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TITLE CONSTITUTIVE LAWS FOR DEFORMATION AND DYNAMIC RECRYSTALLIZATION IN CUBIC METALS

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CONSTITUTIVE LAWS FOR DEFORMATION AND DYNAMIC RECRYSTALLIZATION IN CUBIC METALS†

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ABSTRACT

We describe two cases in which constitutive laws for deformation kinetics are available that are both physically well founded and experimentally well obeyed. New experiments on Al-Mg alloys, in the regime of viscous deformation, fit the solute drift equation very well, with $n=3$ and Q_D unadjustable; they do not fit the solute drag model. The high-stress regime, as well as all data for pure copper, fit the model of hardening and dynamic recovery, at least up to temperatures of 0.6 T_m. In both cases, dynamic recrystallization occurs at high temperatures. It seems to follow rather than determine the deformation kinetics.

1. INTRODUCTION

Constitutive Relations describe the response of a material to imposed boundary conditions; in the present context, to imposed stresses or strain rates, at a prescribed temperature. While the material microstructure is often implicit in the physical basis that led to the formulation and lends it some degree of confidence even beyond the measured range, the constitutive relations themselves are not concerned with the metallurgical end product. In this sense, recrystallization is considered here only to the extent it influences the mechanical response. Grain size does not enter the macroscopic relations in the regimes we are describing; texture does, and we will use its effects without treating them in detail. Dislocations densities and arrangements do contribute to kinetics directly, as well as to hardening and softening processes; the latter will, however, be treated by the equivalent macroscopic state parameters.

All experiments reported in the following were undertaken by prescribing a constant true strain rate in compression, under conditions where friction did not affect either the shape of the sample or the stress strain behavior.[1,2] The equivalence, in principle, between imposed-strain rate and imposed-stress ("creep") behavior is now generally accepted[3], differences that may be important are more in the nature of evaluation procedures (such as the extrapolation to "steady state") and of the range of variables examined. We will emphasize behavior in the "intermediate stress range" (around 10^{-4} to 10^{-3} of the shear modulus) and its connection to the high-stress regime.

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2. THE REGIME OF VISCOUS DISLOCATION MOTION

This case is a modeler's dream come true: it is possible, on the basis of well founded microscopic mechanisms, to derive a quantitative constitutive relation that is verified experimentally over a substantial range of variables; and it should hold for a whole class of materials. It is a well-worked case[4-12]; the only reason it is presented here is that we have a new set of experiments on Al-Mg alloys (originally undertaken to study texture development[2]); and our quantitative evaluation clearly favors one variant of the extant kinetic laws.

2.1 Kinetic Law

The class of materials and conditions for which this model applies (which has been labeled "Class A" behavior[10] is distinguished, on the microscopic scale, by continuous motion of dislocations accompanied by diffusion of solute. The assumption of essentially *straight* dislocations (since one type of dislocation, the edge, moves much more slowly than the other), all of which are moving, justifies the kinematic equation

$$\dot{\varepsilon} = b\,\rho\,v \tag{1}$$

for the whole assembly of dislocations.[13] ($\dot{\varepsilon}$: strain rate; b: amount of Burgers vector; ρ: dislocation density; v: dislocation velocity.) It is of little effect on the result whether the dislocations only glide or also climb; we leave open, for now, what the tensor character of $\dot{\varepsilon}$ may be.

The only interaction between dislocations is that between parallel edges moving in opposite directions which, in consort with Eq. 1, leads to a steady-state dislocation density

$$\rho \propto (\sigma/\mu b)^2 \tag{2}$$

(σ: applied stress; μ: shear modulus). The proportionality factor depends on the details of the dislocation arrangement and on elastic anisotropy[13]; it is of order 10.

The dislocation velocity is, by assumption, linked to a diffusive mobility:

$$v \propto (D/kT)\,(\sigma b)\,b \tag{3}$$

(D: diffusion coefficient; kT: temperature in energy units.) σb is the driving force per unit length of dislocation, and the extra b is the unit advance distance. The relation (3) may further depend on the detailed assumptions made about the solute dislocation interaction; it is an old argument whether there should be an extra term proportional to $(kT/\mu b^3)^2/c$ (c: concentration), which would invert the explicit temperature dependence in Eq. 3.[4,8,9,11]

Combination of Eqs. 1, 2, and 3 gives the kinetic law

$$\dot{\varepsilon} = A \left(\frac{\sigma}{\mu}\right)^3 \frac{\mu b^3}{kT} \exp\left(-\frac{Q_D}{kT}\right) \tag{4}$$

where Q_D is the activation enthalpy for self-diffusion of the diffusing species (which is 131 kJ/mol for Al-Mg.[14]) Our constant A contains the pre-exponential diffusion constant, which is typically, and in the present case[14], not known to an accuracy better than 20%. It is for this reason that we elect to have all other constants lumped into the single adjustable parameter A. The two other constants that enter are from the relation between steady-state dislocation density and (shear) stress (Eq. 2, of order 10), and the relation between local (resolved) stress and strain rate on the one hand and the macroscopic values σ and $\dot{\varepsilon}$ on the other (Eqs. 1, 2, and 3, of order 1/10). Thus, A, while adjustable, should be of the order of $D_0/b^2 \approx 10^{15}\,s^{-1}$. The only significant question is: Is it constant at all, or does it depend on temperature and concentration?

2.2 Viscous Deformation in Al-Mg

There have been many investigations of this case[10,12,16-20]. Figure 1 presents new data for two Al-Mg alloys, 2.3 and 5.2 at%, both prepared in the same way by ALCAN. They are essentially pure binary alloys, with the addition of 0.007 mol% of TiB_2, which is in a particulate dispersion at the grain boundaries. The grain size was about 100μm, the samples were 12.7 mm in diameter, 17.8 mm high.[2] The data are plotted as suggested by Eq. 4. The principal point of a kinetic law is that it combines the temperature and strain-rate dependence of the stress into a single variable: all points for different T and $\dot{\varepsilon}$ combinations for one material must fall on a single curve. This variable is, for the present case, essentially the Zener-Holomon parameter, which is plotted (with a small correction according to Eq. 4) logarithmically along the abscissa.

We first notice that indeed the points for different combinations of T and $\dot{\varepsilon}$ overlap and form a single curve. This attests primarily to the fact that the activation energy of diffusion, as independently measured, not adjustable, serves to unify the data. Second, we emphasize that the line is straight over most of the regime plotted, and a stress exponent of 3 fits all data, except at the highest stresses, very well. This is the most non-negotiable aspect of Eq. 4: the stress exponent must be 3. Minor, but nevertheless crucial aspects of the plot in Fig. 1 are that we did take account of the temperature dependence of μ – which is essential for high-temperature deformation even more than for low temperatures: it has helped bring about the good match between experiment and theory.[7] As the diffusion coefficient, it can be independently measured. We have taken for μ the relevant combination[13] for pure aluminum[21], in the assumption that at least the temperature dependence should be similar.

Finally, the abscissa incorporates the explicit temperature dependence of the pre-exponential in Eq. 4 and this, too, helped the fit. Some models in which the dislocation is presumed to drag a cloud of solutes behind it[4,9,11] arrive at a pre-exponential term that depends on temperature in a different way. To test the sensitivity of the plot to this difference, Fig. 2 shows the data for the 2% alloy plotted both ways: it serves as a nice illustration of how the data do *not* fall on a single curve when you do not have the right law. In addition, a "fit" through the "wrong" plot would give an exponent noticeably higher than 3. We must thus conclude that this type of *drag* model is not applicable to the present case, and that the simple *drift* model[4] of Eq. 4 (which has also widely been favored in experimental overviews[8]) is in fact better. This assertion is confirmed by the observation that the two compositions appear to fall on the *same* curve in Fig. 1. While we did find a 6% stress difference between our two alloys in one direct comparison under identical conditions, this falls far short of the dependence claimed by the drag models. (The horizontal distance between the curves would have to be 0.4 in the units of Fig. 1).

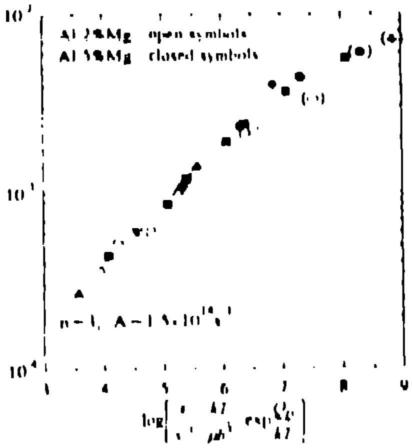

Fig. 1 - Steady state flow stress plotted according to the simple solute drift law.

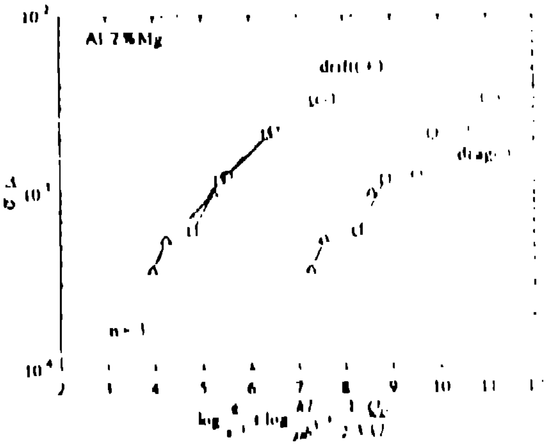

Fig. 2 – Comparison of the drift and drag laws.

2.3 Dynamic Recrystallization in Al-Mg

And now comes the surprise: every one of the specimens shown in Fig. 1 showed clear signs of dynamic recrystallization: at the higher stresses, partial, at the lower stresses, full. (The one in brackets is marginal.) This has been documented by both microstructure and texture observations, and by the difference to the behavior in static annealing after deformation. Humphreys and coworkers[20] had observed the partial recrystallization as serrated grain boundaries, and they saw what they termed progressive lattice rotations rather than "conventional" dynamic recrystallization.

The most striking effect of dynamic recrystallization in Al-Mg is the development of a cube fiber texture: partial at the higher stresses (marginal at the highest); full, and only, at the lower ones.[2] We do not believe that this texture could be a consequence of deformation alone. We had expected a preference for (100) in tension, on the basis of the single crystal experiments by Otsuka and Horiuchi[22]: we have indeed observed this. But it is inconceivable that the same mechanism of glide (or glide-and-climb) can lead to the same texture under opposite signs of deformation.

Another usual aspect of dynamic recrystallization is oscillations in the stress strain curves. Close scrutiny in fact reveals mild oscillations, at least in some cases. Figure 3 shows some of the stress strain curves (even at the lower temperatures, which we will discuss later) for Al-2%Mg. Look at the curve just below the one with the (o) sign: the clear, more-or-less periodic oscillations in the superimposed smoothed curve may well be caused by periodic deviations in a process of dynamic recrystallization that is not really steady.

If the dynamic recrystallization had been observed first, one might well have surmised that the steady-state deformation was a consequence of a balance between hardening and softening mechanisms. The kinetics would have involved some diffusion coefficient (although presumably not the same one); but it stretches credulity to perceive that the complete and quantitative fit with Eq. 4 (with a single adjustable constant that does come out in the right magnitude) is a complete accident: that its real cause is dynamic recrystallization. The conclusion we see ourselves driven to is that dynamic recrystallization, at least in this case, is an ancillary effect; it does not *cause* the stress/strain kinetics. Incidentally, dynamic recrystallization did also occur in the samples at the highest stresses displayed in Fig. 1, even though Eq. 4 no longer holds.

One more feature of Fig. 3 deserved mention: some of the curves (all direct traces) are smooth, some are jerky. At the high stresses, the jerky curves belong to the dynamic strain-aging regime, surrounded by a smooth one above and one or two smooth ones below. But why are the low-stress curves jerky? We cannot rule out a machine effect (due to imperfect servo-control), but it seems unlikely, since some curves are indeed smooth. Is there some jerky aspect to viscous deformation – or to this particular kind of dynamic recrystallization?

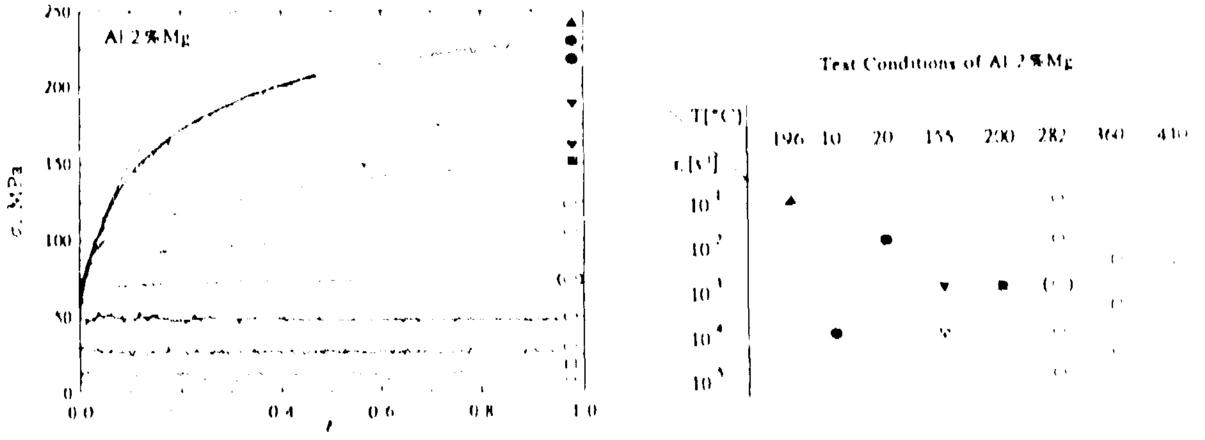

Fig. 3 – Most of the original traces, converted to stress strain, for Al-2%Mg

3. THE STRAIN HARDENING REGIME

There is another part of the mechanical response of single-phase polycrystalline cubic metals for which a well-founded and verified constitutive relation exists. Its defining feature[23] is that the initial hardening rate is a constant, small fraction of the shear modulus (associated with a geometrical-statistical storage of some of the moving dislocations), followed by dynamic recovery (the thermally activated rearrangement of all the previously stored dislocations). This regime is normally identified with low-temperature behavior; the reason we review it here is that it holds to rather high (or "intermediate") temperatures.[24] We will discuss it in some detail for pure Cu, but first use the insight gained there to describe the high-stress regime in Al-2%Mg.

3.1 Strain Hardening and Dynamic Strain Aging in Al-Mg

Figure 3 exhibits the feature of a constant initial strain hardening rate clearly for all the higher-stress curves. (This initial value of $d\sigma/d\varepsilon$ is 0.06μ for the curves that do not exhibit dynamic strain aging, $0.08\ \mu$ for those that do.[25]) There is just one curve, (o), that has a finite but different initial hardening rate; it is the same curve that did not fit within the viscous deformation model and was marginal in the dynamic recrystallization aspects.

The rate and temperature dependence in this regime is still due to thermal activation, but now at discrete points. The activation energy is not related to diffusion, and it is inherently stress dependent. If we call quantify this stress dependence of ΔG by a normalized function $g(\sigma/\mu)$, the kinetic law is:

$$\dot{\varepsilon} = \dot{\varepsilon}_0 \exp\left[- \frac{\mu b^3}{kT}\, g\left(\frac{\sigma}{\mu}\right)\right] \qquad (5)$$

where $\dot{\varepsilon}_0$ may safely be assumed constant[13]: it takes the place of Q_D in the Zener-Holomon paramter inasmuch as it is chosen such as to unify the rate and temperature dependencies. It is again important to incorporate the T-dependence of μ, this time also in the temperature scale[13,3]. Figure 4 is a plot based on this equation: the actual observed curve describes the stress dependence of the activation energy, which is not usually known with sufficient confidence.

Figure 4 shows the yield stress σ_Y for the hardening regime in Al-2%Mg, with the characteristic hump in the dynamic strain-aging range. At the lower stresses, in the no-hardening regime, σ_Y is the same as the steady-state flow stress. The points there are the same as those in Figs. 1 and 2 (open symbols): they demonstrate again that the $(T,\dot{\varepsilon})$-dependence is *not* unified if you have the wrong plot. The stresses labeled σ_S scale the temperature and rate dependence of dynamic recovery processes (rather than yield); they are obtained in a way that will be described in detail below. It is seen that both the σ_Y and σ_S values fall on a single curve for all the different $(T,\dot{\varepsilon})$-combinations in the hardening regime: equation (5) describes the data well, with $\dot{\varepsilon}_0 = 10^7 s^{-1}$, which is a reasonable value according to the dislocation mechanisms.[13,26]

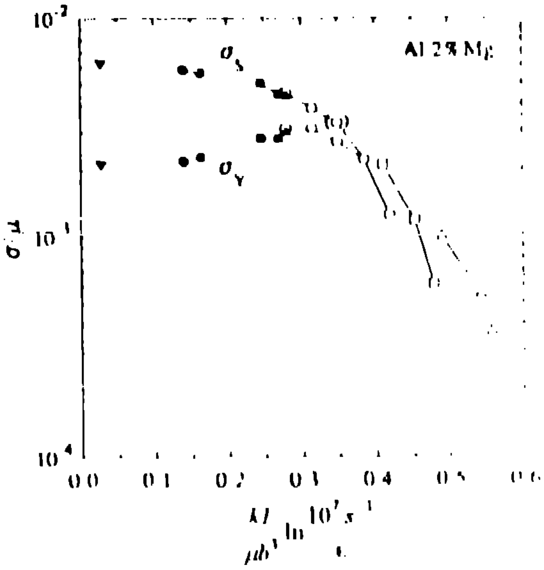

Fig. 4 – Plot under the assumption of a stress dependent activation energy (on the abscissa) and constant $\dot{\varepsilon}_0$

In summary, the behavior of Al-Mg alloys can be described in terms of one or the other of two well-known mechanisms over the entire stress range down to $10^{-4}\mu$. The single curve that does not fall precisely in either regime (at 282°C = 0.6 T_m, and $10^{-3}s^{-1}$) does not warrant a separate mechanism, such as a power law with adjustable n and Q; in fact, it fits well enough in the hardening type.

3.2 Strain Hardening and Dynamic Recovery in Pure Copper

Figure 5 shows a set of stress strain curves from compression experiments on OFE copper (grain size ≈100μm) for a range of temperatures at a strain rate of $10^{-3}s^{-1}$; other strain rates, from 10^{-4} to 1 s^{-1} were also investigated.[1] The only curve in Fig. 5 for which the initial hardening rate is not the same is the absolutely lowest one, at 700°C. (The ones, not shown, at 700°C but higher strain rates do fit with others.) For some of the high-temperature curves, one might not have discovered this constant initial hardening rate if one did not have the backdrop of the low-temperature behavior; but with this experience, the high-temperature data can be honestly analyzed in terms of the same initial hardening rate. One may use this as a *definition* of the "yield stress" at high temperatures and comes out with pleasing results.

The other common feature of this regime is that the hardening rate decreases *linearly* with stress, at least initially. If this were true for the entire range of stresses, it would correspond to the Voce law. At hardening rates less than about half the initial value, clear deviations from the Voce law have been frequently observed[27]; their description and interpretation is still being widely investigated, and is not of concern here. What is of interest is that the *rate* of initial decay of the strain-hardening rate depends sensitively on temperature and strain rate. One may quantify this dependence in terms of a "scaling stress" σ_S: the saturation stress that would be reached if the Voce law held throughout the range. It is a measure of the rate of "dynamic recovery", and the quantity we plotted in Fig. 4.

In interpreting these strain-hardening data for polycrystalline materials, one should take account of the changing texture, which influences the geometric relation between the observed macroscopic behavior and the "physical" response of the representative grain. For copper, we have done so by postulating a certain (differential) hardening law in terms of resolved shear stress vs. shear strain on all slip sytems in all grains, calculating the macroscopic curve on the basis of the Los Alamos polycrystal plasticity (LApp) simulation code[28], and then varying the locally assumed response until the macroscopic experiments are matched. We used measured initial textures as input to the simulation.[1]

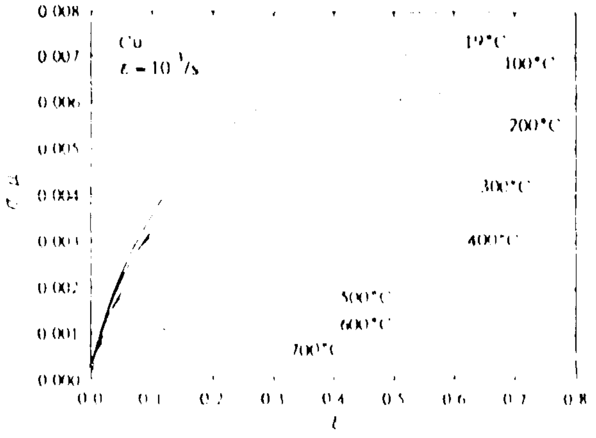

Fig. 5 – Normalized stress strain curves for pure Cu polycrystals.

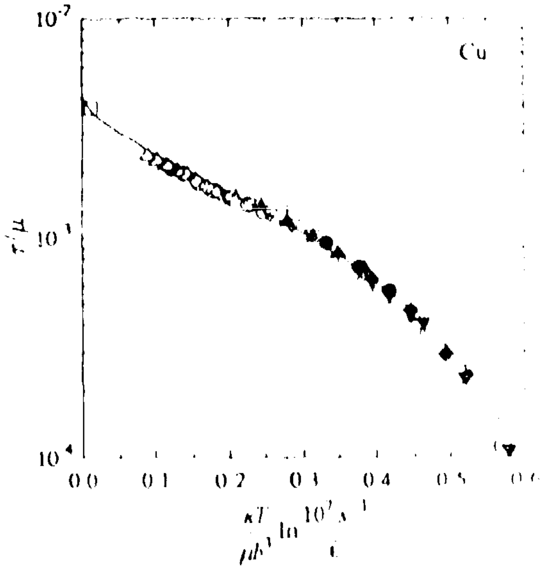

Fig. 6 (right) – Scaling stress for dynamic recovery, with semi-phenomenological fit (line). Also: steady state flow stresses in dynamic recrystallization.

With a shear hardening rate was 0.0045 μ, the resulting values of the Voce stress τ_V (the equivalent of σ_S above, but in shear, and in excess of the yield stress), normalized by μ, is plotted logarithmically as the ordinate in Fig. 6. For the abscissa, we again plot the normalized activation energy, assuming a constant value of $\dot{\varepsilon}_0 = 10^7 s^{-1}$. It is seen that this achieves the desired goal of putting all combinations of temperature and strain rate on the same curve. The point in brackets, at the lowest stress, is the one for which evaluation of τ_V was not straight-forward because the initial hardening slope was clearly less than the standard for this regime. Even disregarding this point, the highest temperature for which the single constitutive law holds is still 700°C, or 0.65 of the melting temperature T_m, at a strain rate of $10^{-2}s^{-1}$ or higher.

If the curve in Fig. 6 were a straight line, one could describe the relation as a power law, albeit with an exponent that continuously varies[29] in proportion to $\mu b^3/kT$. But the line is not straight even for the part of the range in which there are many overlapping points (thus confirming the utility of the chosen value of $\dot{\varepsilon}_0$). The variation in fact expresses the stress dependence of the activation energy ΔG. There is no reason to expect the line to be straight: it would correspond to an exact proportionality of ΔG to $\log(\tau_V)$ over the entire range, for which there is no physical basis.[30]

A semi-phenomenological expression for the activation energy of glide (not dynamic recovery) has been proposed previously and successfully used in many applications.[15,31] We recently discussed that the dislocation processes in dynamic recovery may well be closely linked to those that control the flow stress, since they are essentially delayed forward flow.[28] Using the same type of expression, we have found that the curve in Fig. 6 is well represented by the superimposed line,

$$\Delta G = kT \ln\left[\frac{10^7 s^{-1}}{\dot{\varepsilon}}\right] = 0.66\ \mu b^3 \left[1 - \left(\frac{\tau_V}{0.014\ \mu}\right)^{3/4}\right]^2 \qquad (6)$$

where the numerical values hold for Cu, but the exponents may be more general.

3.3 Dynamic Recrystallization in Cu

As is seen by the oscillations in Fig. 5, dynamic recrystallization occurred at all the lower stresses. A quantitative description of the oscillations has eluded us so far[32]; but the steady-state stress that they approach exhibits an interesting behavior. We have determined it directly from the experiments, then applied the same ratio of τ/σ that had proven successful for τ_V, and plotted the resulting τ_f/μ in Fig. 6 (symbols +,x,*,#). They follow the same kind of law as τ_V for most of the regime. (The first +, which is noticeably higher than τ_V, corresponds to the only single-peak curve in the lot.)

This means that certainly the appearance of dynamic recrystallization is not controlled by the diffusion coefficient, and should therefore not be unified according to the Zener-Holomon parameter. We first assumed[32] that a stress dependent activation energy must be controlling dynamic recrystallization, as it controls dynamic recovery. The interesting alternative, in view of the above results in Al-Mg alloys, is that dynamic recrystallization is, even in Cu, an ancillary effect, not an independent mechanism that controls the mechanical response in any regime.

4. CONCLUSIONS

There are two regimes of behavior for which a quantitative constitutive relation for the kinetics of plasticity can be given that is based on reasonable physical mechanisms and does describe experiments well. One is for "hardening" materials (cubic, single-phase, and in the absence of dynamic strain aging): it holds for temperatures up to about 2/3 T_m (depending on strain rate) or, better, down to (shear) stresses of the order of $10^{-4}\mu$. The other holds for alloys that do exhibit dynamic strain aging at high stresses, but viscous deformation at low stresses. (At lower stresses yet, they may exhibit Harper-Dorn creep, with a short transition in between[15,10]). In cases where both mechanisms should hold, the transition region is quite narrow: there seems little need for any additional mechanism – especially since all these materials show dynamic recrystallization starting in just about that temperature range.

Dynamic recrystallization itself, while clearly occurring in both cases, does not seem to control the mechanical response.

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