

Chapter 2

General Mechanisms of Creep

Abstract The general creep mechanism is discussed in this chapter, which is classified as: (i) dislocation slip; (ii) climb; (iii) grain-boundary sliding; and (iv) diffusion flow caused by vacancies. The relevant relations and illustrations are included. These provide the basic understanding of creep.

Before discussing creep specifically in brittle materials, ductile materials, single crystals and polycrystals, it is important to consider the general creep mechanisms observed acting in materials. There are several basic mechanisms that may contribute to creep in materials (including ceramics). The various classifications of these mechanisms are not always the same and sometimes they are more detailed or combined, depending on the points being emphasized. The classification used here is somewhat arbitrary, but follows a pattern commonly found in the literature:

- (i) dislocation slip;
- (ii) climb;
- (iii) grain-boundary sliding; and
- (iv) diffusion flow caused by vacancies.

(i) Creep by Dislocation Slip

Actually, creep types (i) and (ii) may be combined under the general heading of ‘dislocation creep’, but there is merit in separating the two different types: slip (glide) and climb. Creep takes place as a result of dislocation motion in a crystalline specimen by movement known as ‘slip’ (glide). As a result of such dislocation motion through a crystal, one part of the dislocation moves one lattice point along a plane known as the ‘slip plane’, relative to the rest of the crystal. The slip plane along which the dislocation motion takes place separates both parts of the crystal. For dislocation motion to occur, the bonds between the atoms (ions, in the case of nonmetallic materials) must be broken during the deformation. Deformation by creep, which can be an important contributor to overall deformation, occurs only in certain circumstances. Creep by dislocation glide occurs over the entire range of

temperatures, from low (basically absolute zero) to the melting temperature, although at low temperatures its contribution may be insignificant. The aforementioned factors expressed by Eq. (1.2) are important factors in creep deformation. At high temperatures, however, creep can occur at stresses less than the yield stress. The stresses needed to drive dislocation glide are on the order of a tenth the theoretical shear strength of $\sim G/10$. Glide-by-slip strengthens materials as they deform. In primary creep, stress is constant, while strain increases to a certain extent (see Fig. 1.1a) over time, but the strain rate decreases (Fig. 1.1b), until a minimum strain rate is achieved. This minimum strain rate, on a strain-time plot, represents steady-state creep.

At lower stresses, the creep rate is lower and becomes limited by the rate at which the dislocations can climb over obstacles by means of vacancy diffusion. Since a dislocation may be pinned by various obstacles, further deformation that of creep, must also occur by means of climb (discussed in the next section).

(ii) Climb

During dislocation motion, the creep rate is limited by the obstacles resisting dislocation motion. The obstacles resisting the motions of dislocations harden (strengthen) the material. High temperatures acting during deformation induce recovery processes. During steady-state creep, strain increases over time. The increased strain energy stored in the material, due to deformation, together with the high temperature, provide the driving force for the recovery process. As such, there is a balance between the processes of work hardening and recovery. Recovery involves a reduction in dislocation density and the rearrangement of dislocations into lower energy arrays, such as subgrain boundaries. For this to occur, dislocations must climb, as well as slip, and this, in turn, requires atomic movement or self-diffusion within the lattice. Hence, it is often said that the activation energies for self-diffusion and for creep are almost the same. Vacancies must be located at a site where climb is supposed to occur, to enable climb by means of a vacancy-atom exchange. As the temperature increases, the atoms gain thermal energy and the equilibrium concentrations of the vacancies in the metals increase exponentially. The number of vacancies, n , (see, for example, Damak and Dienes) is given as:

$$n = N \exp\left(-\frac{E_F}{kT}\right) \quad (2.1)$$

This same concept of the steady-state creep-rate mechanism of dislocation climb was suggested by Mott. He assumed that the rate-controlling process is the diffusion of the vacancies. It is assumed in Mott's analysis that the rate-controlling process is the diffusion of the vacancies between the dislocations creating vacancies and those destroying them. The concentration of vacancies along a dislocation line is determined by setting the change in the free energy caused by a decrease or increase in the number of vacancies equal to the change in the elastic energy occurring during dislocation climb. The creep equation that results from this analysis is:

$$\dot{\epsilon} = C\sigma^\alpha \exp\left(-\frac{Q}{kT}\right), \quad (2.2)$$

where C and α are constants, Q is the activation energy (equal to the self-diffusion) of creep, σ is the stress, and kT has its usual meaning. A value for α is indicated as $\alpha \sim 3$ to 4. This same creep-rate relation was given in Eq. (1.9), reproduced here as:

$$\dot{\epsilon} = B\sigma^n \exp\left(-\frac{Q}{kT}\right) \quad (1.9)$$

No theoretical treatment of creep seems to exist that leads to a creep rate as given by Eqs. (2.2) or (1.9). Mott had developed a theory leading to Eq. (2.2) in which he stated that Eq. (2.2) is valid in the stress range from the critical shear stress to a stress about equal to 108–109 dynes/cm². At larger stresses, the creep rate increases much more rapidly with stress. For a derivation of this equation, the reader is referred to Mott's work on the subject.

In Eq. (2.1), N is the number of lattice sites and E_F is the energy of vacancy formation. The activation energy, Q , for the jump rate, J , is given by the sum of the energy of vacancy formation and the vacancy's energy for motion, E_M , ($Q = E_F + E_M$):

$$J = J_0 \exp\left(-\frac{Q}{kT}\right) \quad (2.3)$$

J_0 represents the respective entropies. The diffusion coefficient, D , may be given as:

$$D = D_0 \exp\left(-\frac{Q}{kT}\right) \quad (2.4)$$

D_0 , the pre-exponential factor, is equivalent to J_0 , and Q is the overall activation energy for self-diffusion. The rate of steady-state creep increases with temperature, as does the essential number of vacancies for effective vacancy-atom exchange for climb.

(iii) Grain-Boundary Sliding (GBS)

Different grains and grain-sizes play significant roles in the strengthening (work-hardening) mechanisms given by the Hall–Petch relation as:

$$\sigma_y = \sigma_0 + \frac{k_y}{\sqrt{d}}, \quad (2.5)$$

where σ_y is the yield stress, σ_0 represents the resistance to dislocation glide, k_y is a measure of the dislocation pile-up behind an obstacle (a grain boundary, for example) and d is the size of the grain. The various grains and their sizes are

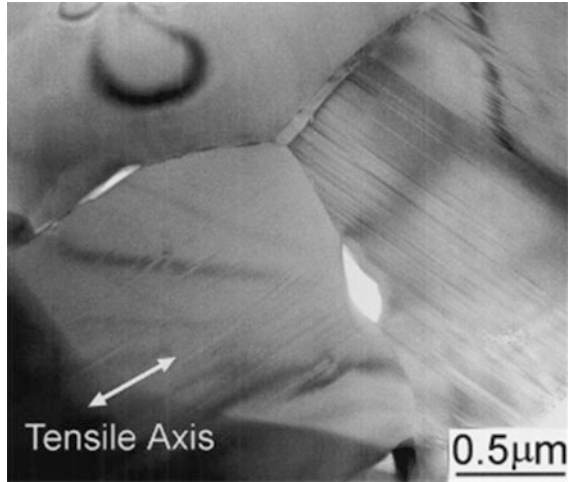
important variables characterizing the microstructure of polycrystalline materials. Grain-boundary movement plays a significant role in the characteristic behaviors of materials in regard to creep. Basically, grain-boundary sliding (GBS) is a process in which grains slide past each other along their common boundary. It has also been observed that sliding may occur in a zone immediately adjacent to the grain boundary (Wadsworth et al.).

In primary creep, the required stress increases due to work hardening (which also acts in steady-state creep, but is balanced by various recovery processes). Decreasing the grain size should strengthen a material, according to the Hall–Petch relation. Thus, for continued deformation, higher stress is required. It may be expected that materials with small grain sizes will show better creep resistance, while increasing grain size should cause an increased creep rate (for example, the secondary-creep rate). This is attributed to the decrease in boundary barriers with increasing grain size (less strengthening media exists, because there are less grain-boundary obstacles). However, this is true as long as no undesirable processes occur at the grain boundaries. For instance, large-grained materials with a small number of grain boundaries are low sources of vacancies and, therefore, dislocation climb will be reduced compared to small-grained materials. Thus, one can see that grain size in creep has a dual effect, because a small grain size strengthens the ceramics, since the large number of grains act as barriers to dislocation glide. Nonetheless, in large-grained ceramics with fewer boundaries, fewer vacancies are emitted, which are prerequisites for creep deformation by climb; therefore, this situation has reduced creep. Note that a suitable choice of grain size in ceramics is critical for achieving the best compromise regarding good creep resistance.

Major structural changes occur at the start of tertiary creep. Damage is initiated by the formation of multi-shaped cavities (in metals, either wedge-shaped or rounded cavities are observed). Wedge-shaped cavities are primarily seen at grain boundaries and their coalescence is the unmistakable sign that creep rupture will occur. It is believed that GBS is a prerequisite for the nucleation of voids and cavities and that it occurs when a sufficiently high stress concentration develops to create new surfaces. Cavity formation increases with increasing strain at high temperatures. The stresses causing GBS are the shear stresses acting on the boundaries. Whether void formation is associated with/or a consequence of GBS has not yet been completely determined, since the experiments found in the literature seem to support both concepts. In Fig. 2.1, cavities at two-grain boundary junctions may be seen in ABC-SiC. The term ‘ABC-SiC’ refers to SiC which has been hot-pressed with additions of Al, as well as B and C. This material has been shown to have an ambient temperature fracture toughness as high as $9 \text{ MPa m}^{1/2}$ with strengths of $\sim 650 \text{ MPa}$ (among the highest strength property reported for SiC).

One of the concepts regarding GBS is associated with the presence of an amorphous grain-boundary film along the boundaries between the grains. More specifically, this film has often been termed a ‘glassy phase’ and considered responsible for GBS. This glassy film may be fully crystallized after heat treatment at high temperature for an appropriate time. Clearly, such crystallization of the

Fig. 2.1 Transmission electron micrograph of ABC-SiC showing grain-boundary cavities at two-grain junctions on the tensile edge of a specimen crept at 1400 °C for 840 h under 200 MPa. Chen et al. [5]. With kind permission of Elsevier



grain-boundary phase would minimize softening and GBS, which would, in turn, cause an increase in strength. As stated previously, the microstructure has a major impact on the creep properties. Creep cavitation may appear at high temperatures in grain boundaries. Cavities are observed on the tensile side, but not on the side under compression. Cavities usually form both at two-grain and multiple-grain junctions. GBS induces cavitation during creep. Thus far, there are no conclusive data proving that GBS is the driving force for the nucleation and growth of creep cavities, although a number of studies have concluded that cavity nucleation is, in fact, induced by GBS. GBS has been the subject of numerous investigations, in light of the importance of grain boundaries for many aspects of material applications. Understanding the physics of the complex behavior of grain boundaries is of great interest in regard to: grain growth, crystallization and recovery deformation, to mention just a few topics. A general review of the properties of grain boundaries may be found, for example, in the work of Valiev, et al. Here, GBS is of interest in order to gain better practical and theoretical understanding. Illuminating research results on GBS may be observed in metals. The instructive photo below (Fig. 2.2) was taken of a Mg-0.78%Al alloy strained to 2.49% at a temperature of 473 K and under an applied stress of 17.2 MPa.

The evidence of GBS is the displacement of the scratch lines during creep testing. The above figure shows scratch lines displaced across a grain boundary; transverse markings are inscribed perpendicular to the tensile axis. Clear offsets may be seen in the transverse marker line in this Mg-0.78%Al alloy. The tensile axis in this experiment is horizontal. An alternate method for evaluating GBS is by means of interferometry. An example of the offsets of the same alloy, as revealed by interferometry, is visible in Fig. 2.3. Chan and Page have developed a model describing creep-induced transient-cavity growth by assuming that cavity growth is governed by the two competing processes—transient creep and sintering. According to this model, the rate of cavity growth is described as:

Fig. 2.2 Grain-boundary sliding revealed by the boundary offsets in a transverse marker line in a Mg-0.78%Al alloy tested under creep conditions at 473 K under a stress of 17.2 MPa. From Bell and Langdon [1], reproduced from Langdon [9]. With kind permission from Springer Science and the author

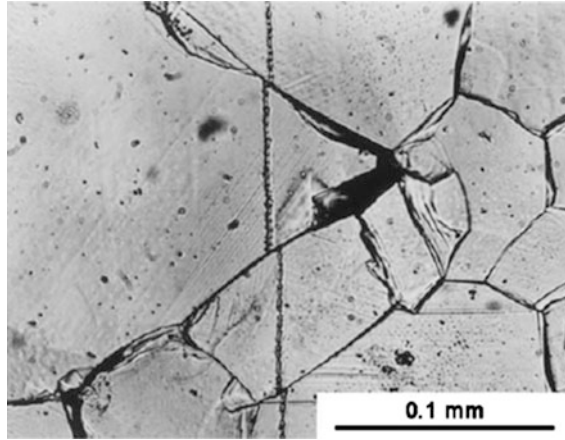
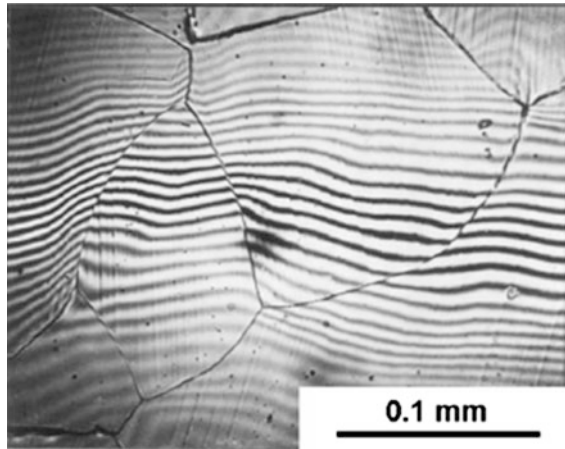


Fig. 2.3 Offset revealed by interferometry in a Mg-0.78% Al alloy pulled to an elongation of 1.5% at 473 K under a stress of 27.6 MPa. From Langdon, Mater. Sci. Eng., A166, 67 (1993), reproduced from Langdon [9]. With kind permission from Springer Science and the author



$$\dot{R} = \frac{33RG(\xi)}{4\pi^2} \left[\dot{\epsilon}_{ss}(t/t_c)^m - \frac{4\pi}{33} \left(\frac{\gamma_s}{\eta l} \right) (1/\xi - 0.9\xi) \right] \quad (2.6)$$

with

$$\xi = R/l \quad (2.7)$$

$$G(\xi) = \frac{2\sqrt{3} - 0.667\pi\xi^2}{0.96\xi^2 - \ln \xi - 0.23\xi^2 - 0.72} \quad (2.8)$$

In Eq. (2.6), R is the cavity radius, $\dot{\epsilon}_{ss}$ is the steady-state creep rate, t is the creep time, t_c is the characteristic time, m is an exponent ranging from -0.5 to -0.6 , γ is the surface energy, η is the viscosity parameter, and $2l$ is the center-to-center cavity spacing. Note that the first term within the bracket in Eq. (2.6) is the transient creep

rate, $\dot{\epsilon}_{tr}$, while the second term is the sintering rate, \dot{s} . From Eq. (2.6), it is evident that the transient creep rate, $\dot{\epsilon}_{tr}$, drives cavity growth, whereas the sintering rate term, \dot{s} , drives cavity shrinkage. In addition, imposing parameters to reach a state of equilibrium between $\dot{\epsilon}_{tr}$ and \dot{s} would result in a condition of zero cavity growth. Therefore, a critical value of $\dot{\epsilon}_{ss}(\dot{\epsilon}_{cr})$ may be determined by setting $R = 0$ in Eq. (2.6), which defines no-growth behavior as follows:

The viscosity parameter is given by:

$$\frac{\eta l \dot{\epsilon}_{cr}}{\gamma_s} = \frac{4\pi}{33} (1/\xi - 0.9\xi) \quad (2.9)$$

$$\eta = \frac{1}{132} \frac{d^3 kT}{h D_b \Omega} \quad (2.10)$$

This no-growth boundary is shown in Fig. 2.4 as the solid line. In addition, cavities exhibit continuous growth in region I, where $\dot{s}_{cr} > \dot{s}_{tr}$, and the cavities will shrink when the opposite is true (region II).

Equation (4) in Fig. 2.9 is given here as Eq. 2.9.

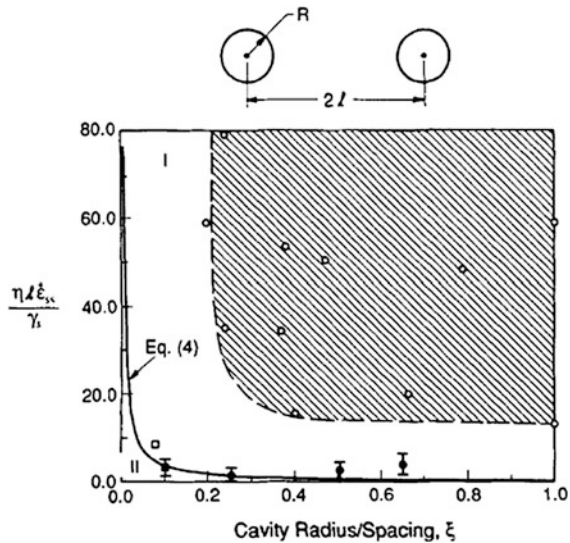
A quantitative estimate of the contribution of GBS to overall strain, ξ , used by Tan and Tan following Langdon's proposal, is:

$$\xi = \frac{\epsilon_{GBS}}{\epsilon_t} \quad (2.11)$$

ϵ_t is the total strain at high temperatures, expressed as:

$$\epsilon_t = \epsilon_g + \epsilon_{GBS} + \epsilon_{dc} \quad (2.12)$$

Fig. 2.4 Comparison of the predicted (--) and experimentally observed conditions for zero cavity growth: (□) Lucalox, 1600 °C; (■) AD99, 1300 °C, and for cavity growth; (○) Lucalox, 1600 °C; (●) AD99, 1150 °C. Region I represents cavity growth and region II, cavity shrinkage. Blanchard and Chan [2]. With kind permission of John Wiley and Sons. (Lucalox and AD99 are aluminas)



ε_g is the strain in the grain, due to processes taking place within the grain; ε_{GBS} is the strain due to GBS; and ε_{dc} is the strain due to diffusion creep. In practice, experiments are often performed with a negligible contribution of diffusion creep and, thus, Eq. (2.12) reduces to:

$$\varepsilon_t = \varepsilon_g + \varepsilon_{\text{GBS}} \quad (2.13)$$

Damage leading to failure, in the form of stress rupture, is initiated by void and crack formation. The tertiary creep, per se, is a sign that some sort of structural damage has occurred. Round or wedge-shaped voids, known as ‘r-type cavities’ and ‘w-type cavities’, are seen at first along grain boundaries and, when they coalesce, creep fracture occurs. As indicated above, the mechanism of void formation is associated with GBS and occurs due to shear stresses acting along the boundaries.

A commonly used illustration of a w-type crack initiation by GBS, its formation and growth (first presented by Chang and Grant, and found in almost every publication) is shown in Fig. 2.5. Another configuration for the initiation of intergranular cracks (somewhat more complex) is shown in Fig. 2.6.

A number of w-crack configurations have been experimentally observed at triple points. Wedge-type crack formation at triple points was initially suggested by Zener as early as 1948. According to Zener, at sufficiently high temperatures, grain boundaries behave in a viscous manner and, when near triple points under an applied tensile stress, wedge-type cracks develop due to the high stress concentration. Specifically, Zener was among the first to suggest the concept that fracturing is a consequence of plastic deformation, which is required for crack formation. His schematic illustration is shown in Fig. 2.7, where a crack can be nucleated at a dislocation site.

In Fig. 2.7b and c, the coalescence of two or three dislocations is illustrated, producing an increase in the size of the crack. The concept of crack origin at dislocation sites has been addressed and modified by various researchers. In essence, Zener suggested that cracks nucleate at dislocation pile-ups, where sufficient stress develops for the nucleation of cracks.

A dislocation model for spontaneous microcrack formation was also presented by Stroh, who calculated the elastic energy associated with wedge deformation. Stroh also determined that the nucleation of a wedge crack was due to the pile-up of dislocations on a slip plane. In Fig. 2.8, the 2D crack dislocation of a giant Burgers vector, nb , with length, c , extending to a barrier, may be seen. His expression for the elastic energy associated with wedge deformation is:

$$W_e = \frac{Gn^2b^2}{4\pi(1-\nu)} \ln \frac{4R}{c} \quad (2.14)$$

in which G is the shear modulus (modulus of rigidity), nb is a giant Burgers vector, with n being the number of dislocations comprising the giant vector, and R —the bounding radius in the stress field. The surface energy term, $2\gamma_s c$, may be added to obtain the total energy of the system as:

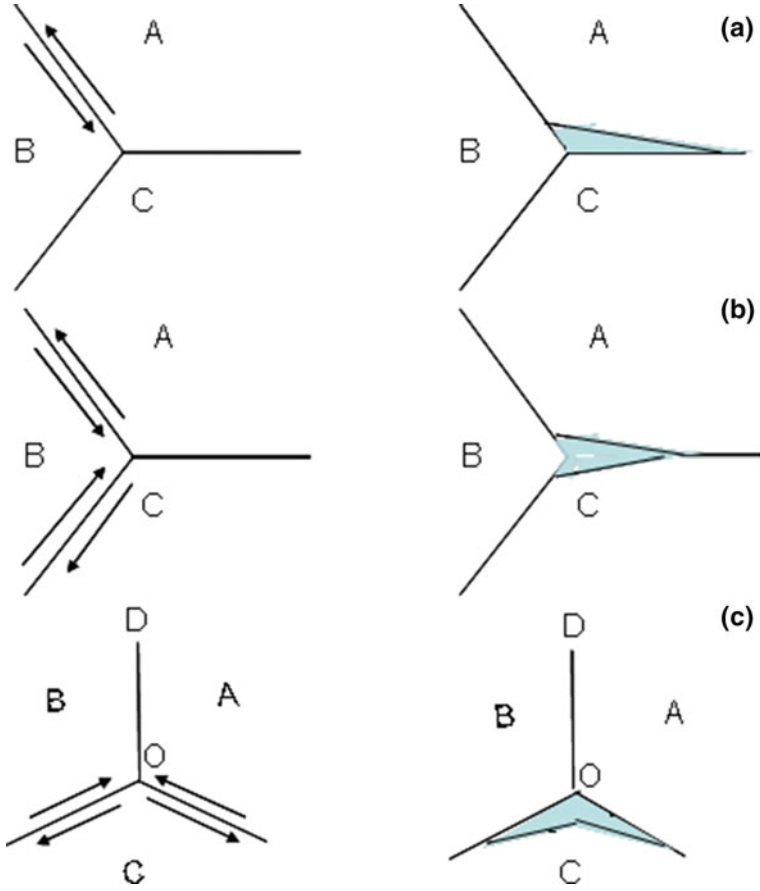


Fig. 2.5 Schematic representation of a w-type crack formation initiated by GBS. From Ref. [4]

$$W_s = \frac{Gn^2b^2}{4\pi(1-\nu)} \ln \frac{4R}{c} + 2\gamma_s c \quad (2.15)$$

By differentiating Eq. (2.15) with c , the critical length, c_{\min} , may be found:

$$\frac{\partial W_s}{\partial c} = 0 \quad (2.16)$$

$$c_{\min} = G \frac{n^2b^2}{4\pi(1-\nu)} \frac{1}{2\gamma_s} \quad (2.17)$$

In polycrystalline solids, the typical values of b , G , ν and γ_s are, respectively, (Sarfarazi and Ghosh.): $b = 2 \times 10^{-8}$ cm; $G = 1012$ dynes/cm²; $\nu = 1/3$; and $\gamma_s = 103$ dynes/cm, which gives for c_{\min} :

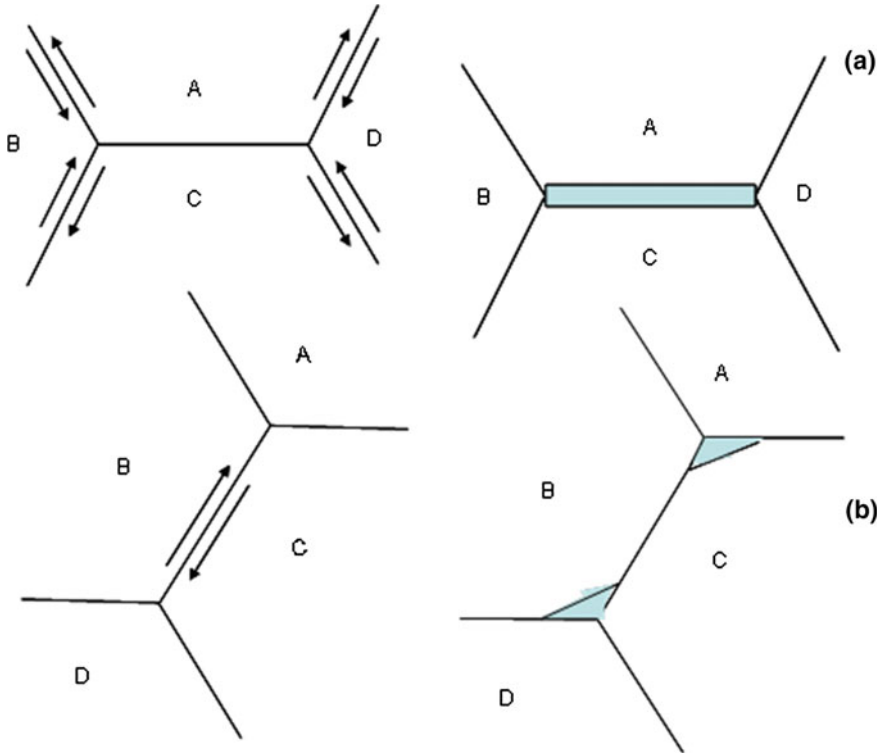


Fig. 2.6 Schematic views showing a more complex intergranular crack initiation by GBS. From Ref. [4]

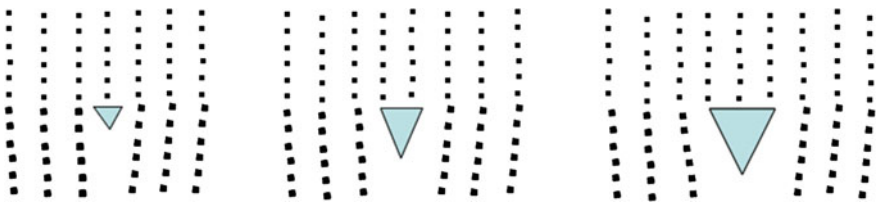
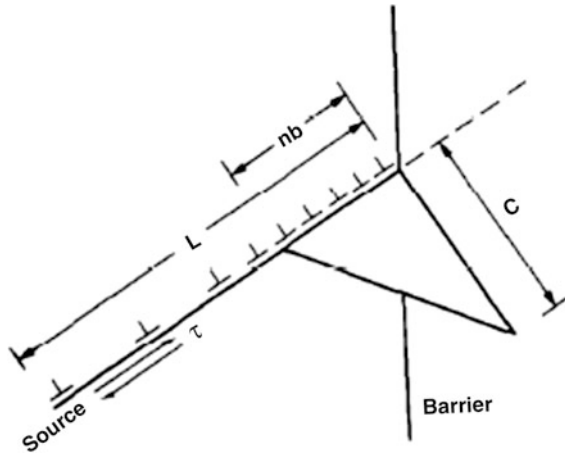


Fig. 2.7 A schematic illustration of Zener's idea, explaining how a crack of atomic dimensions can nucleate at dislocation sites; here, the growth of a crack is initiated by the coalescence of two or three dislocations (See: *Mechanical Properties of Ceramics*, Joshua Pelleg, Fig. 6.81)

$$c_{\min} = 2.4 \times 10^{-8} \quad (2.17a)$$

According to the theoretical presentation of Wu et al., a wedge crack may be formed by the insertion of extra material to create the head of a crack. An extra

Fig. 2.8 Nucleation of a wedge crack due to pile-up dislocations on a slip plane (Stroh's model). Sarfarazi and Ghosh [15]. With kind permission of Elsevier



plane, present above a positive-edge dislocation, may serve as the source of a wedge crack. This is Stroh's idea, based on Zener's original concept.

GBS may be considered as a deformation mechanism above $0.5T_m$. The strain rate is important to the type of failure caused by GBS. It has been shown that r-type cavities transform into w-types with increased strain rate, leading to transgranular fracture with increasing strain rate (Gandhi and Raj).

Alloying additions may decrease the tendency for w-type cavity formation. Both cavity types are the results of GBS (Raj). GBS may produce grain-boundary (intergranular) cracking when the grain's interior is stronger than its boundaries. GBS can be reduced by adding intergranular particles or by serrated grain boundaries. These serve as obstacles to GBS, apparently due to an increase in friction between the boundaries. Cavities have been seen to form at grain and phase boundaries preferentially at interfaces or triple points. The process of cavitation, associated with GBS and cavity nucleation, probably occurs at points of stress concentration in the sliding boundaries or interfaces. Creep failure occurs by the nucleation, growth and coalescence of creep cavities at the boundaries predominantly perpendicularly oriented to the applied stress. An increase in the number of cavitated boundaries over creep-exposure time supports a mechanism of continuous cavity nucleation and growth. Some believe, on the basis of experimental observations, that there are probably preexisting cavities, voids or pores, previously introduced by the forming processes that are actually responsible for creep cavitations in engineering alloys during long-term service at low stresses and elevated temperatures. Many experiments show that GBS is a necessary condition for cavity nucleation. GBS is a key factor not only in the growth of preexisting voids, but also in nucleating voids for cavity formation.

In many polycrystalline ceramics at elevated temperatures, GBS contributes significantly to the total strain. GBS can be markedly reduced by introducing additional phases, which form precipitates (such as nitrides, carbides, borides, etc.) at the grain boundaries. Another method for improving creep resistance in

materials is by the formation of serrated grain boundaries. Serrated grain boundaries are effective in improving creep-strength properties and do not permit continued creep by GBS when stress is applied at high temperatures. The effect of serration is equivalent to the ‘self-locking’ of the sliding process, resulting from creep deformation. Thus, materials with irregular, serrated grain boundaries have improved resistance to creep-crack growth when compared to those with smooth grain boundaries. This is explained as a consequence of the difficulty of GBS and the increase in the path of grain-boundary diffusion. The strengthening mechanisms of serrated grain boundaries are principally the result of: (1) the inhibition of GBS; (2) the retardation of grain-boundary crack initiation, caused by the decrease in stress concentration at grain-boundary triple points as a result of the decrease in GBS length and; (3) dynamic recovery at the serrated boundaries.

To summarize this section, note that GBS may account for 10–65% of the total creep strain, depending on the alloy and the conditions of its use in service (temperature, load, etc.). Its contribution to creep strain increases with rising temperature and stress and with reduced grain size. Above $\sim 0.6 T_m$, the grain-boundary region is thought to have lower shear strength than the grains themselves, probably due to the looser atomic packing at the grain boundaries. GBS may be reduced by introducing precipitates or grain-boundary serrations, which resist GBS and significantly reduce cavity formation of the types indicated above (which is a major factor in creep failure).

(iv) Diffusion Flow Caused by Vacancies

This mechanism of creep involves diffusion and various models have been suggested for diffusion-assisted creep. As such, it will be discussed in Chap. 3.

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