

# Creep at Very Low Rates

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The creep rate in a land-based power station must be less than  $10^{-11} \text{ s}^{-1}$ . At these low rates of deformation the transport of matter occurs by the migration of vacancies rather than by the glide of dislocations. A quantitative understanding of these diffusional processes is, therefore, important. First type of diffusional creep (Nabarro–Herring (N-H)): the sources and sinks of vacancies are grain boundaries. The vacancies may diffuse through the bulk of the grain or along the grain boundaries (Coble (C)). Second type (Harper–Dorn (H-D)): the vacancies diffuse from edge dislocations with their Burgers vectors parallel to the major tensile axis to those with Burgers vectors perpendicular to this axis. The coherence of the polycrystalline aggregate is maintained by sliding along the grain boundaries. The three mechanisms of vacancy migration, grain boundary sliding, and dislocation glide may all interact. The theories of N-H and C creep in pure metals are established and confirmed, but H-D creep and grain boundary sliding are less well understood.

Practical engineering materials are usually strengthened by precipitates that accumulate on grain boundaries and slow down creep in complicated ways.

## I. INTRODUCTION

A LAND-BASED turbine may have a life of 40 years. If a total creep strain of 1 pct is permissible, the creep rate is about  $10^{-11} \text{ s}^{-1}$ . Extrapolation to this rate from rates measurable in the laboratory is perilous, and we need to understand the physical processes involved. These processes are usually classified as diffusional creep, Harper–Dorn (H-D) creep, and grain boundary sliding.

In diffusional creep, the concentration of vacancies close to a grain boundary normal to an applied tensile stress is greater than that in the unstressed material in equilibrium. The excess vacancies diffuse to the grain boundaries parallel to the tensile axis, where they are absorbed. This leads to a plastic deformation in which the applied stress does work. At high temperatures, the vacancies diffuse through the bulk of the grain, leading to Nabarro–Herring (N-H) creep, according to the analyses of Nabarro<sup>[1]</sup> and Herring.<sup>[2]</sup> At lower temperatures, diffusion is mainly along the grain boundaries, Coble (C) creep, as analyzed by Coble.<sup>[3]</sup>

In H-D creep, vacancies diffuse from the edges of dislocations with Burgers vectors parallel to the tensile axis to the edges of dislocations with Burgers vectors normal to the tensile axis. There is again a distinction between the high temperature process in which diffusion occurs in the bulk (H-D creep), and the process at lower temperatures in which diffusion occurs along the channels of the dislocation network which act as pipes for diffusion (H-DP).

In a polycrystal, the grain boundaries are regions of weak cohesion, and deformation may occur by the sliding of one grain over another (grain boundary sliding). Since the grain

boundaries are not smooth and intersect at the lines where three grains meet, this sliding can produce strains greater than the elastic strain only if there is accommodation by the diffusion of vacancies. Again, this diffusion may occur through the bulk of the adjacent grains or along the grain boundary. The theory of grain boundary sliding is less well developed than those of N-H, C, and H-D creep.

A polycrystal cannot remain coherent if any one of these processes occurs alone. When two processes occur simultaneously, one or the other may determine the rate of the combined process, but the rate-determining process will not necessarily provide the major component of the strain. (For example, in the model of Spingarn and Nix<sup>[4]</sup> for plastic deformation at higher rates, the overall rate is determined by the accommodation processes in the grain boundaries, but the strain is produced essentially by dislocation glide within the grains.)

## II. DIFFUSIONAL CREEP

Early observations, such as those reported in the book by Burton,<sup>[5]</sup> confirmed the theories of N-H and C: creep–strain rate proportional to stress, inversely as square of grain size (N-H) or as cube of grain size (C); appropriate temperature dependence (N-H, C); and correct order of magnitude (N-H, C). Above all, in precipitate-hardened alloys, especially Mg hardened by  $\text{ZrH}_2$ , zones denuded of precipitate were found on grain boundaries normal to a tensile stress and zones enriched in precipitate on boundaries parallel to the stress.

A precise numerical theory turned out to be more complicated than had been expected. The stress was not uniform in the steady state, the load being partly transferred from regions where the diffusion path was short to those where it was long.<sup>[6,7]</sup> The rate depended on the assumed shape of the grains<sup>[8]</sup> and on the distribution of grain sizes.<sup>[9]</sup> These considerations all introduce modest numerical factors into the estimated creep rate. More important was the physical consideration that, in a precipitate-hardened alloy, the accumulation of precipitate particles on grain boundaries parallel to a tensile axis could greatly reduce the rate of creep. Two mechanisms are involved.<sup>[10]</sup> First, as will be discussed

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This article is based on a presentation made in the workshop entitled “Mechanisms of Elevated Temperature Plasticity and Fracture,” which was held June 27–29, 2001, in Dan Diego, CA, concurrent with the 2001 Joint Applied Mechanics and Materials Summer Conference. The workshop was sponsored by Basic Energy Sciences of the United States Department of Energy.

subsequently, diffusional creep is necessarily coupled with grain boundary sliding, and according to Reference 11, the presence of precipitate particles in a grain boundary inhibits sliding. Second, according to References 12 and 13, there is much evidence that the interface between a matrix and a hard precipitate is a much less effective source and sink of vacancies than a boundary between two grains of the matrix. The physical reason for this is clear. Consider a tilt boundary in the matrix. This is a configuration of low energy if it consists of edge dislocations belonging to the two grains. It can absorb vacancies and remain a configuration of low energy if the two sets of dislocations each climb into their own grains. However, if the interface lies between the matrix and a high melting precipitate in which the vacancy mobility is low, the climb of dislocations in the matrix alone destroys the low energy structure of the interface.

More recently, the evidence in favor of the diffusional creep process has come under attack. Many of the objections<sup>[14,15,16]</sup> are summarized in the report of the conference on *Creep Behaviour of Advanced Materials for the 21st Century*. The model failed because the observed creep rates were often higher than those predicted, though sometimes lower, and the creep rate increased more than linearly with the stress. The evidence from denuded zones was unreliable in that they were observed when the creep rate increased more than linearly with the stress, they could appear on one side or the other of a grain boundary, or surrounding it, and they could appear on boundaries not closely perpendicular to the tensile axis. They could be found in samples that had not been exposed to creep. The objections fall into two broad classes, the geometrical and microstructural, and the kinetic.

The geometrical objections fail to consider the work of Bilde-Sørensen and Smith.<sup>[17,18]</sup> The emission and absorption of vacancies in a grain boundary can occur only at grain boundary dislocations. These grain boundary dislocations are confined to the grain boundary. Their motion within the boundary requires both climb and glide. These two motions contribute differently, sometimes even with opposite signs, to the component of strain that accommodates the applied stress. This analysis explains all of the observations. Diffusional creep provides an obvious explanation of the appearance of zones denuded of or enriched in precipitate particles. Alternative models meet insuperable difficulties. If it is assumed that the precipitate particles have been swept by a migrating grain boundary, they should be accommodated on one side of the denuded zone.<sup>[19]</sup> This is not observed, although the slight enrichment of particles on *both* sides of a denuded zone shown, *e.g.*, in Figure 5 of Reference 20, remains unexplained. It has been suggested<sup>[15]</sup> that in the case of Mg hardened by precipitates of ZrH<sub>2</sub>, the precipitates have gone into solution as the migrating grain boundary passed over them. This suggestion is directly contradicted by the article<sup>[20]</sup> that presented the first observation of denuded zones: "Electron microprobe analysis . . . has given firm confirmation that the zones adjacent to grain boundaries, showing no precipitate metallographically, are in fact denuded of zirconium." This observation was confirmed by Karim *et al.*<sup>[21]</sup>

The objection that denuded zones can be observed before the creep test has little strength. The principal axes of fabrication and of testing are usually parallel, and it is not surprising that denuded zones normal to this axis are formed during

fabrication. However, the zones shown in Figure 1(a) of Reference 22 at a magnification of 308 times before the creep test are far less prominent than those shown in Figure 1(d) at a magnification of 160 times after the creep test. There is anecdotal evidence that wide denuded zones are observed after creep only in those samples that showed narrow zones before creep. If this is true, it would support Wilshire's view that the wide zones are formed by dislocation glide in the initially soft regions, but this model is not easily reconciled with the evidence (*e.g.*, Reference 23) of the deposition of material on grain boundaries during the creep of a solid solution. Moreover, in any tensile test, there are necessarily parts of the sample that are in axial compression. Further anecdotal evidence states that the denuded zones in these regions are perpendicular to those in the tensile regions, although these regions would have been subjected to the same stresses during the forming operation.

A complete study of the deformation process at a grain boundary requires five measurements: two direction cosines of the grain boundary normal and the three components of displacement across the boundary. One of the first two is obtained by examining an external surface parallel to the tensile axis; two of the last three are obtained by examining the offsets of a grid placed on this surface. The third is measured by the relative displacement of the grains normal to the surface. The final measurement must be made by examining a second surface parallel to the tensile axis. Since diffusional creep is necessarily accompanied by other mechanisms of deformation, it is expected that the denuded zones will represent only a substantial component of the total creep extension. In Reference 22, under conditions of diffusional creep, this fraction was about 0.5, in Reference 23, the contribution of grain boundary sliding in an individual boundary was of order unity but negative, in Reference 21, diffusional creep predominated at low strains and strain rates, and, in Reference 51, grain boundary sliding accounted for 15 pct of the total strain. Clearly, there is no detailed understanding, but diffusional (N-H or C) creep represents a substantial part of the total strain. Further evidence of the addition and depletion of material at grain boundaries in C creep has been provided by McNee *et al.*<sup>[26,27]</sup> A survey<sup>[60]</sup> of superplasticity in Region 1, which appears to be controlled by diffusional creep, showed that grain boundary sliding contributed between 10 and 60 pct of the total strain.

The kinetics of creep in the diffusional regime are less clear. The original theory assumed that all grain boundaries would be ready sources and sinks of vacancies. It is now clear that this is not the case. Twin boundaries and boundaries with a dense coincidence-site lattice (CSL) are not effective sources and sinks. A boundary that is close to a CSL may require a finite excess concentration of vacancies before it will absorb them.<sup>[24,25]</sup> The closer the orientation to that giving a CSL, the higher this excess. This has two effects. It slows the rate of creep. Because more grain boundaries become effective as the applied stress increases, the strain rate increases more than linearly with the stress.<sup>[23]</sup> A similar effect arises when precipitates in the grain boundaries introduce threshold stresses. The importance of these considerations is enhanced by the observation that grains tend to form clusters with a preponderance of near-CSL boundaries, surrounded by random boundaries.<sup>[18]</sup> It has been pointed out that the apparently clear evidence<sup>[28]</sup> for a transition

from N-H creep in polycrystalline copper at a rate inversely proportional to the square of the grain size at 0.82 of the melting temperature,  $T_m$ , to C creep at a rate inversely proportional to the cube of the grain size at 0.61  $T_m$  depends crucially on the two points at the lowest stresses. These points are subject to the experimental problems of very slow creep rates and probably grain growth. If these two points are disregarded, the slopes of the plots become almost identical.

There remains a gross discrepancy in the experimental observations of what appears to be N-H creep in copper. Burton and Greenwood<sup>[28]</sup> found a creep rate 4.5 times faster than that predicted for N-H creep in copper with a grain size of  $3.5 \cdot 10^{-5}$  m at 1093 K in the stress range 0.1 to 7 MN m<sup>-2</sup>. Pines and Sirenko<sup>[29]</sup> used a stress of  $9.76 \cdot 10^{-5}$  G  $\approx 4.7$  MN m<sup>-2</sup>, in the same range, and studied a range of grain sizes, of which the smallest was just greater than  $10^5 b \approx 2.6 \cdot 10^{-5}$  m, roughly the same as that observed by Burton and Greenwood. At a temperature of 1313 K, they found a creep rate that<sup>[30]</sup> is several hundred times faster than that calculated for N-H creep. Since these observations were made at a higher temperature than those of Burton and Greenwood, a transition to C creep is not possible. As will be seen subsequently, H-D creep, which is typically a thousand times faster than N-H creep, occurs only in materials with a low dislocation density, it is, therefore, possible that the samples used by Burton and Greenwood, which were annealed for 0.5 hour at about 1200 K, had too high a dislocation density to show H-D creep, while those of Pines and Sirenko, tested at 1313 K, had annealed to the necessary low dislocation density.

Diffusional creep results in an increase in the area of the grain boundary. At very small grain sizes, the corresponding increase in surface energy leads to an important reduction in the strain rate.<sup>[31]</sup>

### III. H-D CREEP

Understanding of H-D creep is much less satisfactory than that of N-H or C creep. Sometimes the observations of different observers show quite remarkable agreement, as in Figures 1 and 9 of Reference 31, but sometimes the majority of observers agree, while others find a much lower strain rate, as in Figure 2 of Reference 33. Many observers (*e.g.*, References 34 and 35) find no trace of the phenomenon. The theory<sup>[36,37]</sup> has great predictive power, as in Figure 2 of Reference 38 and Figure 1 of Reference 39, yet, it is fundamentally flawed.

The basic mechanism of H-D creep is the migration of vacancies from the edges of dislocations with their Burgers vectors parallel to a tensile axis to those with Burgers vectors perpendicular to this axis. This leads to a strain rate proportional to the stress, independent of grain size, and proportional to the dislocation density. The process occurs within the grain, and there is no accumulation or depletion of matter at the grain boundaries. Unlike other creep processes, H-D creep is characterized by a dislocation density that is independent of the applied stress. (The term H-D creep is sometimes extended to include any creep process in which the creep strain is produced predominantly by dislocation climb, but here, we restrict it to those processes in which the dislocation density is essentially independent of the applied stress.) The essence of the theory is that this density,  $\rho_{H-D}$ , is such that the external stress,  $\alpha \mathbf{b} \mu \rho_{H-D}^{1/2}$ , with  $\alpha$  of order one-third,

$\mathbf{b}$  the Burgers vector, and  $\mu$  the shear modulus, is of the order of the Peierls stress,  $\sigma_p$ . The flaw in the theory is that H-D creep is observed at high temperatures where, in most metals, double-kink formation should be easy, and the Peierls stress should be suppressed.

In discussions at the workshop, a different model of H-D and H-DP creep was developed. While the details are not yet clear, the essence is that these are not truly steady-state processes, but transient processes of very long duration. The model postulates that when the load is applied to a sample with a low dislocation density, a relatively high density of dislocations is produced almost instantaneously by glide, giving a creep transient. There has been no time for recovery, which is controlled by climb, to take place. Then, in accordance with experimental evidence, the dislocation density decreases logarithmically with time during annealing under load. This recovery process is controlled by the climb of dislocations, together with glide over short distances. It does not in itself lead to plastic strain. Because dislocation segments no longer lie in one glide plane, they cannot multiply rapidly by glide, even though the applied stress exceeds the flow stress corresponding to the current dislocation density. They can move only by the activation of short segments or by dragging jogs. However, creep occurs because the applied stress biases the climb and short-distance glide processes, which would otherwise be isotropic.

This model removes the problem that the Peierls stress should not be relevant at the temperatures at which the experiments are carried out. However, it is not yet able to determine the numerical value of the quasiequilibrium dislocation density. Still less is it able to show why this equilibrium density in a given material appears to be closely proportional to the square of the estimated Peierls stress in that material. One might perhaps build a link by saying that, in the present model, the stress that a dislocation exerts on its neighbor in the quasiequilibrium state is  $b^{-3}$  times the energy of a dislocation jog, while in the model of References 36 through 38, it is  $\mathbf{b}^{-3}$  times the energy of a kink, and that in the hard materials on which much of the evidence of References 37 and 38 depends, these energies may be similar.

Figure 2 of Reference 35 shows a rather sharp dichotomy between those observations that display H-D creep at low stresses and those which display much lower creep rates. The explanation may lie in the observation<sup>[40]</sup> that H-D creep is observed in samples of low initial dislocation density but not in those of high initial density. This is in accordance with the theoretical model. In any experiment, three stresses are involved. These are the Peierls stress,  $\sigma_p$ , the internal stress,  $\alpha \mathbf{b} \mu \rho^{1/2}$ , corresponding to the actual dislocation density,  $\rho$ , and the applied stress,  $\sigma_a$ . The dislocations can move by glide, with the operation of Frank–Read sources, or by climb, with the operation of Bardeen–Herring sources. Consider first the case of a low dislocation density,  $\rho < \rho_{H-D}$ . When  $\sigma_a < \sigma_p$ , extensive glide cannot occur, but climb occurs. Dislocations multiply by the operation of Bardeen–Herring sources until  $\rho$  reaches  $\rho_{H-D}$ , when the rate of multiplication is balanced by the rate of recombination, and H-D creep is observed. If  $\sigma_a > \sigma_p$ , rapid glide and multiplication of dislocations occur, and the creep rate is more than linear in the applied stress. This transition is independent of grain size.<sup>[41]</sup> Now consider the case in which  $\rho > \rho_{H-D}$ . Unlike the ordered array of edge dislocations observed in H-D creep,

these dislocations have a random array that forms an effective forest opposing the operation of Bardeen–Herring or Frank–Read sources. Deformation is very slow until  $\sigma_a$  exceeds  $ab\mu\rho^{1/2}$ , when power-law creep sets in. (Some annealing occurs during a prolonged creep test. Thus, the prestrained sample of Figure 3 of Reference 42, which was annealed for 36 hours at 923 K, began to creep after testing for about 25 hours at the same temperature and crept at a steady rate after 80 hours.). Numerically, Reference 32 estimated the equilibrium dislocation density,  $\rho_{H-D}$ , in Al 5 pct Mg at  $4.7 \cdot 10^3 \text{ m}^{-2}$ , while References 40 and 43 found that H-D creep in Al was suppressed at densities,  $\rho$ , above  $10^3$  to  $3 \cdot 10^4 \text{ cm}^{-2}$ . In high-purity Fe,<sup>[44]</sup> the transition from pipe-diffusion (H-DP) creep to power-law creep occurs at a stress that increases slowly with decreasing temperature and is quite independent of grain size, as the theory would predict. The transition from C to power-law creep occurs at stresses that are not well-defined functions of temperature, as is to be expected when a wide range of grain sizes (82 to 478  $\mu\text{m}$ ) is observed. While a coherent explanation of the published observations can be given along the lines suggested here, it is disturbing that a number of experimenters have failed to observe H-D creep in aluminium samples prepared in such a way that they would be expected to have dislocation densities at which other observers have made very concordant measurements.

#### IV. GRAIN BOUNDARY SLIDING

Lifshitz<sup>[6]</sup> was the first to point out that, in a continuum model of diffusional creep, grain boundary sliding would necessarily occur. If the centroids of two adjacent grains move toward or away from each other, a grain spanning both will necessarily slide over them. The contribution of this sliding to the total strain would be of the same order as the contribution of the atomic diffusion. Langdon<sup>[45]</sup> showed that this effect would be numerically largest on boundaries parallel or perpendicular to a tensile axis. The considerations of References 17 and 18 show that this will be true only in a statistical sense. This effect would be absent in H-D creep. Wang<sup>[46]</sup> has argued that, when allowance is made for the movement of grains in three dimensions, grain boundary sliding on the model of Ashby and Verrall<sup>[47]</sup> can contribute strains an order of magnitude larger than those directly produced by the diffusion of matter. Agreement with experiment is good, but the argument fails to recognize that even if all grains were initially in the orientation of Figure 2(a) of Reference 46, and finally in the orientation of Figure 2(e), the total strain would be finite, and the process would then come to an end. The model is inconsistent with the experimental observation that the denuded zones represent a major part of the total strain.

This accommodation creep is entirely different from primary grain boundary sliding. The theory of primary grain boundary sliding is not clear. It is probably best expounded in References 48 and 49. The strain rate varies as the square of the stress and inversely as the square of the grain size. Thus, at low stresses, processes linear in the stress will predominate over primary grain boundary sliding. This does not mean that accommodation grain boundary sliding is unimportant at low stresses. Raj and Ashby<sup>[50]</sup> analyzed this process in detail. Whereas accommodation sliding is most

prominent on boundaries parallel or perpendicular to a tensile axis, primary sliding is concentrated on boundaries at 45 deg to this axis, where the shear stress is greatest. The observations of Reference 51 at stresses just at and well above the transition from H-D to power-law creep both show the greatest sliding on boundaries near 45 deg to the tensile axis. The observations of Reference 52 on the tensile deformation of a superplastic Mg alloy show that the grain boundary ridges formed on the lateral surfaces are strongest on those boundaries at about 90 deg to the tensile axis at low rates of strain where accommodation sliding is expected, and strongest on those boundaries at about 45 deg to the tensile axis at high rates of strain where primary sliding is expected.

There seems to be a serious discrepancy in the estimates of the absolute rate of primary grain boundary sliding. Ball and Hutchison<sup>[48]</sup> show that the observations are fitted by the formula

$$\frac{\varepsilon kT}{D_b \mu b} = 100 \left(\frac{b}{L}\right)^2 \left(\frac{\sigma}{\mu}\right)^2 \quad [1]$$

where  $\varepsilon$  is the strain rate,  $k$  is Boltzmann's constant,  $T$  is the temperature,  $D_b$  is the grain boundary diffusion coefficient,  $\mu$  is the shear modulus,  $L$  is the grain size, and  $\sigma$  is the applied stress. Mukherjee<sup>[49]</sup> obtained a coefficient of 2 instead of 100, also by pure theory. This discrepancy could be resolved by geometrical considerations. However, Lüthy *et al.*,<sup>[53]</sup> similar to Reference 30, use semiempirical formulas in which  $\varepsilon$  is proportional to  $D_b/L^3$  or to  $D_L/L^2$ , where  $D_L$  is the lattice self-diffusion coefficient. These formulae cannot be reconciled with Equation 1.

It would be instructive to have molecular dynamics simulations of grain boundary sliding. Unfortunately, it is possible to perform these only for very small grain sizes at high stresses and extremely high rates of deformation. For example, the computations of Schiøtz *et al.*<sup>[54]</sup> consider Cu polycrystals with grain sizes in the range 3 to 13 nm deformed at 300 K under stresses of order 2 GPa at strain rates of order  $5 \cdot 10^8 \text{ s}^{-1}$ .

#### V. SELECTION OF THE DOMINANT CREEP MECHANISM

The transition between N-H and C creep occurs when

$$D_b b \approx D_L L \quad [2]$$

The corresponding transition between H-D creep by lattice diffusion and H-DP creep by pipe diffusion depends on the pipe-diffusion coefficient,  $D_p$ . The length scale of the equilibrium dislocation array is  $\rho_{H-D}^{1/2}$ , where

$$\rho_{H-D} \approx \frac{10 \sigma_p^2}{b^2 \mu^2} \quad [3]$$

If we consider two neighboring unit lengths of edge dislocations with perpendicular Burgers vectors and suppose that the vacancy concentrations per unit volume in their neighborhoods differ by  $\Delta c$ , then the number of vacancies transferred per unit time per unit length of dislocation by lattice diffusion is of order  $D_L \Delta c$ . The current of vacancies along the dislocation pipes is of order  $D_p b^2 \Delta c \rho^{1/2}$ , and the rate of accumulation or depletion per unit length of dislocation is of order

$D_p \mathbf{b}^2 \Delta c \rho$ . The transition from high-temperature H-D creep to medium-temperature H-DP creep thus occurs<sup>[55]</sup> when

$$D_p \mathbf{b}^2 \approx D_L / \rho_{H-D} \quad [4]$$

The observations of References 44 and 56 on pure iron are consistent<sup>[55]</sup> with the process of H-DP creep.

There are two constraints on the operation of H-D creep.<sup>[57]</sup> First, the mechanism assumes the presence of a steady-state array of dislocations with a mean spacing of  $\rho_{H-D}^{-1/2}$ . This is only possible if the grain size,  $L$ , is appreciably greater than  $\rho_{H-D}^{-1/2}$ , leading to the necessary condition:

$$L/\mathbf{b} > \mu/2\pi\sigma_p \quad [5]$$

Even if this condition is satisfied, H-D creep may not be faster than diffusional creep. The condition that H-D creep will be faster than N-H creep is<sup>[58,59]</sup>

$$L/\mathbf{b} > 7 \mu/\sigma_p \quad [6]$$

The second requirement is that Bardeen–Herring sources should be able to operate within the grain in order to multiply dislocations at the same rate as they are destroyed by pairwise annihilation. The condition for this is

$$L/\mathbf{b} > 2\mu/\sigma \quad [7]$$

where  $\sigma$  is the applied stress, which in H-D creep is necessarily less than  $\sigma_p$ .

All of these considerations assume that dislocations in one grain are unable to trigger dislocation motion in a neighboring grain. This is governed by the Hall–Petch criterion, and numerical considerations<sup>[57]</sup> show that the critical value of  $L^{1/2} \sigma$  required for slip transmission to occur is usually well above those encountered at low strain rates.

## VI. CONCLUSIONS

The existence and theoretical interpretation of diffusional creep in pure metals are well established, although there are many details still to be resolved. However, power stations are not made from pure metals, and the understanding of diffusional creep in materials hardened by precipitates or by transformation strains is very incomplete.

It is certain that creep can occur by the climb of dislocations within the grains. There is strong evidence both for and against the existence of a regime of H-D creep in the strict sense, which requires a dislocation density independent of the applied stress and of the previous treatment of the sample. It is difficult to observe in pure soft metals because it leads to very low rates of creep. It is suppressed if the initial dislocations density is considerably higher than the equilibrium density.

It is important to distinguish between primary grain boundary sliding that is accommodated by diffusional processes and secondary grain boundary sliding that accommodates diffusional processes. While there is a widely accepted theory of primary grain boundary sliding, use is sometimes made of empirically based formulae that are incompatible with this theory.

The present state of knowledge reveals specific questions that call for experimental investigation. Theory will contribute, but atomic computation, with a time scale of  $10^{-11}$  s, will not handle processes that take  $10^{11}$  s. Computation on the dislocation scale may perhaps become helpful in a few years time.

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