

# RECENT DEVELOPMENTS IN UNDERSTANDING THE CREEP OF ALUMINUM

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**Abstract.** This paper summarizes the recent creep research by the author on pure aluminum including classic five power-law creep, Harper-Dorn creep, ambient-temperature creep and an investigations of long-range internal stresses in creep-deformed aluminum. Many of the models and theories for these phenomena persisted for a relatively long period of time. More recent developments in these phenomena are discussed that may lead to new interpretations of creep in aluminum, as well as creep in other crystalline materials.

**Keywords:** creep, Harper-Dorn, five-power law creep, dislocation hardening, long-range internal stress, low-temperature creep

## 1. Introduction

This paper reviews recent advances in the basic understanding of creep in pure aluminum over a very large temperature range. It updates the general thinking of the creep community over the past ten to twenty years with particular attention directed toward the two years since the publication of reviews by this author [1, 2]. The topics discussed are: Harper-Dorn creep (highest temperature) followed by five-power law creep (down to about  $0.6T$ ) and low temperature creep, (less than about  $0.3T_m$ ). Internal stresses are discussed in a separate section as they may impact creep at all temperatures.

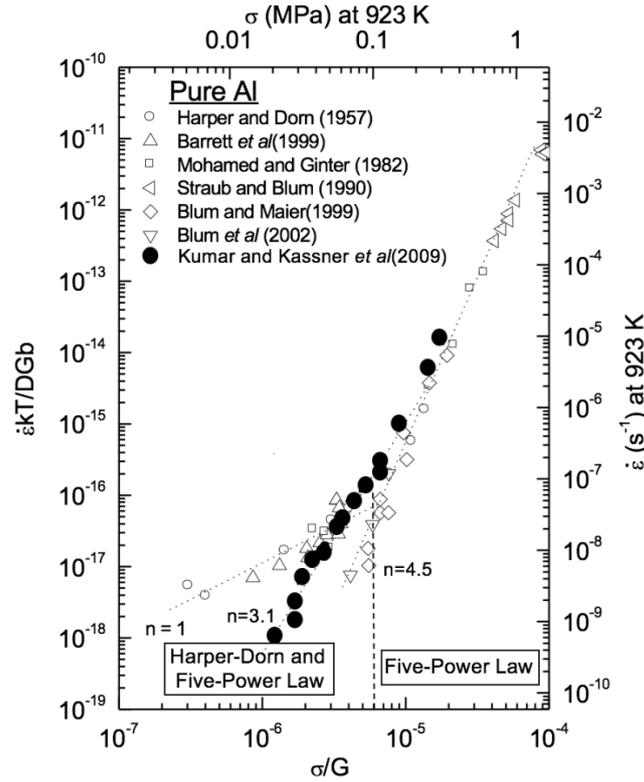
## 2. Harper-Dorn Creep

Harper-Dorn creep has been described by the equation

$$\dot{\epsilon}_{ss} = A_{HD} \left( \frac{D_{sd} G b}{kT} \right) \left( \frac{\sigma}{G} \right)^n, \quad (1)$$

where  $A_{HD}$  is the Harper-Dorn coefficient,  $D_{sd}$  the lattice self-diffusion coefficient (with  $Q_{sd}$  being the activation energy for lattice self-diffusion),  $G$  is the shear modulus,  $b$  is the Burger's vector, and  $\sigma$  is the stress.

Classic Harper-Dorn creep has been observed at higher temperatures, very near the melting temperature [1, 3, 8, 12]. Harper-Dorn creep has been generally suggested at high temperatures for a wide variety of crystalline materials. The generally accepted characteristics of Harper-Dorn creep include  $n=1$ ,  $Q_c$  approximately equal to the activation energy for lattice self-diffusion,  $Q_{sd}$ , but many have suggested that unlike five power-law creep, the steady-state dislocation density is independent of the (modulus-compensated) stress.

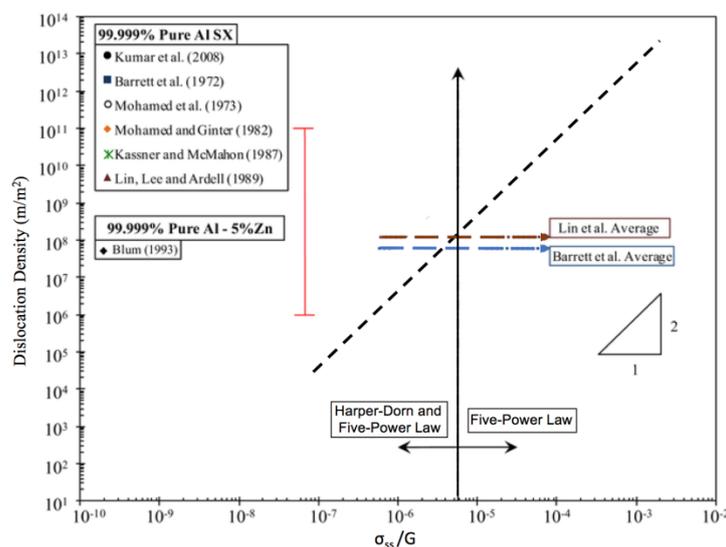


**Fig. 1.** Steady-state creep of high-purity aluminum at high temperatures [3-9] Dark symbols are from the authors earlier work [3]. Interestingly, *both* five-power-law creep and Harper-Dorn Creep have been observed at the same low stresses

Interestingly, subgrains are generally not observed in the Harper-Dorn regime, and the Frank dislocation network would, by default, be the microstructural feature associated with the rate-controlling process for creep, as often suggested for five power-law creep. More recent [3] works suggest that Harper-Dorn may not be observed in Al at these very high temperatures, but rather five-power-law creep extends into this temperature/stress range uninterrupted by Harper-Dorn creep. Fig. 1 shows classic five-power law ( $n = 4.5$ ) behavior at higher stress, but at lower stresses, such as below  $\sigma/G = 10^{-6}$ , either  $n = 1$  (Harper-Dorn) *or*  $n = 3-4.5$  (five-power law) behavior is observed. Recent work by the author [10] demonstrated that if the initial dislocation density is high (e.g. above) then the Frank dislocation network may frustrate at lower stresses, as suggested by Ardell and coworkers [11], and the dislocation density is stable at these values within the so-called Harper-Dorn regime. Thus, the dislocation density is constant and  $n = 1$  is observed. It can be shown that if the dislocation density is constant, the creep rate would be expected to be proportional to the stress since the climb rate is proportional to the stress.

On the other hand, if the initial dislocation density is low, such as at values below those suggested in Fig. 2 by extrapolation of the dislocation density versus stress at low stresses into the so-called Harper-Dorn regime, then normal five-power law creep behavior may be observed. The dislocation density can increase to values expected by on the trends of Fig. 2 extended into the Harper-Dorn regime. Typical starting dislocation densities are indicated by the large vertical bracket in Fig. 2. The figure illustrates that the observed starting dislocation densities can either be below or above the "transition value" of  $10^8 \text{ m}^3$ . It has been observed that initial high dislocation density values (e.g.  $> 10^8 \text{ m}^3$ ) are stable at temperatures near  $T_m$ , for periods over *one year*. Thus, dislocation frustration appears to be a genuine phenomenon. Analysis of the Orowan bowing stress (i.e  $\tau = Gb/l$ , where  $l$  = average network link-length) suggests that at dislocation densities of  $10^8 \text{ m}^3$ , the bowing stress for

multiplication ( $\sigma/G$ ) is about  $4 \times 10^{-6}$ . Interesting, this is the stress at the transition point to Harper-Dorn Creep in Fig. 2. If the stress is below this value, then multiplication (e.g. Frank-Read) may not occur for this initial dislocation density since the stresses are insufficient for multiplication since the average link-length is too small for the corresponding stress. However, if the dislocation density is relatively low at  $10^6 \text{ m/m}^2$ , the stresses are sufficient within the Harper-Dorn regime to cause dislocation multiplication and yield five-power law behavior; the dislocation density can change with the applied stress.



**Fig. 2.** Steady-state dislocation density versus the modulus-compensated steady-state stress at an elevated temperature of 923 K ( $0.99 T_m$ ) based on earlier work and the authors' previous work [1]. The data of Lin et al. [13] and that of Barrett et al. [5] suggest a lower limit of the dislocation density ( $\rho$ ) with creep with decreasing stress at these very high temperatures. However, the work by Barrett et al. [5] and Kumar et al. [3] may suggest a continual decrease in the dislocation density with decreasing stress. Note that the initial dislocation densities (vertical bracket on left-hand side of the figure) are of the same order as the steady-state dislocations densities at low (e.g. Harper-Dorn) stresses. Other data in the figure are also included [8, 13-15]

In summary, Harper-Dorn may be observed when the starting dislocation density is relatively high, but (normal) five power-law creep is observed with initial dislocation densities that are relatively low and frustration does not occur.

### 3. Five-power-law

**Diffusion (dislocation climb) control.** A well-accepted equation [1] for five power-law creep is:

$$\dot{\epsilon}_{ss} = A_1 \left( \frac{\chi}{Gb} \right)^3 \left( \frac{D_{sd}Gb}{kT} \right) \left( \frac{\sigma_{ss}}{G} \right)^5, \quad (2)$$

where  $\chi$  is the stacking fault energy,  $A_1$  is a constant,  $D_{sd}$  is the self-diffusion coefficient, and  $G$  is the shear modulus. It is fairly well established that steady-state creep in pure metals and class M alloys is controlled by dislocation climb. Some, however, have suggested that dislocation cross-slip is the rate controlling mechanism [16-19]. Recent experiments on large-strain softening in aluminum by the author add additional support for the climb contention [20,21]. The strain softening (roughly 17%) after large strain (generally  $>2$ ) deformation in torsion has generally been attributed to a decrease in the average Taylor factor associated with the texture development. That is, the Schmid factor of dislocation glide decreases. However, it was recently demonstrated that the texture also predicts that softening

will occur due to an increase in the climb stress as the torsional texture develops (roughly 7%). This is an important finding as unless the climb stress increased with texture formation, leading to softening, then some question would be placed on the viability of dislocation climb as the rate-controlling mechanism for five-power-law creep. The climb stress increase may not account for all of the softening and some of the observed flow stress decrease may be attributed to slight microstructural changes as some (e.g. Nes and Pettersen, Myshlyaev et al. and Perdrix et al.) have suggested [22-24]. Interestingly, elevated-temperature compressions tests immediately following the elevated temperature torsion tests into the softened state evince an absence of softening which is perfectly consistent with the predictions of climb-control. Glide control predicts a 10% increase in stress following compression which was not observed. In summary, the softening behavior appears much more likely due to a change in the climb stress associated with texture development rather than dislocation glide control.

**Rate-controlling process.** The precise mechanism for creep in the five-power law regime by dislocation climb is not well-established. There are two general lines of thinking with respect to the details of the rate-controlling process: 1.) subgrain-based theories and 2.) Frank-network theories. Analysis by the author suggests that the steady-state stress in aluminum is predictable by the Taylor dislocation-hardening equation,

$$\sigma_y|_{\epsilon, T} = \sigma'_o + \alpha M G b \sqrt{\rho}. \quad (3)$$

This equation very accurately predicts the elevated-temperature flow stress independent of the subgrain size.  $\alpha M G b \sqrt{\rho}$  is the athermal term. The associated constant,  $\alpha$ , in the predictive equation is within the range of values observed at lower temperature where dislocation hardening is undisputed [25-26]. One important note is that even in high purity Al, the values of  $\sigma'_o$  (the thermally activated term) are a significant fraction of the total stress (e.g. roughly one-third). This implies that the small amounts of impurities have a significant influence on the flow stress. Thus, the solute strengthening may complicate the analysis of dislocation hardening in that it is not completely clear how the separate strengthening variables combine to contribute to the flow stress. Here, a simple linear superposition is assumed. In summary, there is strong evidence that elevated temperature hardening is completely explainable by Frank-network hardening rather than strengthening by subgrains as commonly suggested.

#### 4. Long Range Internal Stresses (LRIS)

Long range internal stresses (LRIS) that are 3-20 times larger than the applied stress in creep [27] are often believed to exist in creep-deformed aluminum and other crystalline materials. These high levels of LRIS have been proposed to be associated with dislocation heterogeneities such as subgrain boundaries and are associated with the rate-controlling process for creep. Convergent-beam electron diffraction (CBED) and synchrotron x-ray micro-diffraction experiments suggest that LRIS values are much lower than suggested by the early experiments and theoretical projections by others. However, one difficulty with CBED is that the regions of TEM foils used for CBED are very thin and dislocations, particularly in metals as Al where the stacking fault energy is relatively high, may escape from the foils and relax any internal stresses. The foils may also bend in these thin sections and complicate the stress analysis. Static recovery, including recrystallization, at ambient temperature is also possible [28]. Finally, the CBED analysis is performed in an unloaded specimen at ambient temperature. Of course, *in-situ* assessment (i.e. under load) would be preferred. Some *in-situ* x-ray work was performed by Blum and coworkers on a Cu alloy [29] where the details of x-ray peaks were analyzed from reflections by a large number of grains in the specimen. The x-ray peak asymmetry was related to the long-range internal stress. It was shown that on unloading, most (e.g. 2/3) of the LRIS remain at ambient temperature in the unloaded state.

This suggests some validity in assessing LRIS by us in unloaded specimens in our CBED work. The problem with pure Al is that the stresses are very low in creep and the lattice-parameter change measurement (for LRIS) uncertainly with the x-ray peak asymmetry experiments is of the order of the applied stress. Thus, LRIS may not be detected. In fatigued pure single crystal Cu where the stresses are much higher than in Al, LRIS were also not observed by CBED [20]. At higher stresses in heavily deformed commercial-purity Al by equal channel angular pressing (ECAP) (impurities preclude ambient temperature recrystallization [28]) LRIS, of the order of the applied stress, by both CBED and by x-ray microbeams at a synchrotron facility [30], are observed.

In summary, there is no evidence for significant long range internal stresses in creep deformed metals within the five power-law regime as investigated by convergent beam electron diffraction. In recent cases where LRIS are confirmed by x-ray microbeams in ambient temperature severely deformed aluminum [30], the CBED consistently finds the same magnitude of long range internal stress as the more reliable x-ray microbeams (bulk samples) at a synchrotron facility. Thus, the absence of significant LRIS in creep-deformed pure Al by CBED experiments has additional validity.

### 5. Low Temperature Creep in Aluminum

Analysis of low-temperature experiments of creep below  $0.3T_m$  reveals that the activation energy for creep is relatively low and cannot be easily explained by a dislocation-climb controlled process. The observed activation energies below  $0.3T_m$  range from about 0.75 down to  $0.1Q_{sd}$ , the value of lattice self-diffusion activation energy [31, 32]. At low temperatures we observe primary creep at the small strains rather than steady-state creep as with the discussion in the above sections. Metallic primary-creep can usually be described by either a power-law or a logarithmic equation:

$$\varepsilon_p = at^b. \quad (4)$$

$$\varepsilon_p = a \ln t + c_2. \quad (5)$$

Our analysis revealed that most of the metals and alloys deformed at low temperature under constant stress conditions exhibit logarithmic behavior. The analysis of aluminum shows that at  $0.32 T_m$  the creep behavior is better described by the power law equation. Our analysis showed that, in general, logarithmic behaving metals approached power-law behavior as the temperature increased into the intermediate temperature range (i.e.  $T > 0.3T_m$ ). At a lower temperature  $\{77K (0.08 T_m) \tau_G$ , the backstress, is assumed to be approximately equal to zero [33, 34]}, both behaviors were evident in Al but at lower stresses, logarithmic behavior was observed.

The mechanism for creep plasticity at  $T < 0.3T_m$  is unclear. While for metals such as copper, (lower stacking fault energy), the classic Seeger rate-equation

$$\dot{\gamma} = NAbv_o \exp\left\{\frac{-\Delta H_o - v(\tau - \tau_G)}{kT}\right\}, \quad (6)$$

where  $\dot{\gamma}$  = strain-rate,  $N$  = number of dislocation segments per unit volume held up at the intersection points of mean spacing,  $l$ ,  $v_o$  is an atomic frequency of the order of the Debye frequency,  $\Delta H_o$  = energy required for the intersection process, (i.e. the energy for jog formation  $\approx \frac{Gb^2}{10}$ ) describes the creep behavior and measured creep rates of our Cu data fairly well, the Seeger equation fails to properly describe the behavior of Al.

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