Work Softening in Shock-Loaded Nickel

MARC A. MEYERS

It is shown that nickel shock loaded to a pressure of 20 GN/m^2 (200 Kbar) and pulse durations of 1.2, 2.4 and 10.1 μsec exhibits work softening upon subsequent uniaxial tensile deformation at ambient temperature and nominal strain rate of 10^{-1} sec^{-1}. When the tensile deformation on the preshocked sample is conducted at 77 K, work softening is inhibited. It is proposed, in accordance with the findings of Longo and Reed-Hill (Metallography, vol. 7, p. 181, 1974) that work softening is due to dynamic recovery that during conventional deformation replaces the shock-loading substructure (small cells, with ill-defined walls) by the substructure characteristic of the conventional deformation at the imposed conditions. Since the quasi-static deformation substructure at 77 K is quite similar to the shock-induced one, work softening is inhibited at that temperature.

THE term "work softening" was originally introduced by Polakowski[1] to designate the peculiar yielding behavior of a metal in a tensile test, in opposition to the normally encountered "work hardening." It is solely due to substructural rearrangements taking place during deformation, produced by dynamic recovery or dislocation elimination through dynamic recrystallization. In contrast, the more general term "flow softening"[2] encompasses both geometric softening (due to the rotation of single crystals towards an orientation with a larger Schmid factor) and adiabatic softening (due to adiabatic heating during deformation), in addition to work softening.

Work softening was initially found[3] when aluminum was first deformed at a temperature and then at a higher temperature, the two temperatures having different characteristic substructures. More recently, Longo and Reed-Hill[4-6] have shown that a number of metals can exhibit work softening. Luft *et al*[7] found work softening in predeformed Mo single crystals. Work softening has also been found when both deformations are imparted at the same temperature, but different strain rates.[8-11] Specifically, it takes place when dynamic recovery is inhibited during prestraining (the first deformation stage). This is possible at high strain rates. Of particular importance to the present study is the work of Gülec and Baldwin.[11] They predeformed nickel 270 in a Magneform unit, achieving strain rates of the order of 10^2 sec^{-1}. Upon restraining at conventional strain rates for tensile tests, they were able to detect decreases in the work hardening rates and yield point formation; these changes were, however, modest in comparison to the ones obtained by straining at two different temperatures.

Since shock loading provides the fastest possible means of deformation—strain rates of 10^5 sec^{-1} or more can be reached—and since substructures differing substantially from conventional deformation substructures can be achieved, it was thought worthwhile to verify whether preshocked metals could exhibit work softening upon subsequent deformation.

MARC A. MEYERS, formerly with the Department of Materials Science and Engineering, Instituto Militar de Engenharia, Rio de Janeiro, Brazil, is now Assistant Professor, Department of Metallurgical Engineering, South Dakota School of Mines and Technology, Rapid City, SD 57701.

Manuscript submitted February 14, 1977.

EXPERIMENTAL PROCEDURE

Nickel was chosen for the following reasons:

a) Longo and Reed-Hill[6] showed that predeformation at 77 K and reloading at 370 and 510 K promoted work softening.

b) The substructure of shock-loaded nickel does not present any complicating features such as deformation twins (up to 25 GN/m^2) and phase transformations.

A 6 mm thick nickel plate (impurities in wt pct: Si: 0.04, Fe: 0.13, S: 0.09, C: 0.05, Cu < 0.01) was cold rolled down to 3 mm. Rectangular strips, with 100 × 15 mm and having the largest dimension along the rolling direction, were machined therefrom. These strips were annealed at 1023 K for 1 h in vacuum, inside sealed quartz capsules; the resulting grains were equiaxial, and their mean diameter, determined by the linear intercept method, was 55 μm. They were then shock loaded, as described in the next paragraph; tensile samples with useful gage dimensions of 26 × 7 × 3 mm were machined from the shocked strips. For the tensile samples tested in the annealed condition the machining was conducted prior to annealing. The reason for this is that the annealed samples were extremely soft, so that machining without introducing changes in the tensile properties was difficult to accomplish.

Shock loading was performed by plate impact; the plate was accelerated by explosive means using a mousetrap-type assembly with the driver plate and system being initially parallel. Plastex-P, a PETN base, laminated explosive fabricated in Brazil and similar to Detasheet C, was used both for the main charge and for the line and plane-wave generators. The line-wave generators were made by sandwiching the explosive between two especially built dies, in which the characteristic pattern of holes had been drilled, and inserting a punch in each hole. Five nickel strips were positioned side-by-side in each system, and protected by a 3.2 mm thick 304SS cover plate, by an anvil (100 × 125 × 20 mm) and spall plate (180 × 210 × 20 mm) on the bottom, and lateral bars on the sides. The purpose of the lateral bars and spall plate was to minimize wave reflections. Copper driver plates were used. The system dimensions were calculated according to Orava and Wittman.[12] The author has used similar systems in previous investigations.[13,14] The experiments were conducted at a peak pressure of 20 GPa (200 Kbar) and pulse durations of 1.2, 2.4 and 10.1 μsec; the associated mean rarefaction rates were −52, −33, and −11 GPa/μsec, respectively. This was accomplished by varying the driver plate thickness, but maintaining the same ratio between explosive and driver plate masses. The separation between system and driver plate prior to explosion varied according to driver plate thickness, so that it could be accelerated to its steady-state velocity; the initial gaps were 5, 10, and 20 mm for the 1.2, 2.4 and 10.1 μsec events, respectively. The explosions drove the system into a water tank with sand bottom that served both as a cooling and deccelerating medium.

For tensile testing, an Instron TT-DM universal testing machine, equipped with a 20,000 lb cell was used. All tests were conducted at the nominal strain rate of 10^{-4} sec^{-1}. The room temperature tests were conducted in air; the one at 77 K was done in a liquid nitrogen bath. The transmission electron microscopy was done in a JEOL JEM-100B unit, on properly thinned discs having the surface perpendicular to the direction of propagation of the shock wave.

RESULTS AND DISCUSSION

Figure 1 shows the engineering stress-strain curves at ambient temperature for the shock-loaded samples. The striking feature common to the three curves is the near coincidence between yield stress and U.T.S. Other differences between the curves with regard to pulse duration difference (yield stress increase with pulse duration) will be discussed in a separate paper. Necking was initiated at the onset of yielding and, since no work hardening—but work softening—took place in that region, it became increasingly prone to deformation. For this reason almost all the deformation was confined to the neck, and a true-stress, true-strain plot would be meaningless. Work softening is clearly evident for the three different pulse durations. The curves are similar to the ones presented by Longo and Reed-Hill.[6] Particularly, the third curve in Fig. 3 of that reference, where the sample underwent a predeformation of 0.26 at 77 K, has a yield stress of around 400 MN/m^2, a total eng. elongation of 0.12, and approximately the same shape as the curve of the 2.4 μsec shock-loaded sample.

On the substructural level, what would then be the changes responsible for work softening? The substructure of shock-deformed nickel is well characterized in the literature[15-19] and consists of tangled dislocations in cellular arrays. Table I shows the cell diameters found by the different investigators and by this study (1.2 μsec pulse duration event). The cell size decreases as the shock pressure and consequently the amount of deformation, increases. This is consistent with conventional deformation. In spite of the different experimental conditions, and therefore different pulse durations and rarefaction rates used by the various authors, there is a satisfactory agreement of cell diameter at the pressure used in this investigation: approximately 0.2 μm. This was confirmed by measurements made in this study, as reported in Table I. Murr and Huang[19] measured the cell sizes for samples shock-loaded to the same pressure and different pulse durations, and found a slight decrease of cell size with increasing pulse duration.

On the other hand, conventional deformation by tension at room temperature generates a substructure consisting of larger cells with better defined walls. After tensile strain of 0.20, Nolder and Thomas[15] report a mean cell diameter of 0.825 μm. Longo and Reed-Hill,[6] after tensile deformations inducing strains of 0.14 and 0.25 at 510 K, found cells with diameters of 1 and 0.88 μm, respectively. These cell sizes are about four times the cell size for nickel shocked at 20 GN/m^2. So, the best explanation for work softening is the following. Once the preshocked sample is loaded up to its yield stress in the tensile test, the dislocations start moving and the shock-induced substructure is replaced by the substructure that would normally form at that stress level, temperature and strain rate. In other words, the plastic deformation will not generate new dislocations and induce them to interact (as in the classic work hardening situation) but will reorganize the existing substructure, transforming it into the one that is more stable under the imposed conditions. This process does not take place uniformly over the length of the tensile samples. Once initiated at a certain region, it induces "softening" there, with consequent concentration of deformation, and therefore necking. The increase of stress due to necking only accelerates the "softening," and the tensile sample presents two apparently contradictory features: a very small total elongation with a very large reduction in the area. This can be seen in Fig. 2: the total elongation at ambient temperature is reduced from 0.8 to 0.1 by preshocking, but both curves

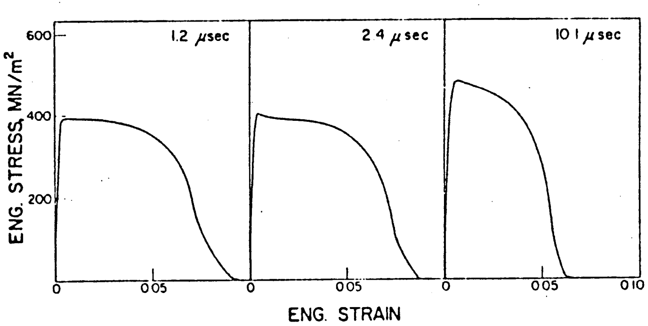

Fig. 1—Engineering stress-engineering strain curves for nickel shock-loaded at a pressure of 20 GN/m^2 (200 Kbar) and pulse durations of 1.2, 2.4 and 10.1 μsec.

Table I. Cell Diameters for Shock-Loaded Nickel

Pressure (GN/m^2)	Mean Cell Diameter (μm)	Investigators
7	0.53	Nolder and Thomas[15]
13	0.315	Nolder and Thomas[15]
25	0.167	Nolder and Thomas[15]
7	0.25 (estimate)	Kressel and Brown[16]
7	0.5	Trueb[17]
32	0.14	Trueb[17]
8	0.8	Murr, Vydyanath and Foltz[18]
18	0.3	Murr, Vydyanath and Foltz[18]
24	0.2	Murr, Vydyanath and Foltz[18]
46	0.1	Murr, Vydyanath and Foltz[18]
25(0.5 μsec)	0.19	Murr and Huang[19]
25(6 μsec)	0.15	Murr and Huang[19]
20(1.2 μsec)	~0.2	Present Study

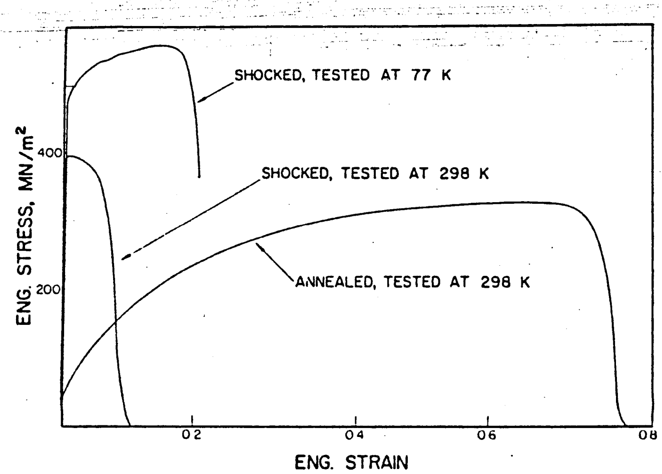

Fig. 2—Engineering stress-engineering strain curves for annealed nickel at ambient temperature and for preshocked nickel (20 GN/m^2 pressure, 1.2 μsec pulse duration) at 77 K and ambient temperature.

terminate at zero load. This means that the separation of the tensile samples into two parts occurs at zero load, or when the cross-sectional area at the neck has been reduced to zero. Therefore, the fracture in the preshocked sample retains the chisel-like appearance when conventionally tested in tension at ambient temperature.

Longo and Reed-Hill[6] describe the low-temperature deformation substructure, and it is strikingly similar to the shock-induced one. The cells are smaller and the cell walls are less well defined than the substructures generated at 510 K for the same strains. This is confirmed by Nolder and Thomas,[15] who found the quasistatically (77 K) and shock-induced substructures very similar. With this in mind, a critical experiment was conducted: if the quasistatic (77 K) and the shock (ambient temperature) substructures are not very different, then work softening would be inhibited in a tensile test at 77 K conducted on a preshocked sample. This is shown in Fig. 2. No work softening is observed (there is only a slight irregularity in the curve) and the material exhibits a reasonable work-hardening rate. No major substructural adjustments are necessary, once the yield stress is reached, and the substructure build-up proceeds almost normally. Of course, one has to recognize that shock loading induces point defect concentrations well in excess of those generated by conventional deformation.[16] However, according to Longo and Reed-Hill[6] the presence of excess point defects would render the material even more amenable to work softening.

CONCLUSIONS

It is shown that nickel shock loaded at a pressure of 20 GN/m^2 (200 Kbar) and pulse durations of 1.2, 2.4 and 10.1 μsec undergoes work softening upon being deformed in tension, at ambient temperature, at a nominal strain rate of 10^{-4} sec^{-1}. It is proposed that the work softening is due to substructural rearrangement occurring at the onset of plastic deformation in the tensile test; the shock-induced substructure (small, not very well defined cells) is replaced by the substructure characteristic of the conditions imposed by the tensile test (larger cells with well defined walls). Consequently, necking starts immediately at the yield point. The small difference between tensile and yield stress found by Dieter[20] in shock-loaded nickel can be explained by work softening. Appleton and Waddington[21] shock loaded copper, nickel and aluminum at various pressures, subjecting the samples to subsequent tensile tests at various strain rates. They found tensile curves that clearly indicate work softening. They found that nickel and copper have the same shape of tensile curves but only present the ones for copper. They interpret the results in terms of dislocation multiplication, failing to associate them with work softening. It is probable that work softening is a more general phenomenon in shock-loaded metals, and it could be used in rationalizing their mechanical response. For instance, shock-loaded Inconel 718 (Ref. 22) exhibited consistently lower elongations than the same material cold-rolled to thickness reductions providing approximately the same yield strength. However, the shocked conditions tended to exhibit higher reductions in area at fracture. This combination of lower elongation and higher reduction in area seems to indicate work softening.

ACKNOWLEDGMENTS

This work was supported by the Brazilian Army, FINEP and BNDE. Thanks are due to Professors J. R. C. Guimarães and W. P. Longo for helpful discussions. The aid of the Marambaia Proving Grounds personnel and of C. A. O. Soares are gratefully acknowledged.

REFERENCES

1. N. H. Polakowski: *J. Iron Steel Inst.*, 1952, vol. 169, p. 337.
2. J. J. Jonas and M. J. Luton: Proc. 21st Sagamore Conference, Syracuse University Press, 1974.
3. A. H. Cottrell and R. J. Stokes: *Proc. Roy. Soc.*, 1955, vol. A233, p. 17.
4. W. P. Longo and R. E. Reed-Hill: *Scr. Met.*, 1970, vol. 4, p. 765.
5. W. P. Longo and R. E. Reed-Hill: *Scr. Met.*, 1972, vol. 6, p. 833.
6. W. P. Longo and R. E. Reed-Hill: *Metallography*, 1974, vol. 4, p. 181.
7. A. Luft, J. Richter, K. Schlaubitz, C. Loose, and C. Muhlhaus: *Mater. Sci. Eng.*, 1975, vol. 20, p. 113.
8. B. Langenecker: *Acta Met.*, 1961, vol. 9, p. 937.
9. F. H. Hammad and W. D. Nix: *Trans. ASM*, 1966, vol. 52, p. 94.
10. S. Sakui, K. Sato, N. Abe, T. Kakuma, and T. Mori: *Supplement Trans. JIM*, 1968, vol. 9, p. 899.
11. A. S. Gülec and D. H. Baldwin: *Met. Trans.*, 1973, vol. 4, p. 1315.
12. R. N. Orava and R. H. Wittman: Proc. 5th Int. Conf. on High Energy Rate Fabrication, U. of Denver, CO, June, 1975.
13. M. A. Meyers and R. N. Orava: *Met. Trans. A*, 1976, vol. 7A, p. 179.
14. H.-J. Kestenbach and M. A. Meyers: *Met. Trans. A*, 1976, vol. 7A, p. 1943.
15. R. L. Nolder and G. Thomas: *Acta Met.*, 1964, vol. 12, p. 227.
16. H. Kressel and N. Brown: *J. Appl. Phys.*, 1967, vol. 38, p. 1618.
17. L. F. Trueb: *J. Appl. Phys.*, 1969, vol. 40, p. 2976.
18. L. E. Murr, H. R. Vydyanath, and J. V. Foltz: *Met. Trans.*, 1970, vol. 1, p. 3215.
19. L. E. Murr and J.-Y. Huang: *Mater. Sci. Eng.*, 1975, vol. 19, p. 115.
20. G. E. Dieter: "Response of Metals to High Velocity Deformation," P. G. Shewmon and V. F. Zackay, eds., pp. 409-45, Interscience, New York, 1961.
21. A. S. Appleton and J. S. Waddington: *Phil. Mag.*, 1965, vol. 12, p. 273.
22. M. A. Meyers: Ph.D. Dissertation, University of Denver, Denver, Co., 1974.