Scripta Materialia 135 (2017) 68–71

Contents lists available at ScienceDirect

Scripta Materialia

journal homepage: www.elsevier.com/locate/scriptamat

Regular Article

Analysis of the strain-rate sensitivity and the Bauschinger effect in fine-grained Cu and Cu-Ni alloys

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ARTICLE INFO

Article history:
Received 4 February 2017
Received in revised form 16 March 2017
Accepted 17 March 2017
Available online xxxx

Keywords:
Fine-grained
Strain-rate sensitivity
Bauschinger effect
Backstress
Grain size distribution

ABSTRACT

The strain-rate sensitivity along the load-unload experiments on fine-grained pure Cu and Cu-Ni alloys are presented. For grain sizes lying in the range of 1–10 μm in pure Cu, Cu16at%Ni and Cu50at%Ni, an elevated rate sensitivity at the first few percent of plastic strain is observed. At this elasto-plastic transition, a substantial hysteresis during unloading is developed corroborating to the Bauschinger effect in the fine-grained materials. The strain rate sensitivity results are analyzed along the load-unload response of the materials. A dislocation-based phenomenological model is developed to describe the experimental observations.

There has been a renewed interest on the mechanical behavior of fine-grained materials over the recent years. Major part of the body of the research in this area is concentrated on materials with grain sizes, (d_g), within the range of ultra-fine grained (UFG, 100 nm < d_g < 500 nm) and nanocrystalline (NC, d_g < 100 nm) materials. There seems to be a gap of understanding of the mechanical behavior of materials with grain sizes between conventional coarse-grained and ultra fine grained/nanocrystalline materials. The beneficial effect of refining grain size goes beyond the Hall-Petch effect and could be also realized through enhancement of work hardening capacity of the microstructure allowing the material to undergo complex deformation processes. Refining the grain size below few microns level has an impact on mechanical behavior as observed through the extension of the elasto-plastic transition [2–4] as well as high rate sensitivity [5]. For instance, Carpenter et al. [6] have reported an order of magnitude higher rate sensitivity for nanocrystalline pure Cu and Ni (m = 0.01–0.04) compared to their coarse-grained counterparts (m = 0.001–0.005). Duhamel et al. [7] reported and modeled an inverse Cottrell-Stokes law with strain rate jumps over the first few percent of plastic strain in pure Cu with grain sizes of few nanometers to 300 nm.

Development of the Bauschinger effect in fine-grained materials is also extensively reported. Observation of anelasticity during the early deformation stage, 1–2% plastic strain, through uniaxial load-unload experiments allows one to probe the Bauschinger effect and whether backstresses exists [8–9]. Rajagopalan et al. [10] have presented load-unload experiments on Au and Al films with average grain sizes of 70 nm and 120 nm, respectively. The Al sample with relative homogeneous grain size did not show a large deviation from the elastic unloading response while the Au film with a wide grain size distribution shows a strong deviation from the elastic unloading path and thus development of the Bauschinger effect. This was related to the width of the grain size distribution and partitioning of stress between the grains as some grains would behave elastically while some grains were already plastically yielded. Raeisinia [11] simulated the effect of grain size distribution on the elasto-plastic transition, employing the elasto-plastic self-consistent (EPSC) model and showed that grain size distribution has an effect on this transition, making the transition progress more gradually as the grain size distribution width increases.

On the load-unload experiments at the elasto-plastic transition, a study by Al-Fadhalah et al. [2] is noteworthy. The uniaxial load-unload experiment on pure Ag with a grain size of 2 μm shows a deviation from the elastic behavior at very small strains of 0.0002 which the material would be expected to show an elastic unloading response. Mompiou et al. [12] have also performed load-unload experiments on samples of ultra fine-grained Al with grain size of 500 nm and compared the results against the coarse-grained Al counterpart. The fine-grained Al sample showed development of a large hysteresis during the unloading while the coarse-grained sample showed a linear elastic load-unload behavior with no hysteresis in contrast to the fine-grained samples. It was argued that there are single dislocation sources acting in these fine grains and considering a perfect dislocation with the Burgers

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http://dx.doi.org/10.1016/j.scriptamat.2017.03.024
1359-6462/

vector of $\overrightarrow{b} = \frac{a}{2}[110]$ and $|b| = 0.286$ nm gliding across the grains, this would result in reverse plastic strain of $\varepsilon_p = 0.0007$.

Even in the case of assumption of single source dislocations operating inside fine grains [13], there could be accumulation of dislocations or segments of them at grain boundaries which would then exert backstresses on the arriving dislocations from the opposite direction. Sinclair et al. [14] introduced a backstress model which takes into account scenario of dislocations arriving at grain boundaries and being stored at grain boundaries which then exert backstress to the arriving dislocations on the opposite side. It was proposed that dislocations arriving at slip bands (n) with distance λ are stored at grain boundaries as a function of plastic strain, $(\frac{dn}{d\varepsilon})$, at the rate described as:

$$\frac{dn}{d\epsilon} = \frac{\lambda}{b}\left(1 - \frac{n}{n_A^*}\right) \quad (1)$$

This model takes into account the possibility of dislocation annihilation at a critical parameter n_A^* above which any newly arriving dislocation to grain boundary is annihilated. Meanwhile, with furthering the plastic strain there could be a possibility that the same number of dislocations accumulate at the other side of the grain boundary. This would cause the screening of the long range backstress of the stored dislocations on a given slip band and thus causing the backstress term to drop to zero. A parameter n_S^* is then introduced at which this long range backstress would drop to zero:

$$\sigma_{back} = \frac{M\mu b}{d_g} n\left(1 - \frac{n}{n_S^*}\right) \quad (2)$$

in which μ is the shear modulus and M the Taylor factor. This backstress model will be used as an essence to model the strain rate sensitivity behavior.

Cast and hot rolled sheets of high purity Cu (purity 99.998%), Cu16at%Ni, and Cu50at%Ni were received for this study. Samples were cold rolled in a laboratory rolling mill to a true strain of 2.37 (in thickness) in 20 passes with the material being submerged and equilibrated in liquid nitrogen bath between each pass. Following cold rolling, samples were cut and undergone heat treatment schedules for temperatures ranging between 500 and 1000 °C for 1 h for the alloys and at 200, 360 and 500 °C for 1 h for Cu samples. Backscattered imaging mode of SEM and EBSD was used to confirm the full recrystallization specifically for samples with the finest grain sizes. Grain sizes were established using backscattered images from the lineal intercept method, while EBSD was used to characterize the grain boundary misorientation distribution and texture. The crystallographic orientation and area fraction of grains from the EBSD was used as an input to the visco-plastic self-consistent model (VPSC) simulation code [15]. The simulations were performed for one step of deformation (no hardening) and with a relative activity of 1.0 of critically resolved shear stress on all {111} ⟨110⟩ slip systems. The Taylor factor in this case is given as the computed average flow stress after the first step of deformation.

Strain rate sensitivity tests were performed at 298 K with a hydraulic Instron testing frame (load capacity of 25 kN). Jumps were performed every 1% true strain at pre-determined strain level. Considering the definition of true strain as $\varepsilon = \ln \frac{l}{l_0}$ in which l is the instantaneous length and l_0 the initial length, the extensometer values that jumps were performed had been predetermined. The initial jump was performed at $\varepsilon = 0.005$. Strain rate tests were performed up to true strain of 0.3 corresponding to 30 jumps. Each jump cycle consisted of low-high-low strain rates, each at nominal strain rates of $10^{-3}s^{-1}$ and $10^{-2}s^{-1}$, respectively. For each jump cycle, the flow curve over approximately 0.3% plastic strain before and after the low to high strain rate jump was fit by a linear regression. It was determined that there was less noise in the experimental data in the downward jump and thus this part of each cycle was used to report the flow stress state (σ) and change in the flow stress, $\Delta\sigma$. This data was used to construct the Haasen plots ($\frac{d\sigma}{d(\ln \dot{\epsilon})}$ versus σ). (Fig. 1).

Load-unload experiments were designed to reveal the Bauschinger effect in the early stage of deformation, i.e. $\epsilon \leq 0.008$ as well as at higher levels of plastic strain up to 0.3. For the load-unload experiments, tensile samples with an attached extensometer with a gauge length of 12.5 mm were used. The testing schedule was designed in such a way that unloading ceased while the sample was held slightly in tension to avoid inducing compressive stresses. Following each unloading sequence, the test was ceased for 30 s to settle the machine. For each tensile sample of each grain size, 16 load-unload sequences were performed. The Bauschinger stress is defined as $\sigma_B = \frac{1}{2}\langle\sigma_f - \sigma_r\rangle$ in which σ_f is the stress just prior to unloading while σ_r is the reverse yield strength [16] here defined as an intercept of the unloading curve with a line with an offset of 0.01% from the elastic unloading path.

The Haasen plots derived from the strain jump experiments are presented in Fig. 1(a, b) in which the behavior of fine-grained material is superimposed on the coarse-grained counterpart. The Haasen plots show a point of high rate sensitivity followed by a transition to the Cottrell-Stokes law. In the case of fine-grained Cu16at%Ni, 5.3 μm, the value for $(\frac{d\sigma}{d(\ln \dot{\varepsilon})})$ for the first data point shows 4 times higher value than for the coarse-grained sample found to be 0.5 (MPa·s).

Using the Haasen plot will help us, to the first approximation, answer whether there is a remnant grain size dependent backstress in the microstructure of fine-grained materials after the elasto-plastic transition [17–18]. The total flow stress, σ, is described as the

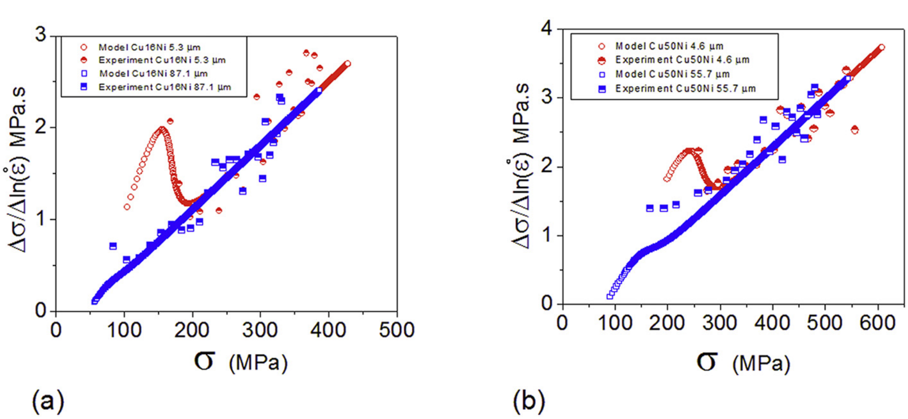

Fig. 1. The supersposition of the proposed strain rate sensitivity model on the experimental results of (a) Cu16at%Ni and (b) Cu50at%Ni.