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Influence of Ni Content on Internal Friction and Acoustoplastic Effect in Cu-Ni Single Crystals

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Abstract. Influence of Ni content in a range 1.3 - 7.6 at.% on amplitude-dependent internal friction and Young's modulus defect of Cu-Ni single crystals oriented for single slip has been studied by composite oscillator technique at frequencies of about 100 kHz in a wide strain amplitude range before, during, and after plastic deformation. The acoustoplastic effect was registered simultaneously with *in situ* internal friction measurements. Effects of the composition and temperature ranging from 6 to 255 K on damping and anelastic strain was studied and allowed to suggest multicomponent strain amplitude-temperature spectra of the amplitude dependent internal friction. A functional form of the amplitude dependence of the internal friction and Young's modulus defect is discussed. Data on *in situ* measurements allow to associate the amplitude-dependent internal friction and acoustoplastic effect with different mechanisms: dislocation - point obstacle and dislocation - dislocation interactions, respectively.

1. INTRODUCTION

Despite numerous theoretical and experimental studies of amplitude-dependent internal friction (ADIF), a number of problems, concerning its frequency, strain amplitude, and temperature dependences has not been studied explicitly so far. An absence of a time dependence of the ADIF in Cu-Ni crystals and the fact that Ni additions in Cu form substitutional solid solution in a wide range of Ni content render Cu-Ni alloys very convenient to study the ADIF. The acoustoplastic effect (APE) appears during active deformation as a decrease in flow stress when an oscillatory component is superimposed on a static mechanical load. Simultaneous measurements of the ADIF and APE *in situ* during plastic deformation of crystals with different impurity content allow to highlight the role of dislocation - point obstacle interaction in the APE.

2. EXPERIMENTAL DETAILS

A computer-controlled setup based on piezoelectric ultrasonic composite oscillator technique was used for excitation of ultrasonic vibrations and measurements of the ADIF and Young's modulus defect (YMD) at frequencies of about 100 kHz. Cu-Ni single crystals with different Ni content (1.3, 2.3, 3.2, and 7.6 at.%) were oriented for single slip (for details see [1]). Schmid factor $\sin\chi \cdot \cos\lambda$ was 0.452, 0.445, 0.492 and 0.497, respectively. Rod-shaped samples of about 1.5*2.0 mm² cross section and one or three half-wave lengths of ultrasound in length were spark cut from large crystals. They were annealed in vacuum at 973 K for 20 hrs, and then furnace cooled.

The strain amplitude dependences of the ADIF and YMD were measured at room temperature before, during, and after deformation of the alloys and at different temperatures between 6 and 290 K for an as-annealed sample of Cu - 1.3 at.% Ni. In so doing, the oscillatory strain amplitude was first increased with a preset step from a low amplitude-independent level to a maximum value and then decreased in reverse sequence. The time to measure an amplitude dependence was about 2 min. Since the forward and reverse runs coincide rather well everytime, the forward runs will be treated hereafter. This fact also allows to neglect dislocation multiplication by ultrasound.

To compare the results, obtained by different techniques, the ADIF and YMD of the undeformed samples were recalculated to the intrinsic values. Each experimental curve was fitted by a set of polynomials. Then, the oscillatory elastic strain dependence of the intrinsic ADIF $\delta_h(\epsilon)$ was derived using procedure [2]:

$$\delta_h = \sum_{j=0}^n K_j \varepsilon^j \frac{\sqrt{\pi}}{2} \Gamma\left(\frac{j+4}{2}\right) / \Gamma\left(\frac{j+3}{2}\right), \quad (1)$$

where K_j are the coefficients of the polynomial fitting the experimental curve, n - the degree of the polynomial, Γ - the gamma function. The same procedure was used to derive the intrinsic YMD values.

In situ measurements were taken during three-point bending of the three-half wave length samples so that the loading points were coincident with the vibration nodes of the ultrasonic standing wave. Quasistatic deformation of the samples was performed at constant deforming fixture movement rate of 10^{-5} mm/s. The bending load was applied to the surface of the samples close to the (111) plane. Additional plastic sag Δd of a specimen with a rectangular cross section under superimposition of the vibrations gives rise to the elastic unloading: $\Delta P = 4 E w \Delta d (h/s)^3$, where s is the span between the lower deforming fixtures, w and h are the width and height of the specimen, respectively. To account for differences in specimen's geometry and Young's modulus, we used the following conversion coefficient from the APE value for the i -th specimen ΔP_i to the APE value for arbitrary reference specimen ΔP_r , providing $\Delta d_i = \Delta d_r$:

$$k_i = \Delta P_r / \Delta P_i = (w_r/w_i) (E_r/E_i) (h_r/h_i)^3 (s_i/s_r)^3 \quad (2)$$

3. RESULTS AND DISCUSSION

Figure 1 shows the ADIF, YMD, and ADIF to YMD ratio for prestrained Cu - Ni samples with different Ni content. The functional form of the ADIF and YMD is rather complicated. The ADIF and YMD can be fitted by a power function only within limited strain amplitude ranges. Three stages are revealed in the ADIF and YMD. The first stage has nearly the same slope in the logarithmic scale for all the crystals (this stage appears only slightly in Cu -1.3 at.% Ni at room temperature). The second stage is steeper than the first one and steepens further with Ni concentration increase. The ADIF and YMD tend to saturate at the third stage. The stress amplitude dependence of the ADIF to YMD ratio exhibits maximum for all the crystals. This maximum and transitions between the stages shift to the higher strain amplitudes with Ni content increase. It should be noted that the strain amplitude dependences in Fig.1 do not fit a straight line in the Granato-Lücke coordinates $\ln(\delta_h \varepsilon_m) - 1/\varepsilon_m$ even within limited strain amplitude ranges.

Figure 2 represents an example of the recalculation of experimental ADIF data for as-annealed Cu - 1.3 at.% Ni alloy at room temperature to the intrinsic values. The intrinsic ADIF curve obtained for the same alloy at a sonic frequency [1] is shown as well. Significant difference in the intrinsic ADIF values, measured in sonic and ultrasonic ranges, implies the existence of pronounced frequency dependence of the ADIF for this alloy. Provided time

Figure 1: Amplitude-dependent decrement δ_h Young's modulus defect $(\Delta E/E)_h$ and their ratio $r = \delta_h / (\Delta E/E)_h$ vs oscillatory strain amplitude ε_m at RT for prestrained Cu-Ni single crystals with different Ni concentration: 1.3 at.% (a), 2.3 at.% (b), 3.2 at.% (c), and 7.6 at.% (d).

Figure 2: Strain amplitude dependence of decrement of a Cu -1.3 at.%Ni sample at RT (curve 1) recalculated to intrinsic values (curve 2) according to equation (1). Resonant frequency was about 100 kHz. The approximation of curve 1 by two polynomials is shown by the solid line 3. The dashed line 4 is the intrinsic decrement of the same alloy at a frequency of about 800 Hz [1].

Figure 3: Intrinsic amplitude-dependent decrement δ_h (a), Young's modulus defect $(\Delta E/E)_h$ (b), and oscillatory anelastic strain $\epsilon_{an} = \epsilon (\Delta E/E)_h$ (c) of Cu - 1.3 at.% Ni vs oscillatory elastic strain ϵ at different temperatures during heating.

dependence of the ADIF is lacking, frequency dependence of the ADIF indicates that thermally activated motion of dislocations is involved.

Effect of temperature on the oscillatory strain dependences of the intrinsic ADIF (a), YMD (b), and oscillatory anelastic strain derived from the YMD (c) of Cu - 1.3 at.% Ni is shown in Fig.3. Different temperature dependence of the ADIF and YMD is evident from Fig.3. Thus, generally speaking, YMD rather than ADIF is needed to derive anelastic dislocation strain in a wide temperature-strain amplitude range. The low-amplitude stage of the strain amplitude dependences becomes more pronounced at low temperatures. Since crystals with different Ni content show similar rather weak amplitude dependence at the first stage, it is reasonable to attribute this stage to the hysteretic low-amplitude background damping [3]. According to [3] the mechanism of hysteretic background damping is efficient in the range of impurity contents, used in the present study. The second stage of the strain amplitude dependence, exhibiting rather strong temperature and concentration dependence, can be attributed to the thermally activated pinning-depinning of dislocations from point obstacles distributed in glide planes. The third stage, similar to the first one, exhibits rather weak temperature dependence. Two possible reasons for that can be mentioned: i) a manifestation of dislocation interactions at high anelastic strains $\sim 10^{-6}$; ii) dynamic effects at high anelastic strain rates $\sim 1s^{-1}$.

Fig. 4 shows temperature dependences of the microyield stress σ_e [4] at different fixed values of anelastic

Figure 4: Temperature dependence of the microyield stress σ_e for Cu-1.3at%Ni crystal in relative units at different fixed levels of anelastic strain ϵ_{an} : 10^{-10} (1), 10^{-8} (2), and $5 \cdot 10^{-7}$ (3). The values of σ_e at T = 255 K were taken as unity.

Figure 5: Stress amplitude dependences of amplitude-dependent decrement δ_h (a), anelastic strain amplitude $\epsilon_{an} = \epsilon_m (\Delta E/E)_h$ (b), and acoustoplastic effect magnitude ΔP (c) simultaneously measured during deformation of Cu-Ni single crystals with different Ni concentration. The acoustoplastic effect magnitude for the i -th sample ΔP_i is recalculated to that of a Cu - 7.6 at.% Ni sample according to equation (2).

4. CONCLUSIONS

- 1) A complicated functional form of the ADIF and YMD strain amplitude dependence observed in wide temperature and strain amplitude ranges argues for their multicomponent nature. Both thermally activated and athermal processes contribute to the ADIF and YMD.
- 2) Different defect structure levels are basic sources of the ADIF and APE: dislocation - point obstacle and dislocation - dislocation interactions, respectively.

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strain. Different σ_e temperature dependence is evident for different ϵ_{an} . Therefore, the similarity between micro- and macroyield stresses [4] may be valid only within limited strain amplitude ranges. The explanation of the similarity by the manifestation of some universal influence of temperature on mechanisms of micro- and macroplasticity [5] appears to be an oversimplification, since the analysis of the specific ADIF mechanisms involved is required.

Magnitudes of the ADIF (a), anelastic strain amplitude (b), and APE (c) simultaneously measured during deformation are represented in Fig.5 vs shear stress amplitude $\tau_m = \sin\chi \cos\lambda E \epsilon_m$. The ADIF and reversible oscillatory anelastic strain decrease drastically with the increase of Ni content. The dependence of the APE magnitude on Ni content is much less pronounced and reverses with the stress amplitude increase. Different impurity content dependence of the reversible oscillatory anelastic strain and irreversible plastic strain giving rise to the APE agrees well with supposition [6] that basic mechanisms of the ADIF and APE are different. The gain in the APE magnitude with the increase of the impurity content at high stress amplitudes can be explained by distinctions between dislocation structures formed during deformation of different alloys.