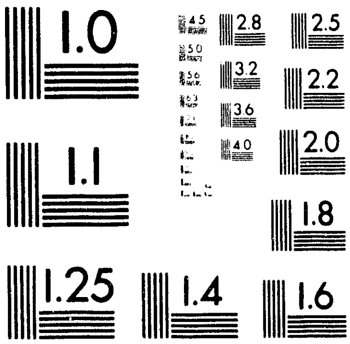
1.0
2.8
2.5
3.2
2.2
3.6
4.0
2.0
1.1
1.8
1.25
1.4
1.6

05-6

Conf-930676--33

LA-UR- 93-2214

Title: **INFLUENCE OF PEAK PRESSURE AND TEMPERATURE ON THE SHOCK-LOADING RESPONSE OF TANTALUM**

Author(s): **G.T. Gray III**

Submitted to: **APS Topical Conference on Condensed Matter Physics**
Colorado Springs, CO
June 28 - July 2, 1993

Los Alamos
NATIONAL LABORATORY

Form No. 836 R5
ST 2629 10/91

DISTRIBUTION OF THIS DOCUMENT IS UNLIMITED

INFLUENCE OF PEAK PRESSURE AND TEMPERATURE ON THE SHOCK-LOADING RESPONSE OF TANTALUM

G. T. Gray III
Los Alamos National Laboratory
Los Alamos, New Mexico, USA 87545

While the deformation response of tantalum subjected to high-rate loading has attracted considerable study, few studies have systematically investigated the influence of peak shock pressure and starting temperature on the shock response of tantalum. In this study the deformation behavior of annealed tantalum has been compared to tantalum shock pre-strained to 7 and 15 GPa at 25°C as well as to 7 GPa after first preheating to 200° and 400°C. The reload yield behavior of shock prestrained Ta was found to exhibit no enhanced shock hardening compared to Ta deformed quasi-statically or dynamically to an equivalent strain level. In addition the reload yield behavior of Ta shock prestrained to 7 GPa at 200 or 400°C was found to exhibit increased hardening compared to the shock prestraining at 25°C. The shock-hardening response of Ta is discussed in terms of defect storage and compared to that seen in fcc metals.

INTRODUCTION

The passage of shock waves through materials has been shown to alter to varying degrees the structure/property response of a broad range of metals and alloys. Specific examples of post-shock material response have been discussed previously[1-3]. These effects have been particularly well documented in a diverse number of face-centered-cubic(fcc) metals such as copper and nickel, and fcc alloys including brass and austenitic stainless steels. Shock response studies on body-centered-cubic (bcc) metals have preferentially focused on iron and ferritic steels due to extensive interest in the $\alpha-\varepsilon$ pressure-induced transition. Fewer studies have been undertaken on other bcc metals such as niobium, molybdenum, tantalum, or tungsten. Shock-loaded fcc metals and alloys have been repeatedly shown to exhibit increased hardening behavior in reload tests after shock recovery compared to the same metal deformed at low strain rate to an equivalent strain level[1-3]. Figure 1 illustrates an example of this substantially increased reload yield strength response for polycrystalline copper and high-purity nickel[4,5].

While fcc, bcc, and hcp metals exhibit a large number of similarities in general physical and

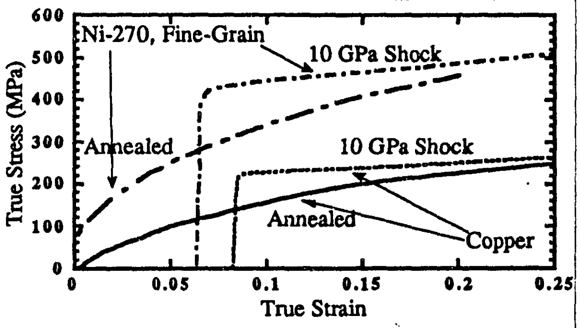

Figure 1: Reload stress-strain response of shock-loaded Cu and Ni compared to that observed during quasi-static deformation.

mechanical properties, significant differences also exist. For example, annealed bcc and hcp metals and alloys exhibit pronounced strain-rate and temperature-dependent properties. This dependence is due to their strong inherent lattice resistance to dislocation motion compared to fcc metals in which the lattice resistance is small Recently, the high-strain rate and shock response of bcc metals, in particular the refractory metals tantalum, niobium, and tungsten, has received increased interest for ballistic applications. While the detailed response of refractory metals to a wide range of deformation paths has been widely examined, systematic studies of the shock response of these metals has received limited attention. Early work by Dieter[1] in 1961 noted that bcc metals other than iron exhibited no or minimal improvement in overall mechanical properties after shock loading. Only iron which displays an allotropic transition and extensive twinning during shock deformation showed a pronounced effect of shock loading on defect storage. Niobium revealed minimal twinning and displayed virtually no increased hardening compared to quasi-static rolling to equivalent strain levels. The purpose of the present study was to investigate the influence of peak pressure and temperature on the shock-loading response of tantalum.

EXPERIMENTAL

Commercial purity tantalum plate 5 mm thick with a measured composition (in wt. %) of Carbon-6 ppm, Nitrogen-24 ppm, Oxygen-56 ppm, Hydrogen-<1 ppm, Iron-19 ppm, Nickel-25 ppm, Chromium-9 ppm, Tungsten-41 ppm, Niobium-26 ppm and balance Ta was used in this study. The plate was studied in an annealed condition and possessing an equiaxed grain structure ~68μm in diameter. Shock recovery experiments were performed utilizing an 80-mm single-stage launcher and recovery techniques as described previously[3]. Tantalum samples were shock loaded in Ta shock-recovery assemblies to 7 and 20 GPa for 1 μs pulse durations under symmetric impact conditions. Elevated temperature shock-loading experiments at 7 GPa were conducted at nominally 220° and 400°C using a resistive heating-element furnace placed circumferentially around the outer momentum trapping ring of the assembly. Compression samples were EDM machined from the as-received Ta and shocked samples, and reloaded at strain rates ranging from 0.001 to 3000 s^{-1}.

RESULTS AND DISCUSSION

The reload mechanical response of shock prestrained tantalum was found to depend on both the peak shock pressure and the temperature at which shock prestraining was performed. Figure 2 presents a plot of the quasi-static reload stress-strain behavior of the annealed starting Ta, as well as the samples shocked at room temperature to 7 and 20 GPa. Figure 3 shows a plot of the same shocked Ta shock prestrained samples reloaded dynamically using a Split-Hopkinson Pressure Bar at a strain rate of 3000 s^{-1}. The reload shock curves in Figures 2 and 3 have been offset with respect to the annealed Ta response at low and high strain rate by the transient strain generated by the shock defined as 4/3 ln (V/V_0), where V and V_0 are the final and initial volumes of Ta during the shock cycle. Contrary to the results reviewed for Cu and Ni in Figure 1, Ta shocked to 7 and 20 GPa does not exhibit an increased shock hardening response compared to Ta quasi-statically deformed to an equivalent strain level. The reload mechanical response following shock prestraining to 7 GPa exhibits a reduced flow stress level compared to the quasi-static loading path. The 20 GPa reload stress-strain curve follows nearly the identical yield and hardening path as the low-rate annealed response at the equivalent strain level. The dynamic reload stress-strain response in Figure 3 shows the same response as that seen in the quasi-static reloads although the overall flow stress levels are higher, consistent with the high rate-sensitivity of Ta.

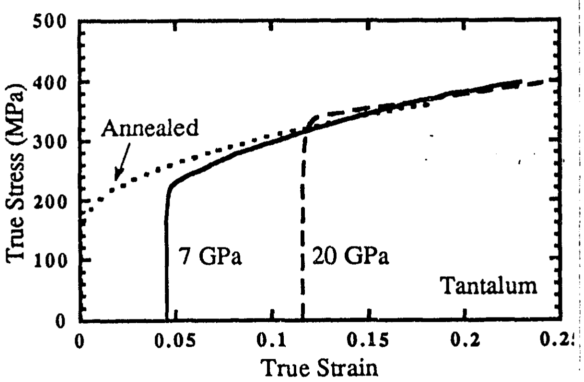

Figure 2: Quasi-static stress-strain response of shock recovered Ta after 7 and 20 GPa shocks. Reload curves are offset by the transient strains due to the shocks as compared to quasi-static deformation of annealed Ta.

The reload flow stress response of shocked Ta does not display the often documented increased

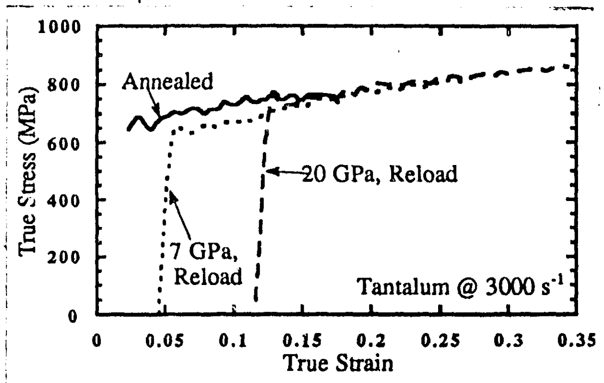

Figure 3: High-strain-rate reload stress-strain response of shock prestrained tantalum.

defect storage response of shock loading in fcc metals compared to conventional strain-rate loading paths[3]. This observation is identical to that seen previously by Lassila and Gray[6]. Increased hardening during shock loading in the fcc case has been qualitatively linked to the subsonic restriction on dislocation velocity, requiring the generation and storage of a larger number of dislocations for a given strain. Observations of a strong dependence of the initial Stage II work-hardening rate[5,7] on strain rate in fcc metals supports this fact and the data for Cu and Ni in Figure 1. This increased hardening response clearly shows that the inherent dislocation-dislocation micro-mechanisms responsible for defect storage are altered in the shock. The exact phenomena responsible for this increased hardening efficiency in the shock has yet to be adequately explained or modeled[5,7]. However, its manifestation in high SFE fcc metals suggests that it is linked to both: (a) increased dislocation interactions resulting from enhanced dislocation nucleation at the higher stress levels achieved at high strain rates, and (b) suppression of dynamic recovery processes, which depend on cross-slip. Cross-slip is made more difficult when deformation occurs at higher strain rates due to reduced thermal activation[8] and more planar slip results. The absence of shock-enhanced hardening in low-SFE Si-Bronze[9] and low-SFE Ni-based 230-alloy[10] is consistent with differences in cross-slip reducing the amount of total defect storage in the shock cycle. More "homogeneous" dislocation nucleation or widespread multiplication from existing sources with increasing stress levels at high/shock strain rates will lead to reduced dislocation slip distances prior to tangling with other defects.

The results in this study on Ta, similar to those described by Dieter for Nb[1], reveal that increased hardening due to shock loading in Ta is not observed. In both Nb and the current Ta, this behavior is consistent with defect storage being dislocation controlled; no significant twinning occurred in the Ta at 7 or 20 GPa. Accordingly, defect storage phenomena may be viewed using the framework of dislocation kinetics as is the case for low-rate plasticity[5,9]. The observation of no enhanced shock hardening is believed to reflect the influence of the large lattice friction (Peierls Stress) component of the flow stress in both Ta and Nb. At low temperatures and high-rates, the Peierls stress(~0.48 GPa shear stress for Ta) is a significant portion of the flow stress.

As such, under high-rate loading dislocation motion in Ta will be significantly restricted and cross-slip inhibited or totally suppressed. These effects will be particularly pronounced on screw dislocations which have a higher Peierls barrier than edge segments in bcc metals[11]. The predominance of long straight screw segments in Ta following low temperature or shock deformation is consistent with this mechanism[6,9]. This disparity in cross-slip ability in a bcc metal where defect generation is dominated by dislocation processes will significantly decrease defect storage in bcc metals at high-rate or low temperatures. The suppression of cross-slip in Ta at shock-loading strain rates will change dislocation motion from 'areal' to 'lineal' glide[12]. This will significantly affect work hardening by suppressing the storage of new dislocation line length. The absence of a shock-enhanced work-hardening rate in Ta is consistent with the lack of a dependence of the Stage-II hardening rate in polycrystalline Ta on strain rate. Recent modeling of shock-loaded Ta shows the importance of rapid dislocation evolution and a very strong dependence of the plastic strain rate on shear stress in the drag regime[13]. Increased viscous-drag effects will also decrease defect storage in the shock by reducing cross-slip.

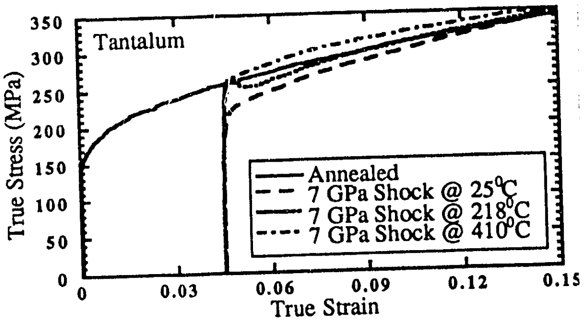

Figure 4: Reload stress-strain response of Ta shock prestrained to 7 GPa at three different starting temperatures.

The lack of an enhanced shock-hardening response in Ta supports the importance of dislocation kinetics in describing the thermally-activated generation and storage of dislocations in bcc metals. In accordance with this idea the influence of temperature on defect storage in the shock process was investigated. Tantalum shock assemblies were preheated to elevated temperatures and shock loaded to a peak pressure of 7 GPa. The reload stress-strain response of the shock-recovered Ta as a function of shock preheat temperature is shown in Figure 4. Increasing the temperature at which shock prestraining was conducted increased the reload yield strength of the shocked Ta. This result suggests that the temperature increase either increases the reload response by: 1) directly affecting defect storage during the shock by increasing thermally activated cross-slip, or 2) post-shock pinning of the dislocations generated in the shock by mobile interstitials; this pinning is aided by thermally-activated diffusion of the interstitials, i.e. assisting static strain-aging. The yield drop evident in the 218°C shock prestrained curve is consistent with at least some strain-aging contribution, supporting the assertion that interstitial effects are important.

Lines to bottom single spacing (for footnotes)

SUMMARY AND CONCLUSIONS

Based upon a study of the variation of peak shock pressure and temperature at which shock prestraining was conducted on the mechanical response of tantalum, the following conclusions can be drawn: 1) the reload yield behavior of shock prestrained Ta exhibits no enhanced shock hardening compared to Ta deformed quasi-statically or dynamically to an equivalent strain level, and 2) the reload yield behavior of Ta shock prestrained to 7 GPa at 200 or 400°C exhibits increased hardening compared to shock prestraining at 25°C.

ACKNOWLEDGMENTS

The author wishes to acknowledge C. Trujillo, W. Wright, and M.F. Lopez for their assistance with the experimental aspects of this study. This work was performed under the auspices of the U.S. Department of Energy.

REFERENCES

[1] G.E. Dieter, in *Resp. of Metals to High Vel.Def.*, Interscience Publ., NY, pp. 409 (1961).
[2] L.E. Murr, in *Shock Waves and High Strain Rate Phen. in Metals*, (ed. by M.A. Meyers and L.E. Murr) Plenum Press, NY, pp. 607 (1981).
[3] G.T. Gray III, in *High Pres. Shock Comp.of Solids*, (ed. by J.R. Asay and M. Shahinpoor) New York: Springer-Verlag,ch. 6, (1993).
[4] G.T. Gray III and P.S. Follansbee, in *Impact Loading and Dynamic Behavior of Materials*, (ed. by C.Y. Chiem, H.-D. Kunze, and L.W. Meyer) DFG, Germany, pp. 541 (1988).
[5] P.S. Follansbee and G.T. Gray III, *Int. J. Plasticity* 7, pp. 651-660 (1991).
[6] D. Lassila and G.T. Gray III, in *DYMAT 91*, J.de Physique IV, vol. 1, pp. C3-19 (1991)
[7] P.S. Follansbee and U.F. Kocks, *Acta Metall.* 36, pp. 81-93 (1988).
[8] G.T. Gray III, in *Modeling the Def. of Crystal.* (ed. by T.C. Lowe, A.D. Rollett, P.S. Follansbee, and G.S. Daehn)TMS, pp. 145 (1991).
[9] G.T. Gray III, *Shock Waves*, (1993) in press.
[10] K.S. Vecchio and G.T. Gray III, this volume.
[11] V. Vitek, in *Disl. and Prop. of Real Matls.*, Inst. of Metals, London, pp. 30-50 (1985).
[12] U. F. Kocks, in *The Mech. of Disl.*(ed. by E. C. Aifanatis and J. P. Hirth) Amer. Soc. for Metals, 1985, pp. 81-83.
[13] J.N. Johnson, this volume.

Lines to bottom 1½ spacing

DATE

FILMED

10 / 13 / 93

END