Tensile fracture and shear localization under high loading rate in tungsten alloys
H. Couque, J. Lankford, A. Bose

To cite this version:

HAL Id: jpa-00248878
https://hal.archives-ouvertes.fr/jpa-00248878
Submitted on 1 Jan 1992

HAL is a multi-disciplinary open access archive for the deposit and dissemination of scientific research documents, whether they are published or not. The documents may come from teaching and research institutions in France or abroad, or from public or private research centers.

L’archive ouverte pluridisciplinaire HAL, est destinée au dépôt et à la diffusion de documents scientifiques de niveau recherche, publiés ou non, émanant des établissements d’enseignement et de recherche français ou étrangers, des laboratoires publics ou privés.
Classification
*Physics Abstracts*
46.30N — 81.40E

**Tensile fracture and shear localization under high loading rate in tungsten alloys**

H. Couque, J. Lankford and A. Bose

Southwest Research Institute, 6220 Culebra Road, San Antonio, Texas 78228-0510, U.S.A.

(Received 7 October 1991, revised 9 July 1992, accepted 15 July 1992)

**Abstract.** — The influence of loading rate and microstructure on the tensile and compressive failure properties of three microstructurally dissimilar tungsten alloys has been investigated. Dynamic tensile fracture properties were characterized through fracture toughness tests performed at a stress intensity loading rate of $10^8$ MPa $\sqrt{m\;s^{-1}}$, and by tensile testing at a strain rate of $10^3$ s$^{-1}$. Shear banding phenomena were investigated by means of compression tests performed at strain rates of $5 \times 10^3$ s$^{-1}$. Under rapid loading conditions, nickel-cobalt-tungsten alloys were found to be tougher than nickel-iron-tungsten alloys; the tungsten/tungsten interface was identified as the governing microstructural factor. Quantitative micromodeling using simple fracture models was found to provide a mean of correlating toughness with microstructures. Compression-induced shear localization was found to be facilitated within systems characterized by either elongated tungsten particles or an adiabatic shear-prone matrix. The shear band width was observed to be proportional to tungsten particle size.

**Introduction.**

Because of their generally high density, strength, and ductility, tungsten heavy alloys (WHA) are used as kinetic energy penetrator materials. However, the development of more efficient WHA has involved high strain rate testing to simulate the rapid loading rates that occur in penetrators [1-5]. While few studies have questioned whether these high strain rate mechanical properties are, in fact, microstructure sensitive [6, 7], the generic micromechanisms that distinguish dynamic tensile and compression mechanical response have recently been clarified [8].

The objective of this paper is to report the influence of microstructure on the tensile (stress-strain/fracture toughness) and compressive failure behavior of three tungsten alloys covering a wide range in ductility and strength. Although tensile properties are known to provide important information with regard to penetrator integrity, knowledge of compressive failure provides insight concerning actual penetration performance. In particular, the latter is thought to be related to shear localization, which can lead to self-sharpening of the penetrator during impact, and thereby facilitate penetration.
Materials.

Two commercial, and one newly developed, liquid-phase, 90 weight percent tungsten alloys were chosen for this investigation. Designations, compositions, and mean tungsten grain size (mean intercept length of the tungsten grain) are given in table I. Since each alloy contains tungsten and nickel, its designation reflects the remaining elemental constituent, i.e., Fe, Co, and Mn, respectively.

Figure 1 shows the three microstructures; it is evident that all are characterized by a contiguous network of nearly pure bcc tungsten grains embedded within a ductile fcc matrix. The swaged Co alloy possesses a directional texture characterized by elongated tungsten grains, and in addition tungsten precipitates are present in the matrix [7]. For the Co material, the mechanical tests were performed with loading and swaging directions parallel. The newly
Table I. — Material conditions.

<table>
<thead>
<tr>
<th>Alloy Designation</th>
<th>Composition (wt.%)</th>
<th>Treatment/Condition</th>
<th>Average Tungsten Grain Size (µm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Fe</td>
<td>90 W-8 Ni-2 Fe</td>
<td>As-sintered</td>
<td>23.5</td>
</tr>
<tr>
<td>Co</td>
<td>91 W-6 Ni-3 Co</td>
<td>Swaged 25 % (*) and aged</td>
<td>22.9</td>
</tr>
<tr>
<td>Mn</td>
<td>90 W-4 Ni-6 Mn</td>
<td>As-sintered</td>
<td>7.0</td>
</tr>
</tbody>
</table>

(*) Reduction in area.

developed Mn alloy [9] has the finest microstructure, with an average tungsten grain (intercept) size of 7 µm, and based on thermomechanical consideration [9], should have an especially high intrinsic propensity for adiabatic shear. As can be seen from figure 1c, this alloy is somewhat porous, and in consequence was only characterized in compression.

Experimental approach.

Compression and tensile tests were performed at strain rates varying from $10^{-4}$ to $5 \times 10^3$ s$^{-1}$ Quasi-static data were obtained by using a servo-controlled hydraulic test machine under displacement control conditions, while dynamic tests were run in a split Hopkinson pressure bar adapted for both compressive [10] and tensile [11] modes of loading. The compression specimens were cylinders 6.35 mm in diameter and 12.70 mm in length; tensile specimens 7.62 mm in gage length and 3.18 mm in diameter were used.

Quasi-static and dynamic fracture initiation tests were conducted at stress intensity loading $K_0$, of 1 and $10^6$ MPa $\sqrt{m}$ s$^{-1}$ using precracked compact specimens. Static specimens of planar size $W = 30.5$ mm, of thickness $B = 15.2$ mm, and of prefatigued crack length $a_0 = 22.7$ mm were tested in a conventional test machine under displacement control. Load and crack opening displacement were monitored, while crack growth was deduced from compliance measurements.

A special coupled pressure bars (CPB) technique was used for dynamic fracture testing. For the background on the design and development of the experimental apparatus the reader is referred to references [12, 13]. Figure 2 shows a schematic of the CPB experiment. The primary components consist of two pressure bars to store energy, a notched, round starter specimen to release the stored energy rapidly, and two prefatigued compact fracture specimens. These experiments were conducted by preloading the pressure bars and starter specimen to a load of 445 kN corresponding to an applied stress of 390 MPa. The test specimens were then inserted into slots in the bars and secured with wedges, as shown in figure 2. Fracture of the starter specimen was subsequently initiated by introducing a sharp cut into the circumferential notch of the starter specimen using a cutter wheel and high-speed air drill. Failure of the starter specimen initiates an unloading (compressive) stress pulse in the pressure bars, which transmits a rapid axial displacement rate to the specimen arms. This stress pulse has a rise time of about 100 µs corresponding to the failure duration of the starter and has a constant stress amplitude part associated with the unloading of the two separated pressure bars. Each specimen have a planar size $W = 30.5$ mm, a height $H = 79.0$ mm, and a thickness $B = 15.2$ mm. Since about the same displacement rate is applied at the load line of the two specimens tested simultaneously [13], each specimen was precracked to a different prefatigued crack length $a_0 = 13.5$ mm and $a_0 = 22.7$ mm as to obtain two dynamic loading rates.
Specifically, a loading rate ratio of 4 was reached with these two specimens. Linear elastic fracture mechanics was used to evaluate the plane strain fracture toughness based on the standard formula for a compact specimen of ratio $H/W = 2.6$ [14]. For the specimen of longer crack length, the toughness was calculated using measured crack opening displacement at 12.5 mm from the load line at the onset of crack growth. This crack opening displacement was measured using an eddy current transducer while the onset of crack growth was deduced from strain measurements performed at a location 10 mm beyond the prefatigued crack tip, see figure 2. For the specimen of shorter crack length, only an estimate of the toughness is provided since no crack opening displacement was recorded. The toughness was calculated using an estimate of the crack opening displacement at load line at the onset of crack growth. The crack opening displacement at load line was approximated from the bar strain history at 127 mm from the starter specimen using one-dimensional stress wave analysis [13], while the
onset of crack growth was deduced from strain measurements performed at a location 10 mm beyond the prefatigued crack tip, see figure 2.

Fracture toughness was validated using the elastic fracture mechanic criterion, i.e., specimen size greater or equal to \(2.5(K_I/\sigma_y)^2\), where \(\sigma_y\) is the yield stress at a 0.2 percent strain offset at strain rates of \(10^{-4}\) s\(^{-1}\) and \(10^3\) s\(^{-1}\) for quasi-static and dynamic loading conditions, respectively. For specimen sizes not satisfying \(2.5(K_I/\sigma_y)^2\), i.e., ductile type fracture, the toughness was calculated using the fracture parameter \(J_{lc}\). This procedure was required only for the Fe microstructure under quasi-static conditions [13].

**Tensile fracture.**

Tensile data for the Fe and Co microstructures, in terms of true stress (\(\sigma_t\)) and natural strain (\(\varepsilon\)), are summarized in figure 3 and table II. These alloys are representative of the extremes in strength and ductility that can be obtained with conventional tungsten alloy systems under quasi-static conditions. With increasing strain rate, the strength differential between the two alloys remains approximately constant. On the other hand, the ductility differential is considerably reduced as a consequence of the proportionately larger ductility loss suffered by the Fe alloy. The as-sintered Fe alloy has significant hardening capability, which decreases somewhat with strain rate. On the contrary, the swaged Co alloy softens to a degree that increases with strain rate. This type of hardening and softening behavior was previously quantified using a simple power law relationship [7], and is reported in table II.

![Fig. 3. — Quasi-static and dynamic tensile stress-strain curves.](image)

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Strain Rate (s(^{-1}))</th>
<th>Yield Stress (MPa)</th>
<th>Hardening (N) (*)</th>
<th>Maximum Strain</th>
</tr>
</thead>
<tbody>
<tr>
<td>Fe</td>
<td>(10^{-4})</td>
<td>665</td>
<td>0.145</td>
<td>33.0</td>
</tr>
<tr>
<td>Fe</td>
<td>(10^3)</td>
<td>1 140</td>
<td>0.014</td>
<td>15.9</td>
</tr>
<tr>
<td>Co</td>
<td>(10^{-4})</td>
<td>1 676</td>
<td>-0.008</td>
<td>9.1</td>
</tr>
<tr>
<td>Co</td>
<td>(10^3)</td>
<td>2 250</td>
<td>-0.012</td>
<td>5.5</td>
</tr>
</tbody>
</table>

(*) \(\sigma = K(\varepsilon^p)^N\), with \(\sigma\) the flow stress and \(\varepsilon^p\) the plastic strain (\(\varepsilon^{yield} < \varepsilon^p < 0.1\)).
Table III. — Measured and calculated toughnesses.

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Rate</th>
<th>Fracture</th>
<th>Initiation Time $t_d$ [μs]</th>
<th>Yield Stress at $r_i^{(a)}$ $(\sigma_y)_p$ [MPa]</th>
<th>Griffith Stress $\sigma_f$ [MPa]</th>
<th>Critical Distance $X$ [μm]</th>
<th>Calculated Toughness $K^m_d$ [MPa√m]</th>
<th>Measured Toughness $K^m_d$ [MPa√m]</th>
<th>Loading Rate $K^m_d/t_d$ [MPa√m s⁻¹]</th>
</tr>
</thead>
<tbody>
<tr>
<td>Fe</td>
<td>Static</td>
<td>Ductile</td>
<td>10⁸</td>
<td>665</td>
<td>12.5</td>
<td>65</td>
<td>71</td>
<td>1</td>
<td>10⁹</td>
</tr>
<tr>
<td>Fe</td>
<td>Dynamic</td>
<td>Brittle</td>
<td>35</td>
<td>1 076</td>
<td>3 034</td>
<td>8.3</td>
<td>20</td>
<td>28</td>
<td>3.2 × 10⁶</td>
</tr>
<tr>
<td>Fe</td>
<td>Dynamic</td>
<td>Brittle</td>
<td>6</td>
<td>1 133</td>
<td>3 034</td>
<td>8.3</td>
<td>22</td>
<td>19</td>
<td>1</td>
</tr>
<tr>
<td>Co</td>
<td>Static</td>
<td>Brittle</td>
<td>10⁸</td>
<td>1 676</td>
<td>4 022</td>
<td>23</td>
<td>47</td>
<td>46</td>
<td>1</td>
</tr>
<tr>
<td>Co</td>
<td>Dynamic</td>
<td>Brittle</td>
<td>19.5</td>
<td>2 235</td>
<td>4 022</td>
<td>23</td>
<td>40</td>
<td>35</td>
<td>10⁹</td>
</tr>
</tbody>
</table>

(a) : yield stress at the limit of the plastic zone $(\sigma_y)_p = A \ln \epsilon + B$ with $\epsilon = 1.22(\sigma_y)_p/(Et_d)$ [18] where $t_d$ is the initiation time, $E$ the Young moduli and for the Fe microstructure $A = 29.5$ and $B = 936.5$, and for the Co microstructure $A = 36.4$ and $B = 2 011.5$.

Figure 4 and table III summarize the toughness data. Under quasi-static conditions, the toughness was found to increase with ductility, a trend which did not prevail at the higher rate, where the less ductile Co alloy is tougher. While the toughness of the Fe alloy decreases dramatically with rising load rate, that for Co material decreases only slightly. The toughness results were interpreted using simple fracture models, the details of which are provided in reference [13]. A summary follows to indicate the different modeling approaches used for the two principally observed fracture modes.

For the Fe microstructure, initiation of fracture was observed to be ductile under static conditions. Figure 5a1 shows the overload region next to the prefatigued crack. Damage evolution within the crack tip plastic zone (during a fracture increment), schematically represented in figure 6, appeared to proceed as follows. Initial fairly uniform deformation at both tungsten and matrix was interrupted at a critical local strain by multiple cracking of the tungsten grains. These cracks inclined at an angle, 20-50°, with regard to the loading direction are shear cleavage cracks resulting from interaction of non-coplanar parallel twins [15]. With the tungsten grains enabled to carry local stresses, the crack tip stress field is transferred to the
Fig. 5. — Scanning electron fractographs adjacent to the prefatigued crack tips of the quasi-static fracture specimens : a1) Fe, b1) Co and of the dynamic fracture specimens : a2) Fe, b2) Co. Schematic of the failure process was deduced from stereographic view. The failure mechanisms are indicated as followed : tungsten = 1, matrix = 2, tungsten/matrix interface = 3, tungsten/tungsten interface = 4.
remaining matrix ligaments, which then suffered disbonding from the adjacent tungsten particles and assumed the configuration of miniature, reduced section tensile specimens. Their failure by void nucleation and coalescence constituted the final (local) fracture step. The toughness was calculated using Rice and Johnson's model [16], which assumes that crack extension occurs when critical strains are achieved over a critical microstructural distance. Using McMeekings stress and strain calculations around a blunted notch [17] for a strain
hardening exponent \( N = 0.10 \) and \( \sigma_y/E = 1/300 \), the crack tip displacement, \( \delta_t \), is equal to
\[
\delta_t = 0.46(1 - \nu^2)K_{1c}^2/(\sigma_y E)
\] (1)
where \( K_1 \) is the stress intensity factor. McMeeking’s results, shown in figure 7, indicate that
the plastic strain (defined as an effective strain) monotonically decreases with the parameter
\((r/\delta_t)\), where \( r \) is the radial distance from the crack tip. As shown in the figure, it is clear that
for strain exceeding 0.1, \((r/\delta_t)\) is approximately unity. Shear cleavage of the tungsten particles
occurs at strain exceeding 0.20 since tungsten is the main component providing the high
hardening and ductility properties of tungsten alloys, see figure 3. Consequently, \((r/\delta_t)\) was
taken to be equal to unity at the onset of crack growth, and using equation (1) the toughness
\( K_{1c} \) can be expressed as
\[
K_{1c} = \sqrt{2.46 \sigma_y Er^*}
\] (2)
where \( r^* \) is the critical microstructural distance over which a critical strain must be achieved.
The critical microstructural distance was chosen to be equal to the width separating the cracked
tungsten particles. This distance is of the order of half the particle size, i.e., 12.5 \( \mu \)m. Based on
this parameter, a toughness of 65 MPa \( \sqrt{\text{m}} \) was computed. The excellent scaling of the
experiment and calculated toughness support the crack growth assumptions, i.e., void growth
and coalescence over a region the size of a tungsten particle.

On the other hand, both dynamic Fe fracture specimens were found to fail at measured crack
opening displacement theoretically characteristic of brittle fracture. Under dynamic loading
conditions the fracture process was dominated by apparent low strain, pure tensile cleavage at
tungsten/tungsten interfaces, as shown in figure 5a2 for the fracture specimen of longer crack
length. The fact that at room temperature, a brittle and ductile fracture was observed under
dynamic and static condition, respectively, is typical of rate-sensitive materials undergoing a
ductile-brittle fracture transition [18]. Precisely, the temperature at which the Fe microstructure
remains brittle increases with the increase of loading rate.

Like the dynamic Fe fracture specimen, the static and dynamic Co fracture specimens were
found to fail at measured crack opening displacements theoretically characteristic of brittle fracture ;
figures 5b1 and 5b2 show the fracture surfaces of the static fracture specimen and of the
dynamic fracture specimen of longer crack length. Compared with corresponding Fe
dynamic specimens, the Co fracture surfaces revealed a larger number of cleaved tungsten
particles, together with a small number of cleaved tungsten/tungsten interfaces. The fact that a
brittle fracture prevails at 23 °C under static and dynamic conditions indicates that the Co
system has a higher dynamic transition temperature (superior to 23 °C) than the Fe system
(inferior to 23 °C).

At both loading rates the Co matrix failed via void growth and coalescence at the sites of
matrix precipitates. There is strong evidence that the low strain fracture process is triggered
primarily by the cleavage of the tungsten/tungsten interfaces rather than by the shear cleavage
of the tungsten particles or by the ductile failure of the matrix. In particular, the matrix of the
Co system should be able to sustain large strain despite the presence of the precipitates since
ductilities in excess of 30 percent along with elevated hardening characteristics have been
measured with precipitate-rich, as-sintered Co alloys [19]. With regard to the tungsten
ductility, the tensile response of the Co specimen at both rates, see figure 3, reveals little
hardening, and failure strains in excess of 5 percent. This suggests that the main alloy
component, tungsten, for worked and unworked systems is active during the overall plastic
deformation, implying that shear cleavage of the tungsten particles does not occur at strains
typical of brittle fracture.

The brittle-type fracture of the dynamic Fe and static and dynamic Co specimens suggests
the use of a cleavage fracture model to calculate these toughnesses. These toughnesses were
calculated using the cleavage model of Ritchie, Knott, and Rice [20], based on the postulate that fracture initiation occurs when the maximum principal stress at the crack tip equals or exceeds the cleavage stress $\sigma_f^*$ (Griffith stress) over some microstructural distance. For all microstructures the critical distance over which the cleavage stress is applied was taken to be the mean spacing between failed tungsten/tungsten interfaces, as indicated schematically in figure 8. For the as-sintered Fe microstructure the cleavage fracture stress $\sigma_f^*$ was estimated to be about three times the yield stress using the Griffith relation [13]. For the worked Co microstructure, the cleavage stress $\sigma_f^*$ was assumed to be constant over the strain rate regime $10^{-4}$ to $10^3$ s$^{-1}$, based on the fact that the strain rate sensitivity of tungsten alloys is primarily due to the high rate sensitivity properties of the bcc tungsten particles. Using this hypothesis and McMeeking's computation for a rigid plastic material, the cleavage stress was taken to be the maximum value of the static normal stress ahead of the crack and equal to 240 percent the static yield stress. Finally, the toughnesses were calculated using (1) Tracey's computations [21] of the crack tip behavior under small scale yielding for a rigid plastic material and (2) the yield stress at a strain rate equal to the strain rate at the limit of the plastic zone [13]. These yield stresses were deduced using a power law fit that expresses the yield stress as a function of strain rate and an expression describing the strain rate at the limit of the plastic zone [13]. These expressions are provided in table III. Figure 9 shows the stress ahead of the crack as a function of the distance ahead of the crack divided by the stress intensity factor $K_I$, both quantities normalized by the yield stress. For a given value of the hardening exponent $N$, a plot is selected and the toughness is deduced using the calculated cleavage stress and the spacing between tungsten/tungsten interface for the stress and the distance ahead of the crack, respectively. No toughness was measured with the dynamic Co specimen of shorter crack length since no local strains were recorded.

Fig. 8.

Fig. 8. — Schematic of the brittle initiation fracture.

Fig. 9.

Fig. 9. — Tracey numerical results [20] of stress $\sigma_{yy}/\sigma_y$ ahead of a crack.
Table III summarized the parameters involved in these calculations along with the measured and calculated toughnesses. The models employed, involving the specific microstructural features identified in the fracture process, result in toughness values that clearly are scaling with the experimental results.

**Shear localization.**

Dynamic compression data for all three microstructures are summarized in Table IV. Compared with the tensile experiments performed at \( \dot{\varepsilon} = 10^3 \text{s}^{-1} \) (Tab. II), the Fe and Co alloys compressed at \( \dot{\varepsilon} = 5 \times 10^3 \text{s}^{-1} \) exhibit a substantial increase in yield strength. Equivalent lower dynamic strain rate data are not available for the Mn alloy, but the results of earlier testing of several conventional alloys \([7,8]\) at \( \dot{\varepsilon} = 5 \times 10^3 \text{s}^{-1} \) indicate that the former material is considerably stronger than the latter ones. After testing, no cracks were discernible on the outer surfaces of the specimens. However, the shear deformation behavior of each alloy was evident following sectioning and polishing.

### Table IV. — Dynamic compression properties.

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Strain Rate (s(^{-1}))</th>
<th>Yield Stress (MPa)</th>
<th>Maximum Strain in Test</th>
<th>Shear Band Width ((\mu\text{m}))</th>
</tr>
</thead>
<tbody>
<tr>
<td>Fe</td>
<td>5 400</td>
<td>1 400</td>
<td>0.47</td>
<td>no shear band</td>
</tr>
<tr>
<td>Co</td>
<td>5 100</td>
<td>2 300</td>
<td>0.42</td>
<td>25</td>
</tr>
<tr>
<td>Mn</td>
<td>5 200</td>
<td>1 700</td>
<td>0.45 (*)</td>
<td>10</td>
</tr>
</tbody>
</table>

(* Lateral deformation restricted after 0.32 axial deformation (9).

Localized shear zones were not observed within the sectioned Fe specimen; instead, the material was characterized by large, uniform, and uncracked shear bands on the order of 500 \(\mu\text{m}\) in width. This result contrasts from previous observations involving an as-sintered alloy differing form the present one only in its nickel/iron weight ratio, 7/3, *versus* the current 8/2 \([8]\). After 45 percent deformation, intense localized shear bands and associated shear cracks had formed within the 7/3 alloy \([8]\). From careful study of the as-processed microstructure, inhomogeneities were found consisting of pores and large matrix zones free of tungsten particles. Such regions may have served as initiation sites of the shear bands. Despite its shear banding capabilities, this particular 7/3 alloy might not be a dependable penetrator material, since other important mechanical properties such as toughness and ductility may be compromised by its microstructural inhomogeneities.

On the other hand, localized shear bands and cracks were observed in the Co microstructure, involving the transformation of tungsten grains into virtual lenses by combined compression and shear loading (Fig. 10a). This process, occurring over a narrow strip only 25 \(\mu\text{m}\) in width, was apparently facilitated by the presence of pre-deformed (by swaging) tungsten particles. This idea is supported by the non-hardening tensile behavior of the Co material (Tab. II), which implies that slip systems involved in the subsequent (to processing) deformation of the tungsten grains are readily re-activated. Strong tungsten/tungsten and tungsten/matrix interfaces are also believed to help the shear band formation, as evidenced by the homogeneous appearance of the heavily sheared zones containing parallel tungsten grains deformed in excess of 300 percent adjacent to the shear band crack (Fig. 10a).

Shear bands and associated cracking were likewise obtained within the dynamically compressed Mn alloy. The result is not surprising, since among the three tungsten systems the
Shear band cracks observed in the a) Co and b) Mn microstructures after 45 percent deformation in compression at a strain rate of $5 \times 10^3 \text{s}^{-1}$.

Mn system has an element, Mn, with a thermal conductivity ten times lower than Ni and Fe and with a melting temperature about 200 °C lower than the melting temperature of Ni and Fe, factors known to favor adiabatic shear banding [9]. In this case, moreover, the shear bands were only about 10 µm in width, which contrasts with the 40 µm wide bands observed in earlier tests of the similar (isotropic as-sintered microstructure) 7/3 alloy, see figure 10. The band width thus appears to scale with tungsten grain size, a conclusion supported by the non-isotropic grain Co alloy results as well.

**Summary.**

Simple models have been used to predict fracture toughnesses for two tungsten alloy systems. Under quasi-static conditions, the high toughness of the nickel-iron-tungsten alloy is attributable to the high matrix ductility. Under dynamic conditions the same alloy displays a low toughness associated with a brittle fracture induced by the cleavage failure of tungsten/tungsten interfaces. Under quasi-static and dynamic conditions, a high strength
nickel-cobalt-tungsten alloy was found to exhibit moderate toughness despite a brittle type of fracture. Again the tungsten/tungsten interface was identified as the main microstructural feature in these brittle fractures. Under dynamic conditions the tungsten/tungsten interface is the microstructural parameter governing fracture for both materials; thus, the higher toughness measured with the Co alloy seems to be related to a stronger tungsten/tungsten interface when compared to the Fe alloy. It was earlier established that the same microstructural parameter governs the dynamic tensile ductility [8]. It seems likely, therefore, that the best hope of enhancing dynamic ductility and, especially, dynamic toughness, is to improve the strength and/or ductility of the tungsten/tungsten interface.

Shear banding under dynamic compression has been previously observed for some as-sintered conventional tungsten systems to originate from either pore or tungsten-free zones [8]. With the need for dependable alloys, cleaner tungsten systems have been developed from which these apparent shear band initiation sites have been eliminated. Instead of shear banding, these « clean » sintered conventional tungsten systems, like the Fe alloy, exhibit large, but stable, shearing zones. However, by introducing new tungsten systems, such as the Mn and Co alloys, dynamic compression failure has been observed to occur via shear banding. For the Mn system, shear banding probably prevails because of matrix thermomechanical properties that favor adiabatic shear banding. For the swaged Co system, the factors favoring shear banding are not well defined. Specifically, the respective roles of both swaging and the matrix precipitates in the initiation of the shear failure remain to be established. It seems possible, however, that dynamic compression failure mechanisms in tungsten systems can be controlled either by altering the matrix characteristics or the processing procedure.

Acknowledgments.

The careful experimental work of V. Aaron, M. A. Griffin, and J. Spencer is greatly appreciated. Support of the Army Research Office under contract DAAG 03-88-K-0204 is gratefully acknowledged.

References