the uranium atoms order antiferromagnetically [4]. In table I we compare the results of magnetic measurements on our powdered samples with literature data [4], which differ somewhat in the effective moments. The marked increase of the literature magneton numbers on going from the sulphide to the telluride may be due to deviations from stoichiometry.

TABLE I Magnetic data of uranium oxychalcogenides: (a) own measurements; (b) according to [4].

	T_N(°K)	θ	$_p$(°K)	n_p(μBohr)
UOS (a)		−55		2.8
(b)	55	−51		2.24
UOSe (a)	72	−30		2.9
(b)		−35		2.44
UOTe (a)	~157	−60		2.9
(b)	162	−56		3.35

Acknowledgements
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   Note that there must be a misprint in the *c*-value given for ThOTe; the correct value should read $c = 7.544$ kX.
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Recovery Behaviour of Nickel Strain-Hardened by Impulse Loading

The experimental results of a hypervelocity-impact study have been reported by Gehring *et al* [1] for several metals, including type-200 nickel, which were subjected to intense impulse loads. The deformation of nickel under such conditions has been amplified in detail more recently by Meyers [2]. The method of strain-hardening employed for that study, as reported in both references, was the impact-loading of four-inch-cube (1.0 in. = 2.5 cm) targets of the metals of interest by aluminium projectiles accelerated to very high velocities. The microstructures within the impact-affected region surrounding each resulting target crater were correlated with the stress levels* effecting them for nickel (figs. 1 and 2) and four other metals studied. In addition, a strain profile within the impact-affected region in the nickel target was determined metallographically and correlated with its micro-indentation hardness profile.

Four-inch cubes of type-200 nickel were prepared for very high impulse loading by projectile impacts at a very high velocity (~ 7.2 km/sec or 23900 ft/sec), as described in detail elsewhere [1, 2]. The microstructure of the nickel targets was predominantly equiaxed (ASTM grain size 2-3) with some twinning

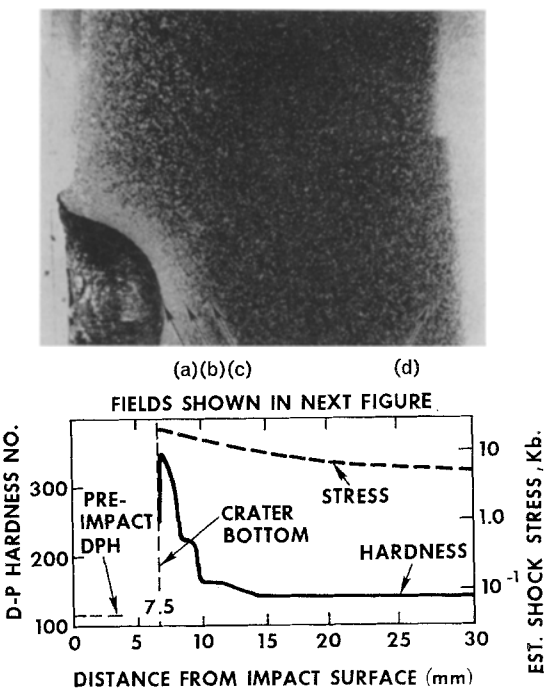

Figure 1 Macrostructure and D-P micro-indentation hardness of impact-affected region in nickel target (~ 3500-joule impact) [1, 2].

*A 0.15 g aluminium-projectile impact into the nickel target at ~ 7.2 km/sec deposited ~ 3500 joule of kinetic energy at the impact site, and produced an initial stress of ~ 1.5 Mb (22×10^6 psi where 1.0 psi = 7.0×10^{-3} kg/cm^2).

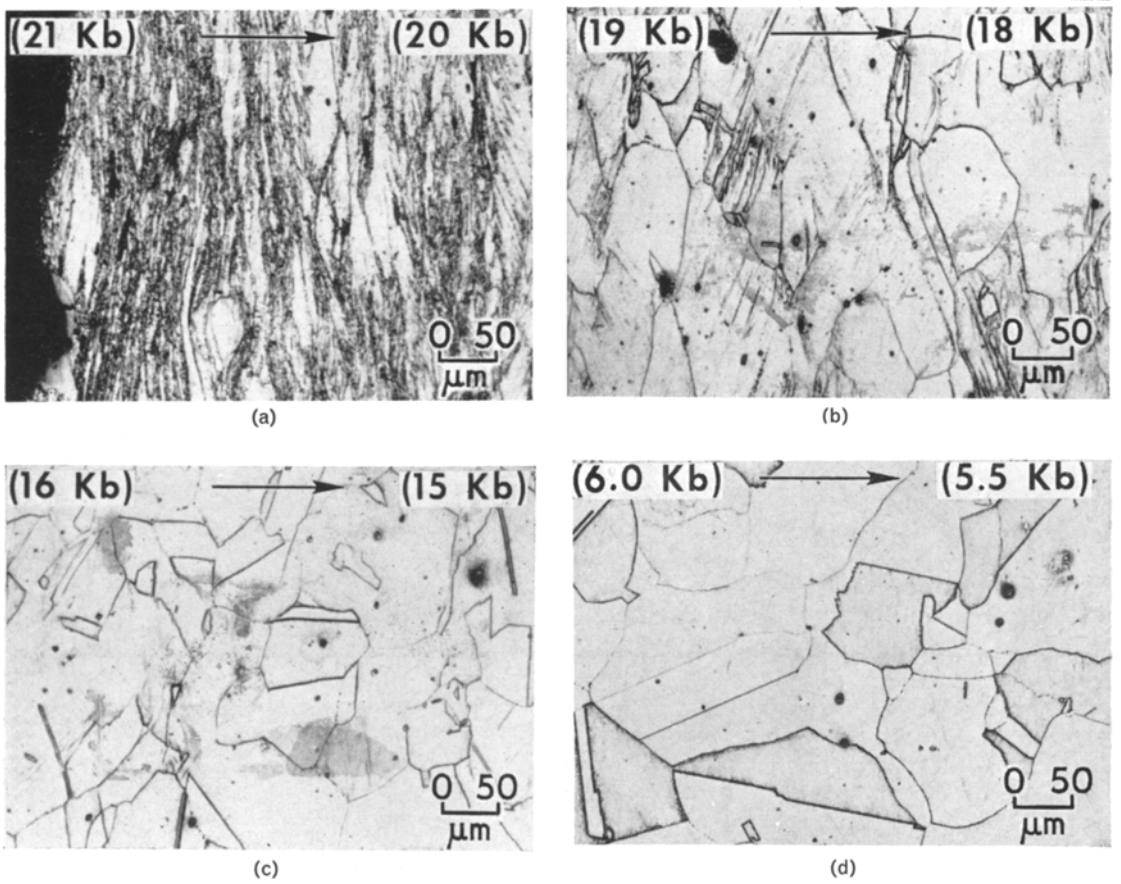

Figure 2 Microstructures of fields (a)-(d) within impact-affected region in nickel target, as indicated in fig. 1 (~ 3500 – joule impact) [1, 2].

present prior to loading. Fig. 2d illustrates the pre-impact microstructure as well as the unaffected material following impact.

After impact loading and crater formation, selected target cubes were sectioned, leaving one full-size hemisphere of the crater, which was sectioned again into two near-quarter-sphere specimens. All sectioning was performed carefully to avoid significant changes in target condition due to localised heating. Each specimen then was subjected to a recovery-annealing programme. Progress of the recovery in each specimen was followed by noting changes in micro-indentation hardness profiles along crater radii through the impact-affected region adjacent to the crater. The effect of surface lapping and other metallographic preparation was evaluated prior to recovery treatment and was found to be negligible.

Four specimens were selected for recovery anneals, each at one of the four following temperatures: 410, 465, 515, and 588° C.

Two thermocouples to monitor specimen temperature were embedded below the surface at two different sites on each specimen, both well away from the impact-affected region. Each specimen was subjected to successive annealing and held within 3° C of the specified temperature for various times at a single setting on the temperature controller. Reproducibility of specimen temperature for successive annealing periods was determined to be $\pm$ 5° C, except for the 515° C recovery specimen for which reproducibility was $\pm$ 10° C. The specimens were protected from reaction with furnace atmosphere during furnace treatment to preserve the polished section as much as possible for photomicrography.

The specimen subjected to recovery treatment at 588° C completely recrystallised after only 0.83 h at temperature, and was not considered of further interest for this study. The specimens

annealed at 410, 465 and 515° C showed no signs of recrystallisation at magnifications up to 1500×. The successive micro-indentation hardness profiles for recovery treatments at 410, 465, and 515° C are shown in figs. 3, 4, and 5, respectively. No significant microstructural

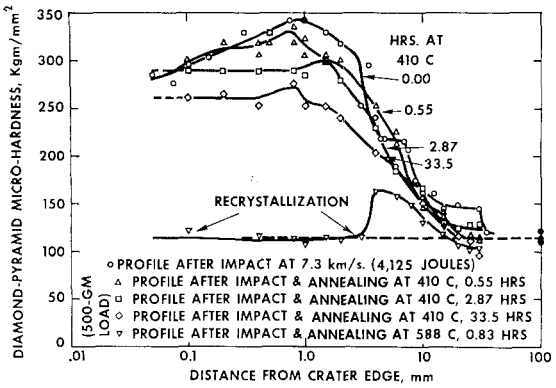

Figure 3 Micro-indentation hardness profile changes during recovery at 410° C within impact-affected region of nickel target (specimen 091-B; 3590 – joule impact).

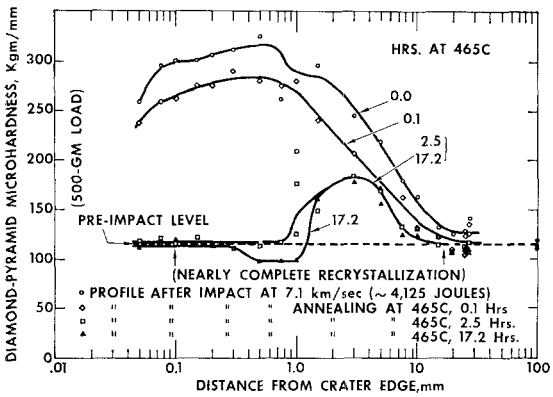

Figure 4 Micro-indentation hardness profile changes during recovery at 465° C within impact-affected region of nickel target (specimen 121-A; 3400 – joule impact).

differences were observed between the 465 and the 515° C specimens. Although no microstructural evidence of recrystallisation was detected by microscopic inspection in the 515° C specimen, the "recovery-factor" curve in fig. 6 (which will be discussed below) indicates that incipient recrystallisation occurs between the 1.9 and the 23.8 h annealing times. Hence, all 515° C data for recovery times longer than ~2 h have been considered affected by recrystallisation in addition to the initially active recovery processes.

The shock or plastic-stress wave resulting from projectile impact may be assumed to have propagated with spherical symmetry outward from the impact site to the other surfaces of the isotropic nickel target. As the stress wave propagated, its peak-stress amplitude dropped progressively from the maximum value at impact as a result of internal friction or energy-absorbing mechanisms within the target material. The progressive attenuation of the stress wave produced a strain gradient radially about the impact site (from the crater wall to the point at which the stress amplitude produced elastic strain only) [3, 4]. This gradient in strain-hardening has been illustrated in figs. 3-5 by the micro-indentation hardness gradient from a maximum value down to the pre-impact hardness level at a distance of ~ 30 mm beyond the crater wall. (Beyond the 30 mm distance the stress wave produced a sonic disturbance only.) Since the maximum hardness value occurs ~ 1 mm from, rather than at, the crater wall, the target material within the region between the wall and the maximum hardness position has already experienced recovery in amount decreasing outward from the crater wall. This pre-annealing recovery resulted from appreciable heating effects produced by the shock front until shock conditions disappeared upon subsequent attenuation of stress-wave energy [5, 6]. (Since the impact energy producing the crater with the unannealed hardness profile shown in fig. 3 was slightly higher than the impact energy producing the crater with the unannealed hardness profile shown in figs. 4 and 5, the hardness maximum in fig. 3 is slightly higher than that in figs. 4 and 5.)

With the foregoing discussion in mind, a position 2.5 mm behind the unannealed hardness maximum was selected for reference to monitor the progress of recovery during successive anneals. To provide a means of expressing recovery quantitatively, a recovery factor, R_f, was defined as follows:

$$R_f = \frac{(H_m - H)}{(H_m - H_o)} \quad (1)$$

where H_o is the pre-impact hardness level; H_m is the post-impact unannealed hardness level at the reference position; H is the post-impact hardness level produced by annealing at the reference position. Values of R_f produced by annealing at 410, 465, or 515° C have been

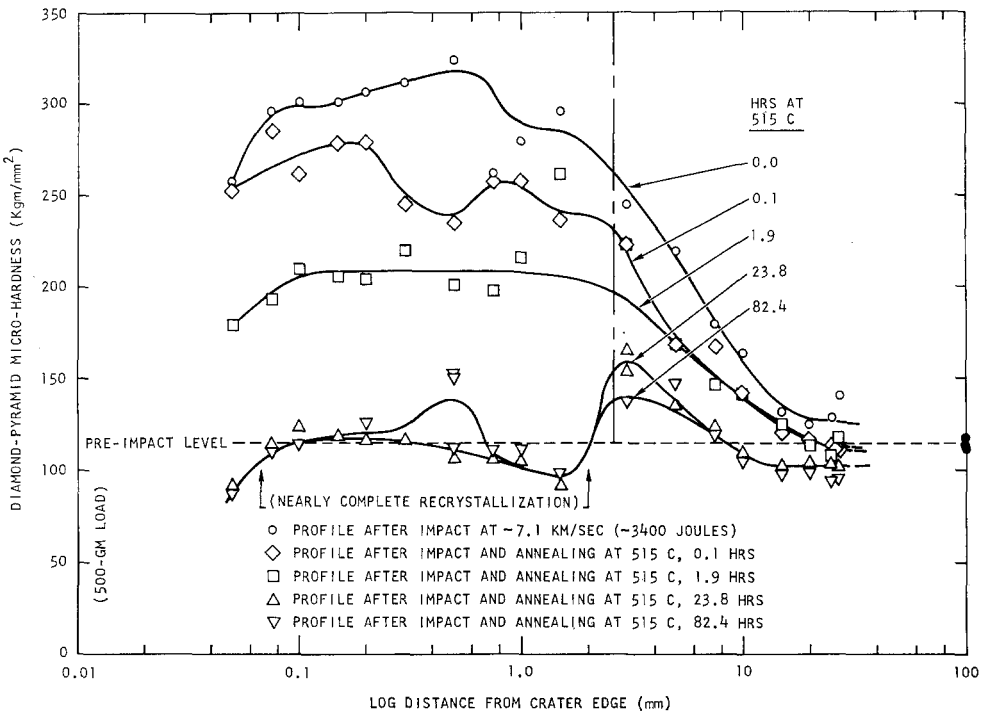

Figure 5 Micro-indentation hardness profile changes during recovery at 515° C within impact-affected region of nickel target (specimen 121-B; 3400 – joule impact).

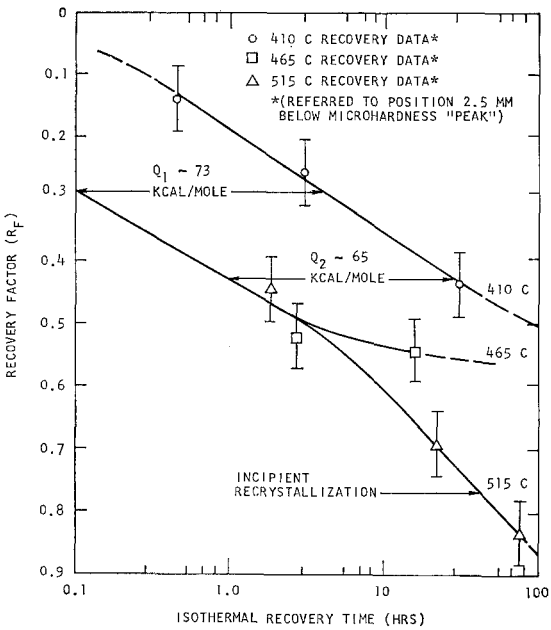

Figure 6 Recovery factor, R_f, curves as functions of time at annealing temperature for nickel targets (~ 3500 – joule impacts).

plotted as a function of time at each temperature in fig. 6. The curves for recovery at 410 and 465°C exhibit different recovery rates for the first hour or so. Then these curves approach slightly different R_f levels between 0.55 and 0.60, respectively, which would appear to remain unchanging with additional annealing time. The 515° C recovery curve essentially seems to coincide with the 465° C curve for about the first hour at annealing temperature. However, additional annealing at 515° C produced such accelerated recovery in the specimen that incipient recrystallisation was suspected to have been superposed upon the initial recovery mechanisms.

If it is assumed that all points having the same value of R_f, in fig. 6, represent equivalent states within the deformed-and-annealed microstructure of the specimens, then it is possible to estimate an activation energy, Q, associated with the mechanism controlling recovery behaviour over the temperature interval studied. The two values of Q shown on fig. 6 were computed by means of the following relation:

$$Q = \frac{R \ln(t_1/t_2)}{(1/T_1 - 1/T_2)} \qquad (2)$$

where: R is the gas constant; T_1 and T_2 are the

recovery temperatures (° K); t_1 and t_2 are the times at T_1 and T_2, respectively, to attain the same value of R_f. Thus, a mean estimate of Q was calculated to be 69 ($\pm$ 4) kcal/mole. This value for Q was found to be very nearly that reported for self-diffusion in nickel (although for a temperature interval between 1085 and 1400°C) [7]. Since the self-diffusion mechanism in nickel would be likely to control recovery behaviour in unalloyed nickel, the above assumption of equivalent states at the same R_f value may be valid.

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   (b) A. YA. SHINYAYEV, *Fiz. Metall. i Metallov.* **15** (1963) 100.

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Book Reviews

An Atomic Approach to the Nature and Properties of Materials

J. A. Pask (editor)

Pp xi + 477 (Wiley, 1967) £8

The advances made in understanding the properties of materials increasingly demand more emphasis in university courses. No doubt many textbooks will be published in the future which attempt to meet the demands of this new emphasis. The volume under review sets out to provide in one 500-page volume a comprehensive account of almost the whole of materials science. In general, the layout is excellent and the publishers are to be congratulated on the quality of both the printing and the illustrations.

The book contains seventeen more-or-less-independent chapters, each written by an authority in that subject. The standard of presentation varies from chapter to chapter but in general it is good. Predictably, the first four chapters discuss atomic structure, interatomic binding, crystal structure, and imperfections in crystals. This leads to an account of the structure and properties of liquids, amorphous materials, and polymeric materials. This is followed by discussion of numerous aspects of electrical, magnetic, thermal, and mechanical properties. The concluding chapter discusses "The Atomic Viewpoint in Materials Application". Inevitably, the scope of the book is so wide that some subjects are inadequately covered. Thus, we find that the discussion of semiconductors, superconductivity, and dielectrical properties is so condensed as to be useless to the reader. In a book with this fashionable title, these are inexcusable shortcomings.

Although many industrial scientists and postgraduate students in technological subjects will find this book useful, few will find room for it on their bookshelves. It seems more likely to become a well-thumbed library reference book, since, at £8, it does not represent good value for money.

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