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Harper–Dorn creep

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Abstract

The fundamental mechanism of time-dependent plasticity, or creep, of crystalline materials at very low stress is, at present, particularly controversial. At ''moderate'' stresses above about 0.6 T_m, over a wide range of stresses, five-power-law creep is observed that is widely believed to be dislocation climb-controlled, with the activation energy for creep to be close to that of vacancy diffusion. Steady-state creep (Stage II or secondary creep) is usually described by the steady-state stress exponent, n, which is typically 4–7. At lower stresses, Harper–Dorn has usually been the mechanism suggested to occur. The relation between the applied stress and the steady-state creep-rate for Harper–Dorn creep is phenomenologically described by

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

where A_{HD} is a constant. Since these early observations, Harper–Dorn creep has been reported to occur in many materials. Recent observations have suggested that in Al (the most extensively studied system) and all other systems, Harper–Dorn may be artificial in that the high creep rates may be a consequence of additional or unusual restoration mechanisms that do not normally occur in five-power-law creep or tensile fracture. These restoration mechanisms include dynamic recrystallization and/or grain growth. Others have suggested that Harper–Dorn is also artificial but because the strain-rates reported were not genuinely steady-state and, consequently, as much as an order of magnitude higher than genuine steady-state behavior. Others also found that they could not repro-

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duce the original Al data of Harper and Dorn, for unknown reasons. The analysis of other materials in which Harper–Dorn has been proposed to occur has been placed in question.
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1. Introduction

The steady-state, time-dependent plasticity, or creep, at high and intermediate temperatures of pure metals, type M alloys and many ceramics and minerals has been studied over a fairly wide range of stress. Classic five-power-law behavior and power-law breakdown which are frequently observed is illustrated in Fig. 1. At low stress (often at high temperatures) the steady-state creep rate is often suggested to evince Newtonian, or one-power, behavior. The figure has some – but not all – of the steady-state creep data of Al but describes the general trends of Al and many other metals and ceramics to start the discussion of this review. Some investigators distinguish steady-state creep and secondary creep. Steady-state generally refers to a fixed creep-rate under a constant stress due to a steady state (i.e., strain independent) substructure and dynamic recovery balances hardening. Often this substructure is defined in terms of the dislocation features such as subgrain-size, (network) dislocation, density, and subgrain misorientation. Secondary creep can appear independent of steady-state, e.g., a minimum creep rate as an inflection between tertiary and primary creep. Often, secondary and steady-state creep are coincident.

A new low-stress mechanism for creep at high temperatures and low stresses in Fig. 1 was originally proposed by Harper and Dorn (1957). This mechanism has since been
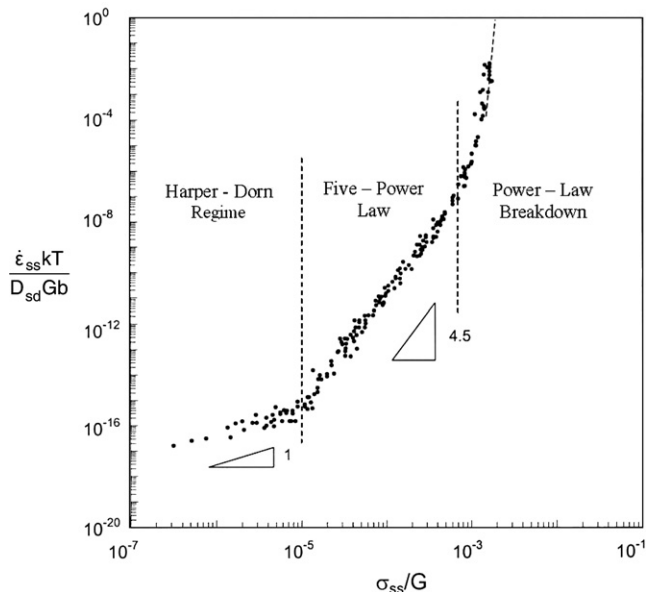

Fig. 1. The compensated steady-state strain-rate versus the modulus-compensated steady-state stress for pure Al, from Straub and Blum (1990).

termed "Harper–Dorn creep". By performing creep tests on aluminum of high purity and large grain sizes, these investigators found that the steady-state creep-rate increased linearly with the applied stress and the activation energy was that of self-diffusion. The observed creep process could not be ascribed to Nabarro–Herring or Coble diffusional creep, a theoretical, 1-power, mechanism. They reported creep rates as high as a factor of 1400 greater than the theoretical rates calculated by the Nabarro–Herring or Coble models. The existence of this *very* low-stress and commercially unimportant creep is under substantial question (Kassner and Perez-Prado, 2004). The observations of Harper and Dorn were reported some years later by Barrett et al. (1972), Mohamed et al. (1973), and Ardell and Lee (1986), as summarized in Fig. 2. The right-hand portion shows a few five-power-law data points, but the left hand portion shows mostly 1-power-law trends. A primary stage of creep was observed, which would not be expected according to the Nabarro–Herring diffusional model since the concentration of vacancies immediately upon stressing cannot exceed the steady state value. Furthermore, grain boundary shearing was reported to occur during creep and similar steady-state creep-rates were observed in aluminum single crystals and in polycrystalline specimens with a 3.3 mm grain size. This evidence led these investigators to conclude that low-stress creep at high temperatures in materials of large grain sizes occurred by a dislocation-climb mechanism. [Note that the applied stress, σ, has not been reduced by a "threshold stress", in contrast to Harper and Dorn.]

The relationship between the applied stress and the steady state creep rate for Harper–Dorn creep is phenomenologically described by Yavari et al. (1982))

$$\dot{\varepsilon}_{ss} = A_{HD} \left(\frac{D_{sd} G b}{kT} \right) \left(\frac{\sigma}{G} \right)^1 \tag{1}$$
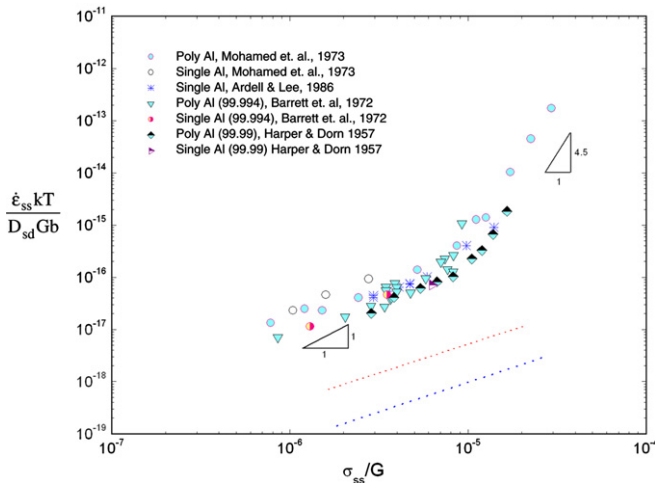

Fig. 2. A comparison between the diffusion-coefficient compensated strain-rate versus modulus-compensated stress for pure aluminum based on early data (Harper and Dorn, 1957; Barrett et al., 1972; Mohamed et al., 1973; Ardell and Lee, 1986), with theoretical predictions for Nabarro–Herring creep (Yavari et al., 1982) (dashed lines) for 3.3 and 9 mm grain sizes. Harper–Dorn creep was presumed in the low stress-exponent regime, a different mechanism than five-power-law creep at higher stresses. The purities are listed, when specified.

where A_{HD} is a constant, k is the Boltzmann constant, T is temperature. The coefficient parameters D_{sd} of self-diffusion, the shear modulus G and the Burgers vector length b are the characteristics of the material. Harper–Dorn creep has been reported to occur in a large number of metals and alloys since 1957 as well as a variety of ceramics and ice (see Table Ia). Interests in ceramics/minerals lies in predicting low-stress creep plasticity in geological systems such as the lower crust, lower mantle and inner core of the Earth (Van Orman, 2004). Fig. 3 illustrates low stress-exponent behavior in Cu. Metallic systems are sometimes chosen on the basis of expected satisfactory service in long-term structural applications. Fracture can be plastic-deformation rate dependent (Kassner and Hayes, 2003). The importance of Harper–Dorn may have been enhanced by the suggestions that diffusional creep does not occur, rather (rare) cases of (Newtonian) diffusional creep are actually Harper–Dorn (dislocation Newtonian creep) (Ruano et al., 1988).

Several studies have been published over the last 25 years that distinguished Harper–Dorn from classic five-power-law creep (and diffusional creep). Yavari et al. (1982) provided more evidence that the Harper–Dorn creep rate is independent of the specimen grain size. Similar rates were observed both in polycrystalline materials and in single crystals (unlike diffusional creep). Yavari et al determined, by etch-pits, that the dislocation density was relatively low and *independent* of the applied stress, unlike five-power-law creep. Owen

Table Ia
Metals for which Harper–Dorn creep has been suggested to operate with relevant references

Material	References	
Al	Ardell and Lee (1986)	McNee et al. (2001)
	Ardell and Przystupa (1984)	Mohamed et al. (1973)
	Barrett et al. (1972)	Mohamed and Ginter (1982)
	Blum and Maier (1999)	
	Harper and Dorn (1957)	Yavari et al. (1982)
Pb	Mohamed et al. (1973)	
α-Ti	Malakondaiah and Rama Rao (1981)	
α-Fe	Cadek et al. (1969)[a]	Fiala et al. (1991b)
	Cadek and Milicka (1968)	Longdale and Flewitt (1978)[a]
	Davies and Williams (1969)[a]	
	Fiala et al. (1983)	Towle and Jones (1976)
α-Zr	Ardell and Sherby (1967)[a]	Novotny et al. (1985)
	Bernstein (1967)	Pahutova and Cadek (1973)
	Fiala et al. (1991a)	Perez-Prado et al. (2005)[a]
	Langdon (2006)	Prasad et al. (1992)
	MacEwen et al. (1981)[a]	
β-Co	Malakondaiah and Rama Rao (1982)	
Sn	Mohamed et al. (1973)	
Cu	Barrett and Sherby (1964)[a]	Pahutova et al. (1971)[a]
	Barrett et al. (1967)[a]	Raj and Langdon (1991)[a]
	Kassner et al. (2002)[a]	Srivastava et al. (2005)[a]
	Lloyd and Embry (1970)[a]	Wilshire and Palmer (2002)[a]
	Muehleisen et al. (1970)[a]	Ya Pines and Sirenko (1963)[a]

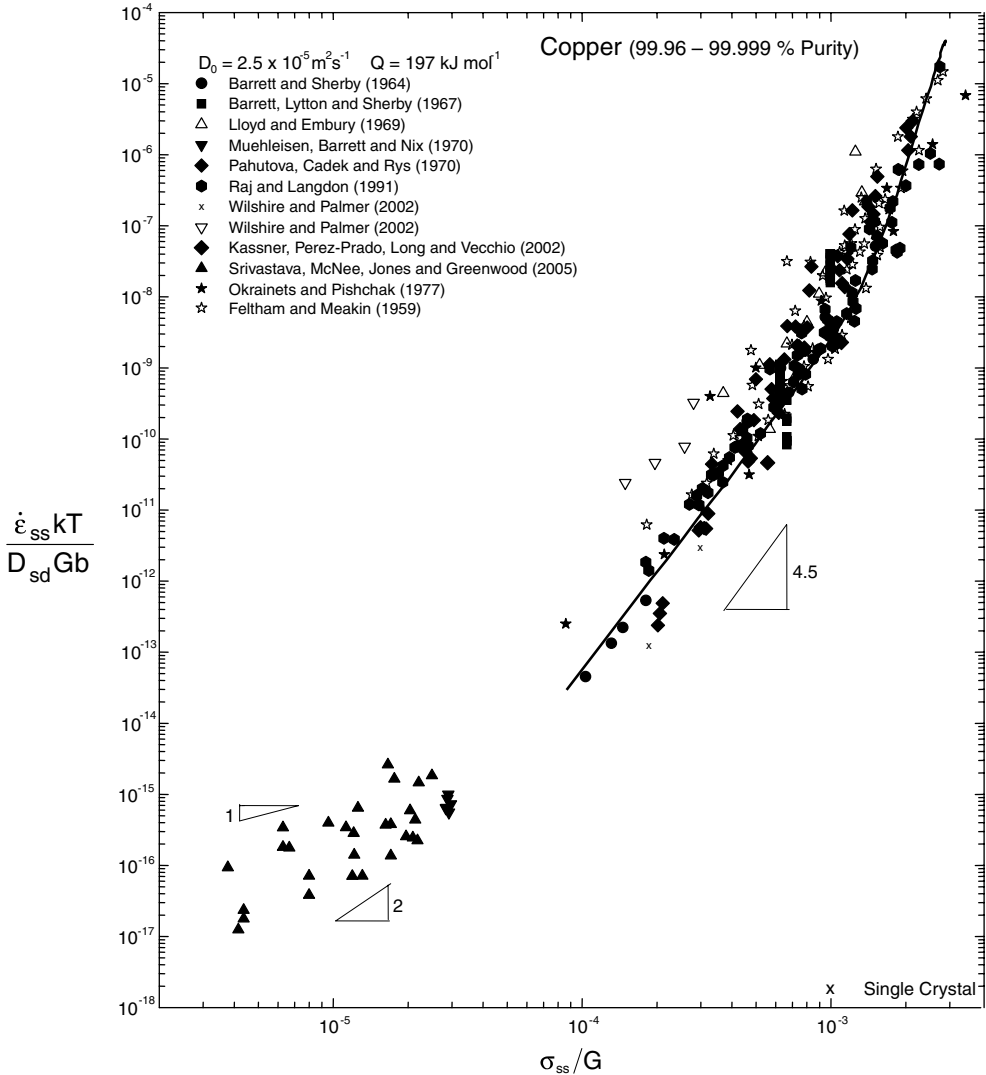[a] For basic creep papers making no reference to Harper–Dorn creep.

Fig. 3. The steady-state creep behavior of copper.

and Langdon (1996) suggested that the values of the dislocation density of Al-5%Mg are near $10^9 \, \text{m}^{-2}$. (Dislocations were found to be predominantly close to edge orientation.) Ardell (1997) suggested using etch-pit analysis that the dislocation density would not reach values less than a "frustration" level, effectively rendering the density independent of stress in aluminum of about $10^8 \, \text{m}^{-2}$ in Al using etch pits. Barrett et al. also found ρ stress-independent in Al at $7 \times 10^7 \, \text{m}^{-2}$ using etch pits. Nes (1998), however, suggested that the dislocation density using X-ray topography apparently showed that ρ was dependent on stress by $\sigma^{1.3}$; however, the test temperature $\leqslant 693 \, \text{K}$ was rather low so that creep was presumably negligible at the low stresses applied.

Fig. 4 displays the average dislocation spacings $\rho^{-0.5}$ as function of normalized stress. It is seen that all data lie in vicinity of bG/σ. Allowing for a typical experimental uncertainty
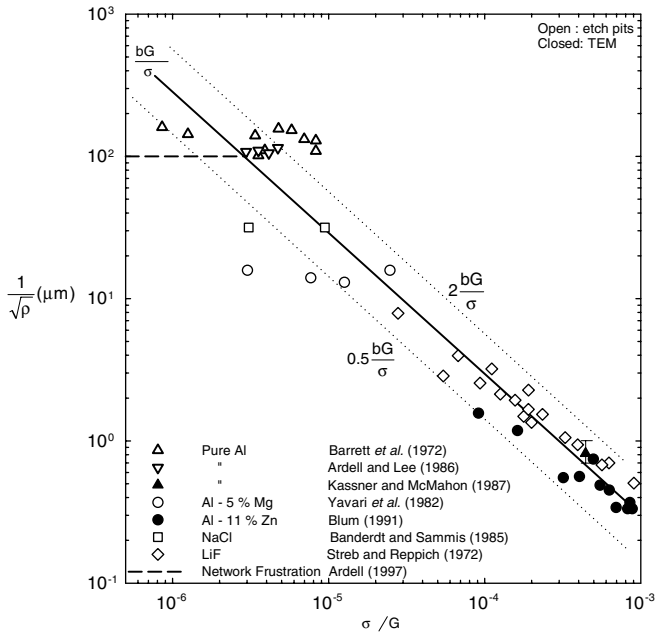
Fig. 4. Dislocation spacings versus normalized stress for pure Al, Al-5% Mg and NaCl in the H–D regime. Data for Al (Kassner and McMahon, 1987), LiF (Streb and Reppich, 1972) and Al-5% Zn (Blum, 1991) in 5-power regime and lines proportional to bG/σ are shown for comparison. The transition to Harper–Dorn is suggested to occur at $\sigma/G = 10^{-5}$ for NaCl and 10^{-6} for Al. *Note.* The values for physical constants (G and b) for Al-5% Mg and Al-5% Zn used for plot is taken to be same as of pure Al.

for determination of dislocation spacings of a factor of 2, the majority of the data points in the HD regime lies in the bG/σ band. Ardell's proposal of frustration of network coarsening (discussed more later) implies that the dislocation spacing would approximately follow bG/σ in the 5-power regime until frustration sets in and the data level off. However, for Al the frustration level given by Ardell is not supported by the Al data to a significant extent, if typical experimental scatter is allowed for. From the results of Barrett et al. (1972) for stress dependent subgrain size, it is also clear that the total dislocation density, i.e. the sum of lengths of free dislocations and subgrain boundary dislocations per crystal volume, will depend markedly on stress (note that the density of boundary dislocations outweighs by far the density of free dislocations at low stresses). Rather, the majority of data lies above bG/σ. Best support for network frustration seems to come from the data for Al–Mg. However, it depends on a single data point and the frustration level is the one order of magnitude in dislocation spacing (two orders of magnitude in dislocation density) below the Al-level. It seems that at the present state of microstructural knowledge there is no conclusive evidence for constancy of dislocation spacings in the HD regime.

The fact that the activation energy for Harper–Dorn creep is about equal to that of self-diffusion suggests that Harper–Dorn creep occurs by climb of edge dislocations. Weertman and Blacic (1984) suggested that creep is not observed at constant temperatures, but only with low amplitude temperature fluctuations, where the vacancy concentration would not be in thermal equilibrium, thus leading to climb stresses on edge dislocations of the order 3–6 MPa. This explanation does not appear widely accepted,

partly due to the observation that H–D creep is consistently observed by a wide assortment of investigators, presumably with different temperature control abilities (Nabarro, 1989).

In summary, the early low-stress experiments, primarily in metals, indicated that Harper–Dorn includes:

1. activation energy about equal to lattice self-diffusion,
2. grain-size independence, with grain boundary shearing,
3. steady-state stress exponent of about one,
4. subgrain boundary formation,
5. dislocation density that appears independent of stress, and
6. primary creep stage.

These combined aspects distinguish the phenomenon from five-power-law creep, and low-stress exponent grain-boundary sliding (superplasticity), and Nabarro–Herring (diffusional) creep.

2. Theories for Harper–Dorn

The theories of Harper–Dorn creep will be briefly reviewed.

Harper and Dorn (Harper and Dorn, 1957; Harper et al., 1958)

As discussed by Langdon and Yavari (1982), Harper–Dorn was initially described by the motion of jogged screw dislocations, analogous to that described by Mott (1956) and Barrett and Nix (1965). This leads to a steady-state strain-rate

$$\dot{\gamma} = 12\pi\rho\ell_{js}b\left(\frac{D_{sd}Gb}{kT}\right)\left(\frac{\tau}{G}\right)^{1.0} \tag{2}$$

where ρ is the dislocation density, ℓ_{js} is the jog spacing in screw dislocations. Langdon (Langdon and Yavari, 1982) criticized this model on the basis that it requires unrealistically small jog spacings.

Friedel (1964)

Langdon also states that Friedel suggested that the so-called "Harper–Dorn" creep is actually diffusion creep (e.g., Nabarro–Herring) where vacancies diffuse between the relatively small subgrain boundaries:

$$\dot{\gamma} = A_{NH}\left(\frac{D_{sd}Gb}{kT}\right)\left(\frac{b}{\lambda}\right)^{2}\left(\frac{\tau}{G}\right)^{1.0} \tag{3}$$

where λ is the subgrain size. As Langdon and Yavari (1982) point out, subgrains are not always observed in the Harper–Dorn region. When subgrains do form, the size, according to Barrett et al. (1972) tends to be stress-dependent and this would increase the stress-dependence in Eqs. (1) and (3) beyond that of the observed value of 1.0.

Barrett et al. (1972)

Barrett et al. (1967) suggested that, as with many other diffusion-controlled creep processes, dislocation generation occurs with dislocation climb (at an assumed fixed number of sources). The rate of dislocation generation is

$$\dot{\rho}_+ = \rho_0 \frac{v_C}{x} \tag{4}$$

where ρ_0 is the fixed dislocation length per unit volume, v_C is the climb velocity, and x is the distance over which climb must occur to create glide dislocations. The climb velocity under a climb stress, σ, is (Hirth and Lothe, 1968),

$$v_C = \frac{Db^2\sigma}{kT} \tag{5}$$

Barrett et al. assume,

$$x = \frac{Gb}{\sigma} \tag{6}$$

which appears to assume Taylor hardening.

Dislocation annihilation is assumed to occur only at subgrain boundaries. (This is a complication as subgrain boundaries may not always exist.) The annihilation rate is

$$\dot{\rho}_- = \frac{\rho v_g}{\lambda} \tag{7}$$

where v_g is the glide velocity. The subgrain size phenomenologically varies with stress as

$$\lambda = \frac{\lambda_0}{\sigma}. \tag{8}$$

It was assumed,

$$v_g = v_0\sigma \tag{9}$$

(and a stress exponent of one for dislocation glide velocity, here, is a critical assumption.) At steady state,

$$\dot{\rho} = \dot{\rho}_+ - \dot{\rho}_- = 0 \tag{10}$$

Combining Eqs. (4)–(10) one find that the steady state dislocation density,

$$\rho = \frac{\rho_0 Db\lambda_0}{kTGv_0} \tag{11}$$

is independent of σ. Combining with the Orowan equation, the steady state creep rate is

$$\dot{\gamma}_{ss} = \rho b v_g = \frac{\rho_0 Db^2\sigma\lambda_0}{kTG} \tag{12}$$

Again, critical to this derivation is a stress exponent of 1 for v_g and the assumption (6) which appear tenuous.

Langdon and Yavari (1982)

Langdon et al. (1982) suggested that the rate-controlling processes for Harper–Dorn is based on the climb of jogged edge-dislocations under vacancy saturation conditions, and

$$\dot{\gamma} = \frac{6\pi\rho b^2}{\ell n(1/\rho^{1/2}b)}\left(\frac{D_{sd}Gb}{kT}\right)\left(\frac{\tau}{G}\right)^{1.2} \tag{13}$$

Wu and Sherby (1984)

The fact that both within the Harper–Dorn and the five-power-law regimes, the underlying mechanism of plastic flow appeared to be diffusion controlled, led Wu and Sherby to propose a unified relation that describes the creep behavior over both ranges. This model incorporates an internal stress which arises from the presence of random stationary dislocations present within subgrains. At any time during steady-state flow, they assume half of the dislocations moving under an applied stress are aided by the internal-stress field (the internal stress adds to the applied stress), whereas the motion of the other half is inhibited by the internal stress. The internal stress is calculated from the dislocation density by the dislocation hardening equation ($\tau = \alpha Gb\sqrt{\rho}$, where $\alpha \cong 0.5$). The unified equation is (Ruano et al., 1988)

$$\dot{\varepsilon}_{ss} = \frac{1}{2} A_{WS} \frac{D_{eff}}{b^2} \left\{ \left(\frac{\sigma + \sigma_i}{E}\right)^n + \frac{|\sigma - \sigma_i|}{(\sigma - \sigma_i)} \left|\frac{\sigma - \sigma_i}{E}\right|^n \right\} \tag{14}$$

where A_{WS} is a constant and σ_i is the internal stress. At high stresses, where $\sigma \gg \sigma_i$, σ_i is negligible compared to σ and Eq. (14) reduces to the (five-power-law) relation

$$\dot{\varepsilon}_{ss} = A_{10} \frac{D_{eff}}{b^2} \left(\frac{\sigma}{E}\right)^n \quad \text{(with } n = 4\text{–}7) \tag{15}$$

At low stresses, where $\sigma \ll \sigma_i$ (Harper–Dorn regime), Eq. (14) reduces to Eq. (1). A reasonable agreement has been suggested between the predictions from this model and experimental data (Wu and Sherby, 1984; Ruano et al., 1988) for pure aluminum, γ-Fe and β-Co. The internal stress model was criticized by Nabarro (1989), who claimed a unified approach to both five-power-law and Harper–Dorn creep is not possible since none of these processes are, in themselves, well understood and unexplained dimensionless constants were introduced in order to match theoretical predictions with experimental data. Also, the dislocation density in Harper–Dorn creep is constant, whereas it increases with the square of the stress in the power-law regime. Thus, the physical processes occurring in both regimes must be different (although Ardell (1997) attempts to rationalize this using network-creep models).

Nabarro (1989)

According to Nabarro (1989), an equilibrium concentration of dislocations is established during steady-state creep which exerts a stress on its neighbors equal to the Peierls stress. The mechanism of plastic flow would be the motion of these dislocations that is controlled by climb.

Wang (Wang, 1993, 1995, 1996; Wang et al., 2002)

The internal stress model of Sherby and Wu (Wu and Sherby, 1984) was also criticized by Wang (1993), who proposed that the transition between power-law creep and Harper–Dorn creep takes place instead at a stress, σ, equal to the "Peierls stress, σ_p", (Wang, 1995, 1996). Wang et al. (2002) suggest that the steady-state dislocation density is related to the Peierls stress. In equilibrium, the stress due to the mutual interaction of moving dislocations is in balance not only with the applied stress but also with lattice friction which fluctuates with an amplitude of the Peierls stress. As a result, the steady-state dislocation density ρ in dislocation creep can be written as:

$$b\rho^{1/2} = 1.3 \left[\left(\frac{\tau}{G}\right)^2 + \left(\frac{\tau_p}{G}\right)^2 \right]^{1/2} \tag{16}$$

where τ is the applied shear stress. When $\tau \gg \tau_p$, the dislocation density is proportional to the square of the applied stress, and five- (or three-) power-law creep is observed. This is consistent with the recent work of Kassner (2004). Conversely, when $\tau \ll \tau_p$, the dislocation density is independent of the applied stress and Harper–Dorn occurs.

Ardell (Ardell and Przystupa, 1984; Ardell and Lee, 1986; Ardell, 1997)

A different and fairly extensive approach to Harper–Dorn is based on the dislocation network theory by Ardell and coworkers (1984, 1986, 1997). The dislocation link length distribution contains no segments that are long enough to glide freely. That is, the longest links of length L_m are smaller than the critical link length to activate a (e.g., Frank-Read) dislocation source. Harper–Dorn is, therefore, a phenomenon in which all the plastic strain in the crystal is a consequence of dislocation network coarsening. The recovery of the dislocation density during Harper–Dorn creep is comparable to static recovery in the absence of an applied stress; climb of nodes is driven by line-tension of dislocation links. The stress-dependence of the $\dot{\varepsilon}_{ss}$ arises because the applied stress biases the collisions, since the lengths of all the links must increase as σ_{ss} increases, thereby increasing the collision possibilities. The climb velocity of the nodes is mostly affected by the resolved force arising from the line tensions of the dislocations at the nodes. Accidental collisions between these links can refine the network and stimulate further coarsening

$$\dot{\varepsilon} = \frac{\pi C b^3 D_{SD}}{2kT} \rho \sigma \tag{17}$$

where

$$C = \frac{\alpha}{\langle u \rangle^2} \int_0^{u_c} u^{2-m}(u-1)\Phi(u)\mathrm{d}u \tag{18}$$

and Φ_u is the scaled-link-length (u) distribution function, $\alpha \cong 0.5$ and m is a phenomenological exponent. The independence of ρ with σ is a consequence of the frustration of the dislocation network coarsening which arises because of the exhaustion of Burgers vectors that can satisfy Frank's rule at the nodes. Others have taken issue with this (Nes et al., 2002).

3. More recent developments

3.1. The effect of strain

Aluminum is clearly the most extensively studied material in the Harper–Dorn regime. Blum and Maier (1999) recently questioned the existence of Harper–Dorn creep, not having been able to observe the decrease of the stress exponent to a value of 1 when performing compression tests with changes in stress in pure aluminum (99.99% purity), in a low temperature-stress regime where Harper–Dorn had been observed by others. These results are illustrated, later, in Fig. 5. Nabarro (2000) responded to these reservations claiming that the lowest stress used by Blum et al. (Blum and Maier, 1999) (0.093 MPa) was still too high to observe Harper–Dorn. Blum et al. (2002) subsequently performed compression tests using even lower stresses (as low as 0.06 MPa), failing again to observe $n = 1$ stress exponents. Instead, exponents close to 5 were measured, indicating normal, five-power-law creep extending into the so-called Harper–Dorn regime. Basically, Blum (Blum
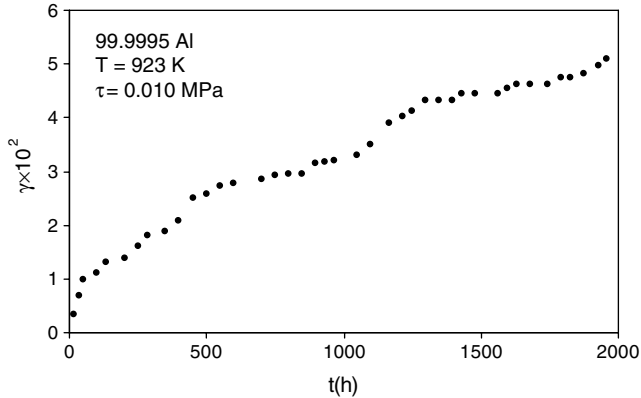
Fig. 5. Creep curve corresponding to very pure 99.9995 Al, from Mohamed (2002). The undulations are suggested to result from new restoration mechanisms.

et al., 2002) suggested that earlier Harper–Dorn studies did not accumulate sufficient strain (hardening) to achieve steady-state prior to tertiary creep. Strain-rates in the Harper–Dorn regime are so low that unusually long testing periods are required to achieve modest strains. In some Al cases, Blum (Blum et al., 2002) is correct that creep-rates reported by other investigators are too high based on a failure to achieve steady-state. However, in other cases where Harper–Dorn is suggested, steady-state appears to have been achieved.

3.2. The effect of impurities

Recently, Mohamed et al. (Ginter et al., 2001; Mohamed, 2002; Ginter and Mohamed, 2002) suggested that impurities may play an essential role in Harper–Dorn (H–D) creep. They performed relatively large strain (up to 10%) creep tests at stresses lower than 0.06 MPa in Al polycrystals of 99.99 and 99.9995 purity. They only observed H–D creep in the latter (most pure) metal. Accelerations in the creep curve corresponding to the high purity Al are apparent in Fig. 5. These accelerations are absent in the less pure 99.99 Al creep curve, at an identical temperature and stress.

Ginter and coworkers (2001, 2002) also reported that the microstructure of the 99.9995 Al includes wavy grain boundaries, an inhomogeneous dislocation density distribution as determined by etch pits, small new grains forming at the specimen surface and large dislocation density gradients across grain boundaries. Well-defined subgrains were not observed. However, the microstructure of the deformed 99.99 Al is formed by a well-defined array of subgrains. These observations led Ginter and coworkers (2001, 2002) to conclude that the restoration mechanism taking place during so-called "Harper–Dorn creep" includes discontinuous dynamic recrystallization (DRX) rather than the contended dynamic recovery. Nucleation of recrystallized grains would take place at the specimen surfaces and, due to the low amount of impurities, highly mobile boundaries would migrate towards the specimen interior. This restoration mechanism would give rise to the periodic accelerations observed in the creep curve, by which most of the strain is produced. Therefore, Ginter and coworkers (2001, 2002) believed that high purity leads to

dynamic recrystallization. It is difficult to accurately determine the stress exponent due to the appearance of periodic accelerations in the creep curves. However, Ginter et al. (2001) claimed that $n = 1$ exponents are only obtained if creep curves up to small strains (1–2%) are analyzed, as was done in the past. Mohamed et al. estimated stress exponents of about 2.5 at larger strains for high purity DRX specimens of Al.

The work by Mohamed et al. (2001, 2002) has received some criticism. Langdon (2002) argues that the jumps in the creep curves are not very clearly defined. Also, Mohamed (2001, 2002) claims that dynamic recrystallization occurs during creep of very high purity metals at regular strain increments, whereas the incremental strains corresponding to the accelerations tend to be relatively non-uniform. Grain growth might be a more appropriate restoration mechanism. It is certainly true that 99.999% pure Al has a greater propensity for (static) recrystallization (and presumably DRX and grain growth) than 99.99% pure Al, but Fig. 2 shows the so-called Harper–Dorn present in just 99.99 pure Al in other studies, in contradiction to the suggestion by Mohamed (2002).

Additional recent experiments were performed by McNee and co-workers (McNee et al., 2001) on, generally, 99.999 pure Al. These investigators generally, consistent with Blum, observed the extension of five-power-law creep into Harper–Dorn regime. A few tests, however, show relatively high strain-rates (creep-rates) and might be more consistent with the early experiments supportive of a low-stress exponent. Despite these new experiments, Langdon (2005) appears to believe that Harper–Dorn is, nonetheless, an independent mechanism.

3.3. Size effects

Raj and co-workers (Raj, 1985; Raman and Raj, 1985) suggested the surfaces are dislocation sources which leads to a size effect. Nes and co-workers (Nes et al., 2002) suggested that, under conditions typical of Harper–Dorn creep, the statistical slip-length may become comparable to or even exceed the specimen diameter (a size effect). That is, it was suggested that under Harper–Dorn conditions, the size influences the rates of generation and loss of dislocations. The rate of dislocation generation is reduced and the loss of dislocations is no longer controlled by dynamic recovery, but by static recovery. The result is that creep rate scales linearly with the applied stress. Size effects with specimen dimensions approaching the obstacles spacing was very recently demonstrated by Uchic et al. (2004), Fredriksson and Gudmundson (2005), and Cheong et al. (2005).

3.4. Recent experiments

Fig. 6 summarizes the data of the more recent experiments just described. The 99.999 pure Al data of Mohamed and co-workers (2001, 2002) show lower stress exponents of 1–2.5 due to the suggested additional restoration mechanisms [e.g., grain growth (GG) or DRX]. The data of Blum (1991,2002) and co-workers and Mohammed et al. (2001, 2002) on lower purity, 99.99% where the impurities presumably suppress DRX and GG, show normal, five-power-law behavior into the so-called Harper–Dorn regime. The polycrystalline data of McNee et al. (2001), which include 99.999% and 99.99% pure Al, do not appear to evince low stress exponent behavior nor does the Ginter et al. (2001) data, although material purity of the McNee et al. (2001) tests may have been sometimes compromised. Rather, some threshold behavior is observed, increasing the exponent
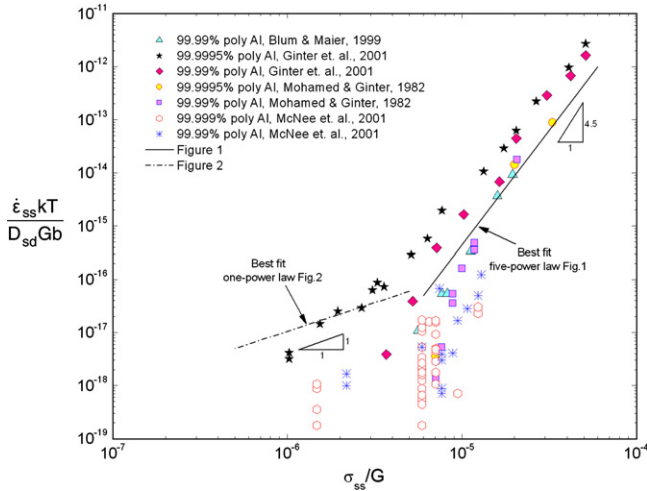
Fig. 6. The trends of Figs. 1 and 2 are plotted with straight lines with the additional, in particular, recent data (mostly 1999–2001) relevant to Harper–Dorn.

above 4.5. Purity did not appear to be a factor in the McNee et al. tests. The curious aspect of the more recent data of Fig. 6 is that the low stress Al data are more consistent with five-power-law behavior than "Harper–Dorn". In summary, the aluminum data are ambiguous as to whether Harper–Dorn is, in fact, a separate creep mechanism. One complication is that polycrystalline specimens were used in these more recent studies and average Taylor factors may be variable in the few coarse-grains (e.g., 5 mm) of the specimens. Typical changes in Taylor factors could rationalize the observed variations in creep rate from Harper–Dorn to five-power-law creep.

4. Other materials for which Harper–Dorn has been suggested

Table Ia lists materials for which Harper–Dorn has been suggested to occur, either by the (original and Table Ib) experiments or by subsequent reinterpretation of original data by subsequent investigators. However, the Harper–Dorn conclusions are ambiguous in several instances. As will be discussed subsequently, the Al work is particularly ambiguous.

4.1. α-Zr

Fig. 7 illustrates the steady-state creep behavior of α-Zr. Harper–Dorn was suggested for helical specimens of α-Ti, α-Fe, and α-Zr and β-Co Malakondaiah and Rama Rao, 1981, 1982; Fiala et al., 1983; Novotny et al., 1985; Prasad et al., 1992). The α-Zr data are particularly ambiguous as there is a grain-size dependence in the purported Harper–Dorn regime. For example, in earlier work by Hayes and Kassner (2002) it was shown that the creep rate of zirconium at low values of σ_{ss}/G varies approximately proportional to the applied stress. The rate controlling mechanism(s) for creep within this regime is unclear. A grain-size dependency may exist, particularly at small (<90 μm) sizes, suggesting a diffusional or perhaps a grain-boundary sliding mechanism. A grain-size independence

Table Ib
Ceramics (and minerals) for which Harper–Dorn creep has been suggested to operate with relevant references

Material	References	
CaO	Dixon-Stubbs and Wilshire (1982)[a]	
	Duong and Wolfenstine (1991)	
	Langdon (1983)	
UO_2	Ruano et al. (1991)[a]	
MgO	Cummerow (1963)[a]	Routbort (1979)[a]
	Hensler and Cullen (1968)[a]	Ruano et al. (1992)
	Langdon and Pask (1970)	Tremper (1971)
	Ramesh et al. (1986)	Wolfenstine and Kohlstedt (1988)
	Rothwell and Neiman (1965)[a]	
TiO_2	Wolfenstine (1990)	
$Mn_{0.5}Zn_{0.5}Fe_2O_4$	Wang (1994a)	
BeO	Wang (1994a)	
Al_2O_3	Wang (1994a)	
$Co_{0.5}Mg_{0.5}O$	Wang (1994c)	
NaCl	Banerdt and Sammis (1985)	Poirier (1972)[a]
	Blum and Ilschner (1967)[a]	Wolfenstine et al. (1991)[a]
	Burke (1968)[a]	
$MgCl_2 \cdot 6H_2O$	Wang et al. (1994b)	
$KZnF_3$	Poirier et al. (1983)	
$KTaO_3$	Beauchesne and Poirier (1990)	
$CaTiO_3$	Wang et al. (1994b)	
Ice	Wang (1994b)	
$CaCO_3$	Wang (1994a)	
SiO_2	Wang et al. (1994a)	
$(Mg,Fe)_2SiO_4$	Darot and Guerguen (1981)[a]	Relandeau (1981)[a]
	Justice et al. (1982)[a]	Schwenn and Goetze (1978)
	Kohlstedt and Goetze (1974)[a]	Wang (1994a,d)
$NaAlSi_3O_8–CaAl_2Si_2O_8$	Wang and Toriumi (1994)	

[a] For basic creep papers making no reference to Harper–Dorn creep.

at larger grain sizes supports, by itself, Harper–Dorn, but the low observed activation energy ($\cong$90 kJ/mol) is not consistent with those observed at similar temperatures at higher stresses in the five-power-law regime (270 kJ/mol) where creep is also believed to be lattice self-diffusion controlled. The stress dependence in this regime is not consistent with traditional grain-boundary sliding mechanisms.

Langdon (2006) suggests that none of the studies have plotted the values of stress–strain behavior of the helical samples at the higher stresses and, hence, it is difficult to have a direct comparison of the results in the case of the helical and the flat samples. Helical samples are of special interest at the lower stress levels and it is mainly due to the fact that various strain rate data can be achieved using the same sample and loading condition,
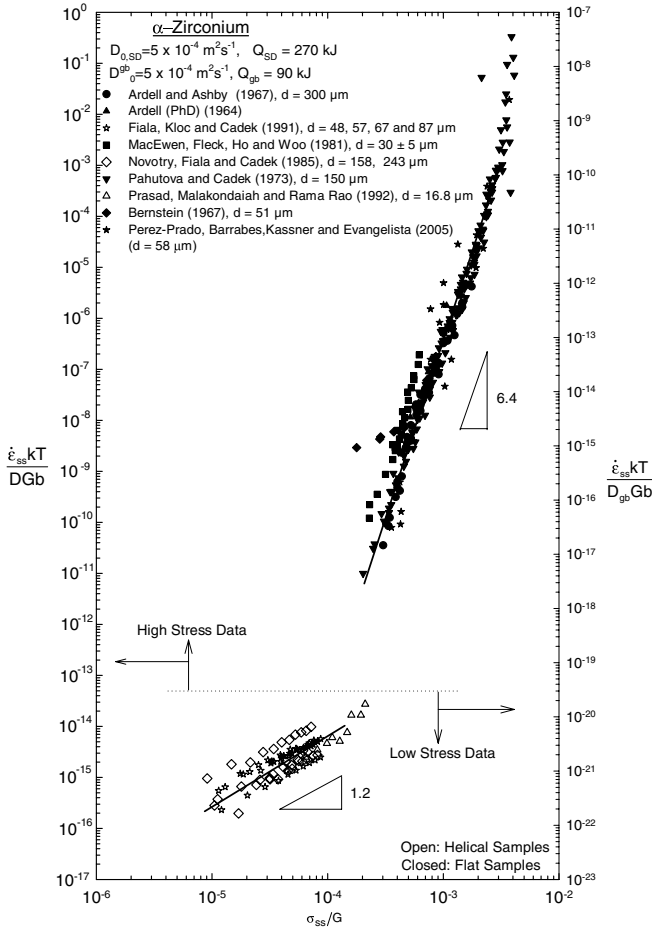
Fig. 7. D_{gb} is used for low stress data whereas D_{sd} is used for high stress data.

at once. Hence, spring samples save time, especially as the probable strain rates are very low at these low stress levels.

Furthermore, Langdon suggests that the main problem with the helical samples arises due to the fact that the strains achieved in these tests are very low (of the order of 10^{-4}–10^{-5}) and, hence, the data do not give information on steady-state creep. The low value of strains may only account for the elastic and anelastic strains, instead of plastic strains. Helical samples may not go for higher strains (roughly less than $<10^{-3}$) due to the instability in the sample (this might be the reason that it has not been used for higher stresses) and, hence, these strains are always insufficient for non-ambiguous steady state data.

4.2. α-Fe

Fig. 8 illustrates the steady-state creep behavior of α-Fe. Low stress, low stress-exponent observation were interpreted (as with α-Zr) to be due to both Harper–Dorn and Coble creep; the latter operating below 123 μm grain sizes according to Fiala et al.
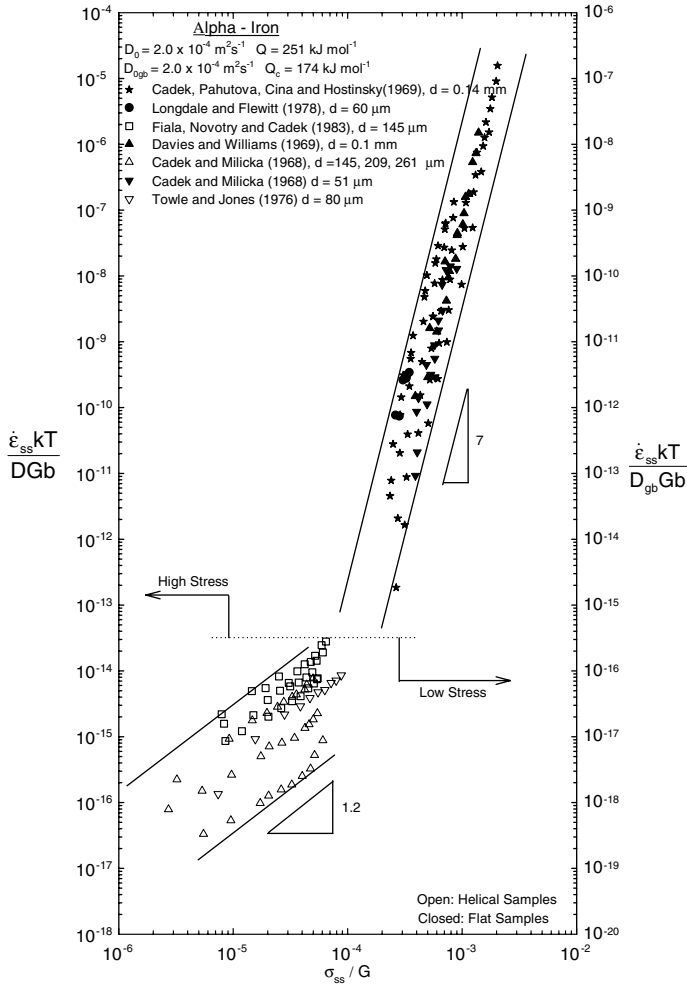
Fig. 8. The data from Cadek and Milicka (1968) at a grain size of 216 μm were interpreted as Harper–Dorn creep by Fiala et al. (1991). The steady-state α-Fe data appear best normalized with D_{gb} at low stress and D_{sd} at high stress to best fit the data.

(1983). However, an examination of all of the data does not reveal a systematic grain size trend. Lack of sufficient number of studies showing Harper–Dorn regime and very low levels of strains (see α-Zr for helical specimen comments) achieved in the studies proposing the Harper–Dorn creep are the reasons for ambiguity in the conclusion about the behavior of α-iron at the lower stress values (Langdon, 2006).

4.3. NaCl

Banerdt and Sammis (1985) suggested Harper–Dorn in NaCl, although NaCl is prone to discontinuous dynamic recrystallization and total strain-levels were rather low, 10^{-3} or less, and required temperature variation corrections were substantial at the lower stresses.

NaCl shows five(3.5)-power-law behavior and one study appears to have shown low stress exponent behavior. Bandert and Sammis, however, appear to have low total strains (<1%) and steady-states are not clearly established.

4.4. Cu

Fig. 3 shows classic five-power-law behavior in copper. Only recently has there been tests at low stress (Srivastava et al., 2005). Shrivastava et al. found a low stress exponent (~2). The Cu was polycrystalline and some grain boundary sliding was reported. The author concluded, however, that the interior dislocations cause slip, although Harper–Dorn is not concluded.

4.5. CaO

CaO is confusing as two studies have been performed within the low strain regime; Dixon-Stubbs and Wilshire (1982) find five-power-law transitions to 1.6 power in single crystals that Langdon (1983) later suggested was Harper–Dorn. Duong and Wolfenstine (1991), however, observe five-power-law also in single crystals over the same stress range that Dixon-Stubbs and Wilshire (1982) observed low stress exponents.

4.6. MgO

MgO appears to have contradictory data. Five (actually 3.2)-power data of Routbort (1979) is of higher strain-rates than other studies, while others appear unusually low (Tremper, 1971). Ramesh et al. (1986) observed a low stress-exponent at lower stresses in single crystals.

4.7. Forsterite (Mg_2SiO_4)

Forsterite shows 3–4 stress exponent behavior in single crystals of different orientation. The low stress tests by Relandeau (1981) on relatively fine-grained material show low stress-exponent behavior.

4.8. $MgCl_2 \cdot 6H_2O(CO_{0.5}Mg_{0.5})O$ and $CaTiO_3$

The Harper–Dorn creep behavior of these were recently questioned by Berbon and Langdon (1996). Discontinuous dynamic recrystallization was suggested in the former, while it was suggested that a transition in mechanism is not evident for the latter on a double logarithmic plot.

5. Summary

Harper–Dorn has been suggested to occur at stresses below classic five-power-law creep. This has been suggested for several metals, minerals, and ceramics. However, the case for Harper–Dorn requires some additional experiments. The steady-states are not always unambiguously determined due to total strains decreasing with decreasing stress and creep-rate, the limitation of testing time and creep rate enhancement by tertiary creep.

Careful microstructural characterization is absent that can quantify dislocation spacings and exclude strain effects from grain-growth, recrystallization, grain boundary sliding, fracture and size effects.

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