UCRL-JC-132348
PREPRINT

Deformation by Grain Boundary Sliding and Slip Creep Versus Diffusional Creep

J. Wadsworth
O. A. Ruano
O. D. Sherby

This paper was prepared for submittal to the
The Minerals, Metals and Materials (TMS) 1999 Annual Meeting
San Diego, CA
February 28-March 4, 1999

November 4, 1998

Lawrence
Livermore
National
Laboratory

DEFORMATION BY GRAIN BOUNDARY SLIDING AND SLIP CREEP VERSUS DIFFUSIONAL CREEP

Jeffrey Wadsworth*, Oscar A. Ruano**, and Oleg D. Sherby***

*Lawrence Livermore National Laboratory
P.O. Box 808, L-001
Livermore, CA 94550, USA

**Department of Physical Metallurgy, CENIM, CSIC
Avenida de Gregorio del Amo, 8
28040, Madrid, Spain

***Department of Materials Science and Engineering
Stanford University
Stanford, CA 94305, USA

Abstract

A review is presented of the debates between the present authors and other investigators regarding the possible role of diffusional creep in the plastic flow of polycrystalline metals at low stresses. These debates are recorded in eleven papers over the past seventeen years. In these papers it has been shown that the creep rates of materials in the so-called "diffusional creep region" are almost always higher than those predicted by the diffusional creep theory. Additionally, the predictions of grain size effects and stress exponents from diffusional creep theory are often not found in the experimental data. Finally, denuded zones have been universally considered to be direct evidence for diffusional creep; but, those reported in the literature are shown to be found only under conditions where a high stress exponent is observed. Also, the locations of the denuded zones do not match those predicted. Alternative mechanisms are described in which diffusion-controlled dislocation creep and/or grain boundary sliding are the dominant deformation processes in low-stress creep. It is proposed that denuded zones are formed by stress-directed grain boundary migration with the precipitates dissolving in the moving grain boundaries. The above observations have led us to the conclusion that grain boundary sliding and slip creep are in fact the principal mechanisms for observations of plastic flow in the so-called "diffusional creep regions."

Introduction

Descriptions of how materials can plastically deform by the stress-directed lattice diffusion of vacancies were first proposed by Nabarro in 1948 (1) and Herring in 1950 (2). In the Nabarro-Herring (N-H) theory, grain boundaries and surfaces were considered as (perfect) sources and sinks of vacancies. It is probably accurate to say that the N-H theory is both the most rigorous as well as the most elegant creep theory that has been proposed. Subsequently, in 1963, Coble (3) added the contributions of grain boundary diffusion, a process that is important at low homologous temperatures compared to N-H creep where only lattice diffusion was considered. The two mechanisms of N-H and Coble creep are called diffusional creep and do not require dislocation motion.

The three principal mechanisms of plastic flow at intermediate-to-high temperatures ($>0.6\ T_M$ where T_M is the absolute melting point) are slip, grain boundary sliding (GBS), and diffusional flow. These mechanisms are considered to be independent, parallel, sources of plastic flow and therefore, they are additive with the fastest dominating. (It is noted that, in some cases, two mechanisms are coupled — that is, they are interdependent, or sequential — such as in the accommodation of grain boundary sliding by either glide or climb of dislocations.) Slip generally dominates at high stresses, grain boundary sliding is known to contribute to plastic flow in materials of all grain sizes, but can dominate flow in fine grained materials at intermediate stresses. Diffusional flow has often been proposed to be the principal mechanism at low stresses, but special mechanisms of slip, such as Harper-Dorn (H-D) creep (4), can also apply at low stresses.

The above mechanisms, and variations of them, have characteristic equations that describe flow, and these lead to specific combinations of stress, temperature, and grain size dependencies. Only the equations for the particular cases of N-H and Coble creep are truly derived from a fundamental mechanistic approach. The others can be appropriately described as "phenomenological"; but, with modifications based on experimental data as well as some mechanistic insights, the phrase "enlightened phenomenological" can be used to describe them. It is appropriate to point out that the N-H and Coble theories also become "phenomenological" when they are modified by utilizing either a pre-constant (e.g., by arbitrary selection of grain boundary efficiencies as sources of vacancies) or through the introduction of a threshold stress of an unpredictable nature.

The simple versions of the equations for slip, GBS, and diffusional creep are shown as equations 1, 2, and 3, respectively, in Table I. The special case of slip in which dislocation pipe diffusion dominates is given as 1a; the case of slip by H-D creep is given as 1b; modifications for GBS controlled by grain boundary diffusion (rather than lattice diffusion), or accommodated by slip in Class I solid solutions, are given as 2a and 2b, respectively; and the modification to N-H creep (equation 3) to account for fine grain size contributions (Coble) is given as 3a.

To make clear the specific dependencies of the various mechanisms on stress (σ), temperature through the diffusivity (**D**), and grain size (**d**), these variables are explicitly noted in Table II.

Table I Equations for Slip, Grain Boundary Sliding, and Diffusional Creep

Mechanism	Equation	Equation No.
SLIP (Lattice)	$\dot{\varepsilon} \propto D_L \sigma^5$	1
SLIP (Dislocation Pipe)	$\dot{\varepsilon} \propto D_P \sigma^7$	1a
SLIP (H-D)	$\dot{\varepsilon} \propto D_L \sigma^1$	1b
GBS (Lattice)	$\dot{\varepsilon} \propto \left(\frac{D_L}{d^2}\right)\sigma^2$	2
GBS (Grain Boundary)	$\dot{\varepsilon} \propto \left(\frac{D_{gb}}{d^3}\right)\sigma^2$	2a
GBS (Class I Solid Solutions)	$\dot{\varepsilon} \propto \left(\frac{D_{sol}b}{d^2}\right)\sigma^1$	2b
Diffusional Creep (N-H) (Lattice)	$\dot{\varepsilon} \propto \left(\frac{D_L}{d^2}\right)\sigma^1$	3
Diffusional Creep (Coble) (Grain Boundary)	$\dot{\varepsilon} \propto \left(\frac{D_{gb}}{d^3}\right)\sigma^1$	3a

$\dot{\varepsilon}$ = true strain rate; σ = true stress; d = mean linear intercept grain size; b = Burgers vector; D_L = lattice diffusion; D_p = pipe diffusion; D_{gb} = grain boundary diffusion; D_{sol} = solute diffusion

Table II Variables Characterizing the Specific Mechanisms

Mechanism	σ	D	d
SLIP (D_L)	5	(D_L)	—
SLIP (D_p)	7	(D_p)	—
H-D (D_L)	1	(D_L)	—
GBS (D_L)	2	(D_L)	2
GBS (D_{gb})	2	(D_{gb})	3
GBS (D_{sol})	1	(D_{sol})	2
Diffusional (D_L)	1	(D_L)	2
Diffusional (D_{gb})	1	(D_{gb})	3

σ = true stress; D = Diffusivity; d = mean linear intercept grain size; D_L = lattice diffusion; D_p = pipe diffusion; D_{gb} = grain boundary diffusion; D_{sol} = solute diffusion

As may be seen, there is no complete match amongst these variables between any two of the mechanisms. Indeed, there is only one specific case in which there is an overlap in both stress and grain size dependencies, i.e., between the very special case of GBS in class I solid solutions and N-H creep. In this case, there is still a difference in the controlling diffusivity between D_{sol} and D_L. Despite these differences amongst the mechanisms, the controversies remain. In part, this is because of the difficulty in measuring the stress dependencies unequivocally at low stresses, especially if the stress range evaluated is very narrow. It can be difficult to be sure that steady state creep rates have been reached at very low stresses, for example. Also, activation energy analyses are notorious for rather large error ranges. Even grain size measurements can involve confusion between mean linear intercepts and true spatial grain size. For the cases of cube dependencies in grain size, this error alone can lead to a factor of 5 in predicted creep rates. It is for these reasons, at least in part, that other direct evidence, such as metallographic observations of denuded zones, takes on such importance.

In this paper, a summary is presented from the prior published work regarding such interpretations of experimental data for the creep of metals and ceramics at low stresses and intermediate-to-high temperatures. In particular, attention is given to those data that have been used as evidence to demonstrate the mechanisms of diffusional creep (i.e., N-H and Coble creep, as well as some derivative mechanisms such as that proposed by Ashby-Verrall (5)). Alternative mechanisms based on GBS and H-D creep are described that can also be used to interpret these data. Coupled with a critical review of the available metallographic observations and of unique microstructures that should accompany diffusional creep, it is concluded that there is no compelling evidence that uniquely confirms the existence of diffusional creep mechanisms.

Discussion

Table III summarizes, in chronological order of publication, the papers which involve differing views on the interpretation of low stress creep results. The left-hand column gives the papers which describe the role of diffusional creep as the rate-controlling process, whereas the right-hand column gives papers by the present authors that propose alternative views based on either a GBS or a dislocation creep model. About half the papers center on alternative views on the N-H creep model (items 5, 6, and 8-13), and the remaining papers center on alternative views on the Coble diffusional creep model (items 1-4, 7, and 11). The pro-and-con details of the thirteen listed items are given in the table segments, and the references for Table III are listed at the end of the table.

Table III. Summary of Papers on Plastic Flow at Low Stresses Related to Diffusional Creep vs. Grain Boundary Sliding and Dislocation Creep

Diffusional Creep Described as Rate-Controlling Process	— Alternative View — Grain Boundary Sliding and Dislocation Creep
1. Coble diffusional creep is proposed as the deformation mechanism for several metals at intermediate temperatures by Jones et al.; these are Cu (1975), α-Fe (1976), Mg (1977a), β-Co (1979), and 304 stainless steel (1980)	Coble diffusional creep is rejected as a deformation mechanism for Co, α-Fe, Cu, Mg, and 304 stainless steel at intermediate temperatures (0.4 to 0.6 T_M); GBS accommodated by slip and rate-controlled by grain boundary diffusion "(GBS-D_{gb})" is instead proposed (1982a)
2. The conclusions of the 1982 paper are rejected by Sritharan and Jones; new data on Co are presented to indicate evidence for Coble diffusional creep (1983a)	The new data on Co is used to show further confirmation for the GBS-D_{gb} model; the range of "GBS-D_{gb}" creep as influenced by grain size is illustrated (1984a)
3. Creep of fine-grained austenitic stainless steel at 0.7 T_M is explained by the Coble diffusional creep mechanism but requires the introduction of a threshold stress by Yamani et al. (1984b)	The grain size dependence and the actual creep rate are shown to not agree with the Coble diffusional creep model; a GBS model for creep at high temperature is introduced in which accommodation by slip is controlled by lattice diffusion (GBS-D_L); dislocation pipe diffusion is also included in the analyses (1985a)
4. Creep of α-Zr at intermediate temperature is attributed to the Ashby-Verrall model for diffusional creep; a GBS model is rejected by Fiala and Câdek (1985b) as not fitting the data	The creep data for α-Zr, in combination with other published data on low stress creep of α-Zr (1967a) reveal that all data are in fact best described by GBS (GBS-D_{gb}) (1986)
5. Several metals are considered to be controlled by N-H diffusional creep at low stresses; these are Cu (1949), Ag (1951), γ-Fe (1964), Cr (1966), Mo (1966), β-Co (1968a, 1982b), Ni (1969a), α-Ti (1981), and α-Fe (1983a)	These metals are shown to deform by a H-D creep mechanism; the creep data are well described by an internal stress-assisted dislocation creep model (1988)
6. Diffusional creep is considered the dominant creep process in fine-grained Al-7475 alloy in the temperature range 400 to 515°C by Hamilton et al. (1982c)	Analysis of data, together with additional studies on the same material, indicates that a GBS process (GBS-D_L) dominates deformation (1989a)
7. Creep of UO_2 at low stresses is "conclusively shown" to be dominated by Coble diffusional creep by Knorr et al. (1989b)	Creep rate of UO_2 is shown to be much faster than predicted by Coble diffusional creep; H-D creep is proposed as the deformation mechanism (1991)

8. Creep in $Mg\text{-}ZrH_2$ showing denuded zones is considered proof of existence of diffusional creep (1963a, 1968b, 1977b, 1985c)	Denuded zones are shown to be only observed at high stress exponents ($n = 4$ to 6); hence slip creep is rate-controlling; soaking time shown to dominate over grain size effect; evidence for diffusional creep of $Mg\text{-}ZrH_2$ at 500°C is shown to be invalid (1993a,b)
9. The conclusions presented by the present authors in their 1988 paper are considered to be in error; new data on creep of Cu (Pines and Sirenko, 1963b) are presented by Fiala and Langdon (1992) to indicate evidence for diffusional creep rather than H-D creep	The new creep data on Cu are shown to be totally unrelated to N-H diffusional creep; data are quantitatively assessed by GBS-D_L and H-D creep; soaking time effect shown to dominate over grain size effect on creep (1993b)
10. The analyses of Jones (1969b) on nine metals (Ni, Co, γ-Fe, Cr, Mo, Ag, Cu, δ-Fe, and Au) are considered by Fiala and Langdon (1992) to be the "widely cited" data which support a N-H diffusional creep model	Seven of the nine metals (Ni, Co, γ-Fe, Cr, Mo, Ag, and Cu) creep faster than predicted by diffusional creep theory and are shown to deform by H-D dislocation creep (1988, 1993b); creep of δ-Fe and Au can be controlled by either N-H diffusional creep or H-D dislocation creep (1993b)
11. Burton and Reynolds present a detailed defense of the diffusional creep theory; these include a discussion of denuded zones, superplasticity, creep of Cu (data of Burton and Greenwood, 1970) and grain boundary grooving in creep of UO_2 (1995)	The arguments of Burton and Reynolds are categorically rejected; GBS accommodated by slip is the principal deformation mechanism for fine-grained materials backed by extensive experimental and theoretical evidence; the Burton and Greenwood data on creep of Cu is shown to be controlled by power-law creep, by GBS, and by H-D dislocation creep (1996a)
12. Greenwood considers the evidence for denuded zones and for anisotropy of creep strength in dispersion-hardened materials is best explained by a diffusional creep process (1994)	The creep flow of dispersion-hardened materials is best explained by dislocation creep with a possible contribution from a threshold stress (the latter is not yet fully understood) (1996a)

13. New evidence is introduced by Langdon to show that diffusional creep is responsible for denuded zones in creep of Mg-0.5 wt% Zr at 400°C; creep experiments of Pickles (1967) at 350 to 450°C are given as proof of N-H diffusional creep as the deformation mechanism (1996b)	The creep rate-stress curve for Mg-0.5 wt% Zr at 400°C shows that denuded zones are observed in the power-law dislocation creep range where solute atoms interact with moving dislocations, and therefore the diffusional creep process is rejected. Pickles observation that denuded zones appear on both longitudinal and transverse boundaries at low stresses are interpreted to indicate that GBS, accompanied by grain boundary migration, is the principal deformation process (1998)

References to Table III

(1949) H. Udin, A.J. Shaler, and J. Wulff, Trans. AIME **185**, p.186.
(1951) E.R. Funk, H. Udin, and J. Wulff, Trans. AIME **191**, p.206.
(1963a) J.E. Harris and R.B. Jones, J. Nucl. Mater. **10**, p.360.
(1963b) B. Pines and A.F. Sirenko, Fiz. Meta. Metalloved. **15**, p.584.
(1964) A.T. Price, H.A. Holl, and A.P. Greenough, Acta Metall. **12**, p.29.
(1966) B.C. Allen, Trans. AIME **236**, p.903.
(1967a) I.M. Bernstein, Trans. AIME **239**, p.1518.
(1967b) B.W. Pickles, J. Inst. Met. **95**, p.333.
(1968a) L.F. Bryant, R. Speiser, and J.P. Hirth, Trans. AIME **242**, p.1145.
(1968b) O.D. Sherby and P. M. Burke, Prog. Mater. Sci. **13**, p.333.
(1969a) E.R. Hayward and A.P. Greenough, Mater. Sci. Eng. **4**, p.106.
(1969b) H. Jones, Mater. Sci. Eng. **4**, p.106.
(1970) B. Burton and G.W. Greenwood, Acta Metall. **18**, p.1237.
(1975) I.G. Crossland, in J.E. Harris and E.C. Sykes (eds.), *Physical Metallurgy of Reactor Fuel Elements*, Metals Society, London, p.66.
(1976) D.J. Towle and H. Jones, Acta Metall. **21**, p.399.
(1977a) I.G. Crossland and R.B. Jones, Met. Sci. **11**, p.504.
(1977b) B. Burton, *Diffusional Creep of Polycrystalline Materials*, Trans. Tech. Pub., p.72.
(1979) T. Sritharan and H. Jones, Acta Metall. **27**, p.1293.
(1980) T. Sritharan and H. Jones, Acta Metall. **28**, p.1633.
(1981) G. Malakondaiah and P. Rama Rao, Acta Metall. **29**, p.1263.
(1982a) O.A. Ruano and O.D. Sherby, Mater. Sci. Eng. **56**, p.167.
(1982b) G. Malakondaiah and P. Rama Rao, Mater. Sci. Eng. **52**, p.207.
(1982c) C.H. Hamilton, C.C. Bampton, and N.E. Paton, in N.E. Paton and C.H. Hamilton (eds.), *Superplastic Forming of Structural Alloys*, TMS-AIME, Warrendale, PA, p.173.
(1983a) T. Sritharan and H. Jones, Mater. Sci. Eng. **61**, p.1.
(1983b) J. Fiala, J. Novotny, and J. Cadek, Mater. Sci. Eng. **60**, p.195.
(1984a) O.A. Ruano and O.D. Sherby, Mater. Sci. Eng. **64**, p.61.
(1984b) T. Yamane, N. Genma, and Y. Takahashi, J. Mater. Sci. **19**, p.263.
(1985a) O.A. Ruano, J. Wadsworth, and O.D. Sherby, J. Mater. Sci. **20**, p.3735.
(1985b) J. Fiala and J. Cadek, Mater. Sci. Eng. **75**, p.117.
(1985c) J.P. Poirier, *Creep of Crystals*, Cambridge Univ. Press, Cambridge, England, p.197.
(1986) O.A. Ruano, J. Wadsworth, and O.D. Sherby, Mater. Sci. Eng. **84**, p.L1.
(1988) O.A. Ruano, J. Wadsworth, and O.D. Sherby, Acta Metall. **36**, p.1117.
(1989a) H.E. Adabbo, G. González-Doncel, O.A. Ruano, J.M. Belzunce, and O.D. Sherby, J. Mater. Res. **4**, p.587.
(1989b) K.B. Knorr, R.M. Cannon, and R.L. Coble, Acta Metall. **37**, p.2103.
(1991) O.A. Ruano, J. Wolfenstine, J. Wadsworth, and O.D. Sherby, Acta Metall. **39**, p.661.
(1992) J. Fiala and T. Langdon, Mater. Sci. Eng. **A151**, p. 147.
(1993a) J. Wolfenstine, O.A. Ruano, J. Wadsworth, and O.D. Sherby, Scripta Metall. **29**, p.515.
(1993b) O.A. Ruano, J. Wadsworth, J. Wolfenstine, and O.D. Sherby, Mater. Sci. Eng. **A165**, p.133.
(1994) G.W. Greenwood, Scripta Metall. Mater. **30**, p. 1527.
(1995) B. Burton and G.L. Reynolds, Mater. Sci. Eng. **A191**, p. 135.
(1996a) O.A. Ruano, O.D. Sherby, J. Wadsworth, and J. Wolfenstine, Mater. Sci. Eng. **A211**, p.66.
(1996b) T. Langdon, Scripta Metall. Mater. **35**, p.733.
(1998) O.A. Ruano, O.D. Sherby, J. Wadsworth, and J. Wolfenstine, Scripta Mater. **38**, p.1307.

One key observation emerging from an analysis of the various activities summarized in Table III is that the experimentally-observed creep rates, in the so-called diffusional creep range, are invariably faster in pure metals than those predicted from the Coble or the N-H creep model. Because of the theoretical rigour underpinning the N-H and Coble mechanisms, this significant discrepancy must mean that another deformation mechanism is more facile in controlling the creep process. An indication of what the alternative dominant creep process could be, can readily be established from deformation mechanism maps (6-8); examples are given in the following figures.

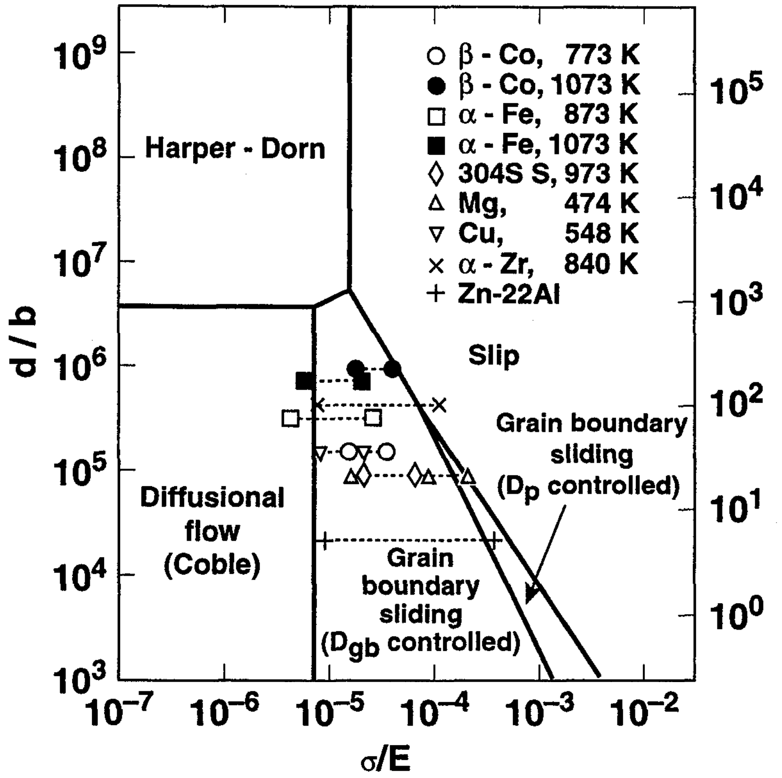

Figure 1
Langdon-Mohamed deformation map (normalized grain size vs. normalized stress) at an homologous temperature $T/T_M = 0.55$ (6, 9). The constitutive equations and material constants used for the construction of the map are given in Tables 1 and 2 of (6).

A deformation mechanism map is shown in Fig. 1 (6) for the particular case of a low homologous temperature ($T/T_M = 0.55$) where Coble diffusional creep is expected over specific grain size and stress ranges (8, 9). In this map, the burgers-vector-compensated grain size is plotted as a function of the modulus-compensated stress, and the regions dominated by the different, rate-controlling, creep processes are indicated. As can be seen, all the creep data that are commonly ascribed to a diffusional creep process, are seen to lie in the GBS region. The materials include β-Co, α-Fe, 304 SS, Mg, Cu, Zr, and a Zn-22%Al alloy. In this region, in addition to the discrepancies in strain rate predictions, the observed stress exponents for these materials are not equal to unity (as required by the diffusional creep theory) but, rather, are typically equal to about two, the value associated with GBS. The spread of the GBS region with a decrease in grain size fits in very well with the pattern of data evaluated. Included in this prediction is the well-studied superplastic Zn-22%Al alloy, containing very fine grains (about 6 µm), which exhibits a stress exponent of two over a region of almost two orders of magnitude of the modulus-compensated stress.

The regions of domination of the various deformation mechanisms shown in Fig. 1 are different when considering the creep characteristics of a ceramic oxide. Experimental evidence indicates that the H-D dislocation creep process, in which n = 1, is a more facile process than Coble diffusional creep. This is shown by the data in Fig. 2 for the case of stoichiometric, polycrystalline, UO_2 (10). These creep data were obtained at low homologous temperatures from a number of sources. The creep rates (normalized by the grain boundary diffusion rate) are observed to exhibit a stress exponent of unity. The creep rates, however, are all significantly above the creep rate predicted by the Coble diffusional creep relation, varying from a factor of 5 to over 1000 above the predicted values. This refutes the commonly stated view that diffusional creep is the rate-controlling process in polycrystalline UO_2 (11). The large range in the experimentally measured creep rates is attributed to the different dislocation densities in the material resulting from the various processing histories used in preparing the UO_2 samples (10). It is well accepted that as the dislocation density is increased, the creep rate is increased when H-D dislocation creep is the rate-controlling process.

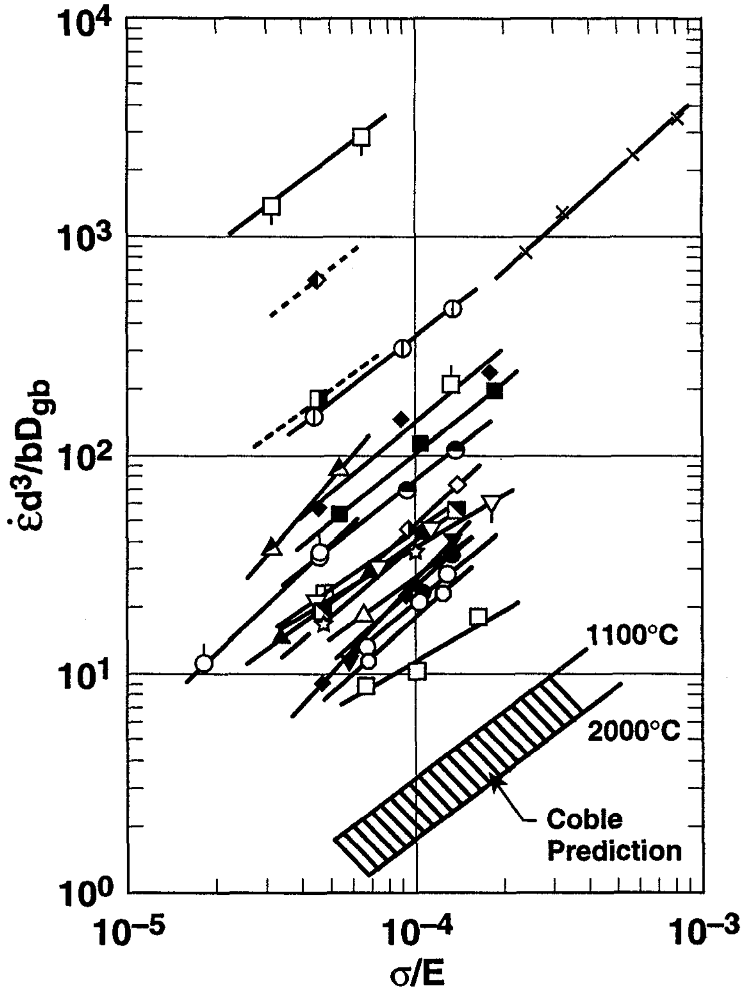

Figure 2
Grain-boundary-diffusion, burgers-vector and grain-size-compensated creep rate vs. modulus-compensated stress for polycrystalline UO_2 deformed in the Newtonian-viscous region. The Coble diffusional creep prediction is shown for comparison. Symbols used are from investigators listed in Table 1 of (10).

The deformation mechanism map shown in Fig. 3 (12) is for a high homologous temperature ($T/T_M = 0.90$). At this temperature, the creep behavior of many metals has been attributed to N-H diffusional creep. The map shows, however, that these metals are not controlled by diffusional creep, but rather are seen to be in the range where H-D dislocation creep dominates the deformation process. These metals include γ-Fe, β-Co, Cr, Ni, Mg, Ag, and Cu. The data on Cu, from Pranatis and Pound (13), obtained at 1040°C, is noted to be closest, on the grain size axis, to the N-H creep regime. These data on Cu have been used as classical examples to show that diffusional creep was the rate-controlling process. It was recently shown by the present authors, however, that an alternative view is required when all the creep data on Cu reported in the literature at 1040°C, where n = 1, are reviewed.

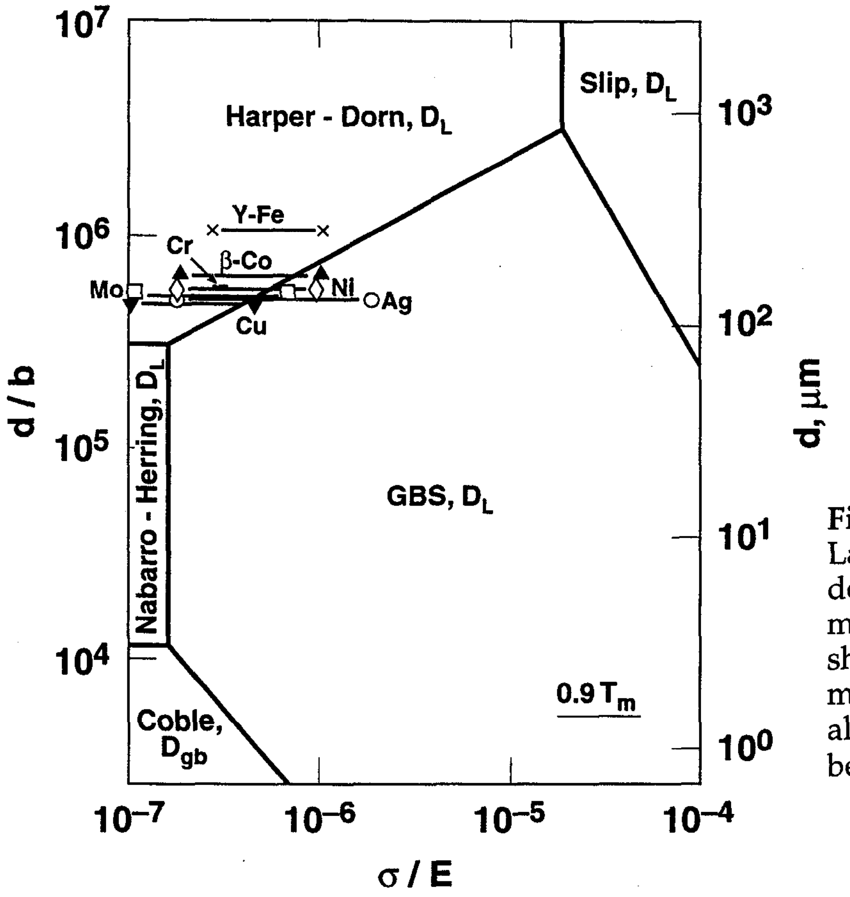

Figure 3
Langdon-Mohamed deformation mechanism map at $T/T_M = 0.9$ showing location of metals exhibiting anomalous diffusional creep behavior. (12)

A critical point is illustrated in Fig. 4. in which the soaking time, prior to testing, is shown to have a major influence on the creep rate of copper at $\sigma = 0.245$ MPa and 1040°C. The data include those from Pranatis and Pound, but also incorporate those from Udin, Shaler, and Wulff (14), and Pines and Sirenko (15). These data are critical because they show that the creep rate decreases with increasing soaking (or annealing) time; this result is to be expected in H-D creep since the dislocation density will decrease with an increase in annealing time. The grain size, in microns, is indicated beside each point. No trend is evident with grain size — a result that is inconsistent with an interpretation involving diffusional creep. The creep rates measured in fine grain size samples, studied by Udin et al., and Pranatis and Pound, are bounded within the creep rates from coarse grained samples from Pines and Sirenko. This correlation represents indisputable evidence that it is the dislocation density that influences creep rate and therefore leads to the conclusion that H-D creep is occurring and refutes the existence of N-H creep. Proponents of diffusional creep need to rationalize these data.

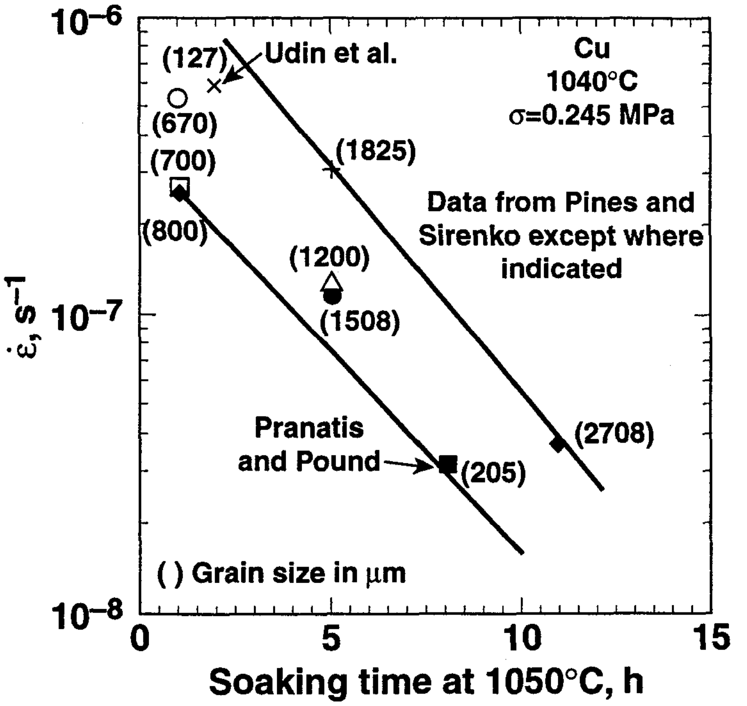

Figure 4
Influence of soaking time at 1050°C on the creep rate of Cu in the Newtonian-viscous range. (16, 17)

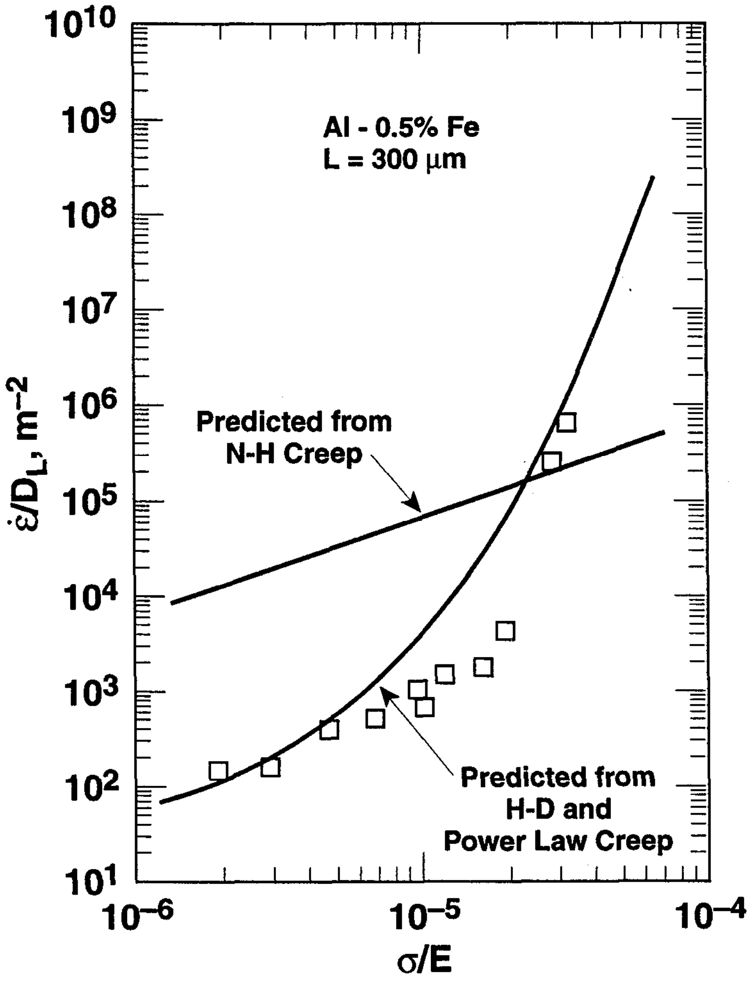

Figure 5
Diffusion-compensated creep rate as a function of modulus-compensated stress for Al-0.5%Fe. Predictions from N-H and H-D models are shown. (16)

In the metals described in Figs. 1 and 3, the experimental creep rates are always faster than those predicted by the diffusional creep theory. It should be pointed out, however, that for the cases of particle-strengthened materials, the opposite effect is observed. That is, the experimental creep rate is slower than the predicted diffusional creep rate. An example is high purity Al-0.5% Fe alloy (the Fe is contained in the alloy as Al_3Fe) studied by Barrett, Muhleisen, and Nix (18). This polycrystalline material (grain size of 300 μm) was studied from 0.95 to 0.99 T_M and at stresses where both power law creep and Newtonian viscous flow were observed. These data are plotted in Fig. 5 as diffusion-compensated creep rate as a function of modulus-compensated stress (16). The creep rate predicted by the N-H model is shown. At the lowest stresses, where $n = 1$, the actual creep rates are 100 times lower than the rate predicted by N-H creep. Thus one observes that creep rates predicted by the diffusional creep model are either slower or faster than the actual creep rate in the low stress range of creep. It is very difficult to create scientifically sound modifications to the diffusional creep theory to account in a consistent way for this range of observed behavior.

The formation of denuded zones at grain boundaries, normal to the tensile stress, during deformation of hydrided Mg-Zr materials, has been cited as direct evidence for the occurrence of N-H diffusional creep. Recently, the present authors refuted this evidence (17, 19) and instead concluded that the deformation process in these materials was one of diffusion-controlled dislocation creep. It was further postulated that the experimentally-observed, denuded zones were created as a result of tensile-stress-directed grain boundary migration during slip creep in which the moving grain boundaries dissolved the hydride precipitates.

The principal support of the present authors' view is three-fold. First, the denuded zones appear at stresses where the stress exponent is much above unity, n equalling from four to very high, threshold-stress-like values. An example is given in Fig. 6 for Mg-0.5%Zr alloys at 500°C for various investigations.

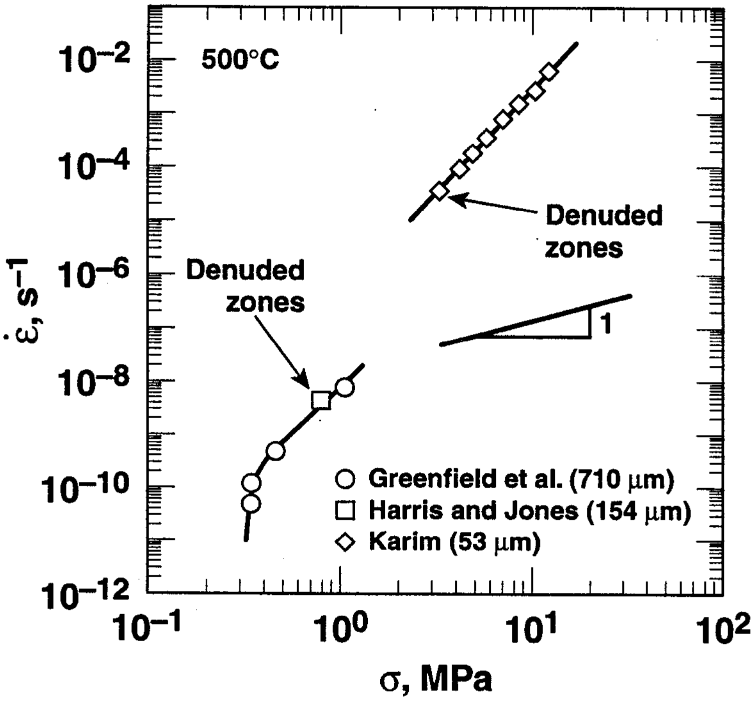

Figure 6
Logarithm of strain rate versus logarithm of stress curves for hydrided Mg-0.5%Zr alloys at 500°C from various investigators. Denuded zones were observed in the stress range in which the stress exponents are on the order of four; this value of stress exponent is found for mechanisms where dislocation creep dominates. (17)

Second, the estimated creep rates from measurements of denuded zone widths have no systematic correlation with the actual, measured creep rates. Third, the details of the observed denuded zones in published metallographic photographs cannot be explained by diffusional creep. Two examples of such micrographs are given in Figs. 7 and 8.

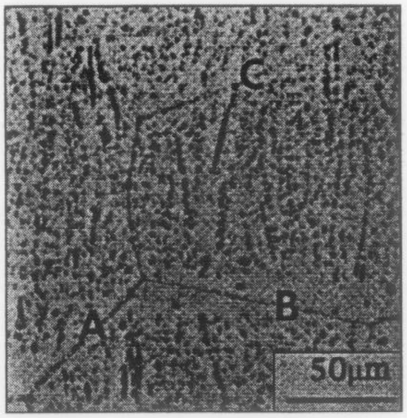

Figure 7
A photomicrograph from (20); it is observed that denuded zones appear at boundaries that are oriented away from the transverse direction and undissolved precipitates remain evident in denuded regions (regions marked A and B) (19). The tensile direction is vertical.

The denuded zones shown in Fig. 7 are from the first, published, observations of denuded zones by Squires, Weiner, and Phillips (20). In this photomicrograph, denuded zones are evident on boundaries that are oriented far from the transverse direction to the tensile axis (for example, region A). This unexpected result is also confirmed by Karim et al. (21, 22). It should also be noted that although denuded zones appear, as one would expect, on both sides of a given transverse boundary (Region B of Fig. 7), there are also examples where denuded zones are found on only one side of a transverse boundary (Region C). The work of Karim et al., in fact, shows that denuded zones appear almost always on only one side of a transverse boundary, an example of which is shown in Fig. 8. These observations are in disagreement with the diffusional creep model in which denuded zones of equal size are required on both sides of transverse boundaries (23, 24). Another observation is that denuded zones are not in fact fully denuded. For example, in Region B of Fig. 7, precipitates are clearly visible and this observation requires an explanation. In some cases, the denuded boundary appears to be serrated, in a manner reminiscent of grain boundary migration processes.

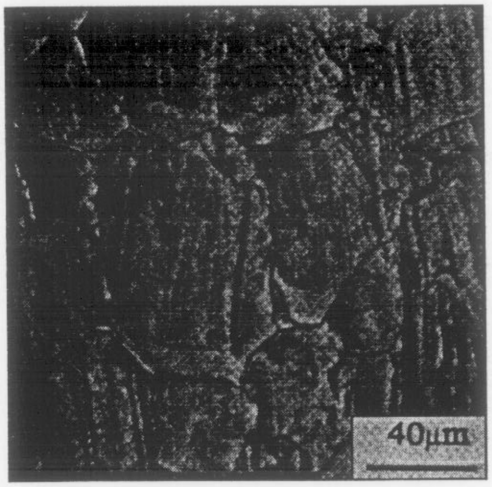

Figure 8
An example is shown of denuded zones from Karim (22). In many cases denuded zones are only found on one side of a transverse boundary in disagreement with the diffusional creep model in which denuded zones of equal size are required on both sides of transverse boundaries (19). The tensile direction is vertical.

It is therefore concluded that the origin of denuded zones lies not in diffusional creep, but rather in grain boundary migration during slip creep (17, 19). That grain boundary migration would be preferred on grain boundaries experiencing tension is sound. (In an analogous example, GBS in superplastic materials is shown to be enhanced when there is a tensile component to the shear stress as opposed to the case when there is a compressive component (8, 25)). The authors' proposal that the hydride particles will go into solution at moving boundaries is reasonable since the grain boundary structure is a more-open, amorphous-like structure than the crystalline matrix structure. The Zr dissolving at the tensile boundaries would migrate along the boundary and redeposit at longitudinal boundaries to give a higher hydride concentration at these locations, as is often observed.

Concluding Remarks

The N-H diffusional creep theory is the most rigorous and elegant theory developed in creep plasticity. However, most of the creep results which have been attributed to diffusional creep are not justified. GBS and H-D creep are competing mechanisms to diffusional creep in the low stress region and can readily explain most of the low stress creep results where low stress exponents are observed. The origin of denuded zones observed experimentally is worthy of further study since, to date, no satisfactory explanation has been made of their origin.

Acknowledgements

This work was performed under the auspices of the U.S. Department of Energy by Lawrence Livermore National Laboratory under Contract W-7405-Eng-48.

References

1. Nabarro, F.R.N., The Physical Society (1948) 75.
2. Herring, C., J. Appl. Phys. 21 (1950) 437.
3. Coble, R.L., J. Appl. Phys. 34 (1963) 1679.
4. Harper, J.C., and J.E. Durn, Acta Met. 5 (1957) 654.
5. Ashby, M.F., and R.A. Verrall, Acta Met. 21 (1973) 149.
6. Ruano, O.A., and O.D. Sherby, Mat. Sci. Eng. 56 (1982) 167-175.
7. Sherby, O.D., and J. Wadsworth in Deformation, Processing and Structure, ed. G. Krauss (Metals Park, Ohio: ASM, 1984), 355-389.

8. Nieh, T.G., J. Wadsworth, and O.D. Sherby, Superplasticity in Metals and Ceramics (Cambridge, United Kingdom: Cambridge University Press, 1997) 273.

9. Ruano, O.A., J. Wadsworth, and O.D. Sherby, Mat. Sci. Eng. 84 (1986) L1.

10. Ruano, O.A., J. Wolfenstine, J. Wadsworth, and O.D. Sherby, Acta Met. Mater. 39 (4) (1991) 661-668.

11. Knorr, K.B., R.M. Cannon, and R.L. Coble, Acta Met. 37 (1989) 2103.

12. Ruano, O.A., J. Wadsworth, and O.D. Sherby, Acta Met. 36 (4) (1988) 1117-1128.

13. Pranatis, A.L., and G.M. Pound, Trans. AIME 203 (1955) 664.

14. Udin, H., A.J. Shaler, and J. Wulff, Trans. Metall. Soc. AIME 185 (1949) 186.

15. Pines, B., and A.F. Sirenko, Fiz. Metalloved. 15 (1963) 584.

16. Ruano, O.A., J. Wadsworth, J. Wolfenstine, and O.D. Sherby, Mat. Sci. Eng. A165 (1993) 133-141.

17. Ruano, O.A., O.D. Sherby, J. Wadsworth, and J. Wolfenstine, Scripta Mater. 38 (8) (1998) 1307-1314.

18. Barrett, C.R., E.C. Muehleisen, and W.D. Nix, Mat. Sci. Eng. 10 (1972) 33.

19. Wolfenstine, J., O.A. Ruano, J. Wadsworth, and O.D. Sherby, Scripta Metall. et Mater. 29 (4) (1993) 515-520.

20. Squires, R.L., R.T. Weiner, and M. Phillips, J. Nucl. Mater. 8 (1963) 77.

21. Karim, A., D.L. Holt, and W.A. Backofen, Trans. AIME 245 (1969a) 1131.

22. Karim, A. in Ultrafine-Grain Metals, eds. J.J. Burke and V. Weiss (Syracuse, New York: Syracuse University Press, 1970), 295.

23. Karim, A., D.L. Holt, and W.A. Backofen, Trans. AIME 245 (1969b) 2421.

24. Harris, J.E., Met. Sci. J. 7 (1973) 1.

25. Wolfenstine, J., G. Gonzalez-Doncel, and O.D. Sherby, Mater. Lett. 15 (1992) 305.