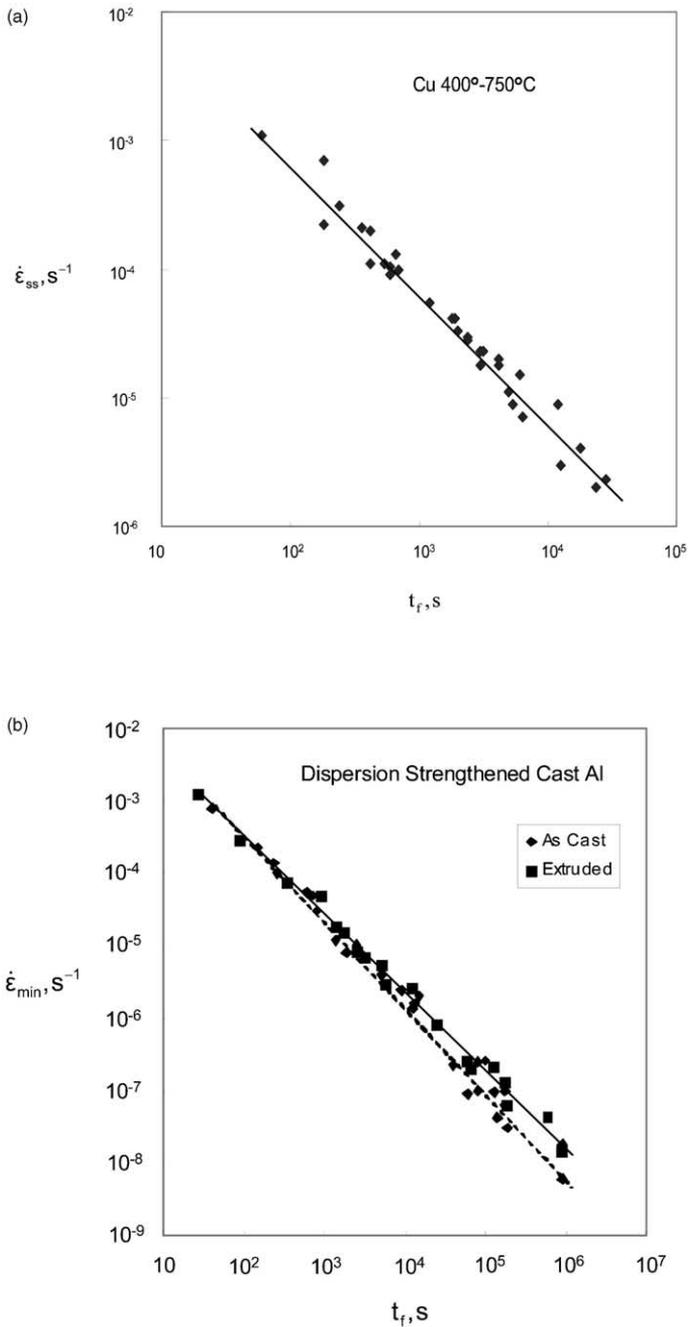


Fig. 4. (a) The steady-state creep-rate (strain-rate) versus time-to-rupture for Cu deformed over a range of temperatures, adapted from Evans (1984). (b) Dispersion strengthened cast aluminum, adapted from Dunand et al. (1999).

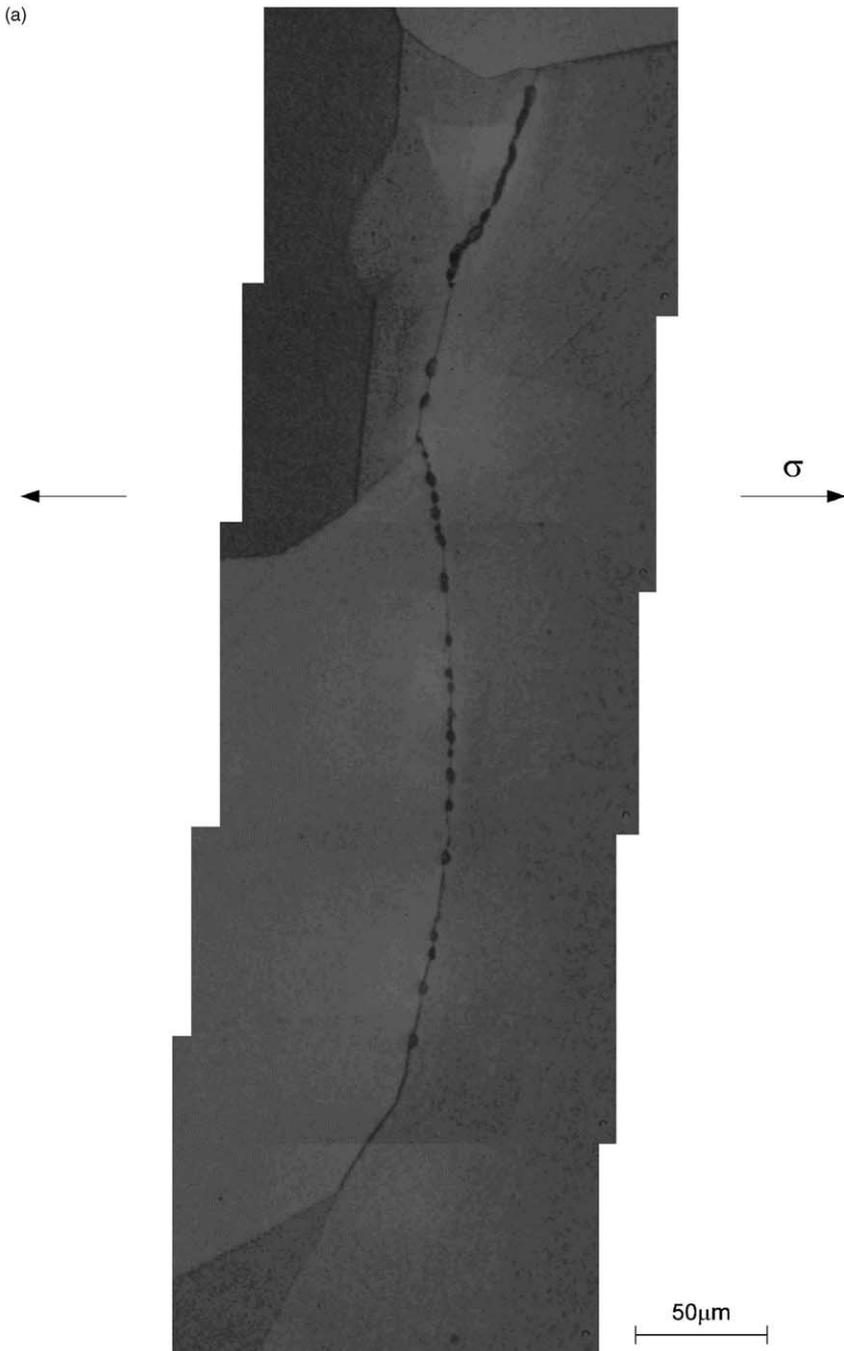
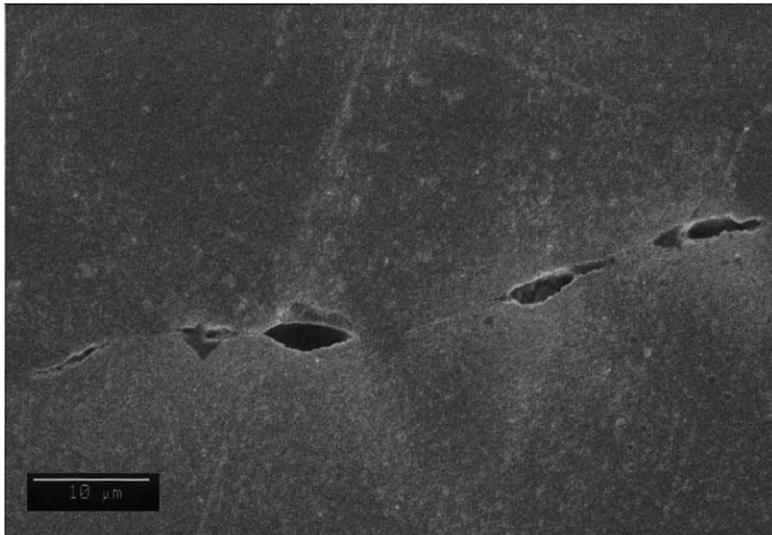


Fig. 5. Micrograph of cavities in Cu deformed at 20 MPa and 550 °C to a strain of about 0.04 (within stage II, or steady-state).



(b)

Fig. 5. (continued).

1984; Lim and Hu, 1994; Arai et al., 1996; Ayensu and Langdon, 1996; Hosokawa et al., 1999). In commercial alloys, the cavities appear to be associated with second phase particles. It appears that cavities do not generally form in some materials such as high purity (99.999% pure) Al. Cavitation is observed in lower purity metal such as 99% Al (Yavari and Langdon, 1983) (in high purity Al, boundaries are serrated and very mobile). The nucleation theories fall into several categories that are illustrated in Fig. 6: (a) grain boundary sliding leading to voids at the head (e.g., triple point) of a boundary or formation of voids by “tensile” GB ledges, (b) vacancy condensation, usually at grain boundaries at areas of high stress concentration, (c) the cavity formation at the head of a dislocation pile-up such as a Zener–Stroh mechanism [or anti-Zener–Stroh mechanism (Weertman, 1986)]. These mechanisms can involve particles as well as Fig. 6(d).

3.1. Vacancy accumulation

Raj and Ashby (1975) developed an earlier (Greenwood, 1952) idea that vacancies can agglomerate and form stable voids (nuclei) as in Fig. 6(b). Basically, the free energy terms are the work performed by the applied stress with cavity formation balanced by two surface energy terms. The change in the total free energy is given by,

$$\Delta G_T = -\sigma\Omega N + A_v\gamma_m - A_{gb}\gamma_{gb} \quad (3)$$

where N is the number of vacancies, A_v and A_{gb} are the surface areas of the void and (displaced) area of grain boundary, and γ_m and γ_{gb} are surface and interfacial

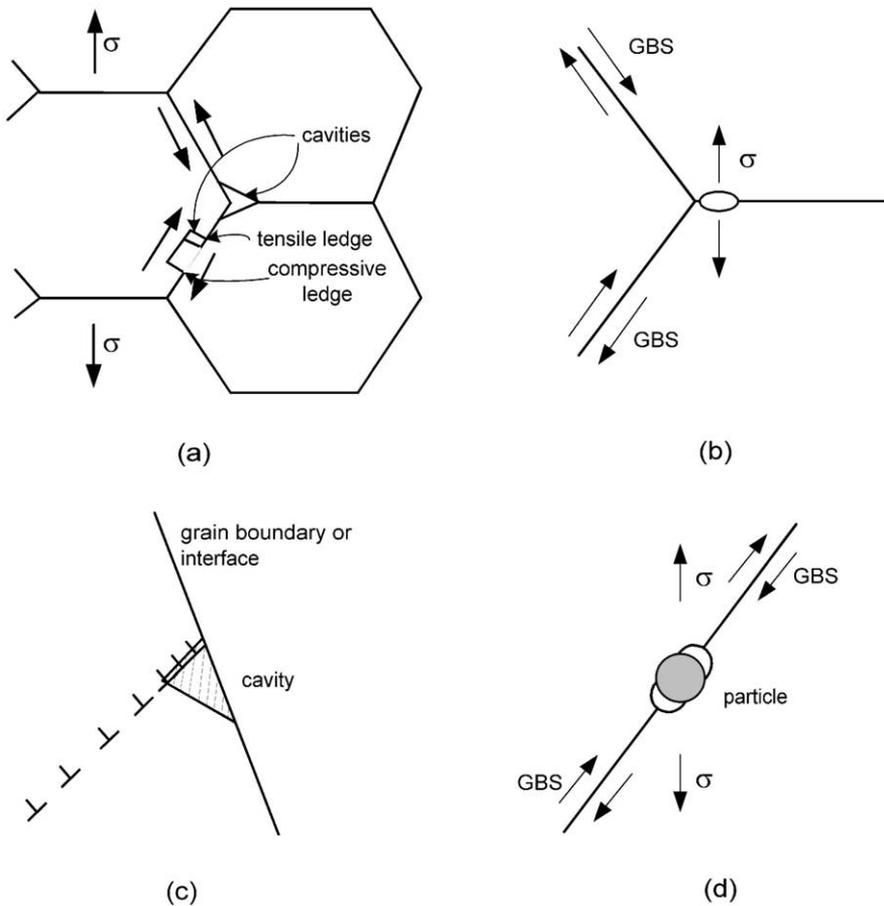


Fig. 6. Cavity nucleation mechanism: (a) sliding leading to cavitation from ledges (and triple points); (b) cavity nucleation from vacancy condensation at a high stress region; (c) cavity nucleation from a Zener–Stroh mechanism; (d) the formation of a cavity from a particle-obstacle in conjunction with the mechanisms described in (a–c).

energy terms of the metal and grain boundary. In all of the equations in this paper, all stresses and strain-rates are normal to the grain boundary.

This leads to a critical radius, a^* , and free energy, ΔG_T^* , for critical-sized cavities and a nucleation rate,

$$\dot{N} \cong n^* D_{gb} \tag{4}$$

where $n^* = n_o \exp(-\Delta G_T^*/kT)$, D_{gb} is the diffusion coefficient at the grain boundary and n_o is the density of potential nucleation sites. (The nucleation rate has the dimensions, $m^2 s^{-1}$.) [Riedel (1987) and Cadek (1988) have included a “Zeldovich” factor in Eq. (4) to account for “dissolution” of “supercritical” nuclei, $a > a^*$.]

Some have suggested that vacancy supersaturation may be a driving force rather than the applied stress, but it has been argued that sufficient vacancy supersaturations are unlikely (Riedel, 1987) in conventional deformation (in the absence of irradiation or Kirkendall effects).

This approach leads to expressions of nucleation rate as a function of stress (and the shape of the cavity). An effective threshold stress for nucleation is predicted. Evans (1984) suggests that the cavity nucleation by vacancy accumulation (even with modifications to the Raj–Ashby nucleation analysis to include, among other things, a Zeldovich factor) requires large applied (threshold) stresses (e.g., 10^4 MPa), orders of magnitude larger than observed stresses leading to fracture, which can be lower than 10 MPa in pure metals (Evans, 1984). Cavity nucleation by vacancy accumulation thus appears to require significant stress concentration. Of course, with elevated temperature plasticity, relaxation by creep plasticity and/or diffusional flow will accompany the elastic loading and relax the stress concentration. The other mechanisms illustrated in Fig. 6 can involve cavity nucleation by direct “decohesion,” which, of course, also requires a stress concentration. Vacancies are often suggested to originate from the grain boundaries.

3.2. Grain boundary sliding

Grain boundary sliding can lead to stress concentrations at triple points and hard particles on the grain boundaries, although it is unclear whether the local stresses are sufficient to nucleate cavities (Riedel, 1984; Cadek, 1988). These mechanisms are illustrated in Figs. 6(a), (b) and (d). Another sliding mechanism includes (tensile) ledges [Fig. 6(a)] where tensile stresses generated by GBS (grain boundary sliding) may be sufficient to cause cavity nucleation (Chen and Machlin, 1956), although some others (Fleck et al., 1975) believed the stresses are insufficient. The formation of ledges may occur as a result of slip along planes intersecting the grain boundaries.

One difficulty with sliding mechanisms is that transverse boundaries (perpendicular to the principal tensile stress) appear to have a propensity to cavitate. Cavitation has been observed in bicrystals (Gandhi and Raj, 1982) where the boundary is perpendicular to the applied stress, such that there is no resolved shear and an absence of sliding. Hence, it appears that sliding is not a necessary condition for cavity nucleation. Others (Chen, 1983a,b; Chan and Page, 1990; Ayensu and Langdon, 1996), however, still do not appear to rule out a relationship between GBS and cavitation along transverse boundaries. The ability to nucleate cavities via grain boundary sliding has been demonstrated by prestraining copper bicrystals in an orientation favoring GBS, followed by subjecting the samples to a stress normal to the previously sliding grain boundary and comparing those results to tests on bicrystals that had not been subjected to GBS (Chen and Machlin, 1957; Dunand et al., 1999). Extensive cavitation was observed in the former case while no cavitation was observed in the latter. Also, as will be discussed later, GBS (and concomitant cavitation) can lead to increased stress on transverse boundaries, thereby accelerating the cavitation at these locations. More recently, Ayensu and Langdon (1996) found a relation between GBS and cavitation at transverse boundaries, but also note a

relationship between GBS and strain. Hence, it is unclear whether GBS either nucleates or grows cavities in this case. [Chen \(1983b\)](#) suggested that transverse boundaries may slide due to compatibility requirements.

3.3. Dislocation pile-ups

As transverse boundaries may not readily slide, perhaps the stress concentration associated with dislocation pile-ups against, particularly, hard second phase particles at transverse grain boundaries has received significant acceptance ([Gifkins, 1956](#); [Dyson, 1983](#); [Yoo and Trinkaus, 1986](#); [Trinkaus and Yoo, 1987](#)) as a mechanism by which vacancy accumulation can occur. Pile-ups against hard particles within the grain interiors may nucleate cavities, but these may grow relatively slowly without short-circuit diffusion through the grain boundary and may also be of lower (areal) density than at grain boundaries.

It is still not clear, however, whether vacancy accumulation is critical to the nucleation stage. [Dyson \(1983\)](#) showed that tensile creep specimens that were prestrained at ambient temperature appeared to have a predisposition for creep cavitation. This suggested that the same process that nucleates voids at ambient temperature (that would not appear to include vacancy accumulation) may influence or induce void nucleation at elevated temperatures. This could include a Zener–Stroh mechanism [[Fig. 6\(c\)](#)] against hard particles at grain boundaries. [Dyson \(1983\)](#) showed that the nucleation process can be continuous throughout creep and that the growth and nucleation may occur together, a point also made by several other investigators ([Ratcliff and Greenwood, 1965](#); [McClintock, 1968](#); [Goods and Nieh, 1983](#); [Wu and Sandstrom, 1995](#)). This and the effect of prestrain are illustrated in [Fig. 7](#). The impact of cavitation-rate on ductility is illustrated in [Fig. 8](#). Thus, the nucleation process may be controlled by the (e.g., steady-state) plasticity. The suggestion that cavity nucleation is associated with plastic deformation is consistent with the observation by [Greenwood et al. \(1954\)](#), [Dyson et al. \(1976\)](#), [Watanabe and Davies \(1978\)](#), and [Nieh and Nix \(1980a\)](#) that the cavity spacing is consistent with regions of high dislocation activity (slip band spacing). [Goods and Nix \(1978a,b\)](#) also showed that if bubbles (nuclei) are implanted, the ductility decreases. [Davanas and Solomon \(1990\)](#) argue that if continuous nucleation occurs, modeling of the fracture process can lead to a Monkman-Grant relationship (diffusive and plastic coupling of cavity growth and cavity interaction considered). One consideration against the slip band explanations is that in-situ straining experiments in the TEM by [Dewald et al. \(1990\)](#) suggested that slip dislocations may easily pass through a boundary in a pure metal and the stress concentrations from slip may be limited. This may not preclude such a mechanism in combination with second-phase particles. [Kassner and Perez-Prado \(2000\)](#) performed creep fracture experiments on high purity Ag at about $0.25 T_m$. Cavities appeared to grow by (unstable) plasticity rather than diffusion. Nucleation was continuous, and it was noted that nucleation only occurred in the vicinity of high angle boundaries where obstacles existed, such as regions of highly twinned metal surrounded by low twin-density metal. High-angle boundaries without barriers did not appear to cavitate. Thus, nucleation (at least on

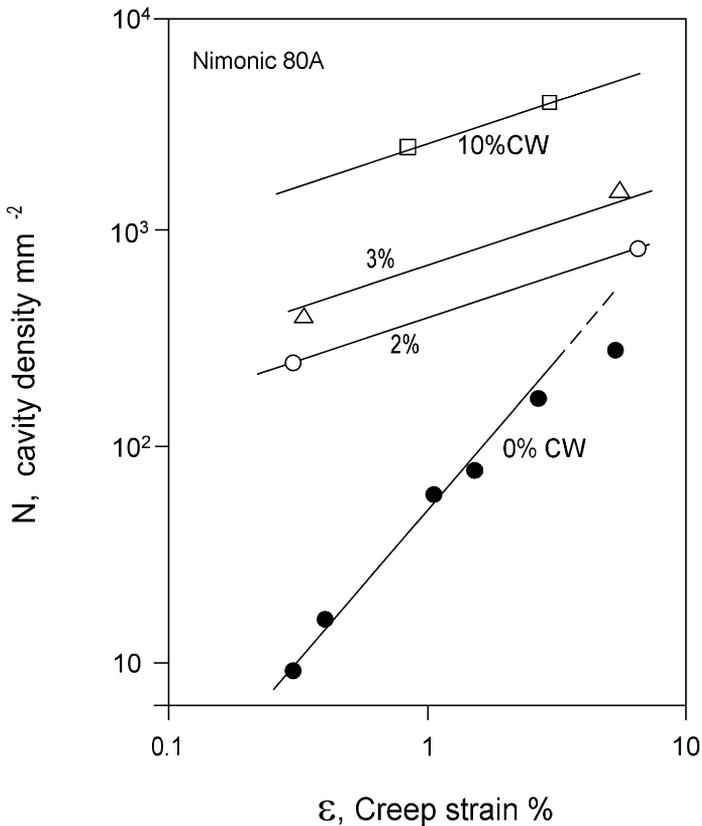


Fig. 7. The variation of the cavity concentration versus creep strain in Nimonic 80A (Ni–Cr alloy with Ti and Al) for annealed and pre-strained (cold-worked) alloy (adapted from Dyson, 1983). Cavities were suggested to undergo unconstrained growth.

transverse boundaries) appears to require obstacles and a Zener–Stroh or anti-Zener–Stroh mechanism appears the most likely mechanism.

3.4. Location

It has long been suggested that (transverse) grain boundaries and second phase particles are the common locations for cavities. Solute segregation at the boundaries may predispose boundaries to cavity nucleation (Argon, 1983). Some of the more recent work that found cavitation associated with hard second phase particles in metals and alloys includes Lombard and Vehoff (1990), Svoboda and Sklenicka (1990), Wu and Sandstrom (1995), Wei et al. (1997), George et al. (1998); Lee and Yu (1999), Oh et al. (1999) and Yousefiani et al. (2000). Second phase particles can result in stress concentrations upon application of a stress and increase cavity nucleation at a grain boundary through vacancy condensation by increasing the

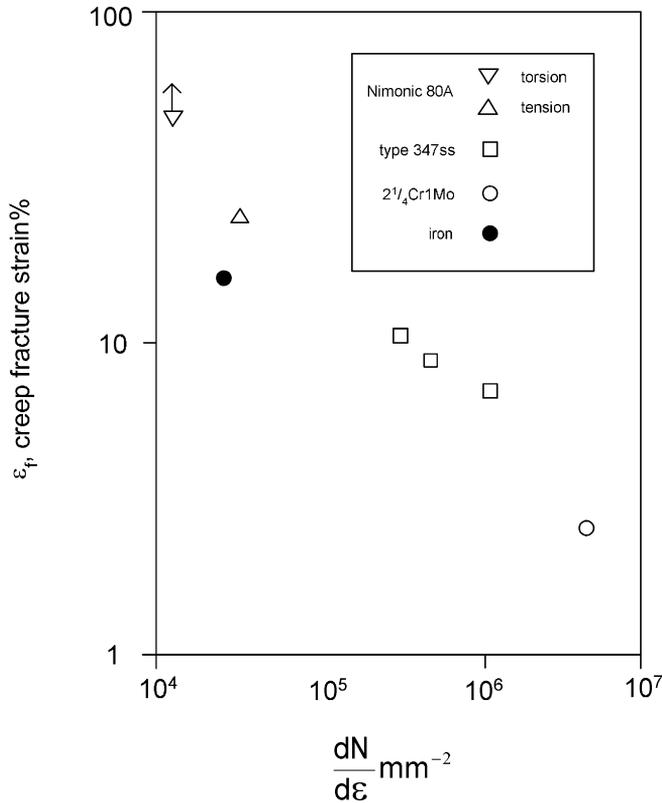


Fig. 8. Creep ductility versus the “rate” of cavity production with strain. Adapted from Dyson (1983) (various elevated temperatures and stresses).

grain boundary free energy. Also, particles can be effective barriers to dislocation pile-ups.

The size of critical-sized nuclei is not well established but the predictions based on the previous equations is about 2–5 nm (Cadek, 1988) which are difficult to detect. SEM under optimal conditions can observe (stable) creep cavities as small as 20 nm (Kassner et al., 1998). It has been suggested the small angle neutron scattering can characterize cavity distributions from less than 10 nm to almost 1 μm (Cadek, 1988). TEM has detected stable cavities to 3 nm (Randle, 1993). Interestingly, observations of cavity nucleation not only suggest continual cavitation but also no incubation time (Yang et al., 1984) and that strain rather than time is more closely associated with nucleation (Cadek, 1988). Fig. 8 illustrates the effect of stress states on nucleation. Torsion, for comparable equivalent uniaxial stresses in Nimonic 80, leads to fewer nucleated cavities and greater ductility than tension. Finally, another nucleation site may be important as damage progresses in a material is the stress concentration that arises around existing cavities. The initial (elastic) stress concentration at the cavity “tip” is a factor of three larger than the applied stress, and even

after relaxation by diffusion, the stress may still be elevated (Anderson and Shewmon, 2000) leading to increased local nucleation rates.

4. Growth

4.1. Diffusion—grain boundary control

The cavity growth process at grain boundaries at elevated temperature has long been suggested to involve vacancy diffusion. Diffusion occurs by cavity surface migration and subsequent transport along the grain boundary, with either diffusive mechanism having been suggested to be controlling, depending on the specific conditions. This contrasts with creep void growth at lower temperatures where cavity growth is accepted to occur by (e.g., dislocation glide-controlled) plasticity. A carefully analyzed case for this is described in Kassner et al. (1998).

Hull and Rimmer (1959) were one of the first to propose a mechanism by which diffusion leads to cavity growth of an isolated cavity in a material under an applied external stress, σ . A stress concentration is established just ahead of the cavity. This leads to an initial “negative” stress gradient (i.e., local stress decreases away from the cavity). However a “positive” stress gradient is suggested to be established due to relaxation by plasticity (Evans, 1984). This implicit assumption in diffusion-controlled growth models appears to have been largely ignored in later discussions by other investigators, with rare exception (e.g., Evans, 1984). The equations that Hull and Rimmer and, later, others (Raj and Ashby, 1975; Speight and Beere, 1975; Riedel, 1987) subsequently derive for diffusion-controlled cavity growth are similar. Basically,

$$J_{\text{gb}} = -\frac{D_{\text{gb}}}{\Omega k T} \nabla f \quad (5)$$

where J_{gb} is the flux, Ω the atomic volume, $f = -\sigma_{\text{loc}}\Omega$ and σ_{loc} is the local normal stress on the grain boundary. Also,

$$\nabla f \sim \frac{\Omega}{\lambda_s} \left(\sigma - \frac{2\gamma_m}{a} \right) \quad (6)$$

where “ a ” is the cavity radius, σ , the remote or applied normal stress to the grain boundary, and λ_s is the cavity separation. Below a certain stress ($\sigma_0 = \frac{2\gamma_m}{a}$) the cavity will sinter. Eqs. (5) and (6) give a rate of growth,

$$\frac{da}{dt} \cong \frac{D_{\text{gb}}\delta \left(\sigma - \frac{2\gamma_m}{a} \right) \Omega}{2kT\lambda_s a} \quad (7)$$

where δ is the grain boundary width. Fig. 9 is a schematic that illustrates the basic concept of this approach.

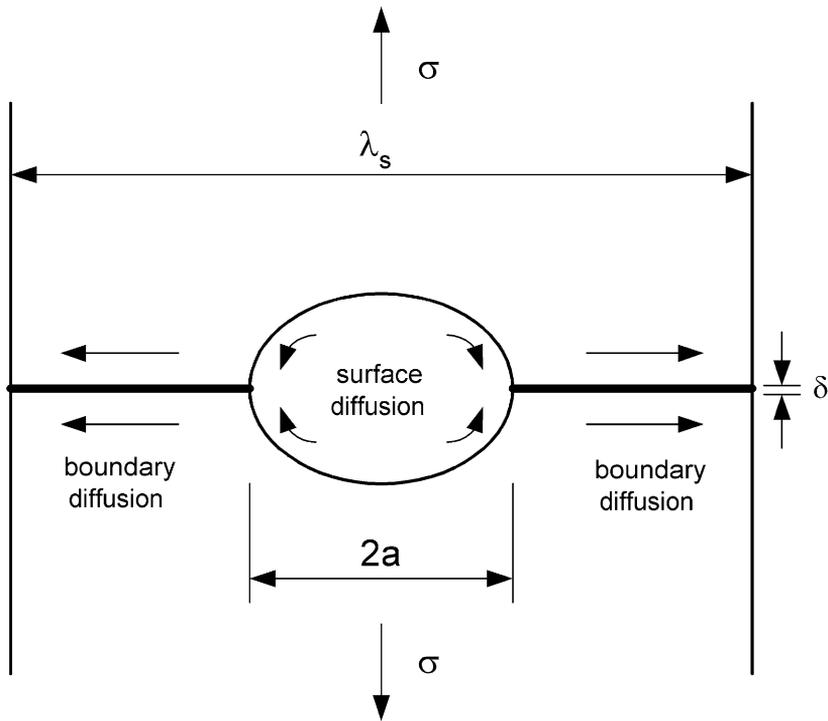


Fig. 9. Cavity growth from diffusion across the cavity surface and through the grain boundaries due to a stress gradient.

By integrating between the critical radius (below which sintering occurs) and $a = \lambda_s/2$,

$$t_r \cong \frac{kT\lambda_s^3}{4D_{gb}\delta\left(\sigma - \frac{2\gamma_m}{a}\right)\Omega} \quad (8)$$

This is the first relationship between stress and rupture time for (unconstrained) diffusive cavity growth. Weertman (1973), Raj and Ashby (1975), Raj et al. (1977), Speight and Beere (1975), and Riedel (1987) later suggested improved relationships between the cavity growth rate and stress of a similar form to that of Hull and Rimmer [Eq. (7)]. The subsequent improvements included modifications to the diffusion lengths (the entire grain boundary is a vacancy source), stress redistribution (the integration of the stress over the entire boundary should equal the applied stress), cavity geometry (cavities are not perfectly spherical) and the “jacking” effect, where atoms deposited on the boundary cause displacement of the grains. Riedel, in view of these limitations, suggested that the equation for unconstrained cavity growth of widely spaced voids is, approximately,

$$\frac{da}{dt} = \frac{\Omega\delta D_{gb}[\sigma - \sigma_o]}{1.22kT\ln(\lambda_s/4.24a)a^2} \quad (9)$$

where σ_0 is the sintering stress. Again, integrating to determine the time for rupture shows that $t_r \propto 1/\sigma$. Despite these improvements, the basic description long suggested by Hull and Rimmer is largely representative of unconstrained cavity growth. An important point here is a predicted linear stress dependence and an activation energy of grain boundary diffusion for Eqs. (7)–(9) for (unconstrained) cavity growth.

The predictions and stress dependence of these equations have been frequently tested (Goods and Nix, 1978a,b; Raj, 1978; Nieh and Nix, 1979, 1980b; Miller and Langdon, 1980; Needham and Gladman, 1980, 1986; Cane, 1981; Hanna and Greenwood, 1982; Svensson and Dunlop, 1982; Chen, 1983a; Mintz and Mukherjee, 1988; Nix, 1988; Cho et al., 1992; Broyles et al., 1996). Raj (1978) examined Cu bicrystals and found the rupture time inversely proportional to stress, consistent with the diffusion controlled cavity growth equations just presented. The fracture time for polycrystals increases orders of magnitude over bicrystals. Svensson and Dunlop (1982) found that in α -brass, cavities grow linearly with stress. The fracture time appeared, however, consistent with Monkman–Grant and continuous nucleation was observed. Hanna and Greenwood (1982) found that density change measurements in prestrained (i.e., prior cavity nucleation) and with hydrogen bubbles were consistent with the stress dependency of the earlier equations. Continuous nucleation was not assumed. Cho et al. (1992) and Needham and Gladman (1980, 1986) measured the rupture times and/or cavity growth rate and found consistency with a stress to the first power dependency if continuous nucleation was assumed. Miller and Langdon (1980) analyzed the density measurements on creep-deformed Cu based on the work of others and found that the cavity volume was proportional to σ^2 (for fixed T , t , and ϵ). If continuous nucleation occurs with strain, which is reasonable, and the variation of the nucleation rate is “properly” stress dependent (unverified), then consistency between the density trends and unconstrained cavity growth described by Eqs. (7) and (9) can be realized.

Creep cavity growth experiments have also been performed on specimens with pre-existing cavities by Nix and co-workers (Goods and Nix, 1978a,b; Nieh and Nix, 1979, 1980b). Cavities, here, were created using water vapor bubbles formed from reacting dissolved hydrogen and oxygen. Cavities were uniformly “dispersed” (unconstrained growth). Curiously, the growth rate, da/dt , was found to be proportional to σ^3 . This result appeared inconsistent with the theoretical predictions of diffusion controlled cavity growth. The disparity is still not understood. Interestingly, when a dispersion of MgO particles was added to the Ag matrix, which decreased the Ag creep-rate, the growth rate of cavities was *unaffected*. This supports the suggestion that the controlling factor for cavity growth is diffusion rather than plasticity or grain boundary sliding. Nix (1988) appeared to rationalize the three-power observation by suggesting that only selected cavities participate in the fracture process. As will be discussed later, it does not appear clear whether cavity growth in the Nix et al. experiments were genuinely unconstrained. That is, it is not clear whether only diffusive flow of vacancies controls the cavity growth rate.

4.2. Diffusion—surface control

Chuang and Rice (1973) and later Needleman and Rice (1980) suggested that surface rather than grain-boundary diffusion may actually control cavity growth (which is not necessarily reasonable) and that these assumptions can give rise to a three-power stress-relationship for cavity growth at low stresses (Nix et al., 1983; Nix, 1988).

$$\frac{da}{dt} \cong \frac{\Omega \delta D_s}{2kT\gamma_m^2} \sigma^3 \quad (10)$$

The result is an increasingly lenticular or crack-like cavity. At higher stresses, the growth rate varies as $\sigma^{3/2}$. The problem with this approach is that it is not clear in the experiments, for which three-power stress dependent cavity growth is observed, that $D_s < D_{gb}$. Activation energy measurements by Nieh and Nix (1979, 1980b) for (assumed unconstrained) growth of cavities in Cu are inconclusive as to whether it better matches D_{gb} versus D_s . Also, the complication with all of these growth relationships [Eqs. (7)–(10)] is that they are inconsistent with the Monkman–Grant phenomenology. That is, for common five-power law creep, the Monkman–Grant relationship suggests that the cavity growth rate ($1/t_f$) should be proportional to the stress to the fifth power rather than 1–3 power. This of, course, may emphasize the importance of nucleation in the rate-controlling process for creep cavitation failure, since cavitation may be controlled by the plastic strain (steady-state creep-rate). Of course, small nanometer-sized cavities (nuclei), by themselves, do not appear sufficient to cause cavitation failure. Dyson (1983) suggested that the Monkman–Grant relationship may reflect the importance of both (continuous) nucleation and growth events.

4.3. Grain boundary sliding

Another mechanism that has been considered important for growth is grain boundary sliding (GBS) (Evans, 1969). This is illustrated in Fig. 10. Here cavities are

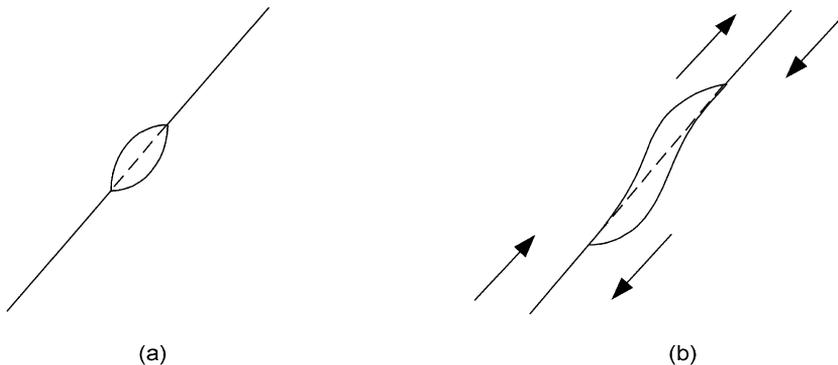


Fig. 10. Cavity growth from a sliding boundary (from Chen, 1983a).

expected to grow predominantly in the plane of the boundary. This appears to have been observed in some temperature-stress regimes. Chen (1983a) appears to have invoked GBS as part of the cavity growth process, also suggesting that transverse boundaries may slide due to compatibility requirements (Chen, 1983b). A suggested consequence of this “crack sharpening” is that the tip velocity during growth becomes limited by surface diffusion. A stress to the third power, as in Eq. (10), is thereby rationalized. Chen (1983b) suggests that this phenomenon may be more applicable to higher strain rates and closely spaced cavities (later stages of creep). The observations that cavities are often more spherical rather than plate-like or lenticular, and that, of course, transverse boundaries may not slide, also suggest that cavity growth does not substantially involve sliding. Riedel (1987) predicted that (albeit constrained) diffusive cavity growth rates are expected to be a factor of $(\lambda/2a)^2$ larger than growth rates by (constrained) sliding. Some more recent work on creep cavitation of dual phase intermetallics suggests, however, that sliding may affect growth (Chakraborty and Earthman, 1997).

4.4. Constrained diffusional cavity growth

Cavity nucleation may be heterogeneous, inasmuch as regions of a material may be more cavitated than others. Adams (1993) and Watanabe (1993) both suggested that different geometry (e.g., as determined by the variables necessary to characterize a planar boundary) high angle grain boundaries have a different tendencies to cavitate, although there was not agreement as to the nature of this tendency in terms of the structural factors. Also, of course, a given geometry boundary may have varying orientations to the applied stresses. Another important consideration is that the zone ahead of the cavity experiences local elongation with diffusional growth, and this may cause constraint in this region by those portions of the material that are unaffected by the diffusion (outside the cavity diffusion “zone”). This may cause a “shedding” of the load from the diffusion zone ahead of the cavity. Thus, cavitation is not expected to be homogeneous and uncavitated areas may constrain those areas that are elongating under the additional influence of cavitation. This is illustrated in Fig. 11. Fracture could then be controlled by the plastic *creep-rate* in uncavitated regions that can also lead to cavity nucleation. This leads to consistency with the Monkman-Grant relationship (Dyson, 1976; Rice, 1981).

Constrained diffusional growth was originally suggested by Dyson and further developed by others (Rice, 1981; Cocks and Ashby, 1982a,b; Tvergaard, 1984; Yousefiani et al, 2000; Delph, 2002). This constrained cavity growth-rate has been described by the relationship (Riedel, 1987)

$$\frac{da}{dt} \cong \frac{\sigma - (1 - \omega)\sigma_0}{\frac{a^2 k T}{\Omega \delta D_{gb}} + \frac{\sigma_{ss} \pi^2 (1 + 3/n)^{1/2} a^2}{\dot{\epsilon}_{ss} \lambda_s^2 g}} \quad (11)$$

where ω is the fraction of the grain boundary cavitated. This is the growth-rate for cavities expanding by diffusion. One notes that for higher strain-rates, where the increase in

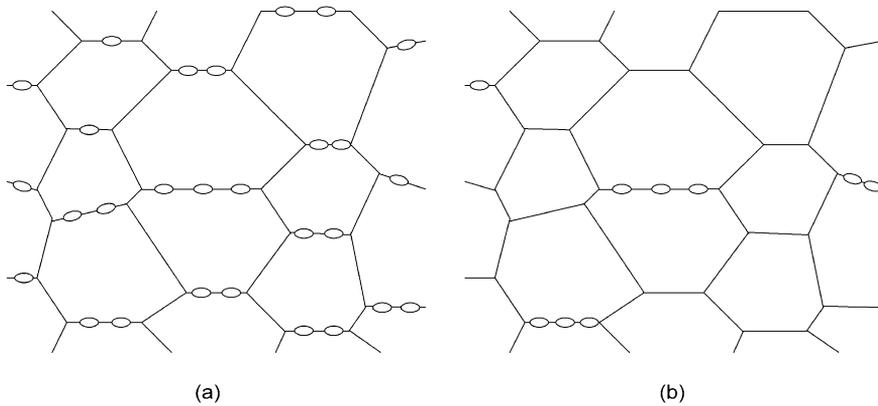


Fig. 11. Uniform (a) and heterogeneous (b) cavitation at (especially transverse) boundaries. The latter condition can particularly lead to constrained cavity growth (Chen, 1983a).

volume from cavity growth can be easily accommodated, the growth rate is primarily a function of the grain boundary diffusion coefficient. If only certain grain boundary facets cavitate, then the time for coalescence, t_c , on these facets can be calculated

$$t_c \cong \frac{0.004kT\lambda^3}{\Omega\delta D_{gb}\sigma_{ss}} + \frac{0.24(1 + 3/n)^{1/2}\lambda}{\dot{\epsilon}_{ss}g} \quad (12)$$

where n is the steady-state stress exponent, g is the grain size, σ_{ss} and $\dot{\epsilon}_{ss}$ are the steady-state stress and strain-rate, respectively, related by,

$$\dot{\epsilon}_{ss} = A_0 \exp[-Q_c/kT](\sigma_{ss}/E)^n \quad (13)$$

where $n=5$ for classic five-power-law creep (Kassner and Perez-Prado, 2000). However, it must be emphasized that failure is not expected by mere coalescence of cavities on isolated facets. Additional time may be required to join facet-size microcracks. The mechanism of joining the facets may be rate controlling. The advance of facets by local nucleation ahead of the “crack” may be important (creep-crack growth on a small scale) and interaction between facets. The nucleation rate of cavities away from the facet may also be important. It appears likely, however, that this model can explain the longer times for rupture (than expected based on unconstrained diffusive cavity growth). This likely also is the basis for the Monkman–Grant relationship if one assumes that the time to cavity coalescence, t_c , is most of the specimen lifetime, t_f , so that t_c is not appreciably less than t_f . Fig. 12, adapted from Riedel (1987), shows the cavity growth rate versus stress for constrained cavity growth as solid lines. Also plotted in this figure (as the dashed lines) is the equation for unconstrained cavity growth [Eq. (9)]. It is observed that the equation for unconstrained growth predicts much higher growth rates (lower t_f) than constrained growth rates. Also, the stress dependency of the growth rate for constrained growth leads to a time to fracture relationship that more closely matches that expected for steady-state creep as predicted by the Monkman–Grant relationship.

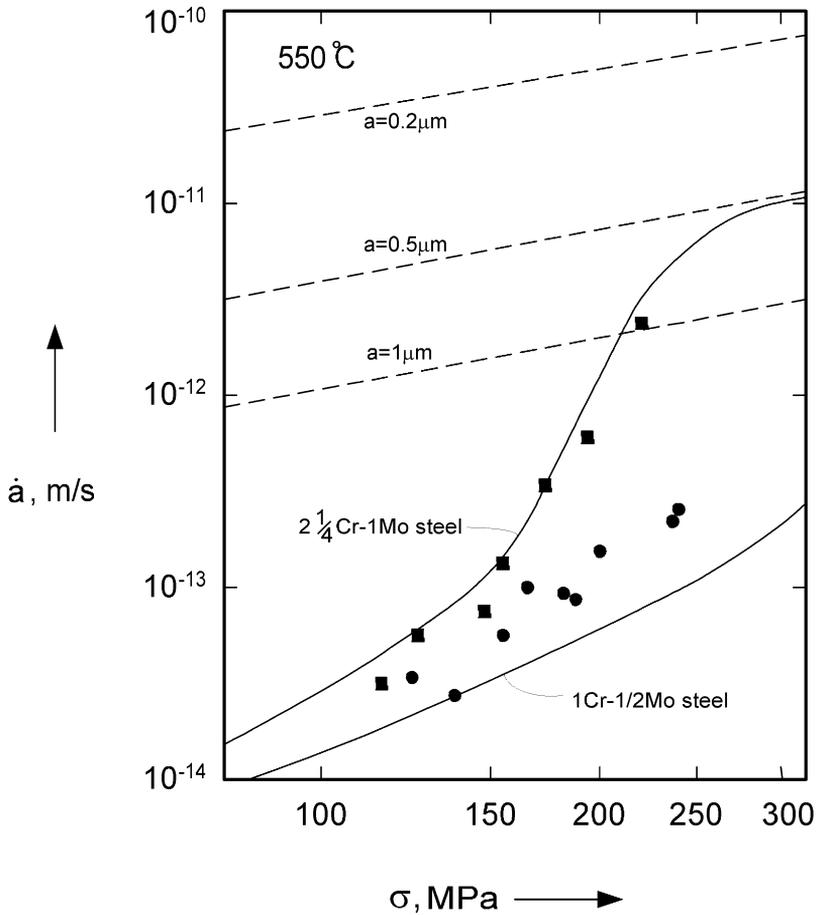


Fig. 12. The cavity growth rate versus stress in steel. The dashed lines refer to unconstrained growth and solid lines to constrained growth (based on Riedel, 1987).

One must, in addition to considering constrained cases, also consider that cavities are continuously nucleated. For continuous nucleation and unconstrained diffusive cavity growth, Riedel suggests:

$$t_f = \left[\frac{kT}{5\Omega\delta D_{gb}\sigma} \right]^{2/5} \left(\frac{\omega_f}{\dot{N}} \right)^{3/5} \quad (14)$$

where ω_f is the critical cavitated area fraction and, consistent with Fig. 8 from Dyson (15),

$$\dot{N} = \alpha' \dot{\epsilon} = \alpha' \beta \sigma^n \quad (15)$$

with $\dot{\epsilon}$ according to Eq. (13).

Eq. (14) can be approximated by

$$t_f \propto \frac{1}{\sigma^{(3n+2)/5}}$$

For continuous nucleation with the constrained case, as Riedel suggests that the time for coalescence on isolated facets is,

$$t_c = 0.38 \left[\frac{\pi(1 + 3/n)}{\dot{N}} \right]^{1/3} \frac{\omega_f}{[\dot{\epsilon}g]^{2/3}} \tag{16}$$

which is similar to the version by Cho et al. (1992). Fig. 13, also from Riedel, illustrates the realistic additional effects of continuous nucleation, which appear to match the observed rupture times in steel. The theoretical curves in Fig. 13 correspond to equations such as (8), (12), (14), (16) (based on Cane, 1979). One interesting aspect of this figure is that there is very good agreement between t_c and t_f for constrained cavity growth. This data was based on the data of Cane (1979) and Riedel (1985), who determined the nucleation rate by apparently using an empirical value of α' . No adjustable parameters were used. Cho et al. (1992) later, in NiCr

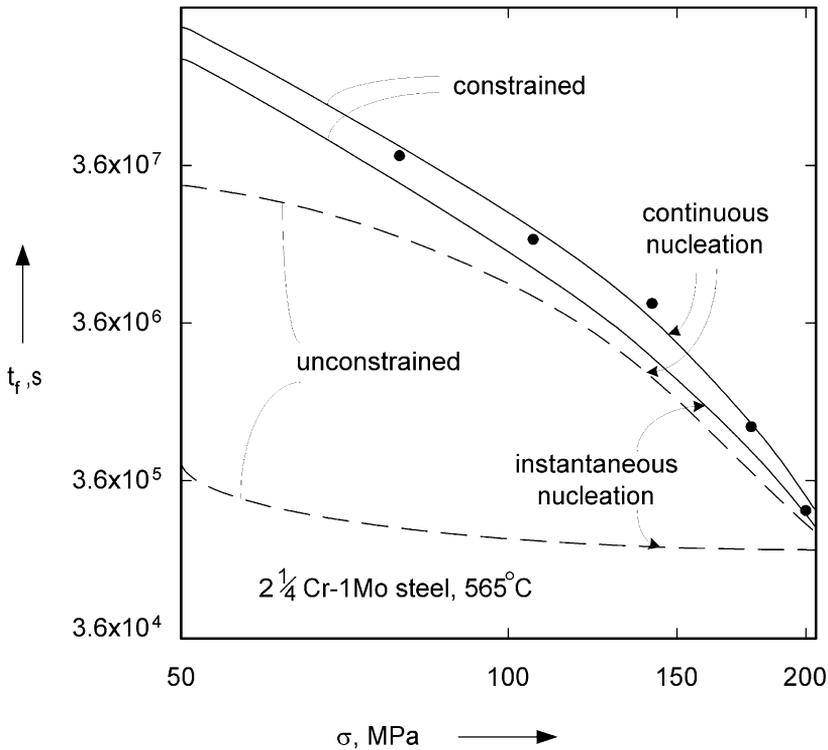


Fig. 13. The time to rupture versus applied stress for (a) unconstrained (dashed lines) cavity growth with instantaneous or continuous nucleation. (b) Constrained cavity growth (t_c) with instantaneous and continuous nucleation. Dots refer to experimental t_f (based on Riedel, 1987).

steel at 823 K, were able to reasonably predict rupture times assuming continuous nucleation and constrained cavity growth. Dyson (2002) suggested that within certain temperature and strain-rate regimes there may be a transition from constrained to unconstrained cavity growth. For an aluminum alloy it was suggested that decreased temperature and increased stress could lead to unconstrained growth. Interestingly, Dyson also pointed out that for constrained cavity growth, uncavitated regions would experience accelerated creep beyond that predicted by the decrease in load carrying area resulting from cavitation.

It should be mentioned that accommodated grain boundary sliding can eliminate the constraint illustrated in Fig. 11 (two-dimensional); however, in the three-dimensional case, sliding does not preclude constrained cavity growth, as shown by Tvergaard (1984) and Anderson and Rice (1985). Nix et al. (1989) and Yousefiani et al. (1997) have used a calculation of the principal facet stress to predict the multi-axial creep rupture time from uniaxial stress states. Here, it is suggested that GBS is accommodated and the normal stresses on (transverse) boundaries are increased. Van der Giessen and Tvergaard (1991) appear to analytically (3D) show that increased cavitation on inclined sliding boundaries may increase the normal stresses on transverse boundaries for constrained cavity growth. Thus, the Reidel solution may be non-conservative in the sense that it overpredicts t_f .

4.5. Plasticity

Cavities can grow, of course, exclusively by plasticity. Hancock (1976) initially proposed the creep controlled cavity growth model based on the idea that cavity growth during creep should be analogous to McClintock's (1968) model for a cavity growing in a plastic field. Cavity growth according to this model occurs as a result of creep deformation of the material surrounding the grain boundary cavities in the absence of a vacancy flux. This mechanism becomes important under high strain-rate conditions where significant strain is realized. The cavity growth rate according to this model is given as

$$\frac{dr}{dt} = r\dot{\epsilon} - \frac{\gamma}{2G} \quad (17)$$

This is fairly similar to the relationship by Riedel (1987) discussed earlier.

It has been suggested on occasion that the observed creep cavity growth rates are consistent with plasticity growth (e.g., Pavinich and Raj, 1977) but it is not always obvious that constrained diffusional cavity growth is not occurring, which is also controlled by plastic deformation.

Khaleel et al. (2001) and Taylor et al. (2002) appear to use a plasticity model to model superplasticity by considering a nucleation rate and growth rate for cavities, as well as strain-rate sensitivity in their continuum approach.

4.6. Coupled diffusional and plastic growth

Beere and Speight (1978), Edward and Ashby (1979), Needleman and Rice (1980), Cocks and Ashby (1982a,b), and others (Chen and Argon, 1981b; Schneibel and

Martinez, 1987; Nix, 1988; Cadek, 1989; Lee et al., 1993; Lu and Delph, 1993) suggested that there may actually be a coupling of diffusive cavity growth of cavities with creep plasticity of the surrounding material from the far-field stress. It is suggested that as material from the cavity is deposited on the grain boundary via surface and grain boundary diffusion, the length of the specimen increases due to the deposition of atoms over the diffusion length. This deposition distance is effectively increased (shortening the required diffusion length) if there is creep plasticity in the region ahead of the diffusion zone. This was treated numerically by Needleman and Rice, and later by van der Giessen et al. (1995). Analytic descriptions were performed by Chen and Argon (1981b). A schematic of this coupling is illustrated in Fig. 14. The diffusion length is described as (Needleman and Rice, 1980),

$$\Lambda = \left(\frac{D_{gb}\Omega\delta\sigma}{kT\dot{\epsilon}} \right)^{1/3} \tag{18}$$

Chen and Argon (1981a,b) describe coupling by

$$\frac{dV}{dt} = \dot{\epsilon}2\pi\Lambda^3 \left[\ln\left(\frac{a+\Lambda}{a}\right) + \left(\frac{a}{a+\Lambda}\right)^2 \times \left(1 - \frac{1}{4}\left(\frac{a}{a+\Lambda}\right)^2\right) - \frac{3}{4} \right] \tag{19}$$

as illustrated in Fig. 15.

Similar analyses were performed by others with similar results (Edward and Ashby, 1979; Lee et al., 1993). It has been shown that when $\Lambda \ll a$ and λ (Edward and Ashby, 1979; Beere and Speight, 1978; van der Giessen and Tvergaard, 1991), diffusion controlled growth no longer applies. In the extreme, this occurs at low temperatures. Creep flow becomes important as a/Λ increases. At small creep rates, but higher temperatures, Λ approaches $\lambda_s/2$, a/Λ is relatively small, and the growth

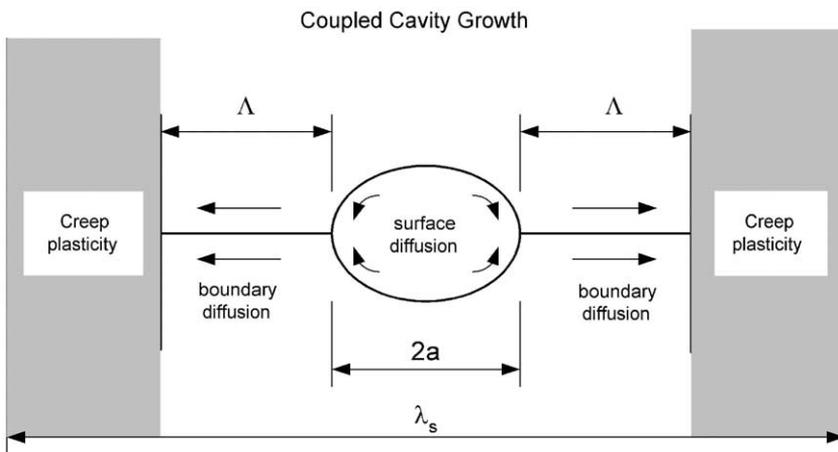


Fig. 14. The model for coupled diffusive cavity growth with creep plasticity. The diffusion length is suggested to be reduced by plasticity ahead of the cavity (based on Nix, 1988).

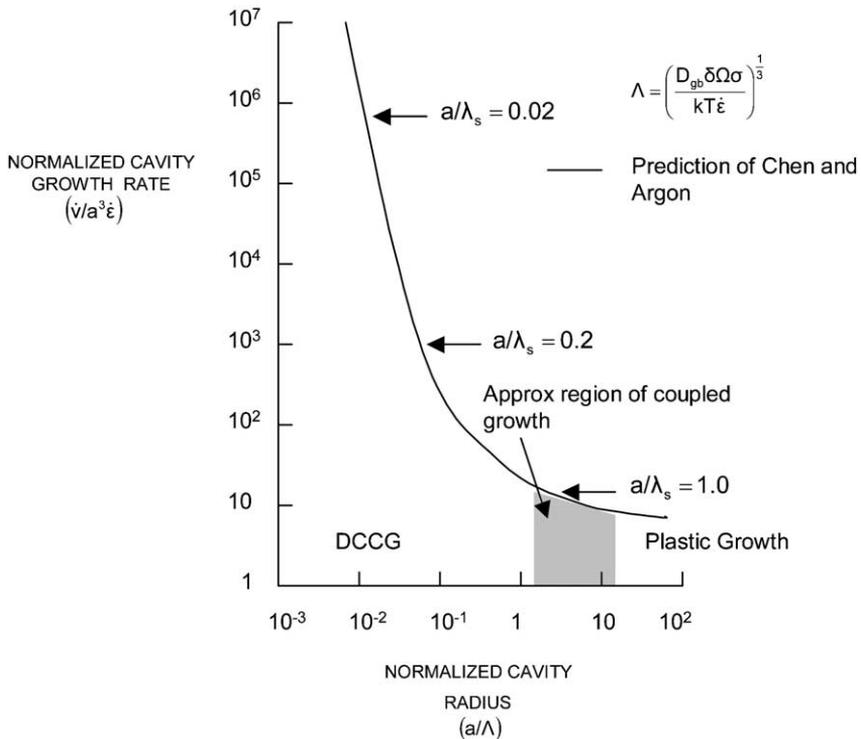


Fig. 15. Prediction of growth rate for different ratios of cavity spacing λ and diffusion zone sizes Λ (from Chen and Argon, 1981a,b).

rate can be controlled by diffusion-controlled cavity growth (DCCG). Coupling, leading to “enhanced” growth rates over the individual mechanisms, occurs at “intermediate” values of a/Λ , as indicated in Fig. 15. Of course, the important question is whether, under “typical creep” conditions, the addition of plasticity effects (or the coupling) is important. Needleman and Rice suggest that for $T > 0.5 T_m$, the plasticity effects are important only for $\sigma/G > 10^{-3}$ for pure metals (relatively high stress). Riedel suggests that for pure metals, as well as creep resistant materials, diffusive growth predominates over the whole range of creep testing. Even under the most relevant conditions, the cavity growth rate due to coupling is, at most, a factor of two different than the growth rate calculated by simply adding the growth rates due to creep and diffusion independently (Cocks and Ashby, 1982b). It has been suggested that favorable agreement between the Chen and Argon analytical treatment is fortuitous because of limitations to the analysis (Schneibel and Martinez, 1987; Lu and Delph, 1993; Delph, in press).

Of course, at lower temperatures, cavity growth occurs exclusively by plasticity (Hancock, 1976). It must be recognized that cavity growth by simple plasticity is not as well understood as widely perceived. In single phase metals, for example, under uniaxial tension, a 50% increase in cavity size require large strains, such as 50%

(Forero and Koss, 1994). Thus, a thousand-fold increase in size from the nucleated nanometer-sized cavities would not appear to be easily explained. Fig. 4(b), interestingly, illustrates a case where plastic growth of cavities appears to be occurring. The cavities nucleate within grains at large particles in the dispersion strengthened aluminum of this figure. Dunand et al. (1999) suggest that this transgranular growth occurs by plasticity, as suggested by Cocks and Ashby (1982a). Perhaps interaction between cavities explains modest ductility. One case where plasticity in a pure metal is controlling is constrained thin silver films under axisymmetric loading where $\sigma_1/\sigma_2 (= \sigma_3) \cong 0.82$ (Kassner et al., 1998; Kassner and Perez-Prado, 2000). Here unstable cavity growth (Kassner et al., 1998) occurs via steady-state deformation of silver. The activation energy and stress-sensitivity appear to match that of steady-state creep of silver at ambient temperature. Cavities nucleate at high angle boundaries where obstacles are observed (high twin-density metal) by slip-plasticity. A SEM micrograph of these cavities is illustrated in Fig. 16. The cavities in Fig. 16 continuously nucleate and also appear to undergo plastic cavity growth. Interestingly, if a plastically deforming base metal is utilized (creep deformation of the constraining base metal of a few percent), the additional concomitant plastic strain (over that resulting from a perfectly elastic base metal) increases the nucleation rate and decreases the fracture time by several orders of magnitude, consistent with Fig. 13. Cavity growth can also be affected by segregation of impurities, as these may affect surface and grain boundary diffusivity. Finally, creep fracture predictions must consider the scatter present in the data. This important, probabilistic, aspect recently has been carefully analyzed (Nix et al., 1983).

5. Creep crack growth

Cracks can occur in creeping metals from pre-existing flaws, fatigue, corrosion related processes and porosity (Ai et al., 1992; Sherry and Pilkington, 1993). In these cases, the cracks are imagined to develop relatively early in the lifetime of the metal. These contrast the case where cracks can form in a uniformly strained (i.e., unconstrained cavity growth and uniform cavity nucleation) metal where interlinkage of cavities leading to crack formation is the final stage of the rupture life. Crack formation by cavity interlinkage in constrained cavity growth cases may be the rate-controlling step(s) for failure. Hence, the subject of creep crack growth is quite relevant in the context of cavity formation. Fig. 17 (from Nix et al., 1977) illustrates a Mode I crack. The stress/strain ahead of the crack leads to cavity nucleation and growth. The growth can be considered to be a result of plasticity induced expansion or diffusion controlled cavity growth. Crack growth occurs by the coalescence of cavities with each other and the crack.

Nix et al. (1977) showed that plastic growth of cavities ahead of the crack tip can lead to a “steady-state” crack growth rate. Nucleation was not included in the analysis. Nix et al. considered the load parameter to be the stress intensity factor for (elastic) metals with Mode I cracks, K_I .

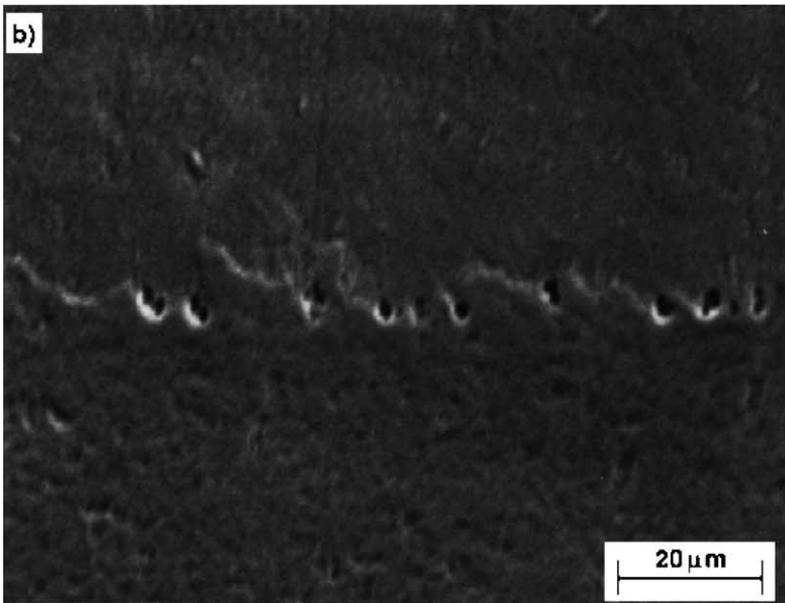
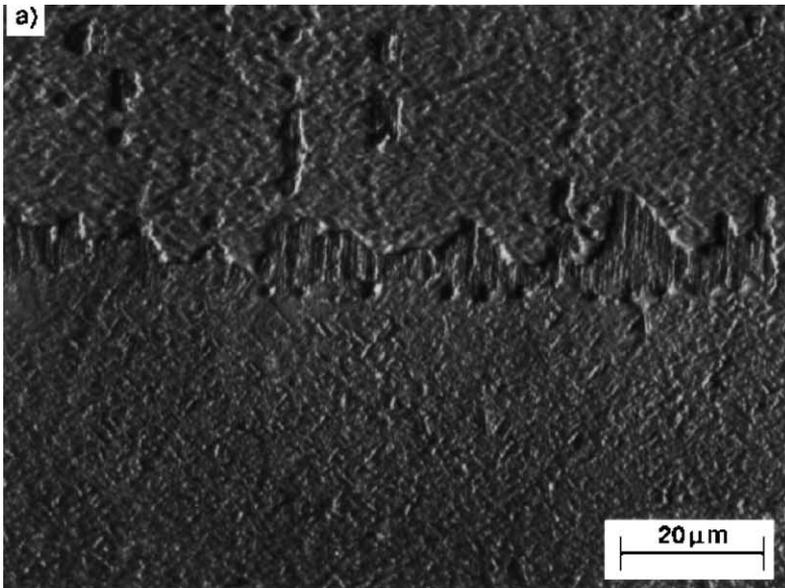


Fig. 16. Creep cavitation in silver at ambient temperature. Cavities grow by unstable cavity growth (Huang et al., 1991), with the rate determined by steady-state creep of silver (Kassner and Perez-Prado, 2000).

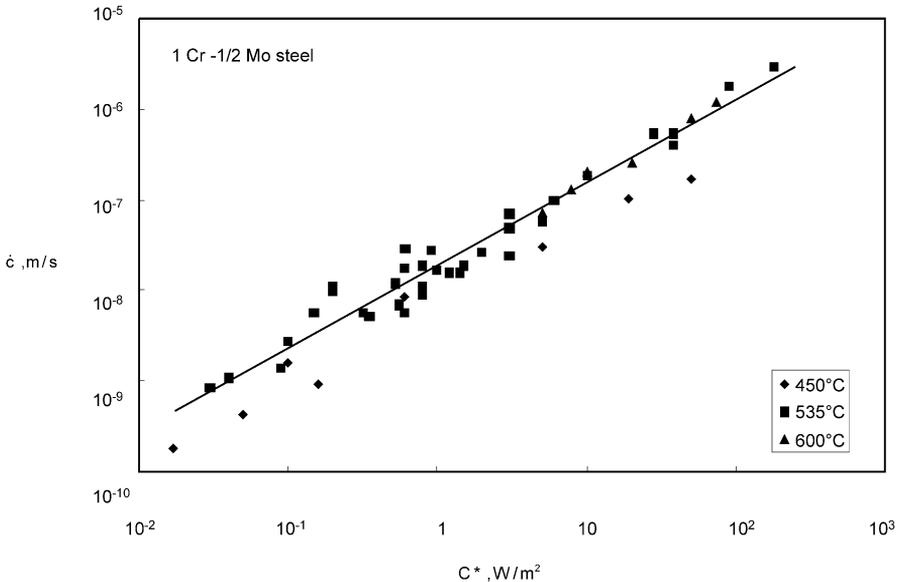


Fig. 18. Crack growth rates in 1Cr–1/2 Mo steel (adapted from Miller and Pilkington, 1980). W is the in-plane specimen dimension.

Riedel similarly argued that if cavity nucleation occurs instantaneously, diffusional growth predicts,

$$\dot{c} = \frac{k' D_b (\Omega \delta)}{2kT\lambda^3} C^* 1/n + 1 (c - c_0)^{n/n+1} \quad (22)$$

where c_0 is the initial crack length and c is the current crack length and $k' - k'''$ are constants. These constants are combined constants from Riedel's original equation and the line in Fig. 18 is based on Eq. (21) using some of these constants as adjustable parameters.

Note that Eq. (22) gives a strong temperature dependence [the “constants” of Eq. (21) are not strongly temperature-dependent]. Riedel also develops a relationship of strain-controlled cavity growth with strain controlled nucleation, which also reasonably describes the data of Fig. 18. Chan (1988) used the C^* to model the crack driving force in a constitutive representation of high temperature creep damage in a Ni alloy.

6. Other considerations

As discussed earlier, Nix and coworkers (Goods and Nix, 1978a,b; Nieh and Nix, 1979, 1980b) produced cavities by reaction with oxygen and hydrogen to produce water-vapor bubbles (cavities). Other (unintended) gas reactions can occur. These gases can include methane, hydrogen, and carbon dioxide. A brief review of environmental

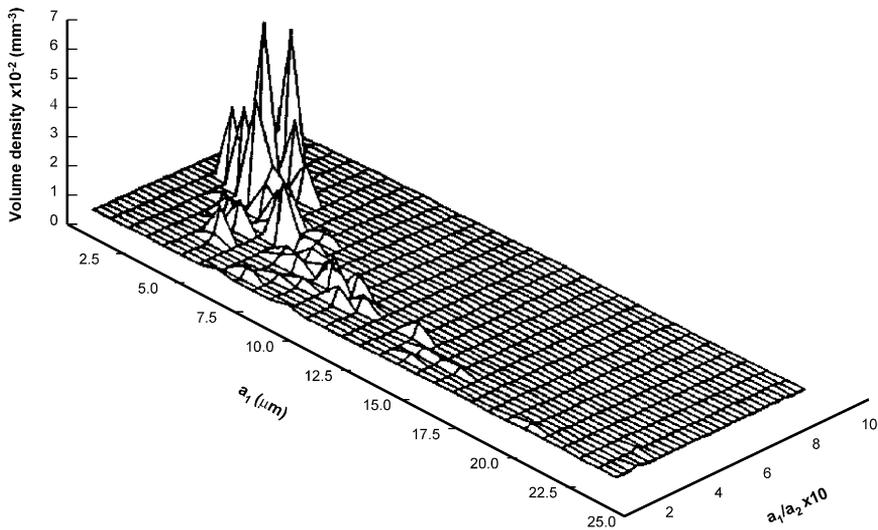


Fig. 19. The cavity density versus major radius a_1 and aspect ratio a_1/a_2 for 304 stainless steel (from Delph, in press).

effects was discussed recently by Delph (in press). The randomness (or lack of periodicity) of the metal microstructure leads to randomness in cavitation and (e.g.) failure time. Fig. 19 illustrates the cavity density versus major radius a_1 and aspect ratio a_1/a_2 . This was based on metallography of creep deformed AISI 304 stainless steel. A clear distribution in sizes is evident. Creep failure times may be strongly influenced by the random nature of grain boundary cavitation.

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References

- Adams, B.L., 1993. Orientation imaging microscopy: application to the measurement of grain boundary structure. *Mater. Sci. Eng.* A166 (1-2), 59–66.
- Ai, S.H., Lupinc, V., Maldini, M., 1992. Creep fracture mechanisms in single crystal superalloys. *Scripta Metall. Mater.* 26 (4), 579–584.
- Anderson, P.M., Rice, J.R., 1985. Constrained creep cavitation of grain boundary facets. *Acta Metall.* 33 (3), 409–422.
- Anderson, P.M., Shewmon, P.G., 2000. Stress redistribution and cavity nucleation near a diffusively growing grain boundary cavity. *Mech. Mater.* 32 (3), 175–191.

- Arai, M., Ogata, T., Nitta, A., 1996. Continuous observation of cavity growth and coalescence by creep-fatigue tests in SEM. *JSME Intl. J.* 39 (3), 382–388.
- Argon, A.S., 1983. Intergranular cavitation in creeping alloys. *Scripta Metall.* 17 (1), 5–12.
- Ayenu, A., Langdon, T.G., 1996. The inter-relationship between grain boundary sliding and cavitation during creep of polycrystalline copper. *Metall. Trans.* 27A (4), 901–907.
- Beere, W., 1983. Models of creep cavitation and their interrelationships. *Scripta Metall.* 17 (1), 13–16.
- Beere, W., Speight, M.V., 1978. Creep cavitation by vacancy diffusion in plastically deforming solid. *Metal. Sci.* 12 (4), 172–176.
- Broyles, S.E., Anderson, K.R., Groza, J., Gibeling, J.C., 1996. Creep deformation of dispersion-strengthened copper. *Metall. Trans.* 27A (5), 1217–1227.
- Cadek, J., 1988. *Creep in Metallic Materials*. Elsevier, Amsterdam.
- Cadek, J., 1989. On the cavity-growth-model-based prediction of the stress dependence of time to creep fracture. *Mater. Sci. Eng.* A117, L5–L9.
- Cane, B.J., 1979. Interrelationship between creep deformation and creep rupture in 2 1/4Cr-1Mo steel. *Metal. Sci.* 13 (5), 287–294.
- Cane, B.J., 1981. Mechanistic control regimes for intergranular cavity growth in 2.25 Cr-1 Mo steel under various stresses and stress states. *Metal. Sci.* 15 (7), 302–310.
- Chakraborty, A., Earthman, J.C., 1997. Numerical models of creep cavitation in single phase, dual phase and fully lamellar titanium aluminide. *Acta Mater.* 45 (11), 4615–4626.
- Chan, K.S., 1988. The constitutive representation of high temperature creep damage. *Int. J. Plasticity* 4 (4), 355–370.
- Chan, K.S., Page, R.A., 1990. Continuous creep cavity nucleation by stochastic grain-boundary sliding. *J. Mater. Sci.* 25 (11), 4622–4629.
- Chen, C.W., Machlin, E.S., 1956. On the mechanism of intercrystalline cracking. *Acta Metall.* 4, 655–656.
- Chen, C.W., Machlin, E.S., 1957. On a mechanism of high temperature intercrystalline cracking. *Trans. AIME* 209, 829–835.
- Chen, I.W., 1983a. Mechanisms of cavity growth in creep. *Scripta Metall.* 17 (1), 17–22.
- Chen, I.-W., 1983b. Cavity growth on a sliding grain boundary. *Metall. Trans.* 14A (11), 2289–2293.
- Chen, I.-W., Argon, A.S., 1981a. Creep cavitation in 304 stainless steel. *Acta Metall.* 29 (7), 1321–1333.
- Chen, I.-W., Argon, A.S., 1981b. Diffusive growth of grain-boundary cavities. *Acta Metall.* 29 (10), 1759–1768.
- Chen, R.T., Weertman, J.R., 1984. Grain boundary cavitation in internally oxidized copper. *Mater. Sci. Eng.* 64 (1), 15–25.
- Cho, H.C., Yu, J., Park, I.S., 1992. Creep cavitation in a NiCr steel. *Metall. Trans.* 23A (1), 201–210.
- Chuang, T.-J., Rice, J.R., 1973. The shape of intergranular creep cracks growing by surface diffusion. *Acta Metall.* 21 (12), 1625–1628.
- Churley, W.E., Earthman, J.C., 1997. High-temperature crack growth in 304 stainless steel under mixed-mode loading conditions. *Metall. Trans.* 28A (3A), 763–773.
- Cocks, A.C.F., Ashby, A.F., 1982a. On creep fracture by void growth. *Prog. Mater. Sci.* 27 (3-4), 189–244.
- Cocks, A.C.F., Ashby, M.F., 1982b. Creep fracture by coupled power-law creep and diffusion under multiaxial stress. *Metal. Sci.* 16 (10), 465–474.
- Courtney, T.H., 1990. *Mechanical Behavior of Materials*. McGraw-Hill, New York.
- Davanas, K., Solomon, A.A., 1990. Theory of intergranular creep cavity nucleation, growth and interaction. *Acta Metall. Mater.* 38 (10), 1905–1916.
- Delph, T.J., 2002. *Metall. Mater. Trans.* 33A (2), 383–390.
- Dewald, D.K., Lee, T.C., Robertson, I.M., Birnbaum, H.K., 1990. Dislocation structures ahead of advancing cracks. *Metall. Trans.* 21A (9), 2411–2417.
- Dobes, F., Milicka, K., 1976. The relation between minimum creep rate and time to fracture. *Metal. Sci.* 10 (11), 382–384.
- Dunand, D.C., Han, B.Q., Jansen, A.M., 1999. Monkman-Grant analysis of creep fracture in dispersion-strengthened and particulate-reinforced aluminium. *Metall. Trans.* 30A (3A), 829–838.
- Dyson, B.F., 1976. Constraints on diffusional cavity growth rates. *Metal. Sci.* 10 (10), 349–353.

- Dyson, B.F., 1983. Continuous cavity nucleation and creep fracture. *Scripta Metall.* 17 (1), 31–37.
- Dyson, B.F., 2002. In: Mishra, R.S., Earthman, J.C., Raj, S.V. (Eds.), *Creep Deformation: Fundamentals and Applications*. TMS, Warrendale, p. 309.
- Dyson, B.F., Loveday, M.S., Rodgers, M.J., 1976. Grain boundary cavitation under various states of applied stress. *Proc. Roy. Soc. London A349* (1657), 245–259.
- Edward, G.H., Ashby, M.F., 1979. Intergranular fracture during power-law creep. *Acta Metall.* 27 (9), 1505–1518.
- Evans, H.E., 1969. The mechanism of void growth and factors affecting the onset of intergranular failure during creep of a Mg-0.8% Al alloy. *Metal. Sci.* J3, 33–38.
- Evans, H.E., 1984. *Mechanisms of Creep Fracture*. Elsevier Applied Science, Essex, UK.
- Feltham, P., Meakin, J.D., 1959. Creep in face-centered-cubic metals with special reference to copper. *Acta Metall.* 7 (9), 614–627.
- Fleck, R.G., Taplin, D.M.R., Beevers, C.J., 1975. An investigation of the nucleation of creep cavities by 1 MV electron microscopy. *Acta Metall.* 23 (4), 415–424.
- Forero, L.E., Koss, D.A., 1994. On the strain-induced growth of neighboring voids. *Scripta Metall. Mater.* 31 (4), 419–422.
- Gandhi, C., Raj, R., 1982. Intergranular fracture in bicrystals. II. *Acta Metall.* 30 (2), 505–511.
- George, E.P., Kennedy, R.L., Pope, D.P., 1998. Review of trace element effects on high-temperature fracture of Fe- and Ni-base alloys. *Phys. Stat. Sol. A* 167 (2), 313–333.
- Gifkins, R.C., 1956. A mechanism for the formation of intergranular cracks when boundary sliding occurs. *Acta Metall.* 4, 98–99.
- Goods, S.H., Nieh, T.G., 1983. Mechanisms of intergranular cavity nucleation and growth during creep. *Scripta Metall.* 17 (1), 23–30.
- Goods, S.H., Nix, W.D., 1978a. The kinetics of cavity growth and creep fracture in silver containing implanted grain boundary cavities. *Acta Metall.* 26 (5), 739–752.
- Goods, S.H., Nix, W.D., 1978b. The coalescence of large grain boundary cavities in silver during tension creep. *Acta Metall.* 26 (5), 753–758.
- Greenwood, J.N., 1952. Intercrystalline cracking of metals. *Bull. Inst. Met.* 1, 104.
- Greenwood, J.N., Miller, D.R., Suiter, J.W., 1954. Intergranular cavitation in stressed metals. *Acta Metall.* 2, 250–258.
- Hancock, J.W., 1976. Creep cavitation without a vacancy flux. *Metal. Sci.* 10 (9), 319–325.
- Hanna, M.D., Greenwood, G.W., 1982. Cavity growth and creep in copper at low stresses. *Acta Metall.* 30 (3), 719–724.
- Hosokawa, H., Iwasaki, H., Mori, T., Mabuchi, M., Tagata, T., Higashi, K., 1999. Effects of Si on deformation behavior and cavitation of coarse-grained Al-4.5Mg alloys exhibiting large elongation. *Acta Mater.* 47 (6), 1859–1867.
- Huang, Y., Hutchinson, J.W., Tvergaard, V., 1991. Cavitation instabilities in elastic-plastic solids. *J. Mech. Phys. Sol.* 39 (2), 223–241.
- Hull, D., Rimmer, D.E., 1959. The growth of grain boundary voids under stress. *Phil. Mag.* 4, 673–687.
- Kassner, M.E., Kennedy, T.C., Schrems, K.K., 1998. The mechanism of ductile fracture in constrained thin silver films. *Acta Mater.* 46 (18), 6445–6457.
- Kassner, M.E., Perez-Prado, M.-T., 2000. Five-power-law creep in single phase metals and alloys. *Prog. Mater. Sci.* 45 (1), 1–102.
- Kassner, M.E., Rosen, R.S., Henshall, G.A., 1990. Delayed mechanical failure of silver-interlayer diffusion bonds. *Metall. Trans.* 21A (12), 3085–3100.
- Khaleel, M.A., Zbib, H.M., Nyberg, E.A., 2001. Constitutive modeling of deformation damage in superplastic materials. *Int. J. Plasticity* 17 (3), 277–296.
- Landes, J.D., Begley, J.A., 1976. In: *Mechanics of Crack Growth*, ASTM STP 590. ASTM, p. 128.
- Lee, Y.S., Kozlosky, T.A., Batt, T.J., 1993. Effects of grain boundary diffusion and power law creep on cylindrical cavity deformation. *Acta Metall. Mater.* 41 (6), 1841–1854.
- Lee, Y.S., Yu, J., 1999. Effect of matrix hardness on the creep properties of a 12CrMoVNb steel. *Metall. Trans. A* 30A (9), 2331–2339.

- Larson, F.R., Miller, J., 1952. A time-temperature relationship for rupture and creep stresses. *Trans. ASME* 74, 765–775.
- Lim, L.C., Lu, H.H., 1994. Effect of substructure on intergranular cavitation at high temperature. *Scripta Metall. Mater.* 31 (6), 723–728.
- Lombard, R., Vehoff, H., 1990. Nucleation and growth of cavities at defined grain boundaries in bicrystals. *Scripta Metall. Mater.* 24 (3), 581–586.
- Lu, H.-M., Delph, T.J., 1993. Models for coupled diffusive/strain controlled growth of creep cavities. *Scripta Metall.* 29 (3), 281–285.
- Mackerle, J., 2000. Creep and creep fracture/damage finite element modelling of engineering materials and structures. A bibliography (1980–1999). *Int. J. Pressure Vessels Piping* 77 (1), 53–77.
- McClintock, F.A., 1968. A criterion for ductile fracture by growth of holes. *J. Appl. Mech.* 35, 363–371.
- Miller, D.A., Langdon, T.G., 1980. Density measurements as an assessment of creep damage and cavity growth. *Metall. Trans.* 11A (6), 955–962.
- Miller, D.A., Pilkington, R., 1980. Diffusion and deformation controlled creep crack growth. *Metall. Trans.* 11A (1), 177–180.
- Mintz, J.M., Mukherjee, A.K., 1988. Creep cavity growth from tritium-induced helium bubbles in nickel. *Metall. Trans.* 19A (4), 821–827.
- Molinie, E., Piques, R., Pineau, A., 1991. Behaviour of a 1 Cr-1 Mo-0.25 V steel after long-term exposure. I. Charpy impact toughness and creep properties. *Fat. Fract. Eng. Mater. Struct.* 14 (5), 531–545.
- Monkman, F.C., Grant, N.J., 1956. An empirical relationship between rupture life and minimum creep rate in creep-rupture tests. *Proc. ASTM* 56, 593–620.
- Murty, K.L., Zhou, Y., Devarajan, B., 2002. In: Mishra, R.S., Earthman, J.C., Raj, S.V. (Eds.). *Creep Deformation: Fundamentals and Applications*. TMS - Miner., Metals & Mater. Soc, Warrendale, PA, VS, 31–40.
- Needham, N.G., Gladman, T., 1980. Nucleation and growth of creep cavities in a type 347 steel. *Metal. Sci.* 14 (2), 64–72.
- Needham, N.G., Gladman, T., 1986. Intergranular cavity damage and creep fracture of 1Cr-0.5Mo steels. *Mater. Sci. Technol.* 2 (4), 368–373.
- Needleman, A., Rice, J.R., 1980. Plastic creep flow effects in the diffusive cavitation of grain boundaries. *Acta Metall.* 28 (10), 1315–1332.
- Nguyen, B.-N., Onck, P., van der Giessen, E., 2000. Crack-tip constraint effects on creep fracture. *Eng. Fract. Mech.* 65 (4), 467–490.
- Nieh, T.G., Nix, W.D., 1979. The coalescence of large grain boundary cavities in silver during tension creep. *Acta Metall.* 27 (6), 1097–1106.
- Nieh, T.G., Nix, W.D., 1980a. A comparison of the dimple spacing on intergranular creep fracture surfaces with the slip band spacing for copper. *Scripta Metall.* 14 (3), 365–368.
- Nieh, T.G., Nix, W.D., 1980b. The formation of water vapor bubbles in copper and their effect on intergranular creep fracture. *Acta Metall.* 28 (5), 557–566.
- Nix, W.D., 1988. Introduction to the viewpoint set on creep cavitation. *Scripta Metall.* 17 (1), 1–4.
- Nix, W.D., 1988. Mechanisms and controlling factors in creep fracture. *Mater. Sci. Eng. A* 103 (1), 103–110.
- Nix, W.D., Earthman, J.C., Eggeler, G., Ilschner, B., 1989. The principal facet stress as a parameter for predicting creep rupture under multiaxial stresses. *Acta Metall.* 37 (4), 1067–1077.
- Nix, W.D., Matlock, D.K., Dimelfi, R.J., 1977. A model for creep fracture based on the plastic growth of cavities at the tips of grain boundary wedge cracks. *Acta Metall.* 25 (5), 495–503.
- Nix, W.D., Yu, K.S., Wang, J.S., 1983. The effects of segregation on the kinetics of intergranular cavity growth under creep conditions. *Metall. Trans.* 14A (4), 563–570.
- Oh, Y.K., Kim, G.S., Indacochea, J.E., 1999. Creep rupture behaviour due to molybdenum rich M/sub 6/ C carbide in 1.0Cr-1.0Mo-0.25V bainitic steel weldment. *Scripta Mater.* 41 (1), 7–12.
- Pavinich, W., Raj, R., 1977. Fracture at elevated temperature. *Metall. Trans.* 8A (12), 1917–1933.
- Raj, R., 1978. Intergranular fracture in bicrystals. *Acta Metall.* 26 (2), 341–349.
- Raj, R., Ashby, M.F., 1975. Intergranular fracture at elevated temperature. *Acta Metall.* 23 (6), 653–666.
- Raj, R., Baik, W., 1980. Creep crack propagation by cavitation near crack tips. *Metal. Sci.* 14 (8-9), 385–394.

- Raj, R., Shih, H.M., Johnson, H.H., 1977. Correction to: 'Intergranular fracture at elevated temperature'. *Scripta Metall.* 11 (10), 839–842.
- Randle, V., 1993. Strain-induced secondary recrystallization. *Mater. Sci. Forum* 113–115, 189–194.
- Ratcliff, R.T., Greenwood, G.W., 1965. The mechanism of cavitation in magnesium during creep. *Phil. Mag.* 12 (115), 59–69.
- Rice, J.R., 1981. Constraints on the diffusive cavitation of isolated grain boundary facets in creeping polycrystals. *Acta Metall.* 29 (4), 675–681.
- Riedel, H., 1984. Cavity nucleation at particles on sliding grain boundaries. A shear crack model for grain boundary sliding in creeping polycrystals. *Acta Metall.* 32 (3), 313–321.
- Riedel, H., 1985. Continuous nucleation of grain boundary cavities in creep rupture. *Z. Metall.* 76 (10), 669–675.
- Riedel, H., 1987. *Fracture at High Temperatures*. Springer-Verlag, Berlin.
- Schneibel, J.M., Martinez, L., 1987. The limitations of the Chen & Argon and Martinez & Nix models for intergranular cavity growth. *Scripta Metall et Mater.* 21 (4), 495–500.
- Sherry, A.H., Pilkington, R., 1993. The creep fracture of a single-crystal superalloy. *Mater. Sci. and Eng.* A172 (1–2), 51–61.
- Speight, M.V., Beere, W., 1975. Vacancy potential and void growth on grain boundaries. *Met. Sci.* 9 (4), 190–191.
- Staley, J.T., Saxena, A., 1990. Mechanisms of creep crack growth in 1 wt.% antimony-copper: implications for fracture parameters. *Acta Metall. et Mater.* 38 (6), 897–908.
- Stiegler, J.O., Farrell, K., Loh, B.T.M., McCoy, H.E., 1967. Nature of creep cavities in tungsten. *Trans. ASM Quart.* 60 (3), 494–503.
- Suri, S., Viswanathan, G.B., Neeraj, T., Hou, D.H., Mills, M.J., 1999. Room temperature deformation and mechanisms of slip transmission in oriented single-colony crystals of an alpha/beta titanium alloy. *Acta Mater.* 47 (3), 1019–1034.
- Svensson, L.-E., Dunlop, G.L., 1982. Growth mechanism of intergranular creep cavities in alpha-brass. *Met. Sci.* 16 (1), 57–64.
- Svoboda, J., Sklenicka, V., 1990. Thermal cavity nucleation at intergranular inclusions in high temperature creep. *Acta Metall. Mater.* 38 (6), 1441–1449.
- Tabuchi, M., Kubo, K., Yagi, K., 1993. In: Wilshire, B., Evans, R.W., (Eds.), *Creep Fracture of Engineering Materials and Structures*. Inst. Metals, London, p. 449.
- Taylor, M.B., Zbib, H.M., Khaleel, M.A., 2002. Damage and size effect during superplastic deformation. *Int. J. Plasticity* 18 (3), 415–442.
- Trinkaus, H., Yoo, M.H., 1987. Nucleation under time-dependent supersaturation. *Phil. Mag.* 55 (3), 269–289.
- Tvergaard, V., 1984. On the creep constrained diffusive cavitation of grain boundary facets. *J. Mech. and Physics of Solids* 32 (5), 373–393.
- van der Giessen, E., Tvergaard, V., 1991. A creep rupture model accounting for cavitation at sliding grain boundaries. *Int. J. Fract.* 48 (3), 153–178.
- van der Giessen, E., van der Burg, M.W.D., Needleman, A.I., Tvergaard, V., 1995. Void growth due to creep and grain boundary diffusion at high triaxialities. *J. Mech. Phys. Solids* 43 (1), 123–165.
- Waddington, J.S., Lofthouse, K.J., 1967. The effect of irradiation on the high-temperature fracture of stainless steel. *J. Nuc. Mater.* 22 (2), 205–213.
- Watanabe, T., 1993. Grain boundary design and control for high temperature materials. *Mater. Sci. Eng.* A166 (1–2), 11–28.
- Watanabe, T., Davies, P.W., 1978. Grain boundary sliding and intergranular fracture behaviour of copper bicrystals. *Phil. Mag.* 37 (5), 649–681.
- Weertman, J., 1973. Hull-Rimmer grain boundary void growth theory—a correction. *Scripta Metall.* 7 (10), 1129–1130.
- Weertman, J., 1986. Zener-Stroh crack, Zener-Hollomon parameter, and other topics. *J. Appl. Phys.* 60 (6), 1877–1887.
- Wei, R.P., Liu, H., Gao, M., 1997. Crystallographic features and growth of creep cavities in a Ni-18Cr-18Fe alloy. *Acta Mater.* 46 (1), 313–325.

- Wiesner, C., Earthman, J.C., Eggeler, G., Ilschner, B., 1989. Creep crack growth and cavitation damage in a 12% CrMoV steel. *Acta Metall.* 37 (10), 2733–2741.
- Wilkinson, D.S., 1981. A model for creep cracking by diffusion-controlled void growth. *Mater. Sci. Eng.* 49 (1), 31–39.
- Wilkinson, D.S., Vitek, V., 1982. The propagation of cracks by cavitation: a general theory. *Acta Metall.* 30 (9), 1723–1732.
- Wu, R., Sandstrom, R., 1995. Creep cavity nucleation and growth in 12Cr-Mo-V steel. *Mater. Sci. Tech.* 11 (6), 579–588.
- Yang, M.S., Weertman, J.R., Roth, M., 1984. In: Wilshire, B., Owen, D.R.J., (Eds.), *Proceedings of 2nd Int. Conf. Creep and Fracture of Eng. Mater. and Structures*, Pineridge, Swansea, 1984, p. 149.
- Yavari, P., Langdon, T.G., 1983. Cavitation in high purity aluminium during fatigue at elevated temperatures. *J. Mater. Sci. Lett.* 2 (9), 522–524.
- Yoo, M.H., Trinkaus, H., 1986. Interaction of slip with grain boundary and its role in cavity nucleation. *Acta Metall.* 34 (12), 2381–2390.
- Yousefiani, A., El-Nasr, A.A., Mohamed, F.A., Earthman, J.C., 1997. In: Earthman, J.C., Mohamed, F.A., (Eds.), *Creep and Fracture of Engineering Materials and Structures*. TMS, Warrendale, p. 439.
- Yousefiani, A., Mohamed, F.A., Earthman, J.C., 2000. Creep rupture mechanisms in annealed and overheated 7075 Al under multiaxial stress states. *Metall. Mater. Trans.* 31A (11), 2807–2821.