On the creep transition from superplastic behavior to nanocrystalline behavior

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Abstract

The present analysis shows that the creep transition from superplastic behavior to nanocrystalline behavior can be defined by an expression, which relates the normalized grain size, d/b (b is the Burgers vector) to the normalized shear stress, τ/G (G is the shear modulus).

Keywords: Boundary sliding, Creep transition, Dislocation climb, Dislocation pileup
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Under conditions of steady-state deformation, the creep rate, $\dot{\gamma}$, may generally be represented by a dimensionless relation of the form [1]:

$$\frac{\dot{\gamma} k T}{D G b} = A \left(\frac{b}{d} \right)^s \left(\frac{\tau}{G} \right)^n \text{ With } D = D_0 \exp \left(-\frac{Q}{RT} \right) \quad (1)$$

Where k is Boltzmann's constant, T is the absolute temperature, D is the diffusion coefficient that characterizes the creep process, G is the shear modulus, b is the Burgers vector, A is a dimensionless constant, d is the grain size, s is the grain size sensitivity, τ is the applied shear stress, n is the stress exponent, Q is the activation energy for the diffusion process that controls the creep behavior, and D_0 is the frequency factor for diffusion.

Transitions in the creep behavior of materials have been the subject of several studies [1-4]. These studies have led to the identification of various materials parameters and experimental conditions that favor the operation of a particular creep mechanism. For example, it has been shown that the mechanism of dislocation climb [1, 5] (stress exponent, $n = 5$, grain size sensitivity, $s = 0$, $Q \sim Q$ for bulk diffusion, and subgrain arrays are formed) controls high-temperature deformation in large-grained solid-solution alloy under the following conditions [3]: (a) a small atom misfit, (b) a low value for the stacking fault energy, (c) a low concentration of solute atoms, and (d) a large elastic modulus. If these conditions are not met, viscous glide [6] (stress exponent, $n = 3$ and grain size sensitivity, $s = 0$, $Q \sim Q$ for solute diffusion and subgrain formation is insignificant) will control the deformation behavior of the alloy [3].

The present communication focuses on examining the creep transition from behavior associated with superplasticity to behavior characterizing nanocrystalline materials.

It is well documented that when some alloys are deformed in tension tests at elevated temperatures, they exhibit superplastic behavior, which is manifested by large elongations, greater than 400 pct and sometimes in excess of 2000 pct. This phenomenon is referred to as superplasticity that is regarded as a creep phenomenon since it has been observed at temperatures at or above 0.4 of the melting point, T_m . The occurrence of superplasticity requires: (a) a fine and equiaxed grain size (less than 10 μm) that does not undergo significant growth during high-temperature deformation, and (b) mobile, high-angled grain boundaries that resist tensile separation. These requirements along with the strong sensitivity of steady-state creep rates measured during superplastic flow to changes in grain size have indicated that boundaries play an important role, which is related to their ability to contribute to deformation through the process of boundary sliding. Over the past four decades, considerable efforts have been made to characterize the nature and significance of such a role in terms of deformation mechanisms. As a result of these efforts, several of deformation mechanisms were developed or speculated [7-11]. The most successful mechanisms are based on grain boundary sliding (GBS) accommodated by dislocation motion [8-11]. These models are different in assumptions and details. Despite various differences in assumptions and details, all models based on GBS accommodated by dislocations predict: (a) a stress exponent, n , of 2, (b) a grain size sensitivity, s , of 2, and (c) an activation energy, Q , that approximates to that of grain boundary diffusion, Q_{gb} .

Detailed experiments were conducted on the superplastic Zn-22%Al eutectoid that contained nanometer-scale dispersion particles [12]. These particles were introduced in the matrix of the alloy via powder metallurgy followed by cryomilling. Transmission electron microscopy observations made on specimens crept at a strain rate near the center of the superplastic region have revealed that the initial microstructure is dislocation free; that after deformation, only some grains contain dislocations, which interact with dispersion particles; and that the configurations of the lattice dislocations in

the interiors of these grains are suggestive of viscous glide and single slip. The above characteristics are consistent with the model of Ball and Hutchison [8], in which a group of grains slide as a unit until blocked by an unfavorably oriented grain; a triple point is present. This process produces stress concentration at the triple point. Then, dislocations are generated and move in the blocking grain under the local stress. When the interior of the blocking grain is free of obstacles, dislocations move and then pile up at the opposite grain boundary until their back stress prevents further generation of dislocations. The dislocations at the head of the pile-up climb into and along the grain boundary [13], and the continual replacement of the dislocations would permit further grain boundary sliding.

Nanocrystalline (nc) materials are characterized by an average grain size that is ≤ 100 nm. Because of this characteristic, grain boundaries, junction lines, and nodes have significant volume fractions— substructural features that can influence properties far more strongly than in conventional materials [14].

An analysis of experimental creep data reported for nc materials [15] has revealed that a successful deformation mechanism is required to account for account for: (a) deformation is rate dependent, (b) an activation volume that has a value in the range of $10 b^3$ - $20 b^3$, (b) an activation energy that is close to the activation energy for boundary diffusion but that decreases with increasing stress, (c) the magnitudes of deformation rates that cover wide ranges of temperatures, stresses, and grain sizes and (d) inverse Hall–Petch behavior. Satisfying one or two requirements are *necessary but not sufficient*. Consideration of several proposed rate-dependent mechanisms [16-18] in light of these requirements has shown [15] that the predictions of the model of dislocation-accommodated boundary sliding (DABS) [19] are consistent with the above requirements.

The basic concept used in developing the model of DABS [15, 19] is that plasticity in nc-materials is the result of grain boundary sliding accommodated by the generation and

motion of dislocations under local stresses, which are higher than applied stresses due to the development of stress concentrations. The model was quantitatively developed using the following sequence of events: (a) as a result of sliding of a group of grains, the shear stress becomes concentrated at a triple point, or a protrusion that obstructs motion of this group; (b) this local high stress, which is higher than applied stress, can then generate dislocations in the blocking grain (or initiate voids); and (c) the generated dislocations move one by one to the opposite boundary where they climb to their annihilation sites, i.e., no dislocation pileup occur. Fig. 2 illustrates the details of the DABS model.

As mentioned earlier, one of the most successful deformation mechanism for superplasticity is the one based on the analysis by Ball and Hutchison [8] in which dislocations glide in the interior of the blocking grain and then pile up at the opposite grain boundary, where they climb into and along the grain boundary. The smallest number of edge dislocations in a dislocation pileup is two dislocations. Under this condition, the equilibrium distance between two edge dislocations, L , is given by:

$$L/b = (G/\tau)/[2\rho(1-\nu)] \quad (2a)$$

where ν is Poisson's ratio ($\nu = 1/3$). By substituting for the value of ν , Equation 4(a) becomes:

$$L/b = 0.25(G/\tau) \quad (2b)$$

Consideration of Fig. 1 and Fig. 2 shows that the transition in creep behavior from superplastic behavior to nanocrystalline behavior occurs when the dislocation pileup at the boundary in Figure 2 reduces to no dislocation pileup. Such a condition is met when the grain size, d , is less than L . On this basis, the transition from superplastic behavior to nc behavior occurs when:

$$d/b < 0.25(G/\tau) \quad (3)$$

Two points are in order regarding Equation (3). First, Equation (3) resembles in form the equation, which defines the transition from creep behavior controlled by dislocation climb ($n = 5$, $s = 0$, and $Q \sim Q_D$) to creep behavior characterizing superplastic flow ($n = 2$, and grain size sensitivity, $s = 2$, and $Q \sim Q_{gb}$), and which is represented by the following expression:

$$d/b < 10(G/t) \quad (4a)$$

The basis of Equation (4a) is the following. When the creep behavior of a material is controlled by dislocation climb, a regular array of subgrains is formed [1]. The subgrains have an average size of Q that is related to the normalized stress by an experimental expression of the following form [1]:

$$d/b = 10(G/t) \quad (4b)$$

It has been suggested [20] that when the grain size of the material, d , is reduced and becomes equal or smaller than the subgrain size, Q , a transition occurs from creep controlled by dislocation climb [1, 3] to creep characterizing superplastic flow, i.e., $d/b \leq d/b = 10(G/t)$. The above suggestion appears to be supported by the observation [20] that the transition stresses from region II (the superplastic region, intermediate stresses) to region III (high stresses) agree reasonably well with those predicted from Equation (6). Despite this agreement, experimental evidence is needed to indicate whether well-developed subgrains are formed in superplastic materials at high stresses where region III is observed.

Second, Equation (3) may be used to identify, to a first approximation, the lower bound of stresses (equivalent to the upper bound of grain sizes) for the transition from superplastic flow to deformation characterizing nc materials. In estimation this bound, two pieces of information are considered: (a) experimental data on the behavior creep of nc materials show [21] that the normalized shear stresses associated with these materials are $\sim 7 \times 10^{-4}$, and (b) data reported for superplasticity in micrograined materials indicate

[22] that the normalized maximum shear stresses at which the material continue to behave superplastically is $< 8 \times 10^{-4}$. Combining the above two pieces of information suggests that the transition to creep behavior characterizing nanocrystalline materials most likely occur at normalized shear stresses $> 7 \times 10^{-4}$. Substituting this value into Equation (5) and using $b \sim 0.3$ nm yields $d \pm 100$ nm, a characteristic of grain sizes associated with nanocrystalline behavior

In summary, by considering the details of the deformation mechanism responsible for superplastic flow and those of the mechanism recently proposed to account for the creep behavior of nc materials, the condition for defining the transition from the formal behavior to the latter behavior may be expressed as:

$$d/b < 0.25(G/\tau)$$

where d is the grain size, b is the Burger s vector, G is the shear modulus, and τ is the applied shear stress. The above equation resembles in form the equation that was suggested to define the transition from creep behavior controlled by dislocation climb to creep behavior characterizing superplastic flow and that is given by:

$$d/b < 10(G/\tau)$$

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Figure Captions

Fig.1. Schematic diagram for the model of Ball and Hutchison for superplasticity [8], showing that as a result of boundary sliding, dislocations are generated at a triple junction, traverse the blocking grain, and pile up at the opposite grain boundary where they climb and are annihilated.

Fig. 2. Schematic diagram for the model of dislocation-accommodated boundary sliding for nanocrystalline materials [19], showing that as a result of boundary sliding, dislocations are generated at a triple junction and then traverse the blocking grain one by one to the opposite grain boundary where they climb and are annihilated (no dislocation pileup)

Fig.1 . Schematic diagram for the model of Ball and Hutchison for superplasticity [8], showing that as a result of boundary sliding, dislocations are generated at a triple junction, traverse the blocking grain, and pile up at the opposite grain boundary where they climb and are annihilated.

Figure(2)