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Theory for very low stress ("Harper–Dorn") creep

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Contrary to the usual characteristic of conventional Harper–Dorn creep, recent experiments by the authors suggest a stress exponent of ~3 in the Harper–Dorn regime and a variation in dislocation density with stress as a direct extension of the behavior anticipated within the five-power creep regime. A model for low stress creep, suitable for the Harper–Dorn regime, is presented in terms of a network-based model that is a modification of earlier work by Evans and Knowles. Our modifications to the model are consistent with new experimental results at very low stresses where subgrains are absent.

Keywords: High temperature creep; Low stress creep; Harper–Dorn creep; Dislocation network; Aluminum

Creep at very high temperatures (~0.9–0.99T_m, where T_m is the absolute melting temperature of the material) and low stresses ($\sigma/G \sim 10^{-5}$–10^{-6}) has often been described as belonging to the Harper–Dorn creep regime. Conventionally, Harper–Dorn creep is associated with a stress exponent, n, of 1 and activation energy, Q, equal to the activation energy for self-diffusion, Q_L, [1]. Recently, Kumar et al. [2,3] proposed a higher value for the stress exponent equal to ~3 for the Harper–Dorn regime based on: (i) their experimental results on [1 0 0]-oriented single crystal Al samples and (ii) critical examination of the various earlier studies supporting conventional Harper–Dorn creep behavior. There are several theoretical models for conventional Harper–Dorn creep [4,5] but none appear to be widely accepted. All such models suggest the independence of the dislocation density and the applied stress within the Harper–Dorn regime. As shown in Figure 1, Kumar et al. [2] reported that the dislocation density varies as the square of the applied stress even in the Harper–Dorn regime and the values of dislocation densities are consistent with the extrapolation of five-power-law regime behavior. Hence, the basic framework of all the earlier Harper–Dorn theories appears in doubt.

It is now well accepted that dislocations in steady-state structures at high temperatures and low stresses form a three-dimensional Frank polyhedron network [5]. Network models generally consider coarsening of the dislocation network due to recovery of sessile links ($l < l_c$ where l is the dislocation-link length and l_c is the critical dislocation-link length required for activating dislocation sources). A discussion of the network-based models by Ardell and Lee [6] in the Harper–Dorn regime (all link lengths < l_c) is presented elsewhere [4]. It predicts, contrary to recent experiments [2,3], a stress exponent of ~1. Evans and Knowles [7] presented a recovery-based model for the evolution of dislocation network that may also be consistent with natural three-power law at low stresses. The Evans and Knowles model [7] is attractive but predicts strain rates that are ~15 times faster than observed strain rates in the Harper–Dorn regime (Fig. 2).

At present, there is a need of a theoretical model for creep which is: (i) based on the dislocation network theory, as subgrains are generally not observed in this regime [2,3]; (ii) consistent with the variation of the dislocation density with the applied stress; and (iii) relevant to the very low stresses of the Harper–Dorn regime. In the present work, the model of Evans and Knowles [7] was carefully examined for any necessity for modification. It was modified in order to develop a dislocation network theory for creep in the Harper–Dorn regime.

It is assumed that the distribution of the dislocation-link lengths is uniform with the smallest length equal to the Burgers vector, b, and the largest link length equal to the critical link length for a Frank–Read source, L_c. It is assumed:

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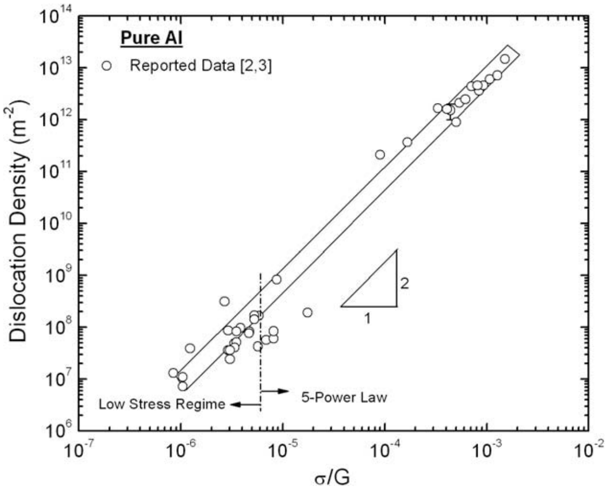

Figure 1. Dislocation density vs. normalized stress showing data measured on pure Al. Graph is reproduced from Refs. [2,3].

$$\bar{l} = \frac{1}{\sqrt{\rho}}, \tag{1}$$

where $\bar{l}$ is the average dislocation-link length and ρ is the dislocation density [8].

Phenomenologically, it is known for polycrystals that [5]:

$$\sigma_{ss} = k'Gb\sqrt{\rho_{ss}}, \tag{2}$$

where σ_{ss} is the steady-state stress value, k' is a constant and G is the shear modulus. Throughout this article, the subscript "ss" represents steady-state. Eq. (2) is shown to be consistent with a verified Taylor equation for five-power-law steady-state creep of polycrystalline Al [9]. The following equation can be written for single crystals, assuming that a similar equation as that of Eq. (2) is also valid for single crystals:

$$\sigma_{ss} = \frac{\alpha'}{S_{ave}} Gb\sqrt{\rho_{ss}}, \tag{3}$$

where α' is a constant and S_{ave} is the average Schmid factor of the active slip-systems. Combining Eqs. (1) and (3) gives the following expression for $\bar{l}$:

$$\bar{l} = \frac{1}{\sqrt{\rho}} = \frac{\alpha' Gb}{S_{ave}\sigma}. \tag{4}$$

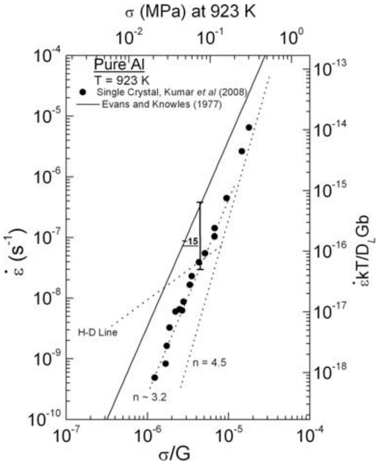

Figure 2. Comparison of the Evans and Knowles model [7] with the experimental data on pure Al single crystals as reported by Kumar et al. [2,3]. Lines with slopes 4.5 and 3.2 are shown in the plot. For the Evans and Knowles model [7] the equation predicting a stress exponent of ~3 is used and the pipe diffusion term is neglected.

By definition, the dislocation density is equal to the product of the average dislocation-link length and the total number of dislocation-links per unit volume, N. Hence, the following expression for N is calculated based on Eq. (4):

$$N = \rho^{\frac{3}{2}} = \frac{1}{\bar{l}^3}. \tag{5}$$

It is assumed that the glide of dislocation-links is rapid and creep is controlled by dislocation climb. For a three-dimensional dislocation network, both nodes and dislocation-links may climb. The following climb velocities are calculated [7]:

$$v_n = \frac{4\pi D_L F b}{\mathrm{k}T}, \tag{6}$$

$$v_l = \frac{2\pi D_L F b}{\mathrm{k}T \ \ln\left(\frac{\bar{l}}{2b}\right)}, \tag{7}$$

where v_n is the climb velocity of nodes, D_L is the lattice diffusion coefficient, F is the total force per unit length of the dislocations, k is the Boltzmann constant, T is the temperature and v_l is the climb velocity of dislocation-links. Eq. (6) assumes that the climb velocity of a single node is equivalent to that of a jog and is taken from Hirth and Lothe [10]. Eq. (7) is originally from Weertman [11]:

$$v_l = \frac{2\pi D_L}{b \ \ln\left(\frac{R}{b}\right)}\left[\exp\left(\frac{\sigma^*\Omega}{\mathrm{k}T}\right) - 1\right], \tag{8}$$

where R is distance from the dislocation to the point at which the vacancy concentration is equal to the equilibrium vacancy concentration, σ^* is the total stress acting on the dislocation that produces a climb force and Ω is the atomic volume. Eq. (8) reduces to the form of Eq. (7) based on the reasonable assumptions that $R = \bar{l}/2$, $\Omega = b^3$, $F = \sigma b$ and the activation volume, $\sigma^*\Omega/kT < 1$ (which is equal to $\sim 10^{-4}$ for the maximum stress used by Kumar et al. [2,3]). Assuming $\sigma^*\ \Omega/\mathrm{k}T < 1$ reduces $\exp(\sigma^*\Omega/\mathrm{k}T)$ to $\sim 1 + \sigma^*\Omega/\mathrm{k}T$.

The ratio of v_n and v_l, as given by Eqs. (6) and (7), is >1 (~20 for stresses used by Kumar et al. [2,3]) and hence the climb of dislocation-links is expected to govern creep in the Harper–Dorn regime. Based on the values of the pipe diffusion coefficient and the lattice diffusion coefficient, it can be concluded that at higher temperatures and lower stresses, as in the case of creep in the Harper–Dorn regime, lattice diffusion will be the rate-controlling processes.

There are two external forces on the dislocations: (i) the applied stress and (ii) the stresses due to other dislocations (elastic interaction). Nevertheless, there is a tendency of a dislocation-link to increase its length to reduce the line tension and hence this also contributes to the total climb force. The following is an approximate expression, similar to that in Evans and Knowles [7], for F: