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Large-Strain Softening of Metals at Elevated Temperatures by Deformation Texture Development

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Abstract: Many (if not a majority) of metals and alloys evince substantial softening with torsion deformation to strains not usually achievable in tension. Of course, softening has long been observed by discontinuous dynamic recrystallization (DDRX) but this paper will discuss cases where softening is associated by texture development with large-strain deformation that is not reliant on changes in the dislocation density. This paper discusses the work of the current authors on FCC metals and alloys and extends to a new discussion of BCC and HCP cases. The analysis of the basis for torsional softening in BCC steel and HCP Zr discussed here is a novel concept that has not been addressed in the literature before.

Keywords: strain-softening; texture; creep

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1. Introduction

Metals and alloys deformed in torsion (pure shear) to relatively large strains, typically greater than 1, often show softening (typically on the order of 20%). Even though observed in torsion, softening (true stress measurements) is not usually evident in tension in the absence of discontinuous dynamic recrystallization (DDRX). Similarly, this dynamic-recovery softening is not observed in compression.

Examples of metals and alloys which have been confirmed to experience this torsional softening include Al [1–5], Al-5.8Mg [6], Fe26Cr–1Mo [7], α-Zr [8,9]. The details of the large strain softening will be discussed. Figure 1 illustrates torsion experiments by one of the authors on aluminum deformed to very large strains [10].

First, Figure 1 illustrates the remarkable ductility in torsion of Al at temperatures of 0.58–0.81 T_m. This corresponds to the five power-law creep regime [11]. The ductility is less impressive at lower temperatures (ambient) and one may observe DDRX [12,13]. The high-temperature rate-controlling mechanism is dislocation climb. Subgrains form homogeneously within the grain interiors. Eventually, "pinching-off" of serrated grains occurs, leading to a refined microstructure. All the observations described above are consistent with geometric dynamic recrystallization (GDRX) accompanying dynamic recovery. There is no evidence of discontinuous dynamic recrystallization (DDRX) or grain growth. Some high angle boundaries may form from shear bands [14], but the classically defined continuous dynamic recrystallization (CDRX) [3,15,16] is not occurring. As discusses in a previous article by one of the authors, a texture develops with a B^1 (112) [110] being the principal component.

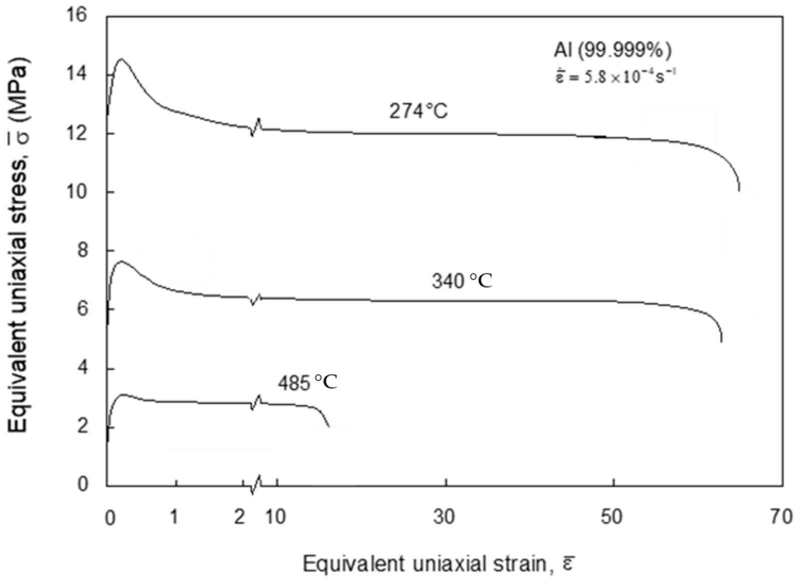

Figure 1. Equivalent uniaxial stress vs. equivalent uniaxial strain behavior of aluminum deformed in torsion at different temperatures [10].

Figure 2 illustrates the high temperature behavior of α-zirconium (HCP) deformed at elevated temperature.

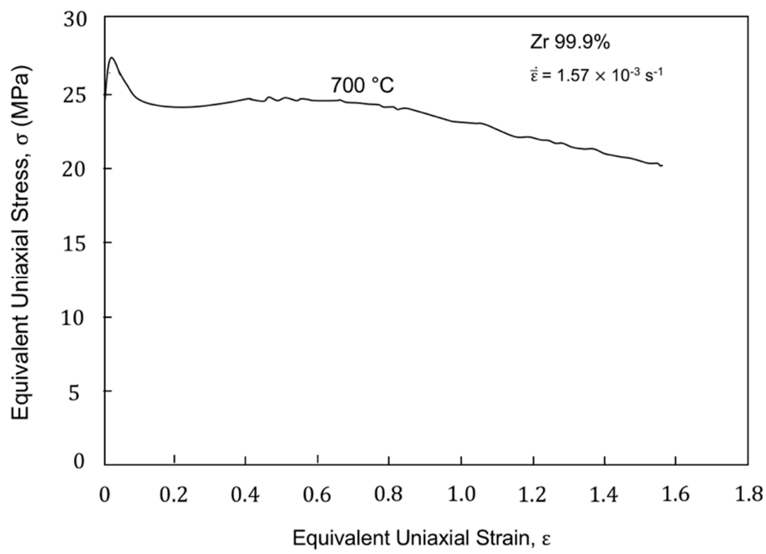

Figure 2. Equivalent uniaxial stress vs. strain behavior of α -zirconium deformed in torsion at 700 °C (0.46 T_m) [9].

The activation energy of creep in Zr is consistent with that of Zr lattice self-diffusion, considering impurity effects [17,18]. The stress exponent is close to the value expected for five-power-law creep. Thus, the creep appears to be dislocation climb-controlled. Optical and transmission electron microscopy (TEM) revealed the formation of grain boundary serrations at the early stages of deformation just as in Al. As the strain increases, grains elongate and subgrain formation is increasingly apparent. Subgrains form homogeneously within the grain interiors. Convergent beam electron microscopy revealed that the mis-orientations of the subgrains formed at lower strains are smaller than 3 degrees and do not increase with higher strains. Eventually, "pinching-off" of serrated grains occurs, leading to a refined microstructure. X-ray analysis revealed the occurrence of noticeable texture evolution during deformation. In particular, fiber texture forms, which is indicative of slip occurring mainly on prism and pyramidal planes. This texture change appears responsible for the modest softening observed during the high temperature tests. All the observations described above are consistent with geometric dynamic recrystallization (GDRX) accompanying dynamic recovery. There is no evidence of discontinuous dynamic recrystallization (DDRX) or grain growth. Again, some high angle boundaries may form from shear bands [14], but the classically defined continuous dynamic recrystallization (CDRX) [3,15,16] is not occurring.

There is a case where dislocation climb is not the rate controlling mechanism for plasticity and that is illustrated in Figure 3 for Al-5.8%Mg [6] deformed at 425 °C and an intermediate strain rate. Viscous glide is the rate-controlling mechanism for the high temperature plasticity where the steady-state exponent is about three (rather than 4.5 for pure Al). Nonetheless, significant softening is observed in the absence of any DDRX [6].

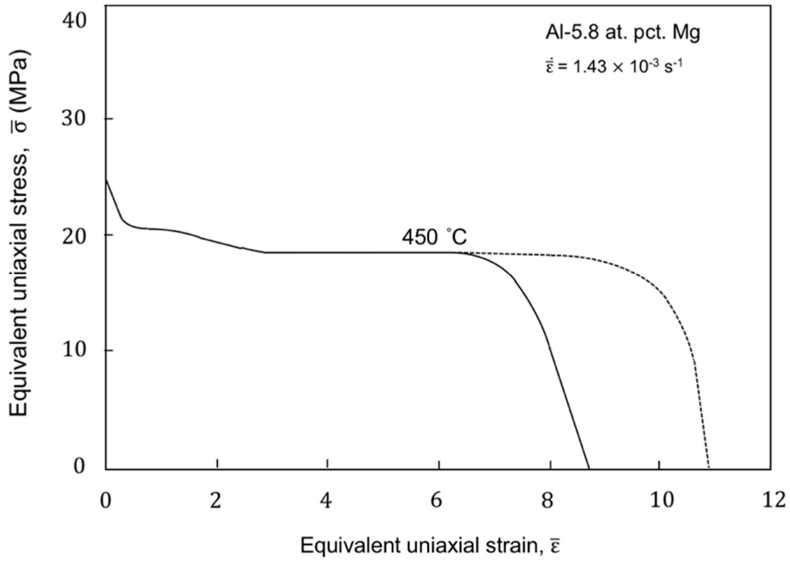

Figure 3. Softening observed in Al-5.8 at % Mg deformed to large strains in the solute drag regime. the solid curve represents the behavior of an L/d = 10 specimen tested using the Rheometrics Spectrometer. The dashed line represents the increased ductility of the L/d = 5 specimens tested on the Stanford torsion machine [6].

Figure 4 illustrates the lower temperature torsional deformation of pure silver [19]. At ambient temperature, there appears to be a mechanical steady-state. TEM did not reveal any evidence of DDRX. Therefore, at ambient temperature, a genuine steady-state that is a balance between hardening and dynamic recovery appears to occur. However, at slightly higher temperatures, there is very noticeable softening with large-strain deformation. TEM revealed the presence of what appeared to be recrystallized grains. So, in this case, softening appears to be due to DDRX.

Figure 5 illustrates the torsion stress versus strain behavior of a 26Cr-1Mo ferritic stainless steel [7]. This BCC alloy was assessed at this temperature and strain-rate to deform by dislocation climb within the five power-law creep regime. A dramatic increase in HAB area was observed which was suggested by the authors of this work to be a consequence of CDRX, although the GDRX mechanism was not considered as it had not yet been formulated in the literature at the time of the ferritic steel study. The authors did not observe the development of a texture which may be an error.

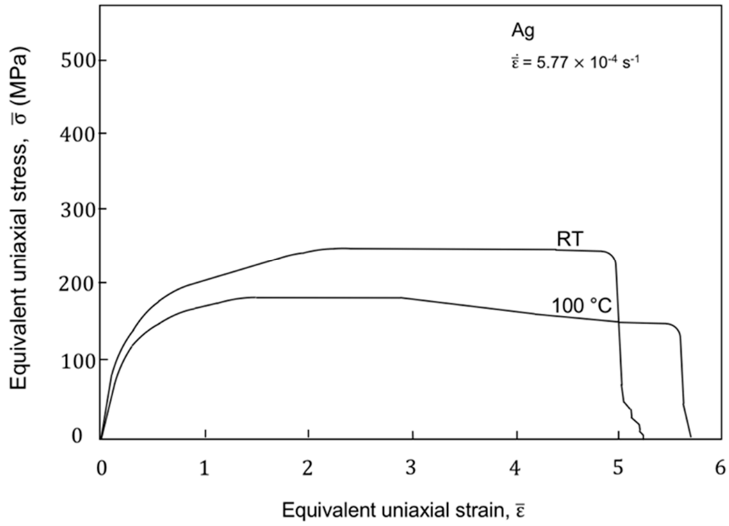

Figure 4. The large strain torsional deformation of pure silver at low temperatures [19].

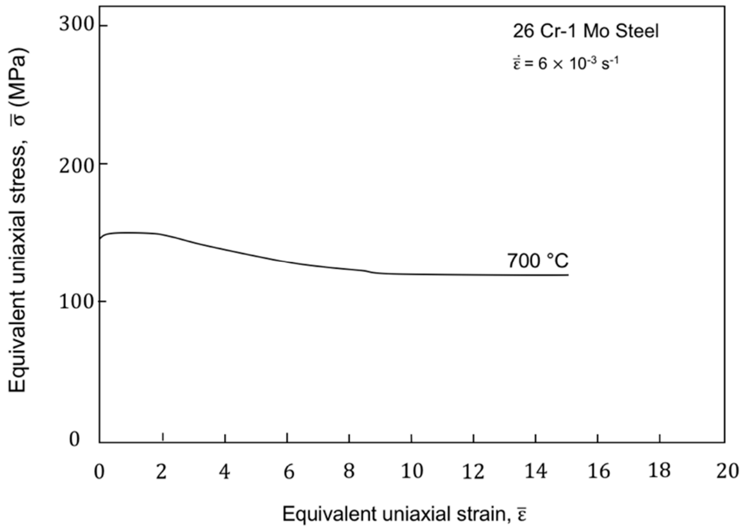

Figure 5. The torsion stress versus strain behavior of a 26Cr-1Mo ferritic stainless steel [7] at 700 °C (approximately 0.55 T_m).

2. Discussion

Therefore, the above four cases suggest that the large-strain softening can be a consequence DDRX, or it can occur during dynamic recovery. Furthermore, for the case of dynamic recovery, softening can occur for different deformation mechanisms, i.e., dislocation climb control or dislocation glide control. Aluminum is the most studied case, and the explanation for the softening has fallen in two basic categories, texture development [3,9,13,15,20] and changes in the hardening microstructure [7,15,16,21] or both [22]. The traditional texture argument is that a decrease in the Taylor factor is observed with the development of the deformation texture. This, it has been argued, leads to the softening. This explanation suggests that the rate controlling process for plasticity is glide. While this may be true for the Al-5.8 at% Mg by viscous drag of dislocations by Mg solute, it is not the case for pure Al, 26Cr-1Mo ferritic stainless steel and, perhaps, also zirconium, where the rate controlling process is dislocation climb.

A recent article by the authors [10] and other work [1] suggested that changes in the microstructure may not be responsible for the softening. It was suggested that the softening is due to the development of a texture as others have also suggested [3,22]. However, in the past, the observed texture softening proposition was based on dislocation glide [3,20,22,23].

With deformation, the Taylor factor (M) developed in torsion [3–5,10,20] led to a calculation of a decrease in the resolved shear stress of about 18%, consistent with the observed softening. However, for the temperatures at which glide was presumed, the controlling process for creep plasticity is dislocation climb. Thus, the change in climb stress is relevant. Climb-control predicts softening in torsion for Al. If B^1 is the primary texture as suggested by McQueen [24] then softening of 11% is predicted based on climb while the Taylor factor (glide) decreases by about 15%. The climb analysis predicts that the elevated temperature flow stress in compression (immediately after torsion) should be approximately unchanged just as observed in the earlier tests. This finding contrasts the predictions of Taylor factor (dislocation glide) calculations that predict a 10% increase in the flow stress which was clearly not observed. Thus, our earlier experimental results appear consistent with the prediction that the texture leads to a higher climb stress, and this leads to the elevated temperature softening. It is important to point out here that for climb to explain softening, the climb stress must increase with the development of a texture. Dislocation glide occurs even if climb rationalizes the softening. A decrease in the Taylor factor (or Schmid factor) for dislocation glide is neither a necessary nor sufficient condition for softening. The texture may lead to a reduction of the shear necessary for glide, but climb may still be the explanation to softening.

For the BCC steel in Figure 5, the drop in the shear stress is about 16%. The resolved shear stress on slip dislocations from a random orientation of the Burgers vectors and shear planes to the {211}<111> texture drops about 50% assuming no starting texture and a perfect deformation texture after deformation. In this work, we calculated the change in the climb stress based on a {211}<111> torsional texture for BCC steel [24]. From the procedure described in [10] there is an increase in the climb stress of 0.88/0.71 = 1.24 or a predicted softening of 24%, which is sufficient to rationalize the softening. The texture development assists dislocation climb.

For HCP Zr, new analysis suggests that there is no change in the climb stress from an initial texture in the torsion specimens with the $(10\bar{1}0)$ in the shear plane with randomly oriented <0001> to the expected texture of $(10\bar{1}0)$ <0001> [25,26]. This is not in line with the proposition that climb is responsible for the softening. For the case of Zr, some [27] have suggested that the rate-controlling process for creep is a glide-controlled mechanism although the present authors have suggested climb in an earlier article [17,18]. This work may suggest support for dislocation glide as the rate-controlling mechanism for plasticity in HCP Zr. The $(10\bar{1}0)$ <0001> leads to the maximum shear in the $(10\bar{1}0)$ with a Burgers vector in the shear plane. The resolved shear is the highest possible [25]. Thus, with deformation, the specimens move from 0.71 of a perfect shear to a perfect shear. Therefore, we expect an increase in the average resolved shear stress with texture development perhaps leading to the observed softening according to a glide mechanism.

3. Conclusions

There has been evidence that the torsional strain softening observed in aluminum that is not a consequence of discontinuous dynamic recrystallization but rather dynamic recovery, is a consequence of the deformation texture leading to higher dislocation climb stresses rather than increases in the dislocation glide stress. The current results for steel (BCC) support the same contention. The zirconium (HCP) analysis is not supportive of climb as the basis of the softening. Torsion followed by compression testing as was performed on Zr may provide addition insights into the contributions of dislocation glide versus climb in this case.

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