

Article

Power Law Breakdown in the Creep in Single-Phase Metals

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Abstract: New analysis provides insight into the basis of power-law breakdown (PLB) in the steady-state creep of metals and alloys. A variety of theories has been presented in the past but this new examination suggests that there is evidence that a dramatic supersaturation of vacancies leading to very high diffusion rates and enhanced dislocation climb is associated with the rate-controlling process for creep in PLB. The effect of vacancy supersaturation may be enhanced by dislocation short circuit diffusion paths at lower temperatures due to the dramatic increase in dislocation density.

Keywords: creep; power-law-breakdown; steady-state

1. Introduction

Steady-state deformation of a material occurs when there is a balance between hardening and dynamic recovery. For constant strain-rate testing, the stress is fixed during steady-state, and for constant stress tests, the strain rate is fixed at steady-state. Generally, at temperatures above about $0.6 T_m$, metals and class M alloys (behave like pure metals) obey a steady-state creep equation of the form [1]:

$$\dot{\epsilon}_{ss} = A_0(\chi/Gb)^3(D_{sd}Gb/kT)(\sigma_{ss}/G)^n, \quad (1)$$

where $\dot{\epsilon}_{ss}$ is the steady-state strain-rate (creep rate), A_0 is a constant, k is Boltzmann's constant, G is the shear modulus, b is the Burgers vector, n is the steady-state stress exponent, D_{sd} is the lattice self-diffusion coefficient and χ is the stacking fault energy. The steady-state creep behavior then has activation energy Q_{sd} that is equal to that for lattice self-diffusion. Hence, it is usually assumed that the five power-law creep is controlled by dislocation climb. Of course, Q_{sd} has two components—about half the value is the activation energy for vacancy formation and the other half is the activation energy for the atomic jumps into the vacancy. Therefore, the creep-rate is proportional to the vacancy concentration. It is important to emphasize that steady-state flow is a balance between the hardening process, which appears to be dislocation network refinement [2], and dynamic recovery; recrystallization mechanisms are not considered. Often a genuine steady-state is confused with a minimum creep-rate which may be due to insufficient strain to reach a true steady-state, or a material proceeding to steady-state but interrupted by stage III creep or tertiary creep in which fracture proceeds. That is Stage I to stage III without Stage II or a steady-state. In the case of 99.999% pure Al in Figure 1, the highest stress points are probably not reflective of steady-state due to DRX, but the author believes (consistent with the reviewer's opinion) the PLB (power-law breakdown) data just about the highest stress of the 5 PL regime are genuine steady-states reflective of a balance between hardening and dynamic recovery.

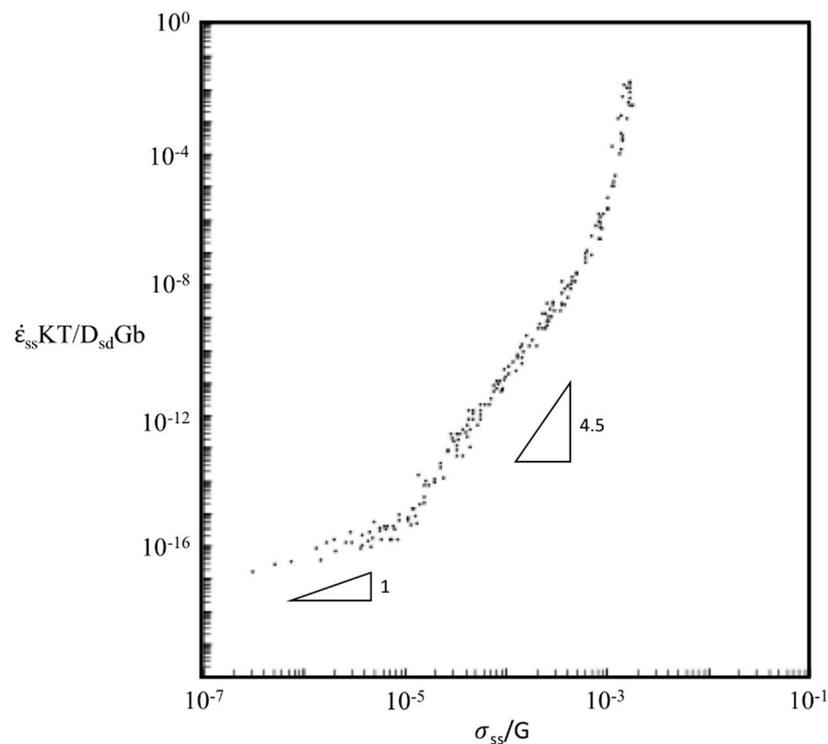


Figure 1. The steady-state creep behavior of high purity aluminum from temperatures ranging from sub-ambient to near the melting point [1,3–5] Figure adapted from [5].

The stress exponent is typically a value close to five, in Equation (1), although variations from four to seven have been observed. Thus, this regime of a constant stress exponent is often referred to as five-power-law-creep. This regime is illustrated in Figure 1 for high purity aluminum where the stress exponent is about 4.5 in the constant stress exponent regime at temperatures typically greater than $0.6 T_m$. At very high temperatures near the melting point, Harper–Dorn creep with $n = 1$ is sometimes observed, but this will not be discussed here. Below about $0.6 T_m$ the stress exponent is no longer constant with changes in stress and increases with increasing stress/strain-rate. This behavior is referred to as power-law breakdown (PLB). The explanation for PLB has been elusive.

Equation (1) can be extended to phenomenologically describe PLB including changes in Q_c , the activation energy for creep, with temperature and stress by the hyperbolic sine function [1,3].

$$\dot{\epsilon}_{ss} = A_1 \exp[-Q_c/kT] [\sinh \alpha_1 (\sigma_{ss}/E)]^5. \quad (2)$$

2. Discussion

Activation energy measurements for steady-state flow within PLB are rare. Luthy et al., [6] attempted to measure Q_c in this regime, but it now appears that the apparent “steady-state” at ambient and near-ambient temperatures in their work was actually influenced by discontinuous dynamic recrystallization (DRX) in addition to dynamic recovery [7]. Thus, it is unclear whether the Q_c , values measured are reflective of the activation energy for steady-state flow. Some of the high-stress data in Figure 1 may be somewhat low with DRX being present.

Some have tried to measure Q_c , at low temperatures by using temperature-change tests, but these were performed at relatively small strains that were probably far from the onset of steady-state [8]. The observed values were, nonetheless, much lower than those above $0.6 T_m$. A genuine steady-state Q_c was measured with very large-strain (torsion) deformation in silver. This is illustrated in Figure 2.

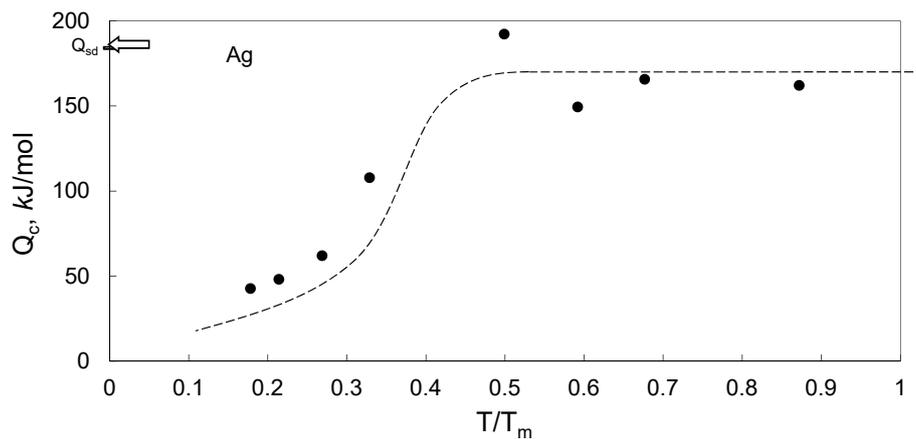


Figure 2. The activation energy for steady-state creep of silver as a function of temperature [9,10].

The strains to reach steady-state were typically greater than 3.0. We note that Q_c decreases with decreasing temperature and the decrease from an approximately constant value within the five-power law regime corresponds to the onset of PLB below about $0.6 T_m$. Robinson and Sherby [11] and Luthy, Miller, and Sherby [6] suggested that as the stress exponent increases into the PLB regime, creep is still dislocation climb controlled, but Q_c may reflect dislocation-pipe diffusion, Q_p [6,11]. Vacancy supersaturation resulting from deformation (moving dislocations with, e.g., jogs), could also explain this decrease with decreasing temperature (increasing stress) and still be consistent with dislocation climb control [11]. These two effects would significantly reduce the activation energy for steady-state creep. Early modeling by Mecking et al. [12] suggested that vacancy supersaturation by deformation in the power-law regime was unlikely, but suggested that supersaturation may occur within PLB. More recently, important modeling and experiments by others [13–15] also suggested that supersaturation may occur with significant plasticity within PLB. Wu and Sherby [16] however, subsequently suggested that internal stress, as did Nix and Ilshner, albeit from different sources, explains PLB behavior and appeared to abandon vacancy superstation effects.

Aside from vacancy supersaturation and short-circuit diffusion by dislocation “pipes”, dislocation glide mechanisms have also been suggested to be important by Arieli and Mukherjee [17], Weertman and Weertman [18] and Ilshner and Nix [19]. Ilshner and Nix [19] elegantly suggested that cellular dislocation structures rather than well-defined subgrains are formed in PLB and this gives rise to lower long-range internal stresses (LRIS) leading to PLB. There may be two difficulties with this approach. First, Kassner [20] and Levine et al. [21] report that both subgrain and cellular structure actually have low (e.g., 0.1σ) LRIS. Phan et al. [22] found that for severely deformed Al at an ambient temperature where the extensive formation of well-defined subgrains is observed, the LRIS, again, may be small. Second, recent studies observe that very well-defined subgrain boundaries form as a result of dislocation reaction even in low stacking fault energy metals such as silver and Zr [23] in PLB, also suggesting that substantial dislocation climb is at least occurring [7,9] in PLB.

Vacancy supersaturation resulting from deformation could explain the decrease in Q_c with decreasing temperature (increasing stress) and still be consistent with dislocation climb control. Again, the diffusion activation energy has two parts—one is the activation energy for vacancy formation, Q_v , and the other is the activation energy for atomic jumps into the vacancy, Q_m . If deformation creates excess (non-equilibrium) vacancies then the Q_m may dominate Q_c . There can be a further decrease in Q_c if short-circuit diffusion through dislocation pipes occurs.

The author published earlier work that illustrated that the increase in the steady-state stress exponent with PLB is coincident with the increase in the (annealed) constant-structure stress exponent defined by

$$N = [\partial \ln \dot{\epsilon} / \partial \ln \sigma]_{T,S'} \quad (3)$$

where “s” refers to the structure and N , thus, reflects the change in creep rate with a change in stress for a fixed structure. The steady-state stress exponent, n , refers to the change in the steady-state creep-rate for a change in stress leading to different structures. N , the constant structure strain-rate sensitivity is sometimes described by m ($= 1/N$). Sometimes the concept of a change in creep rate with changes in stress for a constant structure is described by an activation volume, V ($= Ab$ where A is an activation area and b is the Burgers vector). These terms are described in detail in the author’s earlier manuscript [9]. This manuscript utilized “ N ” to describe the change in creep-rate with a change in stress. N , m , V and A are sometimes utilized in discussions of creep and all are all quantitatively related.

Figure 3 illustrates the described coincidence. Activation area measurements on deformed silver suggest that for strains <0.2 , at low temperatures in the PLB regime, the dislocation intersection mechanism appears reasonable; deformation does not appear to be dislocation climb controlled. However, at much larger strains (e.g., >3.0), where a genuine steady-state is observed, the intersection mechanism was not verified as being rate-controlling [9]. The coincidence of increases in both N for annealed metal and n at similar temperatures and strain-rates implies that PLB is not related to the dislocation structure. This is inconsistent with the PLB explanation based on vacancy supersaturation and short circuit diffusion.

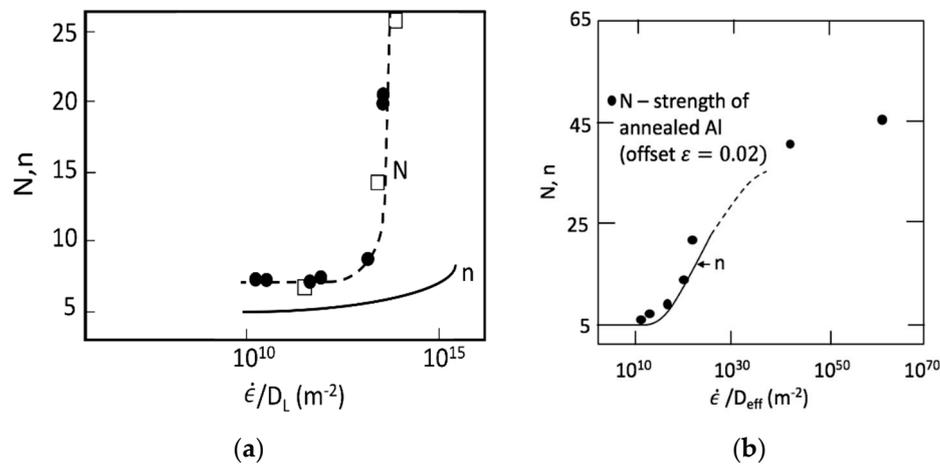


Figure 3. (a) the steady-state (n) and constant structure (N) stress exponents for annealed (broken line) and steady-state (solid line) structures of AISI 304 stainless steel as a function of lattice diffusion coefficient compensated strain-rate. (b) comparison of the constant structure stress exponents (N) of annealed aluminum with the steady-state stress exponent (n) based on earlier work by the author [24] and others [6,25,26]. D_{eff} is the effective diffusion coefficient that is the combined diffusivity considering dislocation pipes as described in [6].

Again, Robinson, and Sherby [11] suggested the possibility that dislocation pipe (fast) diffusion and vacancy supersaturation from plasticity may be responsible for PLB (but as mentioned earlier, Wu and Sherby [16] however, subsequently suggested that internal stress, as did Nix and Ilschner, explains PLB behavior and appeared to abandon vacancy supersaturation and dislocation pipe diffusion effects). If we assume just volume diffusion at ambient temperature for silver, which appears to achieve steady-state after strains greater than 3.0 at ambient temperature, the predicted vacancy concentration (based on the usual equations [27]) is roughly 10^{-17} . With substantial plasticity (e.g., >0.3) the latest experimental and theoretical estimates [12–15,28] of the vacancy concentration at ambient temperature after large strain deformation in metals is of the order of 10^{-3} to 10^{-5} (roughly a factor of 10^{13} higher than the equilibrium concentration). This suggests that the low-temperature diffusion coefficient may be relatively high giving rise to higher than expected creep rates leading to PLB. Thus, plasticity leading to vacancy supersaturation is certainly capable of rationalizing PLB. Also, for Al, the equilibrium vacancy concentration at ambient temperature is about 3×10^{-14} . Vacancy supersaturation by plasticity certainly also rationalizes the increase in creep rate in the PLB illustrated in Figure 1. Also, if dislocation

pipe diffusion rather than volume diffusion control climb (i.e., $D_{\text{eff}} = D_p > D_{\text{sd}}$) then the activation energy for diffusion would be roughly half that of volume self-diffusion and together with vacancy supersaturation, the dislocation climb-rate would be yet higher and Q_c might be a relatively small fraction of Q_{sd} such as in Figure 2 (D_p is diffusion coefficient in the plastically deformed metal at low temperatures by short-circuiting dislocations pipes and D_{sd} is the diffusion coefficient in the absence of short-circuiting). Short circuit diffusion by dislocation pipes [27], at an ambient temperature is also enough to rationalize PLB. The ratio of D_p/D_{sd} at ambient temperature approximately equals 10^{16} . Thus, PLB may be rationalized independently by dislocation pipe diffusion.

In summary, there have been four classes of explanations for PLB: (1) It has been proposed that there is a change in the mechanism of plastic flow from dislocation climb-control to glide-control in PLB. However, this proposition is sometimes based on the presence of internal stress which does not appear to be reasonable as LRIS appears to be low in both the PL and PLB regimes. The glide mechanism at low temperatures may be relevant for low plastic strains (e.g., <0.30), but it does not appear relevant to steady-state deformation. (2) Only recently has vacancy supersaturation been experimentally verified. The levels of supersaturation appear sufficient to create PLB. (3) The coincidence between the increase in N (especially in annealed metals) and the stress exponent, n , for steady deformation in PLB has been suggested to imply that the onset of PLB is not structurally related. But this coincidence has been principally verified for annealed structures for which steady-state is not relevant. Although it may be true that PLB-like behavior is coincidentally observed for n and N , the conclusion that PLB is not structurally related may not be justified. A detailed explanation for the coincidence is not apparent. (4) The changes in the diffusion coefficient from D_{sd} to D_p with large strain plasticity within the PLB regime may independently contribute to the observation of PLB. Another point must be made and that is references [14,15,28] verify vacancy supersaturation at ambient temperature; other, somewhat higher temperatures within PLB were not checked. Of course, higher temperature x-ray diffraction experiments are more difficult than those at ambient temperature. Perhaps future experiments could be performed at other temperatures within PLB to fully verify the coincidence between the onset of PLB and vacancy supersaturation.

3. Conclusions

Vacancy supersaturation and/or dislocation pipe diffusion appear to be the basis for the power-law-breakdown regime for steady-state creep in metals and alloys. Explanations based on changes in the basic mechanism for steady-state flow do not appear to be justified. The suggestion that the transition to PLB is not structurally related also does not appear to be justified.

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