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TITLE THE RESPONSE OF SINGLE CRYSTAL AND POLYCRYSTAL
NICKEL TO QUASISTATIC AND SHOCK DEFORMATION

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THE RESPONSE OF SINGLE CRYSTAL AND POLYCRYSTAL NICKEL TO QUASISTATIC AND SHOCK DEFORMATION

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ABSTRACT

Stress-strain measurements and TEM observations in shock-deformed nickel single and polycrystalline samples are reported. When the stress measurements are normalized by the appropriate Taylor factor, the shock induced strain hardening is shown to be similar in all materials studied.

KEYWORDS

Shock; deformation; nickel; strain hardening; single crystal.

INTRODUCTION

Shock wave deformation in metals remains a fascinating field of study because the shock wave provides a probe of a material's ability to undergo plastic deformation under severe conditions of strain rate, temperature, and pressure. There have been several comprehensive reviews of the influence of shock wave loading on microstructural development in metals (Leslie, 1973; Murr, 1981). While in many previous studies the microstructural observations have been combined with microhardness measurements, there have been fewer investigations of the post mortem stress-strain behavior of shock deformed metals. In this study we report on an investigation of shock-wave hardening and shock-induced microstructural alterations observed in shock deformed nickel single and polycrystals. The objective of the study was to investigate whether crystal orientation or crystallite size had significant influences on the shock-wave response.

EXPERIMENTAL PROCEDURE

The materials studied in this investigation were $\langle 100 \rangle$ and $\langle 111 \rangle$ single crystal nickel and polycrystalline high-purity nickel (Ni 270). The chemical compositions of the materials differed only in the carbon concentrations,

which were 0.43, 0.72, and 0.055 wt % for the $\langle 100 \rangle$, $\langle 111 \rangle$, and Ni 270 materials, respectively. The high carbon concentrations in the single crystals, particularly the $\langle 111 \rangle$, are important because carbon is an effective interstitial strengthener in nickel and because carbon mobility leads to dynamic strain aging. The Ni-270 was cross-rolled at room temperature in multiple passes from 38 mm to 12.5 mm and then recrystallized to form an equiaxed grain structure. Recrystallization at 450°C for 2 hours yielded a 40 μm grain size (hereafter fine grained, FG) while recrystallization at 450°C for 2 hours followed by 8.5 hours at 600°C yielded a 225 μm grain size (hereafter large grain, LG). The texture of the Ni270 samples were measured to be essentially random with a weak texture, consistent with the extensive cross-rolling. The Taylor factors for the Ni270-FG and Ni270-LG for through-thickness deformation were 3.17 and 3.19, respectively. The quasistatic and shock-loading samples were machined with the compression and shock direction in the through-thickness plate direction.

Single crystal samples, with the compression and shock direction (through-thickness direction) parallel to the $[100]$ and $[111]$ directions were cut from the single crystals using Electro-Discharge Machining (EDM) to minimize local damage at the machined surfaces. Taylor factors for deformation in the $\langle 100 \rangle$ and $\langle 111 \rangle$ single crystals are 2.45 and 3.67, respectively.

Shock recovery experiments were performing using a 40-mm single-stage gas gun. A detailed description of the "soft" recovery shock-loading techniques have been previously published (Gray *et al.*, 1989). The nickel samples (4.76 mm thick by 12 mm in diameter) were shock loaded in vacuum to 10 GPa for a pulse duration of 1 μs through the impact of a 2.62 mm thick nickel flyer plate fixed to a projectile traveling at 458 m/s. The recovered shock-loaded samples possessed residual strains of $\approx 2\%$ (final sample thickness divided by the initial thickness). Compression samples were EDM machined from the central portion of the recovered shock-loaded sample with the compression axis parallel to the shock direction. The "post-mortem" mechanical behavior was studied by reloading the samples at a strain rate of 0.001 s^{-1} and at temperatures of 295K and 77K. Compression tests were also performed on samples in the as-received condition. Samples for TEM were sectioned with through-thickness orientations, thinned, punched, electropolished, and examined using a JEOL 2000EX at 200kV.

RESULTS AND DISCUSSION

The substructures of the shock loaded single crystal and polycrystalline nickel samples were similar to those documented by Murr (1981) in polycrystalline nickel, consisting of uniform dislocation cells. The dislocation cells observed in the $\langle 100 \rangle$ and $\langle 111 \rangle$ single crystals were slightly smaller ($\approx 0.4\text{-}0.5\ \mu\text{m}$) than those observed in the polycrystalline nickel ($\approx 0.6\text{-}0.8\ \mu\text{m}$), and there was no readily apparent grain size effect in Ni270. The substructures are similar to those typically observed in nickel (and other high stacking fault energy fcc metals) deformed at lower strain rates, which illustrates that, although the strain rates in a 10 GPa shock wave in nickel exceed those in typical quasistatic mechanical tests by ≈ 10 orders of magnitude, deformation in both shock loading and low strain rate testing occur by the same micromechanical processes.

Figure 1 shows typical stress strain curves for all of the conditions studied. The data are plotted as resolved shear stress (σ/M) versus resolved shear strain (ϵ/M), where M is the appropriate Taylor factor. The data for the shock loaded samples is offset by the transient (total) strain during the

Fig. 1. Stress-strain curves at $\dot{\epsilon}=0.001 \text{ s}^{-1}$, $T=295\text{K}$ (closed symbols), and $T=77\text{K}$ (open symbols) in as-received and shock-deformed samples.

shock, calculated from $[4/3 \log V_0/V_f] \cdot M$ where V_0/V_f is the ratio of the initial sample volume to the volume at 10 GPa. (Before dividing by the Taylor factor, the equivalent uniaxial strain in the 10 GPa shock wave is 6.42%.) When plotted on these coordinates, the shock induced strength increase is i) found to greatly exceed the strength increase in samples deformed quasistatically to the same strain level and ii) roughly similar for all samples tested. The former point is clearer in the polycrystalline samples than in the single crystals. In fact, the measurements at $T=295\text{K}$ in the $\langle 111 \rangle$ crystal show no shock-induced flow stress increase, but these data are complicated by the appearance of jerky flow (present also at $T=77\text{K}$) and the contribution of dynamic strain aging. The rate-dependent hardening shown in Fig. 1 is similar to that observed in copper (Follansbee, 1988) and nickel (Follansbee *et al.*, 1989) and illustrates that strain hardening is a rate-dependent process. Table 1 lists the average yield strengths for all the samples tested. Also listed in Table 1 is the strength differential $\Delta\sigma$, which is the yield strength in shock-deformed samples minus the yield strength measured in the as-received condition. The data in Table 1 show that the strength increase is roughly equivalent in all the materials tested.

Table 1. Summary of compression test results

Material	As-Received		Shock loaded		$\Delta\tau$	
	T=77K	T=295K	T=77K	T=295K	T=77K	T=295K
<100>	61	52	181	145	120	93
<111>	65	54	179	124	114	70
Ni 270 FG	27	21	156	134	129	113
Ni 270 LG	20	12	143	112	123	100

SUMMARY

Compression measurements in as-received and shock deformed nickel single and polycrystals show that when the stress measurements are normalized by the operative Taylor factor, the extent of shock-induced hardening is found to be similar in all materials tested. This result differs from that found by Meyers *et al.* (1983) in Fe-Ni and Fe-Cr-Ni.

The shock-induced hardening greatly exceeds the hardening for an equivalent plastic strain at quasistatic strain rates. While this has been documented previously, we emphasize that this results from the strain-rate dependence of strain hardening rather than from any change in deformation mechanism.

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