Title:

**MICROSTRUCTURE OF DEPLETED
URANIUM UNDER UNIAXIAL STRAIN
CONDITIONS**

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Presented at:

Proceedings for the Shock Compression of Condensed
Matter - 1997

APS Topical Conference in Amherst MA
July 26 - August 1, 1997

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Title: Microstructure of Depleted Uranium under Uniaxial Strain Condition

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To Reviewer* S.R. Chen

From: Author(s): Anna K. Fureh

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MICROSTRUCTURE OF DEPLETED URANIUM UNDER UNIAXIAL STRAIN CONDITIONS

A. K. Zurek, J. D. Embury, A. Kelly, W. R. Thissell, R. L. Gustavsen,
J. E. Vorthman, and R. S. Hixson

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Uranium samples of two different purities were used for spall strength measurements. Samples of depleted uranium were taken from very high purity material (38 ppm of carbon) and from material containing 280 ppm carbon. Experimental conditions were chosen to effectively arrest the microstructural damage at two places in the development to full spall separation. Samples were soft recovered and characterized with respect to the microstructure and the form of damage. This allowed determination of the dependence of spall mechanisms on stress level, stress state, and sample purity. This information is used in developing a model to predict the mode of fracture.

INTRODUCTION

Uranium is a very high density material (19.1 g/cm^3) that is relatively strong and easily cast and formed. It is widely used in nuclear and non-nuclear applications as radiation shields or kinetic energy penetrators. The most common form of pure uranium is the U-238 isotope containing some U-235. The most commonly used is low-temperature α orthorhombic phase depleted uranium. This phase is ductile, but its ductility is very dependent on processing and impurity content. The ductile-to-brittle transition of α uranium occurs around 0°C , but decreasing the grain size and hydrogen content can cause it to vary. Impurities have very low solubility in uranium, and they usually form second-phase particles that may decrease macroscopic ductility. Carbon (C) forms carbide inclusions, which may decrease ductility [1] in a manner analogous to carbides in ferritic steels.

Spallation is one of many experimental configurations that can produce controlled dynamic fracture. Spallation is defined as a dynamic uniaxial strain fracture experiment. Fracture occurs during spallation due to tensile stresses generated

by the interaction of two release (rarefaction) waves [2]. Spallation is a process of damage accumulation and linkage that differs drastically from fracture damage in the uniaxial tensile test by virtue of the stress state and the rate of extent of damage accumulation. In a tensile test, voids and cracks are subject to a nearly uniaxial stress tensile field; homogeneous plastic strain dominates the flow process for most of the strain history. Due to the uniaxial stress deformation field, the voids or cracks grow to form a fracture surface, and the overall change in porosity in the vicinity of failure is small, on the order of 5% [3]. In contrast, in spallation, voids or cracks are subject to extremely high, nearly isotropic, triaxial, hydrostatic stress fields, and high strain rates, which vary spatially in the sample. Voiding or crack growth dominate all stages of the damage process and produce porosity of up to about 30% at the principal spall plane. The growth rate of voids or cracks is very high, and the distribution of damage is dictated by the large gradient in stress and strain rate generated by the interaction of release waves. Porosity, void or crack formation, growth, and coalescence, therefore,

are important variables in descriptions of spallation and the fracture criteria of the material [2, 4, 5].

In this paper we report on microstructural damage development characterization in depleted α uranium deformed under spall conditions.

MATERIALS AND EXPERIMENTS DESCRIPTIONS

Depleted uranium samples containing 280 ppm C were studied [6]. The as-cast material was wrought to specifications by the process of heat treatment, upset forging, re-heating, and finally hot rolling to 58% of the original thickness in four equal reduction passes. Due to this process the resulting microstructure is not uniform across the plate thickness. Large grains (200 μm) dominate the center of the plate while a substantially reduced grain size (down to 40 μm) exist near the top and bottom plate surfaces. Samples were cut from the center of the plate. Similarly, the high-purity depleted uranium that contained only 38 ppm C was wrought; however, the final microstructure was uniform and consisted of equiaxed 10- μm grains.

Samples were spalled and incipiently spalled using a gas gun under uniaxial strain state conditions. The spall tests were performed under nonsymmetric shock conditions; a z-cut quartz flat plate was impacted against the depleted uranium samples at a shock pressure not exceeding 5.3 GPa for 1- μs pulse duration. VISAR traces of the wave interactions were acquired and are described together with all other shock experimental details in a companion paper authored by R. Hixson *et al.* in this volume. Soft recovered samples were stored immediately in 200-proof dehydrated ethyl alcohol to prevent sample oxidation. Metallographic samples, cut through the center of the spalled sample in the direction parallel to the loading direction, were prepared for quantitative analysis [7].

RESULTS AND DISCUSSION

Plastic-Deformation and High-Strain-Rate-Induced Brittle Fracture

Figures 1a and 1b show typical spalled fracture surfaces in depleted uranium. Predominantly brittle fracture was observed in all the spalled uranium samples, with transgranular cracks for the 280 ppm purity sample, and with brittle

intergranular cracks for the 38 ppm purity samples. Some ductility is visible in the 280 ppm purity uranium (Fig. 1a), and in the 38 ppm purity uranium only large grains showed ductility in the form of ductile dimples (indicated in Fig. 1b).

a

b

FIGURE 1. Fracture surfaces of depleted uranium spalled samples. (a) Depleted uranium with 280 ppm carbon showing predominantly cleavage fracture surface. (b) Depleted uranium with 38 ppm carbon showing predominantly intergranular fracture. Both samples show some areas of ductile dimples. Both samples measured comparable spall strength of -1.9 GPa.

Although α uranium is normally a ductile phase, the tests were performed at room temperature, which is close to the ductile-to-brittle transition temperature (DBTT) in uranium. The nature of the spall test, i.e. deformation at high pressure and high strain rate, contributes to the shift of the DBTT to higher temperatures when the strain rate dependence of the flow strength is taken

into account [8]. In addition, the sample is subjected to high strain rate deformation and coincident hardening during the passage of the initial compressive shock wave. This deformation generates a high density of dislocations, and what is more important in the case of uranium, a large number of deformation twins; several twins variants are activated within each grain. Figures 2a and 2b show the highly magnified microstructures of incipiently spalled samples for both purities of uranium.

a

b

FIGURE 2. Cross section of depleted uranium samples incipiently spalled. (a) Depleted uranium with 280 ppm C showing large grains deformed under shock conditions with numerous twins and twin systems. Cracks in this sample are transgranular and frequently run along the twin/matrix interface. (b) Depleted uranium with 38 ppm C showing equiaxed grain structure, deformation twins within the grains, and cracks running along the grain boundaries.

It is evident that the cracks in the 38 ppm purity uranium followed the grain boundaries (Fig. 2b), and the cracks in the 280 ppm uranium are

more correlated with the deformation twin systems developed in the sample during the shock.

The hydrostatic tension in the spall test is expected to aid microcrack nucleation. Work-hardening processes are far more rapid in a spall test than in a tensile quasi-static test, which results in a substantial increase in the yield strength and flow stress. This increase makes the accommodation of plastic deformation at a crack tip more difficult and therefore favors intergranular or transgranular brittle fracture. In addition to the great number of deformation twins, twin intersections and twin-matrix interfaces can serve as preferred nucleation sites for sharp brittle cracks.

During quasi-static loading, DBT transition is dominated by the temperature dependence of the fracture stress. The hydrostatic pressure in a tensile test is about $1/3$ of the flow stress (tensile test $\Rightarrow 1/3 \leq -p/2\tau \leq 3$), which is very small in comparison to the hydrostatic tension developed under spall conditions (spall test $\Rightarrow 7 \leq -p/2\tau \leq 30$). In either case the hydrostatic pressure imposed on the sample during quasi-static or dynamic tests influences a shift in the ductile-to-brittle transition temperature. Figures 3a and 3b schematically illustrate this phenomena by plotting the change in material yield strength with respect to temperature and the fracture stress.

The intersection of the fracture stress level with the yield stress curve marks the DBTT. Figure 3a shows an increase in fracture stress for a tensile quasi-static test with an imposed external compressive hydrostatic pressure.

Fig. 3a

FIGURE 3. A schematic showing the DBT temperature shift resulting from the hydrostatic pressure. In the case of a tension test, compressive hydrostatic pressure imposed on a system shifts the fracture stress to a higher level and thereby the DBTT to a lower temperature (a). In a spall test an inherent to the test—very high tensile hydrostatic pressure—shifts the fracture stress to a lower level and thereby the DBTT to a higher temperature (b).

Figure 3b schematically depicts a spall tensile stress state, with its inherent large tensile hydrostatic pressure, that decreases the fracture stress. In a tensile test, the compressive hydrostatic pressure shifts the DBTT to a lower temperature, while in a spall test, large tensile hydrostatic pressure shifts the DBTT to a higher temperature.

An example of the experimental evidence of this process is shown by Davidson [9] in magnesium tested under a quasi-static tensile stress state. A hydrostatic compression of 0.8 GPa imposed on a Mg sample lowered the DBTT by over 230°C (from 175°C to -55°C) [9]. Likewise, the reverse is expected in a dynamic tensile spall test, where very high hydrostatic tension dominates the fracture process, lowering the fracture stress and therefore increasing the DBTT. This may explain the predominantly brittle fracture exhibited by depleted uranium under spall conditions.

The change in fracture mode from transgranular fracture for the 280 ppm C uranium (large grain) to intergranular fracture mode for 38 ppm carbon content (small grain) uranium can be attributed to the possibility of hydrogen embrittlement in the latter. The amount of 2 ppm of hydrogen (not outgassed) is sufficient to promote the grain boundary decohesion and therefore intergranular fracture.

SUMMARY

Two purities of depleted uranium samples were tested under spall conditions. The spalled samples showed predominantly brittle fracture: transgranular fracture for 280 ppm C uranium and intergranular fracture for 38 ppm C uranium. Deformation twinning was found to be the dominant form of deformation under spall conditions in the pure uranium samples. Both purities of uranium samples had a comparable spall strength of -1.9 GPa. The high hydrostatic tension and very high strain rates inherent to spall testing are thought to increase the DBTT in pure uranium and thus promote brittle fracture.

ACKNOWLEDGMENTS

The authors would like to thank the Spall and Materials Damage Program Manager, Dean Preston of Los Alamos, for financial support. The work was done under the auspices of the US Department of Energy.

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Introduction

- * Uranium samples of two different purity were used for spall strength measurements.
- * Samples of depleted uranium were taken from very high purity material (38 ppm of carbon) and from material that had 280 ppm of carbon.
- * Experimental conditions were chosen to effectively arrest the microstructural damage at two places in the development to full spall separation.
- * Samples were soft recovered and characterized with respect to the microstructure and the form of damage. This allowed determination of the dependence of spall mechanisms on stress level, stress state, and sample purity.

Uranium-Carbon System

“Pure” uranium is a BINARY system

B. Blumenthal, *J. of Nuclear Mat.*, 2 #3, 197 (1960)

Uranium Properties vs. Temperature

Uranium displays unique properties as a function of temperature

K.H. Eckelmeyer, ASM Handbook, vol. 2, pg. 670 (1991).

High Strain Rate Properties of Uranium

K_{IC} dependence of strain rate and temperature

K_{IC} calculation

This is only an estimate, because the mechanism of spall fracture differ from the plane strain Mode I fracture for which this analysis was developed (*Lin et al.*). The difference will lead to differences in the local stress distribution in the vicinity of the crack. In our case the formation of twins add to the process of more "brittle" behavior.

$$K_{IC} = R \frac{1}{\left\{ \sigma_a + \frac{\mu(T) * (\hat{\sigma} - \sigma_a)}{\mu_0} \left[1 - \left(\frac{kT}{g_0 \mu(T) b^3} \log \dot{\epsilon}_0 / \dot{\epsilon} \right)^{2/3} \right]^2 \right\}^2}$$

$\hat{\sigma}$ = temperature dependent fracture stress

σ_a = athermal stress (for uranium about 100 MPa)

$\mu_0 = 91360(\text{MPa})$, shear modulus

$$\mu(T) = \mu_0 - \frac{c_\mu (= 11.729 \text{ GPa})}{\exp\left(\frac{T_\mu (= 239 \text{ K})}{T}\right) - 1}$$

$$\frac{k}{b^3} = 0.5936$$

$$g_0 = 1.6$$

$$\dot{\epsilon}_0 = 10^6$$

$\dot{\epsilon}$ = strain rate

Fracture surfaces of spalled samples

A

B

- (a) Depleted uranium with 280 ppm carbon showing predominantly cleavage fracture surface.
- (b) Depleted uranium with 38 ppm carbon showing predominantly intergranular fracture. Both samples show some ductile dimples areas. Both samples measured comparable spall strength of ~ 1.9 GPa.

Cross sections of depleted uranium; Samples incipiently spalled

- (a) Depleted uranium with 280 ppm carbon showing large grains deformed under the shock conditions with numerous twins and twin systems. Cracks in this sample are transgranular and frequently run along the twin / matrix interface.
- (b) Depleted uranium with 38 ppm carbon showing equiaxed grain structure, deformation twins within the grains, and cracks running along the grain boundaries.

DBT temperature shift due to the hydrostatic pressure

(a) In the case of a tension test a compressive hydrostatic pressure imposed on a system shifts the fracture stress to a higher level and thereby the DBT to a lower temperature.

(b) In a spall test an inherent to the test very high tensile hydrostatic pressure shifts the fracture stress to a lower level and thereby the DBT to a higher temperature.