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AUTHOR(S): M. Stout, P. Martin, D. Helling, and G. Canova

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Los Alamos

Los Alamos National Laboratory
Los Alamos, New Mexico 87545

MULTIAXIAL YIELD BEHAVIOR OF 1100 ALUMINUM FOLLOWING VARIOUS MAGNITUDES OF PRESTRAIN

M. G. Stout and P. L. Martin

Materials Science and Technology Division
Los Alamos National Laboratory
Los Alamos, NM 87545

D. E. Helling

Department of Material Science and Engineering
Stanford University
Stanford, CA 94305

G. R. Canova

Faculte' des Sciences
LPMM
Ile du Saulcy
F-57045 Metz Cedex
France

ABSTRACT

The multiaxial yield and flow behavior of metals has been of interest for many years. Recently, the experimental work of Phillips and Lee [1979], Shiratory et al. [1979] and Ohashi [1982] has been quite notable in this field. These authors have concentrated their efforts in measuring yield loci after small to moderate prestrains (≤ 0.06). In this paper we discuss small strain yield loci we have measured after prestrains between 0.03 and 0.05 in torsion. These experiments on 1100 aluminum are in general agreement with the literature. They show a translation, distortion and expansion of the yield loci. A rounded nose forms in the direction of prestrain with the yield locus flattening opposite the prestrain. We observed that the distortions change to match the strain direction after very small reversals in prestrain.

The subsequent yield locus has also been measured after a large torsional prestrain of $\gamma = 0.5$. Using a 5×10^{-6} offset criterion for yielding, the shape, distortion and translation of the yield locus was very similar to that found after the smaller prestrains. In addition a large-strain yield locus, using a back extrapolation technique, was determined for the same sample. This yield locus exhibited close to von Mises isotropic expansion. The observed deviations, while slight are extremely important. They match those predicted by a polycrystal slip model. Thus, the small-strain yield locus, after a large

prestrain, appears to be determined largely from dislocation considerations only, where as the large-strain yield locus is determined by the developing texture. Finally, aluminum sheet was deformed by rolling to larger prestrains $\varepsilon_{\text{von Mises}} = 0.5, 1.0, 1.5, 2.0$ and 2.5 and subsequently tested in plane strain compression. Two types of compression experiments were done, one such that there was no deformation mode change from rolling, the other rotating the direction of zero strain by 90° producing a stress path change. The large strain yield and flow behavior of these experiments was again predicted using the relaxed constraint polycrystal model of Kocks and Canova [1981]. For these very large prestrains the experiments and texture theory differ. Microstructural observations have shown the presence of micro-shear bands which resulted from the rolling prestrain. We speculate that these features are responsible for the deviation from crystal plasticity theory.

We believe that this work points to several operative mechanisms of deformation. Small-strain yielding (5×10^{-6}) appears to be controlled purely by dislocation mechanisms and interactions even after relatively large prestrains. Large-strain yielding, on the other hand, is controlled by texture after moderate prestrains (at least to $\gamma = 0.5$). After larger prestrains, obtained by rolling, the experiments deviate from texture based predictions. This is possibly the result of microstructural deformation mechanisms, for example micro-shear bands, playing a role in the deformation process.

INTRODUCTION

Investigations of the yield and flow behavior of metals have been typically made over very small ranges of prestrain by individual researchers. Most yield locus measurements are from annealed material or material which has been subject to only small prestrain, typically ≤ 0.10. Stress state effects on flow behavior have been investigated by many authors from yield until sample instability, generally at strains ≤ 0.40, and other scientists have studied the behavior of materials to very large strains. Generally, in these cases, the initial yield behavior is not of concern. The intent of this paper was to select one material, 1100 aluminum, and systematically study its behavior from yield to large strains, > 1.0. The development of microstructure and texture and their effects on yield and flow behavior were of particular interest.

Researchers in the field of metal plasticity have concentrated on the yield and flow behavior of metals for many years. Interestingly, difficulties in definition, for example the definition of the yield point, which were important in the past, are of current concern as well. Some of the experimental papers to date have defined yielding by a particular offset from proportional behavior and have concentrated on measuring very small offset yield loci e.g., Williams and Svensson [1970], Shiratori, Ikegami, Yoshida, Kaneko, and Koike [1976], Shiratori, Ikegami and Yoshida [1979] and Ohashi [1982]. Phillips and his coworkers have chosen to use the "first deviation from proportional behavior" for their definition of yielding: Phillips, Liu, and Justusson [1972], Phillips and Tang [1972], Phillips, Tang, Ricciuti [1974], and Phillips and Lee [1979]. There is a close relationship between the results from these methods as long as a very small offset definition of yielding is used, typically $\varepsilon < 20 \times 10^{-6}$. If a larger offset strain is used the variance of the results generated from these two techniques can be substantial.

The effects of a proportional prestressing on the subsequent yield locus has been investigated quite thoroughly by the authors cited. They have found that a small-strain yield locus ($\varepsilon < 20 \times 10^{-6}$) translates in the direction of the prestress, often to the extent that the origin no longer lies within the measured locus. In addition to the translation there is also an expansion and distortion. A rounded nose forms in the direction of the prestress while opposite the prestress the locus flattens. The generally observed distortion (change in shape) of the yeld locus cannot be described only by the conventional concepts of isotropic and kinematic hardening.

These investigations have been conducted at prestresses which results in strains generally less than 0.06. How might larger prestrains, ones which develop a preferred texture, affect subsequent yield behavior? And, would there be a large difference between a yield locus defined by a small offset from proportionality, (say $\varepsilon < 20 \times 10^{-6}$) and that defined by a large offset (say $\varepsilon \simeq 2000 \times 10^{-6}$. Althoff and Wincierz [1972] have measured the yield loci of heavily textured aluminum and copper tubes using a large offset strain definition of yield (500×10^{-6}, 1000×10^{-6} and 2000×10^{-6}). They compared these measurements to predictions made assuming an ideal texture using crystallographic considerations of both Schmid, and Bishop and Hill and found good agreement. Those experiments were largely with annealed samples, ones which did not have a strongly developed dislocation substructure along with the preferred texture. The effects of dislocation substructure on the yield behavior of textured materials will be considered in the following.

After very large amounts of plane strain deformation aluminum develops small, oriented shear bands within individual grains, in addition to dislocation substructures. Hatherly [1983], Malin and Hatherly [1979], Lloyd, Butryn, and Ryvola [1982], Brown [1972] and Rohr and Hecker [1981a] among others have observed these features for a variety of metals including aluminum and aluminum alloys. The most common manner of producing shear bands is to roll sheet. The shear bands are planar features which contain to the axis of zero strain and lie at an angle of 25° to 35° to the plane of the rolled sheet. For a high stacking-fault FCC material such as aluminum the shear bands begin to appear at a von Mises equivalent strain of approximately 0.6. While microstructural features such as shear bands have been observed and studied using metallographic and electron microscopy techniques and their origins considered theoretically in a recent collection of papers by Hutchinson [1984], little has been done to quantify their possible effects on yield and flow behavior of materials. This will be the final point we will consider.

EXPERIMENTAL PROCEDURES

A. Yield Locus Measurements

The experimental material used in this investigation was 1100 (commercial purity) alumimum which had been annealed at 343°C for 1 hr. Yield locus evaluation was done using tubing precision drawn to 13.41 mm O.D., 10.85 mm I.D.. A reduced section, 6.35 mm long was ground into the tubing giving it a final O.D. in the test section of 12.11 mm and a wall thickness of 1.27 mm. The tubes were tested in combinations of tension and torsion using a MTS servo-controlled hydraulic machine described previously, (Stout, Hecker, and Bourcier [1983]). To prevent buckling in torsion a molybdenum disulfide lubricated mandrel was inserted in the tubing. The mandrel was undercut in the region of the reduced section leaving this region free of constraint. Two levels of torsional prestrain were used $\gamma = 0.02$ and $\gamma = 0.50$. We selected to use a small von Mises equivalent strain (5×10^{-6}) offset definition of yielding to evaluate the yield locus of the starting material and the material after small prestrains. With this definition it was possible to use a single sample for our locus determination. In any single probe we did not exceed an offset of 20×10^{-6}. In the case of the large prestrain we found it again possible to use a single sample. First a yield locus defined by the 5×10^{-6} offset was measured and then one was determined by a back extrapolation technique. To obtain sufficient linearity in the stress-strain behavior to back extrapolate it was necessary to reach offset strains of about 5×10^{-3}. This yield locus is essentially a large offset strain locus consistent with the engineering definition of yielding ($\epsilon = 2000 \times 10^{-6}$).

B. Flow Behavior After Large Plane-Strain Prestrains

Samples of 1100 aluminum sheet were annealed at 343° for 1 hr. and subsequently rolled to von Mises equivalent strains of 0.5, 1.0, 1.5, and 2.0. Following rolling they were tested in plane-strain compression using the knife edge technique of Watts and Ford [1955]. The knife edge compression was conducted two ways: the first such that the strain state remained the same as during rolling, (transverse) and the second with the direction of zero strain rotated 90°(longitudinal). The latter produces a path change experiment analogous to cross rolling. A molybdenum disulfide lubricant was used for these experiments which employed a standard screw driven Instron testing machine, compression load cell and utilized a strain gage extensometer between the knife edges to measure strains.

EXPERIMENTAL RESULTS

A. Yield Loci After Small Prestrains

The initial yield loci and those we have measured after small tensile and torsional prestrains are shown in Figure 1. These are based on a 5×10^{-6}

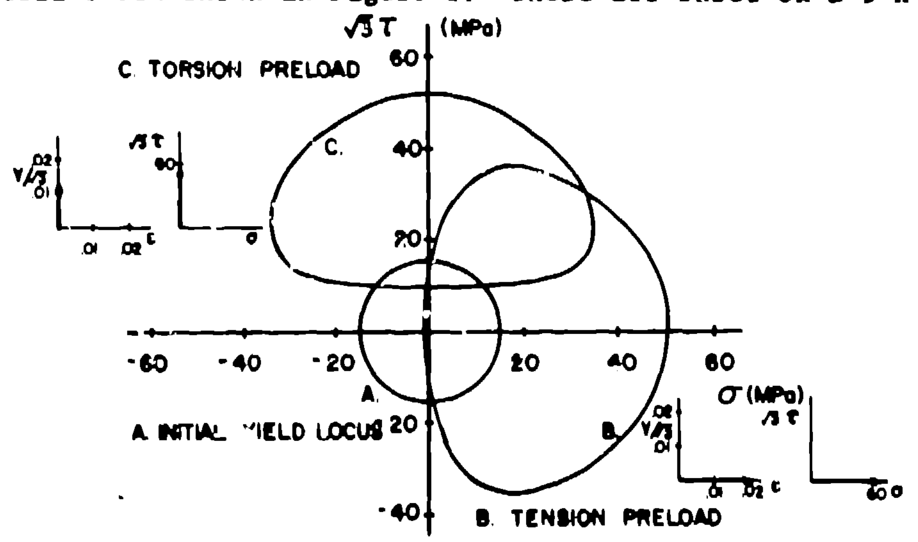


Fig. (1) - Measured yield loci: A. Initial yield locus.
B. Yield locus after a tensile prestrain of $\epsilon = 0.02$.
C. Yield locus after torsion prestrain of $\gamma = 0.02$.

von Mises equivalent offset criterion and are consistent with previous experimental results. After a prestrain the subsequent yield locus distorts, translates and expands. A rounded nose forms in the direction of the prestress while in the opposite direction the locus flattens. The locus translates in the direction of the prestress and for 1100 aluminum after torsion does not encompass the origin. Finally, there is a general expansion. It can be seen that neither classic kinematic or isotropic hardening adquately describe our observations.

Figure 2 shows two yield loci which were obtained after a torsion and reverse torsion preloading. Locus B follows a complete load reversal. It is, to within experimental error, equivalent to the locus following a simple torsional prestrain.

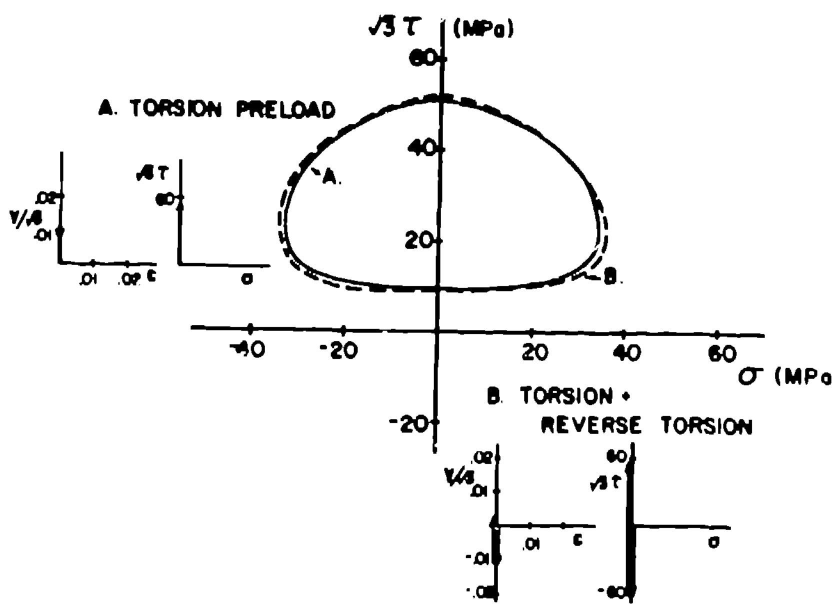


Fig. (2) - Measured yield loci showing the effects of torsion + reverse torsion. A. yield locus after a torsion prestrain of $\gamma = 0.02$. B. yield locus after a complete torsion load reversal.

Figure 2 is for a complete load reversal. In addition we have done a partial torsional load reversal, one which produced only a slight plastic strain reversal after the initial torsional preload. For the stress path and resulting strain path shown in Figure 3 an almost elliptical yield locus results. What is especially interesting is how quickly the yield locus changes shape with very little plastic strain. After an initial prestrain of $\gamma = 0.031$ a reverse plastic strain of only $\gamma = 0.005$ has completely eliminated the rounded nose produced by the first prestress leg and is beginning to "round out" the locus in the direction of current stressing.

TEM observations of Rohr and Hecker [1981b] and [1984] from 1100 aluminum annealed at 343°C for 1 hr. and strained in tension to $\epsilon = 0.05$ show regions having a well developed cell structure. In addition, there are areas which are practically dislocation free. In all cases where dislocations exist they are arranged in cells, the cell walls are diffuse and individual dislocations can

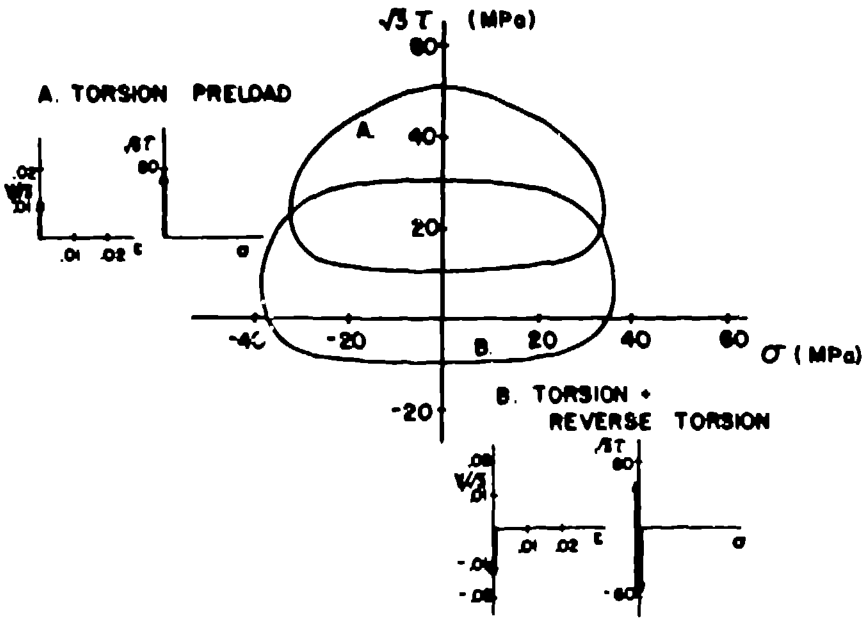
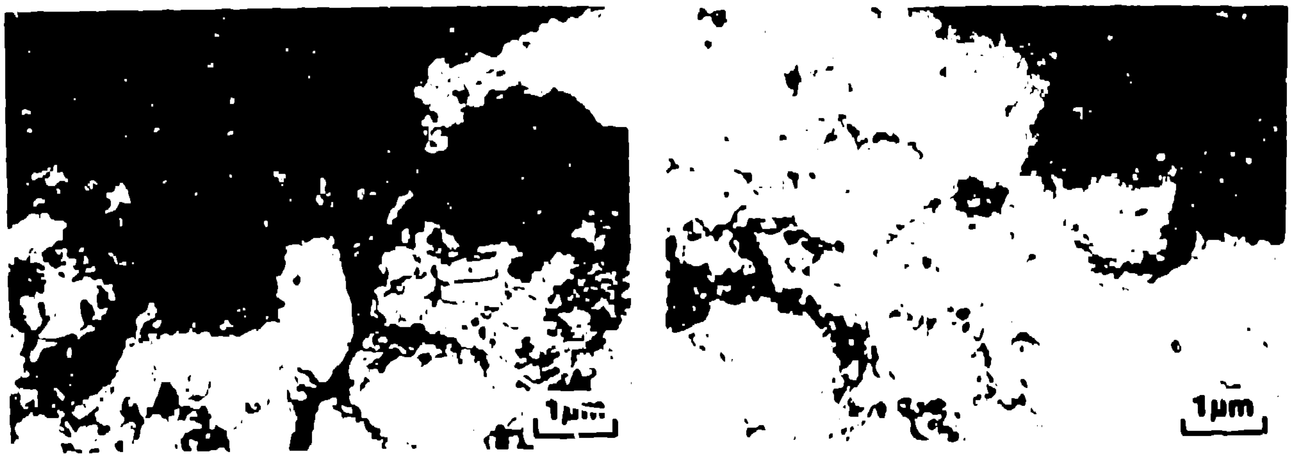


Fig. (3)- A comparison of the yield locus after a torsional prestress A. With that resulting from a slight reverse torsion prestress B.

often be resolved. Figure 4a and 4b are typical micrographs of the developing cell structure (4a) and dislocation free zones (4b). We feel that these micrographs accurately represent the structures which produce the small prestrain yield loci discussed above.



(a) (b)

Fig. (4)- TEM micrographs of 1100 aluminum annealed at 343°C for 1 hr. and subsequently strained to $\varepsilon = 0.05$; (a) bright field of developing cell structure (b) bright field of a structure free region.

B. Yield Loci After Large Prestrains

Both the small offset and back extrapolated yield loci, after a torsional prestrain of $\gamma = 0.50$, are shown in Figure 5. Also drawn is the

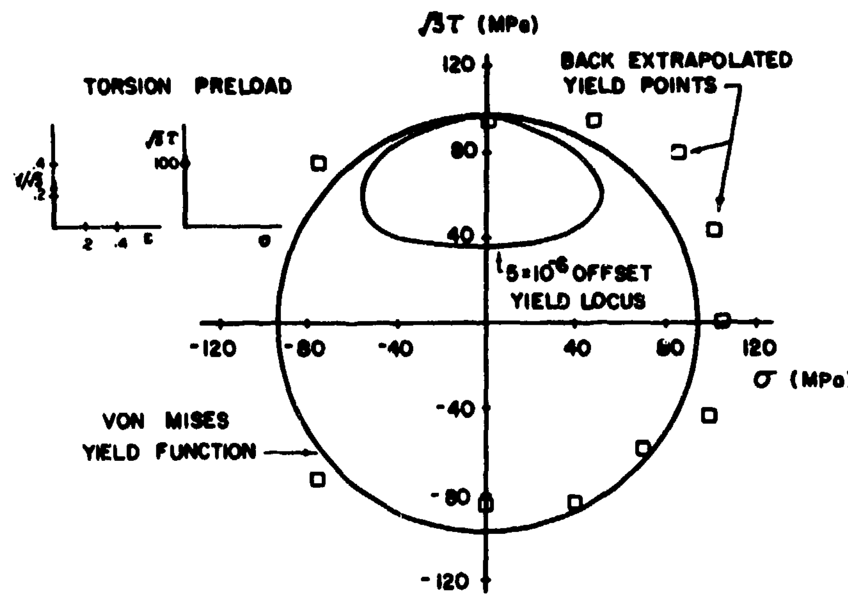


Fig. (5) - The small offset and back extrapolated yield loci after a torsional prestrain to $\gamma = 0.5$. The von Mises yield function is drawn for comparison.

von Mises yield locus to reference the back extrapolated yield points. We observe a continued expansion and translation of the small offset yield locus

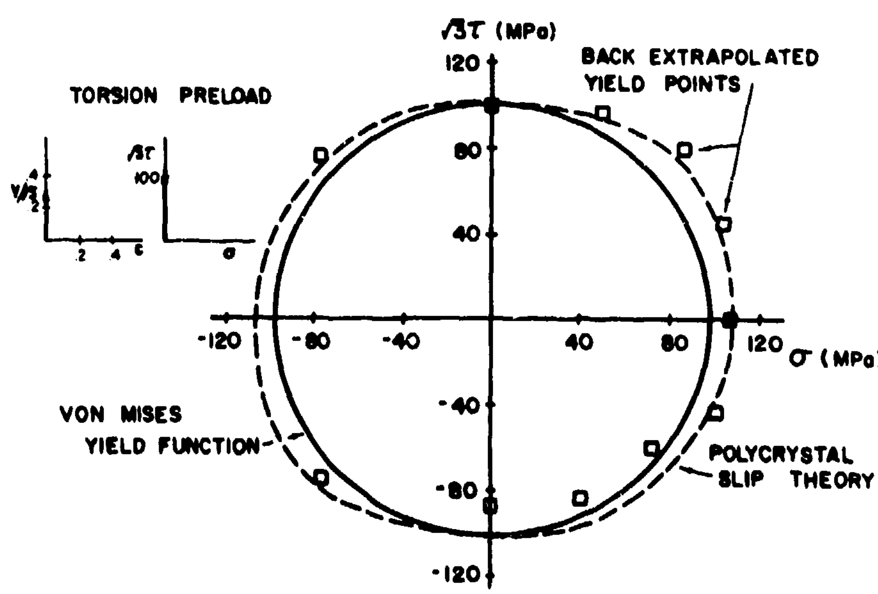


Fig. (6) - A comparsion of the yield locus, based on crystallography, after a torsional strain of $\gamma = 0.5$ compared to the von Mises yield function. The squares locate experimental mesurements.

and the distortion is of the same type as that after very small strains. Again a rounded nose forms in the direction of the prestress while in the opposite direction the surface flattens. The back extrapolated locus (large offset strains) is quite different and approximates the von Mises yield function. However it too is slightly distorted. A slight Bauschinger effect is observed, but more importantly the yield points have expanded outward in the direction of axial stress, past that predicted by von Mises.

Figure 6 shows the theoretical polycrystal slip yield locus for a torsional prestrain of $\gamma = 0.5$. The calculations assume a random initial texture and use a sampling of 300 grains. The grains are permitted to deform in the relaxed constraints manner of Kocks and Canova [1981] producing a slightly preferred orientation. The subsequent yield locus is determined from this calculated texture. It can be seen that the back extrapolation experimental points closely match the predictions of the texture model.

At larger prestrains the microstructure has continued to develop. The cell walls have sharpened and a slight misorientation of $\sim 2°$ exists between individual cells. Although dynamic recovery processes have begun the cells have not yet become subgrains in the classic sense as there is still appreciable thickness to the cell walls. A typical microstructure, after a uniaxial tensile strain of 0.31 is shown in Figure 7. On the basis of von Mises effective strain this corresponds closely to $\gamma = 0.5$.

Fig. (7) - Dislocation substructure after a uniaxial tensile strain of 0.31.

C. Flow Behavior of Rolled Sheet

Figure 8 contains the results of the plane-strain compression on rolled sheet. The data was reduced in terms of Taylor factors calculated from the relaxed constraints crystallographic analysis for tension, rolling, and rolling followed by cross-rolling. We assumed a unique microscopic hardening law independent of stress state, represented by $\tilde{\tau}$ and $\bar{\Gamma}$, which can be averaged over all grains. τ and $\bar{\Gamma}$ defined by the following relationships:

$$\tilde{\tau} = \bar{M}^g \tau_c^g / \bar{M}$$

and

$$\bar{\Gamma} = \bar{M} d\varepsilon = \Sigma d\gamma_c$$

where: The bar denotes an average over all grains, M_g is the Taylor factor for a particular grain, g, $\bar{M}$ is the average Taylor factor, τ_c^g is the critical resolved shear stress for slip and γ_c is the accummulated shear strain on each activated slip system. The axial stress and strain are related to the microscopic law through the average Taylor factor for uniaxial tension:

$$\sigma = \bar{M}\tilde{\tau}$$

and

$$d\varepsilon = (1/\bar{M}) d\bar{\Gamma}$$

Similarly, for plane-strain compression (psc)

$$\sigma^{psc} = \bar{M}_{psc} \tilde{\tau}$$

$$d\varepsilon^{psc} = (1/\bar{M}_{psc}) d\bar{\Gamma},$$

combining the equations in terms of $\tilde{\tau}$ and $\bar{\Gamma}$ we can write the following expressions for crystallographic (cry) stress and strain:

$$\sigma^{cry} = (\bar{M}/\bar{M}_{psc}) \sigma^{psc}$$

$$d\varepsilon^{cry} = (\bar{M}_{psc}/\bar{M}) d\varepsilon^{psc}.$$

As seen in Figure 8 a wide spread exists between the data from the knife edge

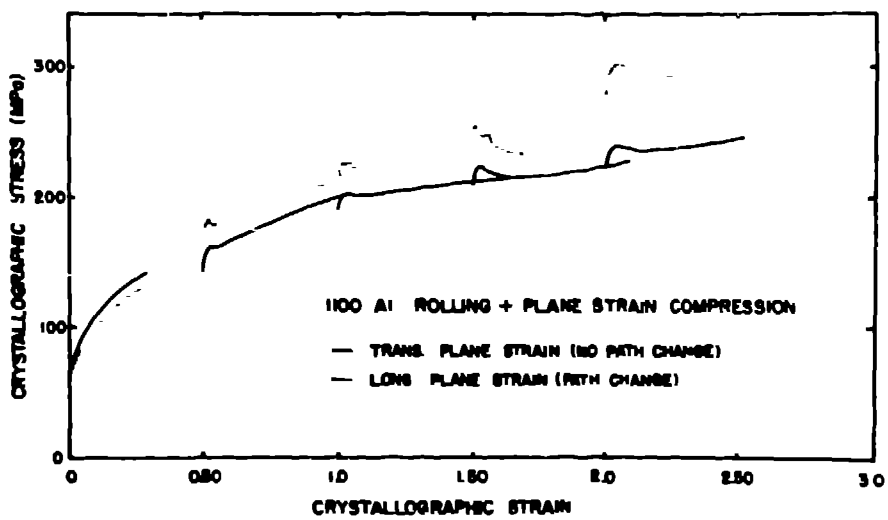


Fig. (8)- Knife edge compression on rolled 1100 aluminum sheet. Data is normalized to uniaxial tension using Taylor factors calculated by relaxed constraints.

compression experiments with the same strain state as the rolling prestrain (transverse) and those with a path change ie. cross-rolling (longitudinal). This difference increases with initial rolling prestrain. Interestingly when the data is compared using the von Mises yield function as an effective stress-strain criterion, Figure 9, agreement is much better. This is exactly the opposite of what we found after a smaller prestrain in torsion of $\gamma = 0.5$.

At von Mises equivalent strains above 0.5 aluminum shows many interesting microstructural features after rolling. On a microscopic scale grains become elongated and flattened. When a longitudinal section (the normal to this plane is aligned with the rolls) is viewed the grains develop "waves" around inclusions. Figure 10a an optical micrograph of sheet rolled to a von Mises strain of 1.5 illustrates this. Upon examination near these waves by TEM one observes shear bands running across the individual grains, Figure 10b. The evidence of shear bands initially takes the form of subgrains inclined to the

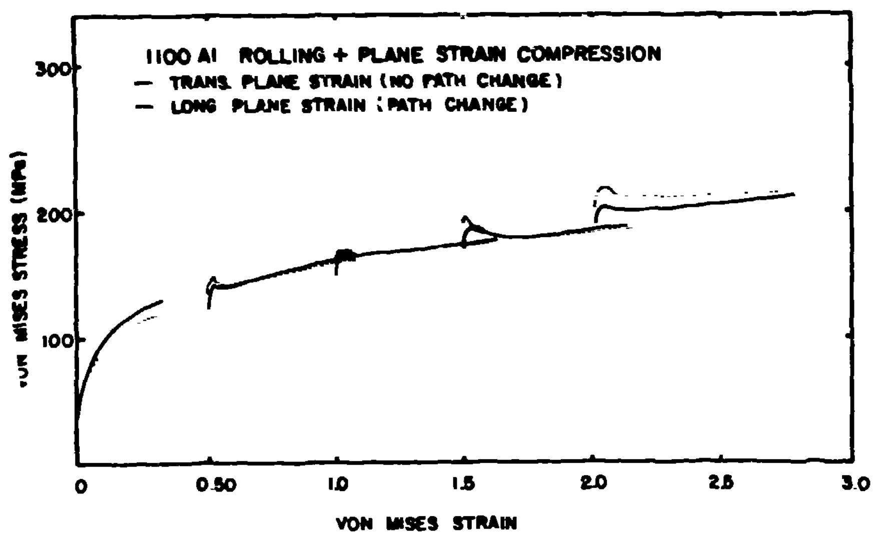


Fig. (9)- Knife edge compression on rolled 1100 aluminum sheet. Data is reduced using the von Mises yield function as an effective stress-strain criterion.

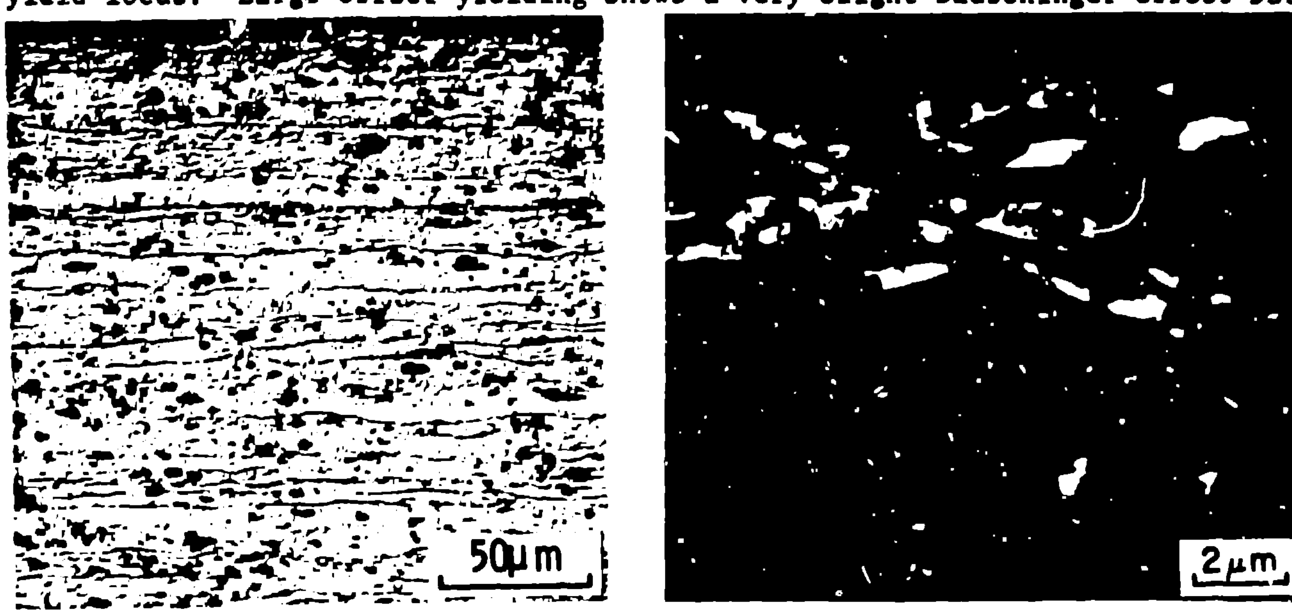
rolling direction. They do not extend beyond single grains, and do not become catastrophic; this was also observed by Rohr and Hecker [1981a].

DISCUSSION OF RESULTS

A. Effect of Prestressing Direction

The effect of the stress probe direction on the small offset yield locus is profound. There is a flattening of the "back" of the locus and the formation of a rounded nose in the direction of the prestressing. In addition the entire locus translates in the direction of the prestressing. The observed dislocation cell structure is consistent with these effects. Due to the diffuse structure of the all boundary the material has a knowledge of the direction its past history. Continued deformation in a forward direction γ forces dislocations into the cell walls, while in the reverse direction dislocations move easly into the open areas. Thus plastic deformation in the reverse direction is quite easy. This results in the yield locus distortion and translation. From the reverse torsion experiments we found that the distortion of the yield locus changes very quickly with small amounts of reverse plastic strain. The cell structure in this case must be very quickly altered to match the sign of deformation. It appears that the shape and position of the small offset strain yield locus is a function principally of dislocation substructure.

Prestressing direction has almost no effect on the back extropalated yield locus. Large-offset yielding shows a very slight Bauschinger effect but



(a) (b)

Fig. (10)- (a) optical and (b) TEM micrographs of 1100 aluminum sheet rolled to a von Mises strain of 1.0. Figure 10b illustrates the presence of microscopic shear bands.

for all practical purposes the yield locus does not exhibit any translation. Rather, this locus simply expands with prestressing and distorts from von Mises behavior in accordance with the developing texture. The effect of prestressing direction is thus only in terms the preferred orientation of grains which it develops, ie. different states of prestressing produce different textures.

B. Effect of the Magnitude of Prestrain

While the direction of prestressing had a very large effect on the small offset yield locus the magnitude of the prestrain had little effect. Qualitatively the distortion and translation of the yield locus were the same for both prestrains, $\gamma = 0.02$ and $\gamma = 0.50$. There was simply a scaling difference. In other words, the entire picture enlarged with greater prestrain. This yield behavior correlates to our microstructural observations. At the large torsional prestrain, $\gamma = 0.5$, dynamic recovery has not progressed to the point that the cells have completely become subgrains. There is still a finite thickness to the cell boundaries which enables the substructure to recognise the direction of deformation. Although an increased misorientation of $\sim 2^{o}$ exists between the cells/subgrains this has not changed the yield behavior.

We have found that the back extrapolated yield behavior is strongly influenced by the amount of prestrain. This is the direct effect of texture. As the magnitude of prestrain increases the material develops a continuously sharpening texture which results in deviations of the large offset yield locus from purely isotropic behavior. The influence of texture is always present. However, it does not tell the entire story. After von Mises equivalent prestrains at and above 0.5, in rolling, we found a discrepancy between experimental plane-strain compression data and polycrystal slip theory. If the theory were accurate the longitudinal and transverse plane-strain compression would have fallen on single curve. We attribute this discrepancy to the formation of micro shear-bands in rolling, speculating that the shear bands account for a substantial amount of plastic strain. The shear bands form with a specific orientation favorable to continued plane-strain deformation in the same mode as the rolling prestrain. They are totally inactive for a "cross-rolling" plane strain making deformation difficult and producing a higher yield stress than predicted crystallographically. We believe that the success of the von Mises model to describe the data is fortuitous. The effects of texture and microstructure (shear bands), neither or which is accounted for, appear to cancel one another evenly.

SUMMARY

We have found that there are several regimes of yield and flow behavior for 1100 aluminum. At strain levels below those which form shear bands, small-scale yield behavior, as reflected by a 5×10^{-6} offset yield criterion, is governed by the developing dislocation cell substructure. The yield locus is distorted and translated in the direction of the applied stress. This is because the cell structure "knows" the directional history of deformation. If a large offset yield criterion is used to evaluate behavior the yield locus is quite different. It is distorted very little and essentially does not translate. What distortions occur match those predicted by a texture based polycrystal plasticity theory. In short, large scale yielding is only a function of texture not microstructure.

After grain scale shear bands form, (at a von Mises effective strain of ~ 0.5 for aluminum in rolling), texture alone no longer controls large offset strain yielding. These shear bands, which possibly form due to the developing texture and relaxed constraints on deformation, have a specific orientation, which produces anisotropies in plastic flow independent of texture.

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