Materials Science Forum Vols. 604-605 (2009) pp 39-51
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Backstress, the Bauschinger Effect and Cyclic Deformation

This paper is dedicated to Prof. Enrico Evangelista in the occasion of his 70th birthday

M.E. Kassner[1, a], P. Geantil[1], L.E. Levine[2], B. C. Larson[3]

[1]Aerospace and Mechanical Engineering, University of Southern California, Los Angeles, CA, 90089-1453, USA

[2]Mater. Sci. and Eng. Lab., NIST, Gaithersburg, MD, 20899-8553, USA

[3]Mater. Sci. Tech. Div., ORNL, Oak Ridge, TN, 37831, USA

[a] kassner@usc.edu

Abstract. Backstresses or long range internal stresses (LRIS) in the past have been suggested by many to exist in plastically deformed crystalline materials. Elevated stresses can be present in regions of elevated dislocation density or dislocation heterogeneities in the deformed microstructures. The heterogeneities include edge dislocation dipole bundles (veins) and the edge dipole walls of persistent slip bands (PSBs) in cyclically deformed materials and cell and subgrain walls in monotonically deformed materials. The existence of long range internal stress is especially important for the understanding of cyclic deformation and also monotonic deformation. X-ray microbeam diffraction experiments performed by the authors using synchrotron x-ray microbeams determined the elastic strains within the cell interiors. The studies were performed using, oriented, monotonically deformed Cu single crystals. The results demonstrate that small long-range internal stresses are present in cell interiors. These LRIS vary substantially from cell to cell as 0 % to 50 % of the applied stress. The results are related to the Bauschinger effect, often explained in terms of LRIS.

Introduction

Stress fields in crystals can be either short range, such as with individual dislocations, or long-range internal stresses (LRIS) which refer to variations in stress that occur over longer length scales, such as the microstructure (often micrometer-range). LRIS appears to be important for a variety of practical reasons including the Bauschinger effect (as a "backstress") and cyclic deformation, springback in metal forming, as well as, of course, understanding plastic deformation, in general.

Bauschinger Effect (BE). The concept of backstresses or LRIS in materials may have been first discussed in connection with the Bauschinger effect. The material strain hardens and on reversal of the direction of the straining, the metal plastically flows at a stress less in magnitude than in the forward direction, in contrast to what would be expected based on isotropic hardening. A Bauschinger effect is illustrated in Fig. 1. Not only is the flow stress lower on reversal but the hardening features are very different as well. The Bauschinger effect is important as it appears to be the basis for low hardening rates and low saturation stresses (and failure stresses) in cyclic deformation. The Bauschinger effect is also central to the difference between kinematic and isotropic hardening models used to predict springback in sheet metal forming.

Non-Backstress Explanation for BE. Sleeswyk *et al.* [1] analyzed the hardening features in several metals at ambient temperature and found that the hardening behavior on reversal can be modeled by that of the monotonic case provided a small (e.g., 0.01-0.05) "reversible" strain is subtracted from the early plastic strain associated with each reversal (in Fig. 1 this is approximately β). This led to the conclusion of an Orowan-type mechanism (long-range internal stresses or "back-stresses" not especially important) [2] with dislocations easily reversing their motion (across cells).

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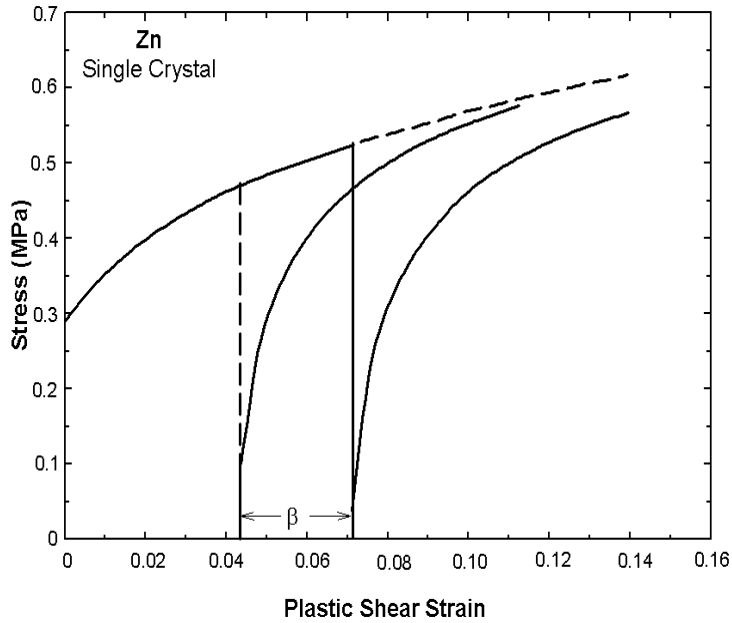

Fig. 1. A zinc single crystal is deformed in tension to a strain of .07, then unloaded from 0.54 MPa and compressed (although only positive stress is reported). The material yields at ≅ 0.1 (negative) MPa and gives rise to a Bauschinger effect. The hardening on reversal approaches that for continued monotonic deformation if a "reversible" strain β is subtracted to the dashed vertical line (From [1]).

Sleeswyk *et al*. suggested gliding dislocations, during work hardening, encounter increasingly effective obstacles and the stress necessary to activate further dislocation motion or plasticity continually increases (work-hardening). On reversal of the direction of straining from a "forward" sense, under σ, the dislocations will easily move past those, non-regularly-spaced, obstacles that have already been surmounted. Thus, the flow stress on reversal is initially relatively low, < σ. There is a relatively large amount of plastic strain on reversal to – (σ + dσ) in comparison to the strain associated with an incremental increase in stress to (σ + dσ) in the forward direction. Backstresses were not believed to be important. This is referred to as the Orowan-Sleeswyk explanation for the Bauschinger effect.

LRIS (Composite Model) BE. In an influential development, Mughrabi [3] advanced the concept of relatively high (long-range internal) stresses in association with heterogeneous dislocation substructures (e.g., cell/subgrain walls, dipole bundles, PSB walls, etc.). He presented the simple case where "hard" (high dislocation density walls, etc.) and "soft" (low dislocation density channels, or cell interiors) elastic-perfectly-plastic regions are compatibly sheared. Each component yields at different stresses and it is suggested that the "composite" is under a heterogeneous stress-state with the high-dislocation density regions having the higher stress. Thus LRIS are present. This composite picture was suggested to rationalize the Bauschinger effect, the basic element of cyclic deformation. As soft and hard regions are unloaded in parallel, the hard region eventually places the soft region in compression while the stress in the hard region is still positive. When the total, or average, stress is zero, the stress in the hard region is positive while negative in the soft region. That is, a backstress is created. A Bauschinger effect may be observed, where plasticity occurs on reversal at a lower average magnitude of stress than just prior to unloading due to "reverse" plasticity in the soft region. The "composite" model for backstress is illustrated in Fig. 2. The concept of heterogeneous stresses has also been widely embraced for monotonic deformation [4,5] including elevated-temperature creep deformation. The long range internal stresses are defined by

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$$\tau_w^y = \tau_a + \Delta\tau_w \,. \tag{1}$$

$$\tau_I^y = \tau_a + \Delta\tau_I \,. \tag{2}$$

where τ_a is the applied stress, and $\Delta\tau_w$ and $\Delta\tau_I$ are LRIS in the composite substructure. Simply stated, LRIS is the local deviation from the average (or applied) stress in a loaded material. It, perhaps, should be stated at this point that the Bauschinger effect may actually involve *both* LRIS and an Orowan-Sleeswyk component, as will be discussed more later.

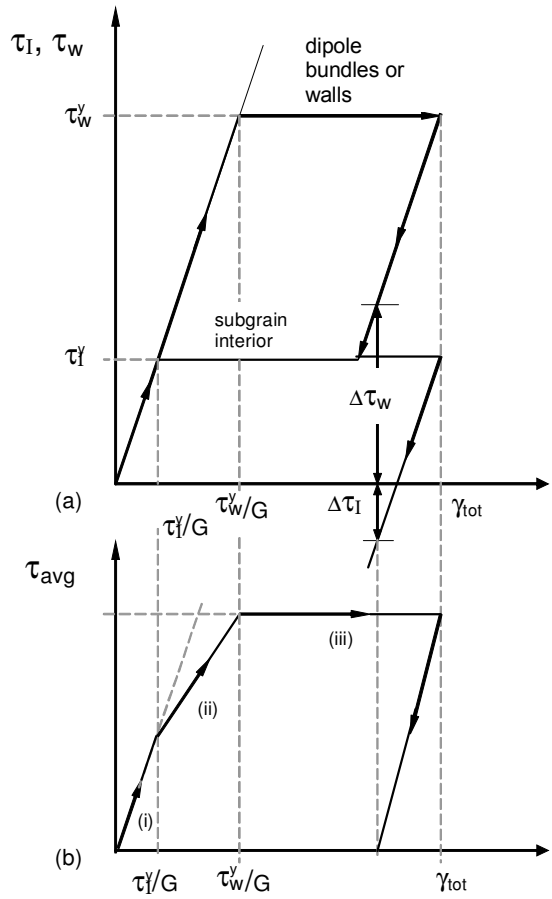

Fig. 2. The composite model illustrating the Bauschinger effect. The different stress versus strain behaviors of the (hard) cell walls and the (soft) cell interiors are illustrated in (a), while the stress versus strain behavior of the composite is illustrated in (b). When the composite is completely unloaded, the low-dislocation density cell interior region is under compressive (back) stress. This leads to a yielding of this "softer" component in compression at a "macroscopic" stress less than $-\tau_{avg}$.

Experimental History of Long-Range Internal Stresses

Bauschinger Effect. In early work by Kassner, a random dislocation arrangement in monotonically deformed stainless steel produced nearly the *same* elevated temperature Bauschinger effect as one with cells and/or subgrains where LRIS should be more substantial from the composite model [6]. Also, it can be noted from the aluminum experiments of [7,8] that a very pronounced Bauschinger effect is evident in the first cycle (just 0.0005 monotonic plastic strain) at 77 K when a cellular substructure is not expected to be evident during the very early Stage I deformation. The Bauschinger effect is comparable to the case where a heterogeneous vein/channel substructure developed after hundreds or thousands of cycles. These suggest either LRIS are not generally

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pronounced in either cellular or non-cellular materials or the principal features of the Bauschinger effect has explanations independent of or in addition to LRIS.

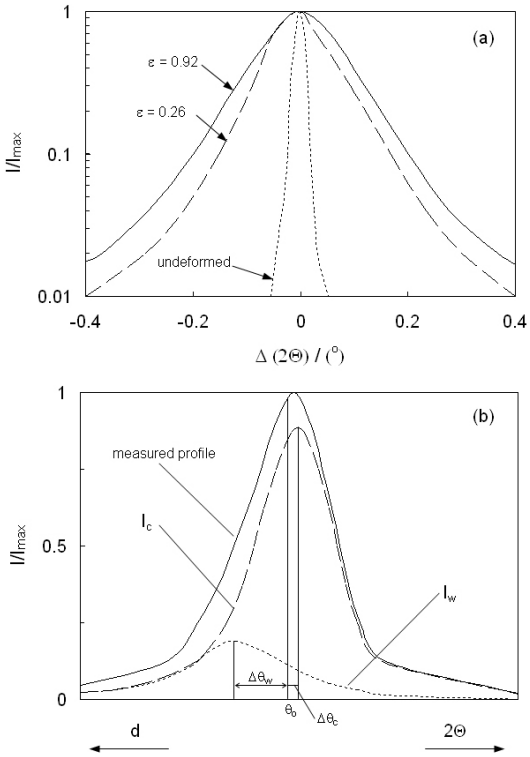

Fig. 3. The x-ray diffraction peak in Cu deformed to various monotonic strains showing asymmetric broadening. A decomposition is performed that leads to two symmetric peaks which has been interpreted as a heterogeneous stress-state [5], with different lattice parameters (stresses) in high and low dislocation density regions of the plastically deformed material.

X-Ray Peak Asymmetry. Indirect evidence for internal stresses has been based on x-ray diffraction (XRD) experiments during both cyclic and monotonic deformation [3-5,9-13]. Basically, x-ray peaks have been observed to broaden asymmetrically with plastic deformation [pronounced asymmetry may not always be observed, e.g., pure Al (although recent work appears to have observed asymmetry [34]) or materials cyclically deformed to presatuation]. A decomposition can be performed on the asymmetric peak profile into two symmetric peaks; one peak is suggested to represent the material in the vicinity of the high dislocation-density heterogeneities such as dipole bundles, persistent slip bands (or PSB dipole walls) or cell walls or subgrain boundaries with high local-stresses. The second peak in deformed metals is suggested to represent the lower dislocation density material (e.g., cell interiors) where the stresses are smaller than the applied stress. This asymmetry is illustrated in Fig. 3. Asymmetry analysis varies, but often suggests that (under load) stresses in the walls that are about a factor of 1.4 to 2.8 larger than the applied stress [4,9,14] with PSB walls in cyclically deformed metals having especially high values. The single crystal Cu compression deformation experiments by Ungar, Mughrabi, *et al.* suggested relatively low values by this technique at +0.10 σ_A in the cell interiors and 0.4 σ_A in the cell walls [15]. Although these experiments are indirect, they have been used to infer *local* stresses based on the irradiation of millions of cells.

In-Situ Experiments. *In-situ* deformation experiments in the TEM by Lepinoux and Kubin [16] and particularly the neutron irradiation experiments by Mughrabi [3] of Fig. 4 have long been cited as evidence for long-range internal stresses in cyclically deformed metals; the stresses are, roughly, a factor of three higher than the applied stress for dipole bundles in Cu persistent slip bands (PSB) walls (i.e., LRIS is about twice the applied stress). These two studies assessed LRIS by measuring

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dislocation loop radii as a function of position within the heterogeneous microstructure in cyclically deformed Cu single crystal. Very recent work by Mughrabi et al. suggests that the actual values may be half the earlier estimates [17] suggesting that the interpretation of the "pinned" dislocation configurations to elastic strains in the lattice is not straightforward. Fig. 5 illustrates *very* high local stresses at subgrain boundaries (up to 20 × the applied stress) in Al-5 % Zn creep-deformed alloys, based on pinned dislocation loops via a precipitation reaction on cooling under load. Other *in-situ* TEM experiments by Kassner were inconclusive [7].

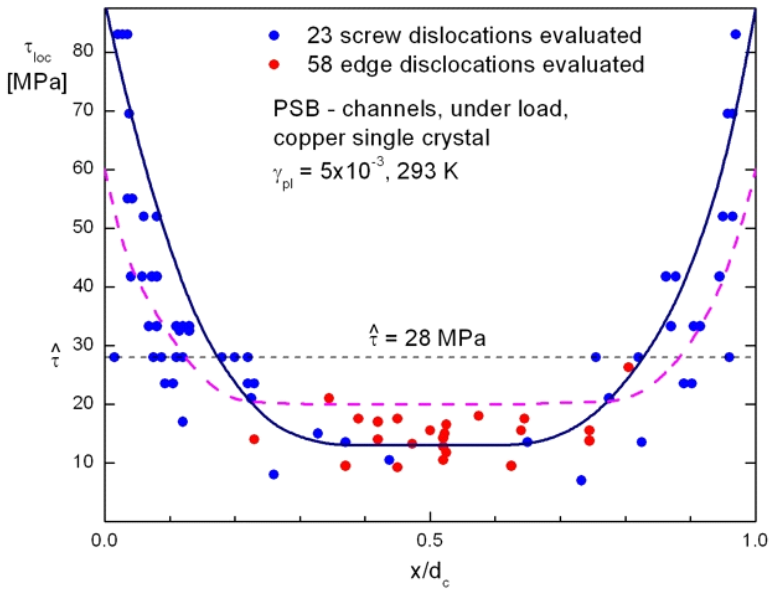

Fig. 4. Dislocation loop-radii-calculated LRIS in neutron irradiated cyclically-deformed Cu single crystals (under load) with persistent slip bands (PSBs). From [3] with recent modifications by the same investigators [17] (dashed line). The high local stresses correspond to high dislocation density regions (PSB walls).

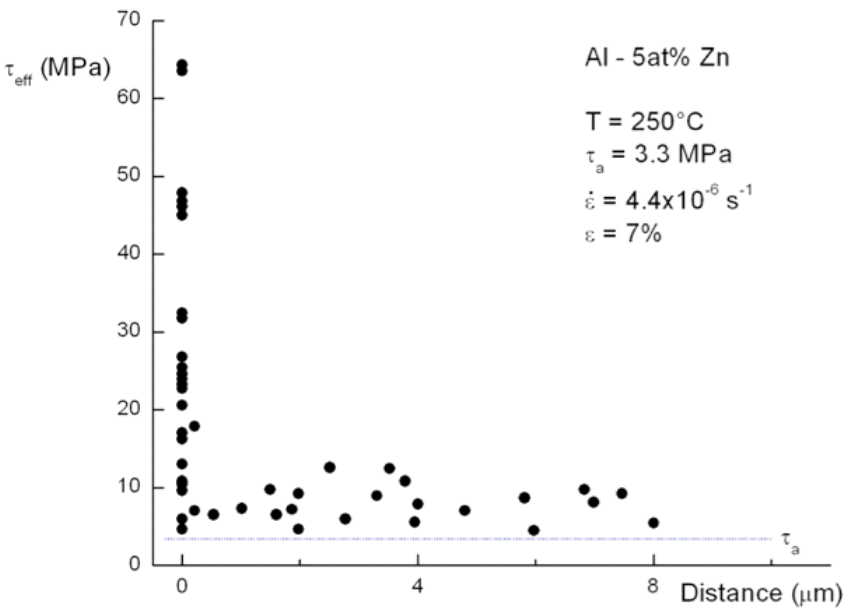

Fig. 5. Dislocation loop-radii-calculated LRIS based on precipitate pinning (under load) in creep deformed Al-Zn alloy with subgrains. The high stresses are very close to the subgrain walls. From [18].

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Dipole Height Measurements. The measurement of *maximum* dipole heights may allow the prediction of local stresses in cyclically deformed materials. The approximate stress to separate a dipole of height "h" can be calculated from

$$\tau_d = \frac{Gb}{8\pi(1-\nu)h}. \quad (3)$$

where G is the shear modulus, b the Burgers vector, and ν is Poisson's ratio. Cyclic deformation experiments on aluminum and copper single crystals (oriented for single slip) by one of the authors [8] showed that the *average* dipole heights in the presaturation (no PSBs [22,23]) microstructure are also approximately *independent of location*, being equal in the dipole bundles (or veins) and the channels, consistent with other Ni work [19,20]. This could suggest a uniform stress state across the microstructure. The *maximum* dipole heights are the widest stable dipoles and suggest, through Eq. (3), a maximum stress in the vicinity of the dipoles. The maximum dipole height translates to a stress, according to Eq. (3), that is about *equal* to the applied (cyclic) stress for Al. Hence, this is evidence of an absence of long-range internal stress in Al single crystals cyclically deformed to presaturation. However, the stress to separate the widest dipoles is within a factor of 2 to 3 in Cu while for Ni is within a factor of four [8,19,20,27] larger than the applied stress. The dipole heights are not expected to change significantly after unloading. Overall, these results are ambiguous in terms of an assessment of LRIS.

CBED. A few years ago, convergent beam electron diffraction (CBED) experiments were used by Kassner to assess internal stresses in unloaded monotonically and cyclically deformed Cu and Al [21,27]. CBED, with a 20 nm to 100 nm beam size, can probe smaller volumes than the x-ray studies referenced so far which irradiate over large portions of the specimens. Lattice parameter measurements were made near (within 80 nm) dipole bundles and within the channels in Cu cyclically deformed to presaturation and near cell/subgrain walls and interiors in creep deformed Al and Cu by Kassner. The errors, however, were equal to the flow stress for the cyclically deformed Cu examined [27] and 1.5 to 2 times the applied stress for the creep tests [21] on Cu and Al. No evidence of long-range internal stress was noticed, although small internal stresses such as those observed by Mughrabi might be undetected by this technique. Of course, another difficulty with these experiments (besides uncertainties) is that the region of the foils examined in CBED is fairly thin, under 100 nm, so relaxation of LRIS could have occurred. Table 1 illustrates the LRIS values from various studies, many already discussed. The table indicates a wide variation of results for strains (LRIS) in deformed materials.

X-Ray Line Broadening Analysis

Introduction. As discussed earlier, it has been argued that a heterogeneous dislocation microstructure can be associated with long range internal stresses (LRIS), often through a two-phase composite model, as illustrated in Fig. 2. Again, a commonly suggested manifestation of LRIS is the well known x-ray line broadening analysis [11], where it is argued that asymmetric diffraction peaks are composed of one larger and another smaller symmetric peak. The smaller peak is assumed to be associated with the (often smaller) volume of high stress, high dislocation density material as illustrated in Fig. 2. A review of the literature relating to x-ray line profile asymmetry shows consistent patterns:

i. *Asymmetry increases with applied stress and strain*

   X-ray line profiles tend to become more asymmetric as the specimen accumulates plastic strain.

ii. *Characteristic asymmetry*

   When comparing asymmetry from the axial to side faces, the asymmetry is in the opposite "sense" or direction. This is expected based on the composite model in terms of a

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Poisson effect. Also, the asymmetry is reversed between tension and compression deformed samples.

iii. Asymmetry Reversal and Saturation in Cyclic Loading

Some recent [44] work indicates that the asymmetric x-ray line profiles reverse after reversed deformation, which is expected from the composite model. Other investigators [9,10,33] have unloaded their samples from various points in the hysteresis loop to obtain these results on single crystal Ni, and polycrystalline Cu. An interesting phenomenon is the observation of asymmetry for saturation (PSBs) but an absence for presaturation (no PSBs). This led Hecker *et al.* [9] and Bretschneider [24] to conclude the presence of LRIS only when PSBs are present in cyclically deformed Ni.

As mentioned earlier, these measurements infer local cell wall and cell interior strains by averaging information from diffracting material comprised of over 10^6 cells.

Material	Ref.	Deformation Mode	Strain	LRIS (σ_A) Walls	LRIS (σ_A) Interior	Observation Method	Temp.	Notes
Cu	Mughrabi [3]	cyclic	saturation w/PSBs	+2.0	-0.5	in-situ neutron irrad.	RT	
Cu	Mughrabi et al. [17]	cyclic	saturation w/PSBs	+1.3	-0.37	in-situ neutron irrad.	RT	reanalysis of above
Cu	Mughrabi et al. [12,15,25,26]	tension		+0.4	-0.1	x-ray peak asymm.		unloaded [001] oriented single crystal
Cu	Lepinoux & Kubin [16]	cyclic	saturation w/PSBs	+2.5	-0.5	in-situ TEM	RT	loaded single crystal
Cu	Kassner [27]	cyclic	pre-sat. no PSBs	–	0	CBED	RT	unloaded [123] single crystal
Cu	Kassner [27]	cyclic	pre-sat. no PSBs	–	0	dipole sep.	RT	unloaded [123] single crystal
Cu	Borbely et al. [4]	creep	steady-state	+1.0	-0.08	x-ray peak asymm.	527 K	loaded
Cu	Kassner et al. [21]	creep	steady-state	–	0	CBED	823 K	unloaded
Cu	Straub et al. [28]	compression		-0.3 – -0.6	+0.05 – 0.08	x-ray peak asymm.	RT – 633 K	
Cu	Straub et al. [28]	compression		–	observed	CBED	RT – 633 K	
Cu	Levine et al. [29]	Compression (ave.)		–	+0.29	x-ray microbeam	RT	unloaded [001] oriented single crystal
Cu	Levine et al. [29]	Tension (ave.)		–	-0.17	x-ray microbeam	RT	unloaded [001] oriented single crystal
Ni	Hecker et al. [9]	cyclic	saturation w/PSBs	±1.4 – 1.8	± 0.16 – 0.2	x-ray peak asymm.	RT	loaded
Ni	Hecker et al. [9]	cyclic	pre-sat. wo/PSBs	0	0	x-ray peak asymm.	RT	loaded
Al	Kassner et al. [7,27]	cyclic	pre-sat wo/PSBs	0	0	dipole separation	77 K	unloaded [123] oriented single crystal
Al	Kassner et al. [21]	creep	steady-state	–	0	CBED	664 K	Unloaded
Al 5% Zn	Morris & Martin [18]	creep	steady-state	+ 25	+1	disl loops from precipitation pinning	483-523 K	
–	Sedlacek et al. [30]	creep		1.5 – 10.0	0.5 – 1.0	theoretical	creep	
–	Gibeling and Nix [31,32]	creep		7.7	~0.1 – -0.2	theoretical	creep	

Table 1. Other Selected Studies of Long-Range Internal Stresses

Synchrotron X-Ray Microdiffraction Experiments.

Background. In 2002, Larson *et al.* developed a polychromatic x-ray technique, white-beam differential-aperture X-ray microscopy (DAXM), that allows measurement of the crystallographic orientation and the internal deviatoric (shear) stress state in relatively small, spatially-resolved

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volumes within a bulk specimen [36]. More recently, this technique has been extended by using a scanned monochromatic X-ray beam [29] to allow absolute lattice parameters and dilatational strains to be measured from buried submicrometer sample volumes. The small spot size is obtained using elliptically-figured crossed Kirkpatrick-Baez focusing mirrors and undulator x-rays at the Advanced Photon Source (APS) at Argonne National Laboratory (ANL) [42,43]. The submicrometer (≅ 0.5 μm) spot size of this new technique is small enough to allow probing within individual dislocation cells within a bulk specimen. Local data acquisition is possible on the length scale that is necessary to assess the stress state in a heterogeneous dislocation microstructure. In scanning-monochromatic DAXM, a submicrometer x-ray beam is scanned through a broad enough range of energies to ensure diffraction from all of the sample volumes along the beam path. The diffracted patterns from various sub-grains or dislocation walls would overlap on the CCD detector, thus convoluting the diffraction pattern. This is illustrated in Fig. 6. The DAXM method decomposes these overlapping profiles by step-scanning a 50 μm diameter platinum wire (i.e., knife edge) through the diffracted beams (in front of the CCD detector), and triangulation determines the location of the scattering volume along the incident x-ray beam using the knowledge of the CCD pixel locations, and the instantaneous stepped location of the leading edge of the wire.

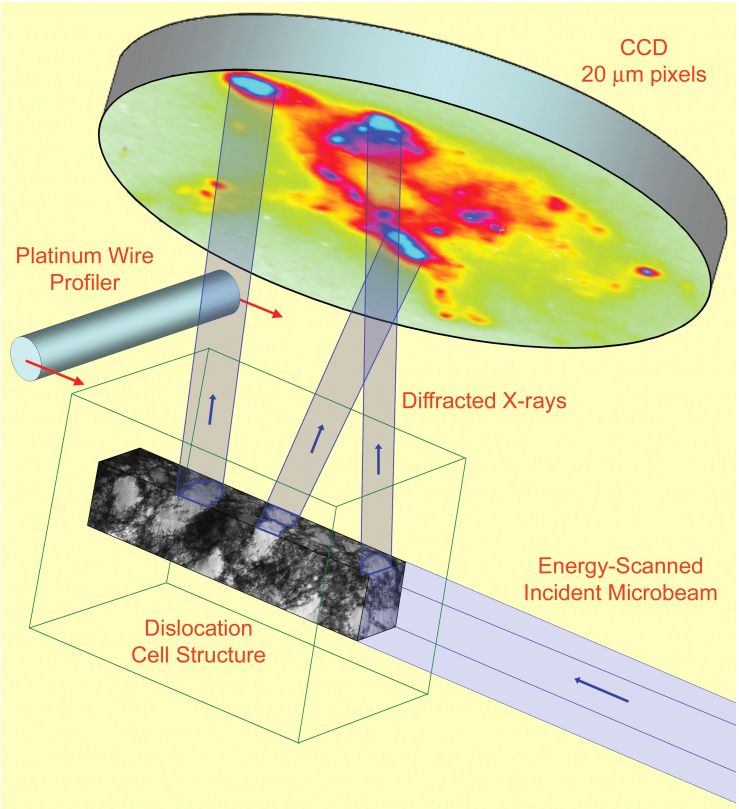

Fig. 6. Depth-Resolved Diffraction Profiles. Energy-scanned DAXM measurements were conducted using an incident X-ray beam focused to ≈ 0.5 μm by crossed Kirkpatrick-Baez (KB) mirrors (60 mm and 30 mm focal lengths) using the XOR/UNI beamline 34-ID at the Advanced Photon Source at Argonne National Laboratory. Diffracted X-rays from the (energy-scanned) incident microbeam were diffracted on a CCD area detector. The axial <006> reflection was used for all depth resolved measurements. Depth resolution was provided by a 60-μm diameter platinum wire profiler that was incrementally translated parallel to the sample surface, blocking diffracted X-rays from the sample; the origin of the diffracted X-rays was then determined by triangulation. The profiler step size (0.7 μm) was selected to give a resolution along the incident beam that matched the width of the microbeam, thus providing a spatial resolution of ≈ 0.5 μm in all three dimensions.

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A wire scan must be conducted at each stepped photon energy. Computer analysis of the ensemble diffraction profiles with the triangulation data for each Pt step position allows the calculation of depth-dependent diffraction line profiles with half-micrometer spatial resolution. Ultimately, this technique allows the 3-D mapping of the local crystal structure, orientation and elastic strains with submicrometer resolution. Said another way, monochromatic beams are diffracted by various cells onto a detector. By knowing the wavelengths, a Pt wire can "back out" from various diffraction spots on the detector, the precise locations of the cells. This process is repeated for other wavelength monochromatic beams (up to 600 eV range for 3 eV steps) and a series of lattice parameter values can be assigned to specific, small, volumes of material leading to measurements of the average lattice parameter in a cell interior.

Cell Interiors. The DAXM technique has been applied to make direct measurements of the strain state of dislocation cells in both tension and compression deformed copper. Copper single crystals with [001] orientations were deformed in either tension or compression to strains of about 30 %. This led to well developed dislocation cell structures such as those shown in Fig., 7, where the microstructure in a 28 % compression deformed sample was characterized using TEM. Comparing the cell sizes with the ≈ 0.5 μm beam sizes available using the DAXM technique, it is clear that synchrotron microbeams make an ideal probe of the lattice strain in the interior of dislocation cells in *bulk* deformed materials.

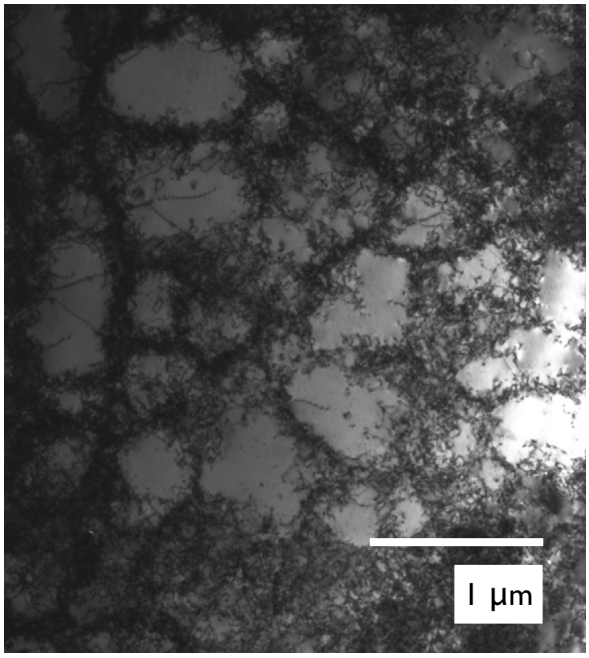

Fig. 7. The dislocation substructure of a [001]-oriented Cu single crystal deformed in compression to a true strain of -0.28.

Fig. 8 (a) and (b) show the lattice strains measured on individual dislocation cells in the compression (-0.28) and tension (+0.31) deformed [100]-oriented Cu single-crystals [39]. Also shown in Fig. 8 for reference are volume average (006) x-ray line profiles for the respective samples. These line profiles show the well known asymmetry that reverses "sense" for tension and compression. When taken together, the tension and compression results for strain in individual dislocation cells are direct evidence that a long range internal stress (LRIS) remains after plastic deformation and elastic unloading, specifically showing that cell interiors are under compressive stress after tensile deformation and vice versa. Further, these results are in qualitative agreement with the predictions of the composite model. While we do not have direct measurements of strain in the cell walls yet, the presence of a stress with a sign opposite to that of the cell interior would be

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consistent with the collected data. Thus, the x-ray line profile asymmetry here (and perhaps in general) appears reflective of LRIS. In these samples, the LRIS in cell interiors for tension and compression is (-)0.17(+/-0.03) and (+)0.29(+/- 0.04) of the applied stress. The uncertainties are one standard deviation of the measured distribution of cell interior elastic strains.

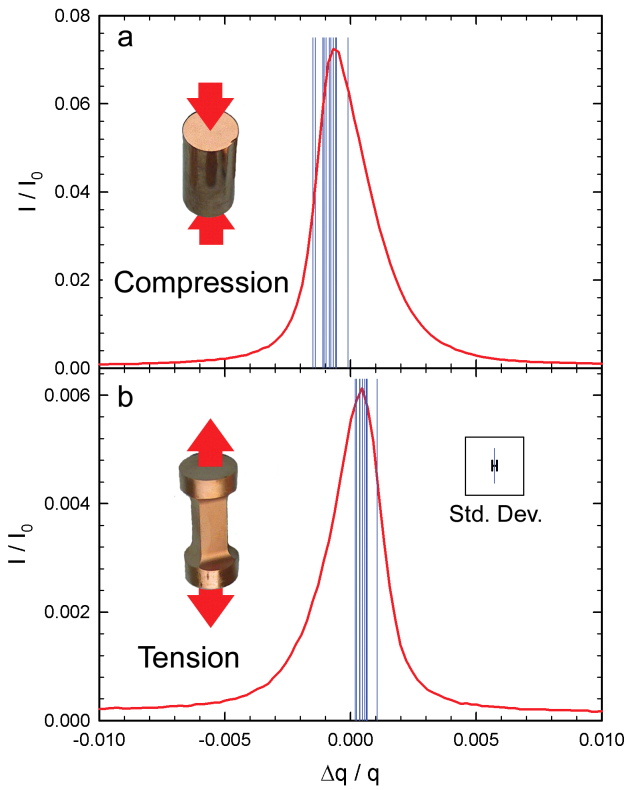

Fig. 8. The (006) x-ray peak profile of (a) compression-deformed, (b) tension deformed [001] oriented Cu single crystal. Asymmetry is observed and the blue vertical lines represent the lattice parameters within various cell interiors as determined by x-ray microbeams. Recent results by the PIs. Note that the asymmetry and the lattice parameter (blue lines) reverse from compression to tension. From [39].

It is important to emphasize that the cell interior stresses (obtained by multiplying the elastic strains by 66.66 GPa, the elastic modulus in the 001 direction) vary widely from cell to cell over the range 0 to -0.5 of the applied stress.

As mentioned above, the microbeam experiments were conducted on unloaded specimens. Straub *et al.* made asymmetry measurements *in-situ*, or under load [41]. The XRD line profiles provided the expected asymmetry under stress at elevated temperature. The *unloaded* specimen retained a majority of the asymmetry despite elevated temperature exposure. If the asymmetry under load is reflective of LRIS, then unloaded specimens appear to substantially "lock in" these stresses.

More recent (and more extensive) microbeam experiments [40] on a different compressed Cu, sample give an average cell interior stress of +0.16 the applied stress (210 MPa) at a true strain of 0.28.

Summary

There is reasonable agreement between our Cu monotonic deformation results and the work of Mughrabi and Ungar *et al*. [12] as corrected in [15]. Our reported values are lower than other x-ray asymmetry measurements and other measurements of LRIS, as discussed earlier. Our estimates for the volume fractions of cell walls at -0.25 strain in [001] oriented Cu is $f_w \cong 0.44$ to 0.58 based on a fairly simple image analysis of TEM micrographs that partition "high" and "low" dislocation density regions. This implies from the composite model that if $\Delta\sigma_I \cong (0.17 \text{ to } 0.28)\ \sigma_A$, then the

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LRIS in the cell walls are about (0.17 to 0.56) σ_A (and of opposite sign). The cell wall estimates of LRIS are in rough agreement with the Mughrabi *et al.* value of 0.4 σ_A based on x-ray diffraction of unloaded specimens that assesses an average of a very large number of cells.

An explanation for the basis of the LRIS that is consistent with the observations is critical. The composite model predicts higher local (LRIS) stresses in association with a higher local strength. With the disparity in yield stress, dislocations accumulate at the interface for compatibility. These create a geometrically necessary boundary (GNB), as was discussed originally by Mughrabi. It seems that LRIS can arise from a summation of the stress fields of individual dislocations in a GNB cell wall or subgrain wall, as discussed by Zbib et al [37], Argon *et al.* [14] and Nix [38]. It would make some sense to eventually model the LRIS as the summation of the stress fields of dislocations based on TEM images.

Bauschinger Effect Experiments and Mechanical Modeling. Another aspect to the X-ray microbeam work is explaining the Bauschinger effect. The sources for Bauschinger strain or "reversible" strain (ε_B or β) on unloading include, but are not restricted to: elastic unloading (γ_e), dislocation loop unbowing (γ_A), plasticity associated with backstress (LRIS) of opposite sign to the strain prior to unloading (γ_B), and recovery on reversal, (γ_C).

A *maximum* (anelastic) unbowing for a "cubic grid" of dislocations can be estimated as

$$\gamma_A \cong \frac{\pi b}{8}\frac{\sqrt{\rho}}{\sqrt{6}}. \tag{3}$$

If $\rho \cong 10^{14}$ m^{-2} to 10^{15} m^{-2} for cyclically and monotonically deformed Cu single crystal (as determined in our earlier research), $\gamma_A \cong 10^{-3}$. The sum of maximum recoverable axial plastic strains from various elements, i, under load from backstresses is

$$\gamma_B = \overline{\sum_{p,LRIS}} = \sum_{1}^{n}\left(\frac{\sigma_i - \sigma_m}{E}\right). \tag{4}$$

where σ_m is the maximum stress (e.g., prior to unloading) and σ_i is a local stress prior to unloading. In this case, for simplicity, we assume $\gamma_B \approx \sigma_m/E$. That portion of β or ε_B due to recovery (γ_C) cannot be easily predicted; this is a contribution to β based on dislocation density changes. However, $\gamma_e + \gamma_A + \gamma_B$ is small for cyclic and monotonic deformation of Cu, roughly 10^{-3} to 3×10^{-3}. The observed values of β and ε_B are about an order of magnitude too large. Thus, it appears the low proportional limit might be explained by LRIS and/or anelastic behavior, but the principal basis for the Bauschinger effect may need to include the Orowan-Sleeswyk explanation in addition to an "early" yielding from LRIS.

Conclusions

Long range internal stresses were confirmed for monotonically deformed Cu single crystals with a cellular substructure. These vary substantially from one cell interior to the next, with an average value of 0.16 to 0.29 the applied stress, with lower values in this range somewhat more probable. The Bauschinger effect is likely a combination of LRIS and an Orowan mechanism.

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Acknowledgments

MEK and PG acknowledge support from the Office of Basic Energy Sciences (BES), DOE. Research at ORNL is supported by DOE Office of Science, Basic Energy Sciences, Division of Materials Sciences & Engineering. The XOR/UNI facilities on Sectors 33 and 34 at APS are supported by USDOE, Office of Science.

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