APPENDIX I

C ONF-740724-3

DEFORMATION MECHANISM MAPS - A REVIEW WITH APPLICATIONS

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Contract AT(11-1)-2408

ABSTRACT

Deformation mechanism mapping has recently evolved as a method for quantitatively expressing complex deformation equations while at the same time allowing visual insight into our understanding of these deformation processes. Techniques for the generation of deformation mechanism maps are described, specific applications in the literature to date are reviewed, and some applications not previously considered are presented. These include the examination of sequential dependent deformation processes, grain boundary sliding, and the use of deformation mechanism maps to interpret pressure-sintering kinetics.

INTRODUCTION

It is always interesting to be able to observe how concepts develop from very unsophisticated qualitative descriptions to more precise quantitative approaches. However, in many areas, as our quantitative ability to handle complex problems increases, our basic physical understanding of these processes is sometimes easily lost. I think that the reason there has been so much interest in deformation mapping over the last year or so is that it overcomes this problem in a very picturesque way. Deformation mapping is an approach that allows us to quantitatively express complex deformation equations and at the same time allows us to visualize our physical concepts about the mechanisms involved.

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My purpose today is to describe what deformation mapping is, to review what has been done in the area to date, to describe some applications not previously considered, and to summarize what appears to be the value of these deformation maps for future problems.

In 1968, at an ASM Symposium on High Temperature Creep, Weertman [1] used a figure called a "creep diagram" to demonstrate in a qualitative manner the regions of stress and temperature in which different creep mechanisms could be expected to dominate. Recently, Ashby and coworkers [2,3,4], in a more quantitative approach, have shown that all of the deformation mechanisms may be combined graphically and their relative contributions predicted from a "deformation map." At any given stress and temperature, one mechanism for creep deformation is dominant. This means that on a graphical plot with stress as one axis and temperature as the other (Figure 1), fields may be determined which show the range of stress and temperature over which a particular mechanism predominates. In addition, contours of constant strain rate may be shown (Figure 1). This means that if any pair of the three variables, stress temperature and strain rate is known, the third may then be predicted.

Figure 1 - Deformation Map for Nickel ($d = 32 \mu$). Elastic boundaries arbitrarily defined as $\dot{\epsilon} = 10^{-8} \text{ sec}^{-1}$ (after Ashby [2]).

The boundaries of the fields are obtained by equating consecutive deformation rate equations (each mechanism thus contributes 50% of the deformation rate at a boundary) and solving for stress as a function of temperature. Contours of constant strain rate can be found by summing all of the strain rate contributions and solving for the stress that results in some constant strain rate value as temperature is varied. These calculations can be easily made and the deformation maps may be plotted by a computer if the necessary material constants are known. Maps may be made, for example, for a wide range of stress, temperature, and structural parameters such as grain size, and the results can be examined visually. Conversely, overlays of experimental data onto the deformation map quickly enable the calculation of the relative contribution of each deformation mechanism to the total deformation rate. Similarly, experimental conditions necessary for the observation of a particular deformation mechanism may be easily predicted.

CONSTITUTIVE EQUATIONS FOR DEFORMATION

In order to obtain rate equations for the deformation of materials, Ashby [2] initially considered the following five distinguishable and independent ways that a polycrystalline body may deform. First, a stress which exceeds the theoretical shear strength will cause deformation even in a defect-free crystal. Second, plastic deformation may be achieved through twinning; in early deformation maps [2] twinning was not included because of its limited contribution to deformation but later maps [3] included a contribution that appeared only at high stresses and low temperatures. Third, the glide motion of dislocations can lead to extensive plastic flow. Fourth, at higher temperatures dislocation creep is made possible as the dislocations increase their ability to climb as well as glide. Fifth, at high temperatures, low stresses and small grain sizes, the creep of polycrystalline materials may be controlled by point defects where deformation is accomplished by the diffusional flow of matter from areas near grain boundaries under compressive stresses to those which are under tensile stresses. This diffusional flow may take place through the bulk lattice [5,6] (Nabarro-Herring Creep) or a similar diffusive flow may occur by point defects moving along the grain boundaries [7] (Coble Creep). Although most of these mechanisms will be described in detail during the course of the next few days, I would like to present a short description for each of them in order to develop the rate equations necessary to produce the deformation maps.

The steady state creep (strain) rate due to Nabarro-Herring creep [5,6] is given by:

$$\dot{\epsilon}_{NH} = \frac{13.3 D_v \Omega \sigma}{kT d^2} \quad (1)$$

where σ is the stress, D_v is the lattice diffusion coefficient, Ω is the molecular volume, d is the grain size, and kT has its usual meaning. For Coble creep [7] the strain rate is given by:

$$\dot{\epsilon}_C = \frac{47.5 \Omega (D_b \delta) \sigma}{kT d^3} \quad (2)$$

where $(D_b \delta)$ is the product of the grain boundary diffusion coefficient and the effective grain boundary width.

Raj and Ashby [8] have developed a more general description of deformation by diffusional flow, including both bulk and lattice diffusion such that

$$\dot{\epsilon}_{DIFF} = 14 \frac{\sigma \Omega}{kT} \frac{1}{d^2} D_v \left[1 + \frac{\pi \delta D_b}{d D_v} \right] \quad (3)$$

Note that if the second term within the brackets is $\ll 1$ this equation reduces to an expression similar to one obtained from the Nabarro-Herring stress-directed diffusional model. Alternatively, if the second term within the brackets is $\gg 1$ this equation yields results very similar to those obtained by Coble for grain boundary diffusion controlled creep.

At high temperatures (above about $0.5 T_M$) and at relatively high stresses creep deformation is apparently diffusion controlled but the strain rate is a non-linear function of stress. This has generally been attributed to the ability of dislocations both to climb and glide in this stress-temperature regime. It has been shown [8] that in this regime the creep behavior of many materials can be well described by the semiempirical equation

$$\dot{\epsilon}_{DISL} = A \frac{D_v \mu b}{kT} \left(\frac{\sigma}{\mu} \right)^n \quad (4)$$

where μ is the shear modulus, b is the Burger's vector. The constants A and n are commonly referred to as the Dorn parameters. The exponential constant, n , has been related to the nature of the specific dislocation mechanism proposed to be operative during creep [9]. Recent work by Stocker and Ashby [10] has shown that n and A do not vary independently, and they derive a relationship that can be used to find A when n is known.

At lower temperatures but at approximately the same stress levels applicable for equation (4), it becomes necessary to include

a term that describes atom transport by dislocation core diffusion [3]. The volume diffusion term (D_v) in equation (4) may be replaced by an effective diffusion coefficient:

$$D_{\text{eff}} \approx D_v \left[1 + \frac{3}{b^2} \left(\frac{\sigma}{\mu} \right)^2 \frac{a_c D_c}{D_v} \right] \quad (5)$$

where D_c is the dislocation core diffusion coefficient and a_c is the cross-sectional area of the dislocation core. Thus, at high temperatures and at lower stresses lattice diffusion is dominant; Frost and Ashby [3] have termed this "High Temperature Creep." At lower temperatures and at higher stresses, dislocation core diffusion is dominant, the strain rate varies as σ^{n+2} rather than σ^n ; this type of dislocation creep has been termed "Low Temperature Creep." While the more recent deformation maps [3] include a regime for low temperature dislocation creep, the early maps [2,11,12] ignore this mechanism. Since most of the applications that will be described here, and that are of interest insofar as steady state deformation in ceramic materials, occur at higher temperature, I have chosen to simplify the maps by eliminating low temperature dislocation creep effects.

At even higher stress levels, the rate of dislocation glide rather than climb usually controls the deformation rate and this is usually obstacle limited. However, materials with strong covalent bonding and many bcc metals demonstrate glide controlled by lattice resistance (Peierls barrier). Above the cut-off stress for glide control, Ashby [2] has adopted a rate equation for obstacle limited flow such that

$$\dot{\epsilon}_{\text{OBST}} = \dot{\epsilon}_0 \exp \left[-\frac{b^2 \ell}{kT} (S - \sigma) \right] \quad (6)$$

where ℓ = obstacle spacing

$$S = \frac{\mu b}{\ell} = \text{flow stress at absolute zero}$$

$$\dot{\epsilon}_0 = \text{strain rate when } \sigma = S \text{ (assumed } \approx 10^{+6} \text{ /sec.)}$$

Later work by Frost and Ashby [3,4] gives essentially the same equation but in slightly different form. As pointed out by Roberts and Voglewede [13], the upper stress limit for deformation in many ceramic materials may be determined by the brittle-fracture stress, σ_F , rather than by the presence of pure dislocation glide. Therefore, on many of the maps presented here, I have included an approximate value of σ_F , rather than including obstacle controlled glide.

Finally, at stress levels above some fraction of the theoretic-

cal strength, the material will fail even in the absence of crystal defects. For failure by this mode, Ashby [2] assumes that

$$\left\{ \begin{array}{ll} \dot{\epsilon}_{TH} = \infty & \text{above} \\ \dot{\epsilon}_{TH} = 0 & \text{below} \end{array} \right\} \quad \tau = 0.039 \mu \quad (7)$$

Another complication that arises in the case of ceramic materials is the dependence of the diffusion constants upon the complex nature of the diffusing species. Not only must the path for diffusion be considered, as in metals, but creep will depend on the nature of the rate controlling species (i.e., anion control versus cation control) and the ambient conditions and purity that delimit both intrinsic and extrinsic regions for each ionic species. An initial attempt to include these factors has been made by Stocker and Ashby [14] for MgO. Recently maps have also been generated for NiO [12] (Figure 2), CoO [11,17], UO₂ [2,13], olivine [14], NaCl [15], and MgAl₂O₄ [16,17]; the interest generated by these maps has been indicated by their rapid inclusion in at least two review papers on deformation in ceramic materials [18,19].

Figure 2 - Deformation Map for NiO ($d = 4.74\mu$). Experimental pressure sintering data [12] has been superimposed.

APPLICATIONS OF DEFORMATION MAPPING

To a certain extent the principal advantage of deformation mapping is purely pedagogic. As one example, it allows for the easy visual comparison of the properties of different groups of materials; the ability of these maps to contrast the mechanical behavior of the FCC, HCP, BCC and diamond cubic elements has clearly been demonstrated [2]. Similarly, Frost and Ashby [3] have compared the normalized constant strain rate contours calculated for a number of FCC metals and have related the difference in behavior of each metal to its stacking fault energy.

In addition to its use as a pedagogic tool, since deformation maps clearly and quantitatively indicate strain rate contours and the regions in which specific deformation mechanisms are dominant, they also provide a very strong predictive tool. In this context, Frost and Ashby [20,21] have investigated their use as a means of solving a serious creep problem for tungsten filaments in light bulbs. The maps that they generated indicate that for fine grained pure tungsten, under the stress and temperature conditions imposed during operation, the filaments would be in a diffusional creep field with a creep rate approximately six orders of magnitude higher than the acceptable limit. Although the maps indicated that changing to a coarser grained material would shift the controlling creep mechanism to high temperature dislocation creep, the creep rate was still too high. As pointed out by these investigators, a successful solution to the problem could only be found by limiting creep by both mechanisms. A proper solution was found in commercial practice where tungsten is doped with impurities that leave a fine dispersion of voids after processing. This fine dispersion stabilizes an elongated grain structure that limits diffusional creep and inhibits dislocation creep in a manner similar to dispersion hardening. It is interesting to note that an alternate commercial solution to the problem has been to use single crystal tungsten filaments [22]; diffusional creep is decreased by the absence of grain boundaries but the approach does not represent as viable an engineering solution as that indicated by deformation mapping.

Thus deformation mapping provides for the prediction of material behavior for which little or no experimental data exists and, since the dominant regimes for each mechanism can be observed, it prevents incorrect extrapolations from control by one mechanism to another. Because the maps increase our physical understanding of observed relations, they therefore increase the confidence with which data may be extrapolated beyond the range of experiments. It is this predictive-extrapolative feature of deformation mechanism maps that makes them extremely interesting, for example, insofar as applications to nuclear materials. This has already been demonstrated by the generation of maps for UO_2 and UO_2 - PuO_2 [13] that are

based on experimental data (including in-pile creep effects) rather than theoretically derived rate equations. Combinations of the information output of deformation maps and the reactor-modeling programs already developed [23] should provide a much more satisfying prediction of mechanical properties of nuclear materials than currently exists.

I would like to turn to three areas of application of deformation mapping in which we have been involved. Hopefully, they will help elucidate the variety of applications for which mapping may be used. These are: the inclusion of interdependent deformation mechanisms in the maps, the inclusion of missing mechanisms that have been indicated to be present through microstructural evidence, and finally the use of maps to jointly model deformation and densification during pressure sintering.

INTERDEPENDENT CONSTITUTIVE EQUATIONS

Each of the mechanisms for deformation considered so far have been characterized as independent processes. Namely, the total strain rate due to the combination of these mechanisms is the sum of each individual contribution:

$$\dot{\epsilon}_T = \dot{\epsilon}_1 + \dot{\epsilon}_2 + \dots \quad (8)$$

These types of mechanisms have been called SIMULTANEOUS-INDEPENDENT [24] or PARALLEL-CONCURRENT [25] modes of deformation. On the other hand, it is possible for two processes to be interdependent in such a way that the one process cannot proceed until the other has taken place to some degree. These types of mechanisms have been called SEQUENTIAL-DEPENDENT [24] or SERIES-SEQUENTIAL DEPENDENT [25] modes of deformation. When additive (independent) processes are involved the fastest is rate controlling; if sequential (dependent) processes are involved the slowest is rate controlling [26,27].

The plastic flow behavior of most materials studied to date indicate agreement with the types of deformation maps previously described. However a few, i.e., α -zirconium and titanium appear to behave in an anomalous manner. For example, the stress dependence of α -zirconium is such that the stress exponent, n , appears to decrease with an increase in stress. Ardell and Sherby [28] have attributed this behavior to the high climb rate that results from the high diffusivity and low elastic modulus of α -zirconium. Dislocation glide could then be assumed as the rate controlling mechanism at lower stresses. Ardell and Sherby define a glide controlled creep rate such that:

$$\dot{\epsilon}_{\text{GLIDE}} = C \exp(\beta\sigma) \exp\{-(Q_k + \eta/\sigma)/RT\} \quad (9)$$

where C , β and η are constants and Q_k is the kink formation activation energy.

We have considered the problem of the sequential dependent interaction between glide and climb for both α -zirconium and Zircaloy-2 and have incorporated our results into deformation mechanism maps [29]. Equation (4) for dislocation climb and equation (9) for glide were incorporated into the mapping program such that if $\dot{\epsilon}_{\text{CLIMB}}$ was greater than $\dot{\epsilon}_{\text{GLIDE}}$, the deformation rate was set equal to $\dot{\epsilon}_{\text{GLIDE}}$; if $\dot{\epsilon}_{\text{GLIDE}}$ was greater than $\dot{\epsilon}_{\text{CLIMB}}$ the deformation rate was set equal to $\dot{\epsilon}_{\text{CLIMB}}$. The total strain rate was found by adding the smaller of these quantities to the diffusional contribution. The maximum temperature considered for the maps was 862°C, the $\alpha(\text{HCP}) \rightarrow \beta(\text{BCC})$ phase transition temperature, and the temperature scales for the maps were normalized with respect to this temperature.

A deformation map for α -zirconium with a grain size of 60 μ is shown in Figure 3. The values for the constants C , β , η and Q_k were taken from Ardell and Sherby [28] whose experiments were made with 300 μ material; superimposed on the map are experimental values of strain rates (the numbers near the points are $\log \dot{\epsilon}$ values) reported by Bernstein [30]. The agreement between the contours predicted by the map and the experimental values is excellent. An especially significant point is that the constants for the equations were obtained by Ardell and Sherby at quite high temperatures, yet the data points are in agreement with the deformation map at lower temperatures.

Gilbert, Duran and Bement [31] have indicated that conceptually the same deformation mechanisms are applicable in Zircaloy-2 (1.5% Sn, 0.14% Fe, 0.10% Cr, 0.05% Ni, balance Zr) as in α -zirconium. We have therefore included a deformation map of Zircaloy-2 having a grain size of 10 μ in Figure 4; new values for the constants in equations (9), (4), (2) and (1) appropriate to Zircaloy-2 have been used. Again, data from Bernstein [30] are in excellent agreement with the deformation map predictions.

MISSING MECHANISMS

One difficulty with the Nabarro-Herring and the Coble creep theories is that they imply that grain elongation should occur during deformation when in fact this is almost never observed to occur. The models also assume the relaxation of shear stresses at the grain boundary by a viscous relaxation phenomenon but the

contribution of this relaxation mechanism is assumed to be unimportant in the deformation process. Creep studies at low stress levels have also revealed that scratch offsetting, grain rotation, and granular surface roughening can occur while an equiaxed grain structure is maintained [32]. These observations are consistent with the description of creep as a grain boundary sliding process. However, in order for sliding to continue it is necessary that an accommodation process occurs. Generally it is the accommodation process which controls the extent and rate of sliding. Raj and Ashby [8] have studied the process of sliding with diffusional accommodation. Their model combined both the grain boundary sliding and the accommodation into a single model where both processes occur simultaneously. Raj and Ashby thus show that diffusional flow and grain boundary sliding are not independent. They are coupled, and the resulting deformation is described either as diffusional creep or as grain boundary sliding with diffusional accommodation [3].

At higher stresses, grain boundary sliding may be accommodated by plastic flow. Actually, two types of grain boundary sliding may be defined. Sliding that takes place on interior grain boundary surfaces must be accompanied by accommodation deformation in adjacent grains (Figure 5a); this type of sliding can be termed accommodated grain boundary sliding. On the other hand grain boundaries that intersect the sample surface (Figure 5b) should slide freely at a rate determined predominantly by the nature of the boundary; this type of sliding can be termed free grain boundary sliding. Raj and Ashby [34] have recently shown that at high temperatures the free sliding rate varies with temperature as D_v , while at sufficiently low temperatures the free sliding rate is controlled by boundary diffusion. Matlock and Nix [35,36] have shown that geometrically accommodated grain boundary sliding and free grain boundary sliding can be directly related. Thus, except for the inclusion of a constant term near unity, equations for both accommodated and free grain boundary sliding should be expected to have the same form. In addition, in a porous sample (Figure 6), the accommodational requirements necessary for sliding are considerably relieved.

In metals, grain boundary sliding associated with plastic flow appears to be associated with strain rates that have fairly high stress exponents. Matlock and Nix [36] have correlated the stress exponents associated with grain boundary sliding and stress exponents associated with grain deformation; they indicate that these may be related such that $n_{\text{gbs}} \approx n_{\text{g}} - 1$. For oxides such as MgO [37], UO_2 [38], $\text{BaO} \cdot 6 \text{Fe}_2\text{O}_3$ [39] and ZrO_2 -30 w/o UO_2 [40] there is considerable evidence that plastically accommodated grain boundary sliding is associated with a stress exponent in the range of 1.5 to 2.5.

Figure 5 - Accommodated Grain Boundary Sliding (5a) and Free Grain Boundary Sliding (5b), after Matlock and Nix [36].

Figure 6 - Sliding in a Porous Compact. Requirements for Accommodation are Considerably Relaxed by Internal Pore Surfaces.

Wire and coworkers [41] have indicated that the region labelled dislocation creep in the deformation mechanism maps should really be subdivided into two regions; one region governed by grain matrix processes and the other governed by the grain boundary sliding contribution through plastic accommodation. Langdon [24] has developed a model for grain boundary sliding by assuming that all the deformation mechanisms, including sliding, operate independently. In this model, sliding occurs by the movement of dislocations along, or adjacent to, the boundary by a combination of climb and glide. The strain rate associated with grain boundary sliding is given as:

$$\dot{\epsilon}_{\text{gbs}} = \frac{\beta b^2 \sigma^2 D_v}{d \mu k T} \quad (10)$$

where β is a constant approximately equal to one.

We have included grain boundary sliding, as a plastic flow accommodating process, in the deformation mechanism maps by incorporating equation 10. This relation has been chosen because of its availability as a quantitative expression, because of the same stress-exponent ($n = 2$) and grain size dependence (d^{-1}) as found in at least two other similar models [42,43], and because of the general agreement with the experimentally observed stress dependence in ceramic materials. In order to emphasize the stress and grain size dependence of this mechanism we have chosen to plot a different form of deformation map in Figure 7. This figure shows a "deformation map" for MgAl_2O_4 at a constant temperature of 1773°K; rather than a plot in σ - T space, this map is in σ - d space and highlights the microstructural variations more effectively. Constant strain rate contours can still be shown; in addition, because of the log-log nature of the plot and the form of the equations, the boundaries all appear as straight lines rather than curves. For the particular example of MgAl_2O_4 , shown in Figure 7, grain boundary sliding does not appear to be a major contributor to creep for a grain size below about one micron or below about 500-1000 psi. The constant strain rate contours on this type of plot also highlight the extreme sensitivity of diffusional creep and the insensitivity of dislocation creep to grain size effects.

APPLICATION TO HOT-PRESSING AND SINTERING

As a final application of deformation mapping, we might look at the compaction of a powder under the application of pressure and/or temperature. In a pressure sintering die, the total mass of the powder and the cross-sectional area are constant while the density and total sample height are variable. If the density is given as

$$\rho = \frac{m}{A\ell}$$

where m is the mass of the sample, A is the cross-sectional area of the die and l is the sample height, then

$$\frac{1}{\rho} \frac{d\rho}{dt} = - \frac{1}{l} \frac{dl}{dt} \quad (11)$$

As shown by Coble [44,45], because the right-hand side of this expression is a linear strain (linear shrinkage) rate which may be obtained from measurement of the ram travel in a pressure-sintering die, the densification rate may be related quantitatively to a kinetic expression of strain rate during creep as a function of stress and temperature if appropriate values for the effective stress in a porous sample can be determined. In addition, Coble has shown [44] that for final stage densification during pressure sintering the driving force should be $[\frac{2\gamma}{r} + \frac{\sigma_a}{\rho}]$, (where γ is surface energy, r is the pore radius, σ_a is the applied stress, and ρ is the relative density) rather than just the surface energy contribution of the pore. In fact, the influence of pressure on changing densification rate can be more significant during most of the closed pore stage than is the surface energy effect.

Figure 7 - Deformation Map for $MgAl_2O_4$ ($T = 1773^\circ K$) in σ -d Space Showing Field for Grain Boundary Sliding.

If the stress axis of a deformation map is converted to the effective stress for densification ($\frac{\sigma_a}{p} + \frac{2\gamma}{r}$) then it could be used

to compare pressure sintering data obtained at constant density levels to the deformation behavior of a material. We have recently used such "densification maps" [17] to jointly model pressure sintering and creep deformation. Superimposed on the map for NiO shown in Figure 2 is the experimental range of stress-temperature-strain rate conditions found during a study of pressure sintering kinetics. It may be seen that the densification rates observed agree very well with those predicted from the map. The pressure sintering data falls within the Nabarro-Herring creep region, and calculations from the experimental data, using the above information, give excellent agreement between effective diffusion constants and cation tracer values with respect to both magnitude and activation energy [12]. Similar results have been obtained with CoO [11]. Markworth [46] has used this same approach to demonstrate the effect of pore radius, grain size, and diffusivity on pressure sintering of UO_2 .

It sometimes becomes difficult to use the simple atomistic models available for pressure sintering [44,45], because of either major microstructural changes that occur during processing, or because of a non-linear stress-strain rate dependence over the range of experimental conditions. Our studies, now in progress, for $MgAl_2O_4$ indicate that such non-viscous behavior is indeed observed [16]. Analysis of the pressure sintering data for $MgAl_2O_4$ according to the phenomenological approach of Rossi and Fulrath [47] indicated that a stress exponent of $n = 1$ was not observed; a hot pressing model developed by Wolfe [40,48] for a solid with non-linear stress-strain rate response was therefore used. According to Wolfe's analysis, for constant applied stress, the change in porosity as a function of time may be given as:

$$\frac{dP}{dt} = A \left[\frac{P(1-P)}{(1-P)^n (1-P^{1/n})^n} \right] \quad (12)$$

where A is a constant that depends upon the stress level, the stress exponent, and the strain rate. Coble's porosity correction

($\sigma_e = \frac{\sigma_a}{p}$) has been incorporated into equation 12. A plot of dP/dt

versus the bracketed term in equation 12 should be a linear function for the effective value of n operative over the range of investigation. Our data for $MgAl_2O_4$ hot pressed at 1500°C and 8000 psi is presented on such a plot in Figure 8; the data produces a straight line fit for $n \approx 1.4$.

Figure 8 - Wolfe's Non-Viscous Phenomenological Model [48] for MgAl_2O_4 at 1500°C and 8000 psi .

Figure 9 - Deformation Map for MgAl_2O_4 ($d = 0.3 \mu$). Experimental Pressure Sintering Data has been superimposed.

The range of stress and temperature investigated during this study has been superimposed on the deformation map for MgAl_2O_4 shown in Figure 9. This shaded region lies within the Nabarro-Herring creep regime but near the boundaries for both dislocation creep and Coble creep and the experimentally observed creep rates are in good agreement with those predicted from the map. Since tentative results show the Coble contribution not to have a major effect, the total strain rate may be expressed as:

$$\dot{\epsilon}_T = D_{\text{eff}} [K_{\text{NH}} \sigma + K_{\text{gbs}} \sigma^2 + K_{\text{DISL}} \sigma^{4.5}] \quad (13)$$

where D_{eff} is the effective diffusion coefficient and where K_{NH} , K_{DISL} and K_{gbs} are the constants found in equations 1, 4, and 10 respectively. At each temperature, a plot of the total experimental strain rate versus the bracketed term in equation 13 should produce a straight line whose slope is D_{eff} ; a plot of $\log D_{\text{eff}}$ versus inverse temperature may then be used to find the activation energy for the process. Analysis in this manner gave good agreement with anion lattice diffusion control both as to magnitude and activation energy.

The variation in total strain rate (the sum of the terms in equation 13) may be related to the stress according to the relation

$$\dot{\epsilon}_T = K_T \sigma^{\text{"n"}} \quad (14)$$

where "n" is a macroscopic effective stress exponent. This effective stress exponent may be evaluated by first calculating the total strain rates (equation 13) at a number of stress levels and then using equation 14 as the basis for a plot of $\log \dot{\epsilon}_T$ versus $\log \sigma$; the slope of this plot is the effective stress exponent for the multimechanism deformation process. The data for MgAl_2O_4 at 1500°C with a density of $\rho = 0.87$ was evaluated in this manner over the stress range 5000-10,000 psi; the resultant value of "n," approximately 1.3, is in good agreement with the value obtained from Wolfe's analysis ($n = 1.4$). A consistent interpretation of the non-viscous behavior observed during pressure sintering is thus made possible through the mapping technique. However, an important point should be made. Although the experimental data (shaded area) in Figure 9 lies entirely within the dominant Nabarro-Herring field, the strain rate contours do not indicate completely viscous response; a mechanism can contribute significantly to the total deformation process as the boundaries of a field are approached, and may even contribute significantly if no regime corresponding to the mechanism appears on the map. This is because the mechanism must contribute at least 50% of the deformation in order to appear. Thus

the grain boundary sliding regime that appears in Figure 7 but not in Figure 8 can contribute to the total strain rate contours that do appear in Figure 8.

Two other recent applications of deformation mapping relevant to sintering phenomena might be mentioned at this point. Stocker and Ashby [14], in order to explain the deformation that occurs in the earth's mantle, have included a strain rate contribution due to fluid phase transport under applied stress for some of their maps. An extension of this idea could lead to densification maps for sintering of a material in the presence of a small volume of a liquid phase (such as $\text{MgO-Li}_2\text{O}$ or mullite). Lastly, Ashby [49] has reported on the generation of sintering diagrams for initial, intermediate and final stage densification for a number of metallic systems and for UO_2 . These maps are in the form of plots of normalized neck radius versus temperature and include contours of either constant normalized sintering rate or constant time. Because of the simple way that they demonstrate regions in which rapid densification would or would not occur, these displays may be used to design commercial sintering schedules. Specimens held under conditions that enhance volume and boundary diffusion will sinter rapidly and could be used where high densities are important; specimens held under conditions that enhance surface diffusion do not densify and these conditions could be used, for example, to produce porous filter materials.

SUMMARY

Deformation mechanism maps provide strong pedagogic, predictive and extrapolative tools for the investigation of the mechanical deformation of materials. They provide a means for quantitatively expressing the behavior of the material while at the same time increasing our physical understanding of the mechanisms involved.

Applications for deformation mechanism maps include the generation of information on mechanical properties under conditions where direct experiment might be extremely difficult; the mapping technique has already been used to try to understand the behavior of in-service nuclear materials. Because of the visual insight that they allow, maps may be helpful in identifying missing or dependent (sequential) mechanisms that are not easily defined, and they can be used to understand complex deformation-related processes such as pressure-sintering.

Deformation maps have been described as "mechanical phase diagrams." In this light, a map for nickel (after Frost and Ashby [3]) is shown in Figure 10. On it are shown the areas that have

Figure 10 - Deformation Map for Nickel ($d = 100\mu$) (after Frost and Ashby [3]). Areas of commercial interest and laboratory experiments are shaded.

been investigated and the areas that are of interest for commercial use; the areas that are of interest for engineering applications, such as forming and structural design for maximum creep life are the least studied in the laboratory. It is for this reason that the maps are of interest and are useful for understanding the behavior of materials.

ACKNOWLEDGEMENT

The author gratefully acknowledges the Atomic Energy Commission for the support of this work under Contract No. AT (11-1) 2408.

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