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Investigation of Ductile Fracture Under Tensile High Rate Loading

by

Gang Liu, M. S., B. S.

A thesis submitted to
the Faculty of Graduate Studies and Research
in partial fulfilment of requirements
for the degree of

Master of Applied Science

in Mechanical Engineering

**Ottawa-Carleton Institute
for Mechanical and Aerospace Engineering
Department of Mechanical and Aerospace Engineering**

Carleton University
Ottawa, Ontario
January 2002

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**INVESTIGATION OF DUCTILE FRACTURE UNDER TENSILE HIGH RATE
LOADING**

submitted by

**Gang Liu, M.S., B.S.
in partial fulfilment of the requirements for
the degree of Master of Applied Science**

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ABSTRACT

This thesis presents a study of ductile failure under high strain rate deformation based on the study of interrupted test specimen microstructures using light microscopy and Scanning Electronic Microscopy (SEM). The materials used in this study are tantalum, tantalum-tungsten, Armco iron, and leaded brass which are of different crystal structures, (i.e. BCC and FCC) and purities. The effects of triaxiality on damage rates are also studied using different specimen configurations, corresponding to uniaxial, c-notch, and e-notch.

The study has demonstrated that ductile failure is dependent upon the material purity, second phase particle size and distribution, and stress triaxiality level. The high purity metals studied (tantalum and tantalum-tungsten alloy) failed by ductile rupture to form a chisel point. Conversely, with the presence of inclusions or second phase particles, the Armco iron and brass failed by ductile fracture through void nucleation, growth, and coalescence. Under high triaxiality, the tantalum-tungsten alloy displayed cleavage fracture triggered by the presence of small inclusions.

The Gurson damage model is also employed to simulate ductile fracture through void nucleation, growth, and coalescence in the iron and brass samples. These calculations were performed using the explicit finite element code LS-DYNA. A comparison of measured and predicted damage rates reveals a reasonable degree of accuracy using the Gurson model.

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Chapter 1

Introduction

The fracture behavior of ductile materials under high strain rate deformation has been widely studied for decades, especially for defense applications, aircraft engine design and personal protection for example. Shock or impact loading may be generated in a machine or structure either by the collision of moving bodies or simply by the sudden application of a force or motion to the structure. Thus, a rigorous understanding of the behavior of materials and structures subjected to the impulsive loading is required. Understanding of material behavior and property changes under dynamic loading is still far from complete, and additional information on fracture toughness, strength, stiffness, and stress concentration effects under impact conditions would be of great help to designers. The understanding of the mechanics and the mechanism of fracture under high strain rate conditions, and the prediction of fracture are of significance, both industrially and academically.

Fracture in metals has been traditionally characterized by two modes: ductile and brittle. The distinguishing feature is whether the broken parts may be refitted together to regain essentially the original dimensions [1]. Ductile fracture is associated with dislocation motion (slip) and residual necking and there is no long-range association of microcracks with crystallographic or interfacial features. In addition, ductile fracture is generally not initiated in the presence of a stress concentration. Cracks that do develop during ductile failure are characterized by plasticity effects which involve the movement and concentration of dislocations at the tip of the advancing crack. When there are inclusions in ductile metals, fracture often happens by the nucleation, growth and coalescence of voids.

Brittle fracture, on the other hand, is associated with elastic stress behavior, that is, a lack of readily recognizable dislocation motion. Microcracks develop along crystallographic planes and are associated with long-range interfacial phenomena. Brittle cracks advance primarily as a result of systematic decohesion, called, cleavage. Thus, a typical ductile metal is soft but tough, since the stress required to move dislocations is small, whereas a brittle metal behaves like a covalently bonded material such as silicon; it is hard but fragile because the stress required to move dislocations is large. Fracture in metals and alloys can therefore be recognized as a matter of cohesion or cohesive strength. Whereas brittle elastic fracture is characterized by cleavage phenomena, brittle plastic fracture is characterized by features similar to ductile fracture, that is, hole growth on a microscopic level resulting in a dimpled fracture surface [2].

Besides the materials themselves, material fracture behavior also depends on the type of load (eg. tension, compression, torsion), the rate of deformation, the temperature, and the size and geometry of the specimen.

An impact load is distinguishable from a conventional load primarily by its suddenness of application and its brevity of duration. In addition, its intensity is usually of sufficient magnitude to produce extensive fracturing and large permanent distortion in the body upon which it acts [3]. The process of fracture of ductile materials under intense dynamic loading can be

mainly characterized by inertial effects (kinetic energy of void growth) which is different from that of quasi-static loading [4-7]. The influence of the thermal effect (adiabatic heating) generated by high rates of deformation, and the rate-dependent effect on the evolution of dynamic damage is also important [5, 7]. In order to simulate an impact process, many approaches and techniques have been developed. Among these techniques, the most widely used is the split Hopkinson bar for evaluation of material response to impulse loads [8]. In the current research, the tensile split Hopkinson bar has been used to impose high strain rate loading at strain rates of 10^3 to 10^4 s⁻¹. Different specimen geometries (uniaxial or notched) and different materials (brass, Armco iron, tantalum, and tungsten tantalum alloy) have been tested using the tensile split Hopkinson bar.

The microstructures of interrupted high strain rate test specimens, on the other hand, record each step of plastic deformation. In order to conduct metallographic examinations, previous researchers have developed momentum trapping techniques for the tensile split Hopkinson bar at Carleton University [9, 10]. A systematic characterization and comparison of the interrupted specimen microstructures for the deformed Armco iron, brass, tantalum and tungsten-tantalum alloy were developed utilizing optical microscopy and electron microscopy techniques.

1.1 Fracture Mechanisms of Ductile Material

Ashby [11, 12] classified three types of fracture: i) cleavage, ii) nucleation and growth of voids, and iii) rupture, as illustrated in Figure 1.1. At low temperatures the material may fail by cleavage (i.e. splitting along natural crystallographic planes), or by brittle fracture along grain boundaries (intergranular fracture), or it may fail in a ductile fashion by void coalescence or by intense shearing. At high temperatures it may fail by a variety of creep fracture mechanisms, some of which are transgranular, and some intergranular.

1.1.1 Cleavage

The main mechanism of brittle damage is the nucleation and growth of microcracks and their coalescence to initiate a crack at the mesoscale which has been defined by Lemaitre [13] as the scale at which the constitutive equations for mechanics analysis are written. It is quite evident that pre-existing flaws are the nuclei of crack. Microcracks can also nucleate through the interaction of dislocations alone. Rosenfield *et al.* [14] classified various possible nucleation mechanisms, as shown in Figure 1.2.

Research on flaw-free materials indicates that cracks are nucleated at grain boundaries by the stress concentrations associated with blocked shear bonds or dislocation pile-ups [15-19]. However, in many metals and alloys, particularly those that fail in a brittle and intergranular manner, there is little evidence of dislocation pile-ups, and fracture appears to originate and propagate within the grain boundaries. In 1956, Gifkins [20] proposed that there was another potential source of microcracks on grain boundaries due to the formation of ledges.

Ledges on the grain boundary are the result of agglomerations, or coalescence of grain boundary dislocations (GBDs) which have been observed under both transmission electron microscopy and field-ion microscopy [16-18]. Ledges can be caused by plastic deformation and can also occur as a result of GBD coalescence during grain growth (or related heat treatment) [2]. Since the grain boundary ledges possess large strain fields [21], they are relieved either by further slip or by microcrack initiation, when subjected to deformation-induced stress [22]. The microcrack nucleation mode through transgranular or intergranular microcracking at a grain boundary ledge, as illustrated in Figure 1.3, will depend primarily on the cohesive strength of the grain boundary as compared with the lattice (grain interior) [2].

The first work in quantitatively measuring fracture strength was done by Griffith [23],

and eventually led to the concepts of fracture mechanics [24]. Based on the premise that a crack is unstable if the stored energy released at fracture is greater than the creation of surface energy due to new surfaces, Griffith [23] in 1920 developed an analytical approach for fracture prediction. The model was set up as a panel of thickness, t , containing a crack of length, $2a$, subjected to a remote stress, s as shown in Figure 1.4. The Griffith equation is expressed as:

$$\sigma_f = \left(\frac{8E\gamma_s}{\pi a(1+\nu)(k+1)} \right)^{1/2} \quad (1.1)$$

where σ_f is the stress to fracture, γ_s the specific surface energy, and $k = 3 - 4\nu$.

Since the Griffith equation can only be applied to brittle materials which do not deform plastically, it is not always sufficient for fracture associated with plastic deformation. Orowan [25] added the effects of plastic deformation by simply noting that the effective surface energy can be substituted for the true surface energy in the Griffith equation. The effective surface energy, γ_e , includes the work of plastic deformation around the fracture surface, γ_p , and true surface energy, γ_s , as expressed in the following:

$$\gamma_e = \gamma_p + \gamma_s \quad (1.2)$$

1.1.2 Void Nucleation, Growth, and Coalescence

When polycrystalline solids do not cleave, they may fail in a transgranular, ductile manner. The mechanism of ductile fracture in metals has been clearly established, by metallurgical research [26-29], to consist of three principal stages which can be simply described as void nucleation, growth and coalescence. In the first stage, voids nucleate at the sites of inclusions or second-phase particles [30, 31], when the plastic deformation in the matrix has elevated the matrix/particle interface stress to the level where the particles either crack or debond from the matrix [28, 30-38]. In the absence of foreign particles, but when plastic flow is easy, ductile fracture can initiate from 'cavity dislocations' produced from pile-ups or intersecting slip planes. The subsequent void growth can be classified into two processes, that is, extensional and dilational void growth during further plastic flow. When the voids are large enough, they coalesce by expanding or flowing unstably into one another, leading to widespread cracking.

1.1.2.1 Void Nucleation

Void nucleation in metals can be defined as the formation of a void or cavity at either (1) a pile up of dislocations, (2) an inclusion, or (3) a grain boundary junctions [1]. The mechanisms of void nucleation in a ductile material can be categorized as homogeneous and heterogeneous. The main characteristic of homogeneous nucleation mechanism is that the void nucleates through the process of dislocation interactions and vacancy migration[39], and cavities are formed in areas characterized by high dislocation densities. Observations by Wilsdorf *et al.* [40-42] have shown the opening up of cavities in front of propagating cracks in high purity single crystal silver foils. On the other hand, the interaction between dislocation and inclusions, second-phase particles or other structural discontinuities is the main characteristic of heterogeneous nucleation.

The factors affecting nucleation at the sites of second phase particles or inclusions include the size, shape, property, and distribution of particles, as well as the properties of the interface between inclusion and matrix, and the stress state. Hirsch and Humphreys [43] have proposed a mechanism of plastic relaxation around small particles. At small strains prismatic loops are punched out and the stresses associated with the particle are reduced. This process is not available to the larger particles and larger particles fracture at lower strains [28]. When the size of inclusion is less than 2 μm [44], dislocation models are required to describe the microvoid nucleation process [30, 45]. If the size of inclusion is larger than 2 μm , plastic continuum models are applicable [32, 33]. Particles which are roughly equiaxed will generally suffer interfacial decohesion after some amount of plastic strain, while irregularly shaped particles or those with large aspect ratios often fail by internal fracture [46]. If the density of inclusions is higher on grain boundaries than within the grains, the voids nucleate on the grain boundaries and the fracture path may follow the boundaries, resulting in a ductile intergranular fracture. If the bonding of particles to the matrix is fairly weak, then voids can nucleate at low stresses and small strains. However, if the particle is strongly bonded to the matrix then large strains can precede nucleation. Gurland [28] determined the effect of stress state on particle fracture in a high carbon steel and found that in tension, torsion and compression, particle cracking occurred preferentially on planes normal to the maximum principal stress direction.

When dislocations pass through second phase particles, three kinds of mechanisms have been observed: (1) the particle is sheared or fractured [47]; (2) dislocation loops are generated around the particles [48]; or (3) the dislocation line passes the particle by cross slip, leaving dislocation segments behind [49]. Large plastic strains are possible both within an inclusion and at the interface between the inclusion with the matrix, depending on the respective mechanical properties [50].

Stress concentrations at hard particles can lead to local plasticity, which in turn can lead to inclusion-matrix separation or inclusion cracking [51-53]. A necessary condition for the nucleation of a microvoid, by decohesion of the particle/matrix interface, is that the elastic strain

energy released by the particle is at least equal to the newly created surface energy [30, 32, 35,45]. Gurland and Plateau [35] proposed a criterion for void nucleation due to particle fracture or decohesion, which is similar to Griffith's [23] brittle fracture criterion, as:

$$\sigma = \frac{1}{\alpha} \left(\frac{E\gamma_s}{l} \right)^{1/2} \quad (1.3)$$

with:

$$\gamma_s = \gamma_m + \gamma_p - \gamma_{mp} \quad \text{for interface decohesion}$$

$$\gamma_s = 2\gamma_p \quad \text{particle being sheared or fractured}$$

where d is the diameter of the particle, l is the length of elongated particle (for a spherical particle $l=d$), α is the stress concentration factor, σ is the matrix stress for void initiation, E is the Young's modulus, γ_s is the surface energy, and γ_m , γ_p , and γ_{mp} represent the surface energies of the matrix, particle, and particle/matrix interface, respectively. However, experimental results have indicated that Gurland and Plateau's approach is not a sufficient condition because plastic strain is required before void nucleation can occur [54]. The sufficient condition for microvoid nucleation by decohesion is the attainment of a critical normal stress at the particle/matrix interface [32].

Ashby [55] proposed a dislocation model for void nucleation around a second-phase particle. In this model, the yielding was assumed to be accomplished by the dislocation line pushing through and leaving loops behind. With the increase of dislocation pile-ups against the particles, the prismatic dislocation loops increase so that vacancy loops will eventually coalesce at the interface and initiate microvoids around the particles. This process is shown schematically in Figure 1.5. The matrix strain required for microvoid nucleation is given by

$$\varepsilon = \frac{CbL\sigma_{cr}}{a} \quad (1.4)$$

where C is a constant, b is the Burgers vector, L is the pile-up size, a is the particle size and σ_{cr} is the critical stress required for nucleation.

Another dislocation model of void nucleation for micron-sized, spheroidal second-phase particles has been developed by Goods and Brown [30, 45]. Considering the energy balance between the elastic strain energy, ΔE_{el} , released through nucleation, and the free energy, ΔE_s , due to the creation of new surfaces,

$$\Delta E_{el} + \Delta E_s \leq 0 \quad (1.5)$$

the critical condition for microvoid nucleation by decohesion of the matrix/particle interface is given by [30, 45]:

$$e_1^n = Kr(s_c - s_m)^2 \quad (1.6)$$

Where e_1^n is the critical strain required to nucleate microvoids by particle decohesion, K is a material constant related to the volume fraction of particles, r is the radius of particles, s_m is the macroscopic mean-normal, and s_c is the critical cohesive strength of the interface. Equation 1.6 shows that the nucleation strain is strongly influenced by the particle radius which has been confirmed by experimental results for the Cu-SiO₂ and Fe-Fe₃C systems[30, 45] and in a number of spheroidized steel specimens [56]. This effect is the direct result of the increased rate of dislocation storage and elevation of the local flow stress as the particle radius decreases.

When the size of the particle is relatively large, plastic continuum models are applicable [32,33]. Argon *et al.* [32,33] developed a continuum-plasticity model as a criterion for microvoid nucleation for relatively large particles of approximately spherical form in which:

$$\sigma_c = \sigma + \sigma_m \quad (1.7)$$

where σ is the equivalent stress, σ_c is the critical cohesive strength of the interface, and σ_m is the macroscopic mean-normal stress. The term $(\sigma + \sigma_m)$ in equation 1.7 represents the maximum stress on the particle/matrix interface, and microvoid nucleation occurs when this stress becomes equal to the cohesive bond strength of the interface σ_c . In contrast to the dislocation models, equation 1.7 for relatively large particles shows that there is no influence of absolute particle size on the magnitude of the interface stress and the corresponding nucleation strain.

1.1.2.2 Void Growth

After nucleation through either debonding or cracking of a second-phase particle or inclusion, microvoids will immediately begin the continuous extensional and volumetric growth process induced by remote applied stress and strain-rate fields [31]. Most of the earliest studies on the mechanism of hole growth [57,58] were largely empirical. Holes were considered to interact from the beginning, while the effect of the triaxiality was neglected. A number of micromechanical models, such as those employed by McClintock [59] and Rice and Tracey [60], make the simplifying assumption that voids are isolated with cylindrical or spherical shapes, respectively. It is possible to develop an adequate model in terms of a single void in an infinite plastic solid and give a reasonable approximation for the early stages of microvoid growth in which the holes are still small enough so that their interactions with each other may be neglected. But near final failure this is usually a very poor approximation due to the interaction of voids.

Among the earliest theoretical analysis, McClintock's approach [59] is the most successful model and has been widely adopted by later researchers. In McClintock's model, the material is assumed to contain three mutually perpendicular sets of cylindrical holes of elliptical cross-section with axes parallel to the principal directions of the applied stress (and strain increment), as shown in Figure 1.6. The mechanics problem then reduces to the generalized plane-strain deformation of a hole in an infinite medium. Taking the dimensions of a cell to be

l_a and l_b , which correspond to the respective hole spacings, and take a and b to the corresponding semi-axes of the holes, two hole-growth factors [61] can be defined as:

$$F_a = \frac{a / l_a}{a_0 / l_a^0}$$

$$F_b = \frac{b / l_b}{b_0 / l_b^0}$$

where a_0 , b_0 , l_a^0 , and l_b^0 are the initial values.

Based on Berg's approach [62] in which the change in shape of an elliptical hole is expressed in terms of the mean radius, $R=(a+b)/2$, and the eccentricity, m , (i. e., a and b are the semi-major and semi-minor axes), McClintock [59] derived an approximate solution for the growth of an elliptical through-thickness hole in an infinite plate as:

$$\ln \left[\frac{R}{R_o} \right] = \frac{\bar{\epsilon} \sqrt{3}}{2(1-n)} \sinh \left(\frac{\sqrt{3}(1-n)}{2} \frac{(\sigma_a + \sigma_b)}{\bar{\sigma}} + \frac{\epsilon_a + \epsilon_b}{2} \right) \quad (1.8)$$

where R_o is the initial average void radius, $\bar{\epsilon}$ is the effective plastic strain in the matrix, $\bar{\sigma}$ is the matrix flow stress, and σ_a , σ_b , ϵ_a , and ϵ_b are the stresses and strains in the principal directions, respectively. The eccentricity m was defined as:

$$m = \frac{\sigma_a - \sigma_b}{\sigma_a + \sigma_b} + \left(m^0 - \frac{\sigma_a - \sigma_b}{\sigma_a + \sigma_b} \right) \exp \left[- \frac{\sqrt{3} \bar{\epsilon}}{(1-n)} \sinh \left(\frac{\sqrt{3}(1-n)}{2} \frac{\sigma_a + \sigma_b}{\bar{\sigma}} \right) \right] \quad (1.9).$$

The parameter n in equation 1.9 is the hardening coefficient, defined in terms of the stress at the

point of maximum strain divided by the average stress over the stress-strain curve up to the maximum strain [63]

$$n = \left(\frac{\bar{\sigma}|_{\bar{\epsilon}_{\max}}}{\bar{\sigma}|_{\bar{\epsilon}_{AVG}}} \right) - 1 \quad ; \quad (1.10)$$

Thus,

$$\bar{\sigma} = K(\bar{\epsilon})^n. \quad (1.11)$$

The McClintock model has been validated in some experimental studies [64-67], but other experimental results indicate that the McClintock model can not be used to predict void growth [68,69], especially when the voids are not cylindrical [70].

A subsequent analytical model considering a spherical void geometry, as shown in Figure 1.7, was developed by Rice and Tracey [60] in terms of a rigid-plastic non-hardening material. Rice and Tracey classified the velocity field as three parts: (1) a velocity field resulting in a uniform strain rate field $\dot{\epsilon}_{ij}^{\infty}$, so as to meet remote boundary conditions; (2) a spherically symmetric velocity field corresponding to a change in volume of the void but no change in shape; and (3) a velocity field decaying at remote distances, which changes the void shape but not its volume. These velocity fields can be expressed as:

$$\dot{u}_i = \dot{\epsilon}_{ij}^{\infty} x_j + D \dot{u}_i^D + E \dot{u}_i^E \quad (1.12)$$

where D and E are constants to be determined, $\dot{u}_i^D$ is a spherically symmetric volume changing

field, and $\dot{u}_i^E$ is a shape changing field which preserves void volume.

The rates of change in the radius of the void $\dot{R}_k$, in the directions x_1, x_2, x_3 of the principal strain rates for the remote strain-rate field, were shown to have the form:

$$\dot{R}_{0k} = \left\{ (1+E) \dot{\varepsilon}_k + \left(\frac{2}{3} \dot{\varepsilon}_L^2 \dot{\varepsilon}_L^2 \right)^{1/2} D \right\} R_0 \quad (1.13)$$

where subscript k , $L=1, 2, 3$, and

$(1+E) \approx 5/3$ for linear hardening, and low values of σ_m for non-hardening,

$(1+E) \approx 2$ for high values of σ_m with non-hardening,

$D = \frac{\sqrt{3}\sigma^\infty}{4\tau^\infty}$ for strong linear hardening,

$D = 0.558 \sinh\left(\frac{\sqrt{3}\sigma^\infty}{2\tau_0}\right) + 0.008\nu \cosh\left(\frac{\sqrt{3}\sigma^\infty}{2\tau_0}\right)$ for a non-hardening material.

The parameter ν is the Lode variable [71] defined by:

$$\nu = - \frac{3\dot{\varepsilon}_2}{\dot{\varepsilon}_1 - \dot{\varepsilon}_3} \quad (1.14)$$

The work by Le Roy *et al.* [65] on low, medium and high carbon steels has confirmed the general validity of the Rice-Tracey equations for void growth. They assumed that materials

display a continuous linear increase in microvoid nucleation with applied strain and that the mean-normal stress increases continuously as external necking develops in tension specimens. Their theoretical results, obtained by carrying out a numerical integration of the Rice-Tracey equations to give estimates for the changing area-fraction of voids on a microsection with increasing plastic strain, gave good agreement for the low- and medium-carbon steels up to a strain of about 0.6.

These first micromechanical studies of void growth focussed on the growth of a single void in an infinite elastic-plastic solid, and the results were used to estimate critical strains for coalescence. Most of the subsequent work has focussed on incorporating the effect of the interaction between neighbouring voids, and there has also been some interest in micromechanical modelling of nucleation and void deformation in the final stages prior to failure by coalescence.

1.1.2.3 Void Coalescence

Metallurgical studies have shown that voids coalesce either by the formation of a shear band between adjacent voids [68,72] or by necking of the inter-void ligament [68,73,74] until complete impingement of adjacent voids is realized [75]. Sun Jun [73] has summarized void impingement under varying degrees of stress triaxiality, as shown schematically in Figure 1.8.

Voids may coalesce at the void-nucleation strain or following void growth depending on the current stress and the plastic limit-load stress of the matrix material [44]. When the current stress is larger than the plastic limit-load stress, microvoids coalesce at the void-nucleation strain, resulting in nucleation-controlled ductile fracture [31,76]. Otherwise, ductile fracture occurs through three phases, that is, nucleation, growth, and coalescence.

Microvoid coalescence at the void-nucleation strain, or nucleation-controlled ductile fracture [31,76] can occur only at very high mean-normal stress levels, where the maximum principal stress is so high that the stress on the inter-void matrix, at the microvoid nucleation strain, has already reached or exceeded the plastic limit-load stress [31,77]. Under these conditions there is no opportunity for dilational-plastic void growth to occur, and the microvoid nucleation strain is then equal to the total ductile-fracture strain [76,78]. This may happen for hard particles which are strongly bonded to the matrix [44].

Both equation 1.6, developed by Goods and Brown for micron-size particles, and equation 1.7, proposed by Argon *et al.* for larger particles indicate that the microvoid nucleation strain is greatly reduced at high mean-normal stress levels. Hence, at very high mean-normal stresses, the microvoid nucleation strains are always likely to be very small. This process is illustrated schematically in Figure 1.9. This two-dimensional plane-strain plasticity model for a square cell with an array of circular particles can be used to estimate the plastic limit-loads for the intervoid matrix between spherical particles. The critical mean-normal stress for nucleation-controlled ductile fracture, based on the analysis of the square unit cell of width $2W$ containing a hole of diameter $2a$, as illustrated in Figure 1.10, can be expressed as

$$\frac{\sigma_m}{2\tau} = 1.43V_f^{-1/6} - 1.41 \quad (1.15)$$

where σ_m is the mean normal stress, τ is the yield shear-stress for a non-hardening rigid-plastic solid, and V_f is the volume fraction defined by

$$V_f = \frac{\pi}{4\left(\frac{a}{W}\right)^2} \quad (1.16)$$

Equations 1.15 and 1.16 show that when the mean-normal stress exceeds a value of $\sigma_m/2\tau$, ductile fracture is likely to occur spontaneously at the microvoid nucleation strain whenever the volume fraction of void-nucleating particles exceeds a value of 0.014.

If the maximum principal stress for nucleation is less than the plastic limit-load stress, further plastic flow is required to bring about extensional and dilational void-growth [56,60,77] and reduce the plastic limit-load stress to the level of the maximum principal stress.

In the McClintock model [59], fracture is assumed to occur when there is a complete loss of cross-section. Voids coalesce as a gradual process involving the thinning down of the ligaments between the voids with increasing strain. Final separation would occur when the ligaments have reduced to zero width. Since it assumed that failure was due to stable void growth, McClintock's model can not successfully match experimental results. The detailed calculations in the McClintock theory showed that fracture strains were an order of magnitude greater than those observed experimentally.

In the Rice and Tracey model [60], since both extensional and dilational void-growth have been considered, an estimate can be made of the plastic void-growth strain needed to bring the intervoid matrix to the point where the limit-load stress has been reduced to the level of maximum mean normal stress. Thus presents the condition for incipient ductile fracture by microvoid coalescence, and the corresponding ductile fracture strain [6,14,15]. The Rice and Tracey model predictions are in better agreement with experimental results, particularly when they are combined with a later correction proposed by Huang [79].

Both the McClintock and Rice and Tracey models show that the rate of void growth is strongly dependent upon the level of hydrostatic tension. These results suggest that fracture by coalescence of voids would be promoted by a high level of triaxial tension. This has been confirmed by tensile test results [44] in which voids grow more quickly and remain more equiaxed at the center, rather than near the specimen surface, after necking.

The Thomason model of microvoid coalescence following void growth has been developed based on the attainment of unstable plastic flow within inter-void ligaments. The two-dimensional plane-strain plasticity model has been schematically illustrated in Figure 1.11 [44], and is expressed as [31,78]

$$\frac{\sigma_{lc}}{2\tau} = 1.49 \left(1 + \frac{1}{N} \right)^{1/3} - 0.91 \quad (1.17)$$

where $N=a/(W-b)$ defines the shape of the intervoid matrix, $2a$ and $2b$ are the major and minor axes, respectively, of the elliptical void, and $2W$ is the width of the unit cell. This model has been shown to give close estimates of the ductile fracture strain observed in real materials [31,78, 80] and to predict void and matrix geometry, at the onset of ductile fracture, which is fully consistent with micrographs of ductile-fracture surfaces and microsections immediately adjacent to the fracture surface [26-28, 31, 81].

1.2 Gurson Model

Gurson's model [82-84] is an approximate yield condition that is the most widely known dilatant plasticity model for the average response of a ductile porous solid. Based on experimental observation, using averaging techniques similar to those used by Bishop and Hill [10] and approximate upper bound solutions for a rigid perfectly plastic material containing a concentric cylindrical void (see Figure 1.12) or spherical void (see in Figure 1.13), Gurson proposed the following continuum model for porous plasticity:

$$\Phi = T_{eqv}^2 + 2f \cosh\left(\frac{\sqrt{3}}{2} T_{rr}\right) - 1 - f^2 = 0 \quad \text{for a cylindrical void} \quad (1.18)$$

$$\Phi = T_{eqv}^2 + 2f \cosh\left(\frac{1}{2} T_{kk}\right) - 1 - f^2 = 0 \quad \text{for a spherical void} \quad (1.19)$$

$$\text{where } T_{eqv} = \frac{\left(\frac{3}{2} \Sigma'_{ij} \Sigma'_{ij}\right)^{1/2}}{\sigma_o}, \frac{1}{3} T_{kk} = \frac{\frac{1}{3} \Sigma_{kk}}{\sigma_o}, \text{ and } \frac{1}{3} T_{rr} = \frac{\frac{1}{3} \Sigma_{rr}}{\sigma_o} \quad (1.20)$$

with Σ'_{ij} being the deviatoric components of stress Σ_{ij} , and σ_o being the yield stress.

These equations can be simply written as

$$\Phi(\sigma^{ij}, \sigma_M, f) = 0 \quad (1.21)$$

where σ^{ij} is the average macroscopic Cauchy stress tensor, σ_M is an equivalent tensile flow

stress representing the actual microscopic stress-state in the matrix material, and f is the volume fraction of voids [85]. When void volume fraction is equal to zero, the Gurson model will reduce to the standard Mises yield condition.

Further studies of the Gurson model have been carried out by Tvergaard and Needleman. Based on detailed numerical solutions for the onset of plastic flow localization in materials containing periodic distributions of cylindrical voids or spherical voids and comparison with predictions using the Gurson model, Tvergaard [86-88] introduced the parameters q_1 , q_2 , and q_3 to improve the Gurson predictions at small void volume fractions. Hence, the modified Gurson model became:

$$\Phi = T_{eqv}^2 + 2f^* q_1 \cosh\left(\frac{1}{2} q_2 T_{kk}\right) - 1 - q_3 f^{*2} = 0 \quad (1.22)$$

where the “calibration” coefficients q_1 , q_2 , and q_3 were suggested by Tvergaard to be

$$q_1 = 1.5, \quad q_2 = 1, \quad q_3 = q_1^2. \quad (1.23)$$

In equation 1.22, f^* is initially equal to the void volume fraction f . However, using equation 1.22 to predict the final failure of a porous ductile solid caused by void coalescence, loss of material stress carrying capacity occurs at the ultimate value $f = 1/q_1$, which has been illustrated in Figure 1.14. This result is much larger than the experimental observation of porosity of void coalescence [89,90]. Consequently Tvergaard and Needleman [85] introduced:

$$f^* = f \quad \text{if } f \leq f_c \quad (1.24)$$

$$f^* = f_c + \frac{f_u^* - f_c}{f_f - f_c}(f - f_c) \quad \text{if } f \geq f_c. \quad (1.25)$$

to accelerate the rate of coalescence once f reaches f_c . The rate of growth of the void volume fraction is due to the growth of existing voids and the nucleation of new voids, and can be written in the form:

$$\dot{f} = \left(\dot{f} \right)_G + \left(\dot{f} \right)_N \quad (1.26)$$

with the growth of existing voids

$$\dot{f}_G = (1 - f) \dot{\varepsilon}_{kk}^{p\infty} \quad (1.27)$$

and the nucleation of new voids

$$\dot{f}_N = Ah \dot{\bar{\varepsilon}}^p = \frac{f_N \dot{\bar{\varepsilon}}^p}{S_N \sqrt{2\pi}} \exp \left\{ -\frac{1}{2} \left(\frac{\bar{\varepsilon}^p - \varepsilon_N}{S_N} \right)^2 \right\} \quad (1.28)$$

where $\dot{\varepsilon}_{kk}^{p\infty}$ is the hydrostatic component of plastic strain rate (due purely to void expansion),

f_N is the volume fraction of void nucleating particles, ε_N and S_N are the average and standard deviation of the strains at which the particles nucleate voids.

1.3 High Strain Rate Deformation

Shock or impact loads may result in states of stress that are significantly more intense than for quasi-static loading. In addition, the material properties may be significantly influenced by rapidly applied loads or displacements [91]. Under high rates of strain, the mechanical behaviour of materials is characterized by an increased strain rate sensitivity, by increasing effects of mass inertia forces, and by the adiabatic character of the deformation process [92]. Experimental results indicate that the yield point generally increases with increased loading rate, as does ultimate strength [93]. On the other hand, there are a number of parameters that affect the material properties at high strain rates. These parameters include the strain actually imposed, the strain-hardening effects, and the thermal process control which will influence the movement of dislocations and their rate of flow. The influence of these parameters varies with the material considered, or at best with the crystallographic type, and it therefore becomes extremely difficult to categorize metals neatly into respective groups [94].

For the relation between stress, strain and strain rate, empirical formulae are now mostly replaced by material laws based on structural mechanical models, whose parameters are to be determined by adequate systematic methods. Also, special effects such as the influence of strain rate on the strain at the lower yield point of bcc-metals can be quantitatively described by simple models. The fracture mode and ductility are highly affected by the strain rate. The elongation at fracture can be increased by the stabilizing effect of the strain rate sensitivity in a manner similar to super plastic behavior. On the other hand, elongation can be reduced by thermal induced instability, by increased sensitivity to internal notches and by multiaxial stress states induced by inertial forces. Strain rate also affects the ductile fracture conditions as well as the transition temperature to cleavage fracture.

The Gurson theory has also been used to model the process of dynamic ductile fracture

by Rajendran and Fyfe [95], Johnson and Addessio [96], and Worswick and Pick [97]. Recently studies of inertial and thermal effects on dynamic growth of voids in ductile materials have been carried out by Wang [7,98]. Theoretical analysis indicates that the inertial effect (kinetic energy of void growth) mainly dominates the behavior of the void growth in temperature-dependent and high-strain-rate cases. Otherwise, the viscoplastic effect dominates and the inertial effect can be neglected. Based on this analysis, Wang [7] proposed a hollow-sphere model for a temperature-dependent viscoplastic material. This model provided the threshold stress for void growth, which depends on the initial porosity, the yield strength, the void surface energy density, the initial temperature, and the melting temperature [7].

Based on the Gurson theory, Wang [98] developed an approximate yield criterion for porous ductile media at high strain rates

$$\Phi(\Sigma_{ij}, \dot{E}_{ij}, \ddot{E}_{ij}, f) = \left(\frac{\Sigma_e - \Sigma_e^d}{\sigma_o} \right)^2 + 2f \cosh\left(\frac{3}{2} \frac{\Sigma_m - \Sigma_m^d}{\sigma_o} \right) - (1 + f^2) = 0 \quad (1.29)$$

where superscript d denote the quasi-static part and the dynamic part respectively. Thus a new concept, that the macroscopic stresses are composed of two parts, representing dynamic and quasi-static components, has been introduced.

Numerical analysis reveals that both the dynamic growth of voids and the loading surface have strong rate dependence and temperature dependence. Experimental data, theoretical analysis, and numerical modelling indicate that the inertial and thermal effects play a very important role in the dynamic behavior of voided nonlinear materials [99]. The dynamic yield function is dependent not only on the rate of deformation but also on the distribution of initial micro-damage, which is different from that of the quasi-static condition. It is evident that inertial effects play a very important role in the dynamic behavior of the yield function. However, it is also shown that when the rate of deformation is low, less than or equal to 10^3 s^{-1} , inertial effects become negligible small, and the dynamic yield function reduces to the Gurson model [98].

1.3.1 Material Constitutive Behaviour at High Strain Rates

Material behaviour at high strain rates has been studied for nearly a century [106]. The micromechanism of plastic deformation is the slip of dislocations in the slip systems which correspond to the particular type of crystal structure. When a FCC crystal is under plastic deformation at room temperature, dislocations mainly glide on $\{111\}$ planes along $\langle 110 \rangle$ directions. For a BCC structure, glide occurs in $\{110\}$, $\{112\}$, and $\{123\}$ planes along $\langle 111 \rangle$ directions. Research [107-113] has shown that dislocations in FCC and HCP metals glide more easily with lower rate sensitivity compared to BCC metals and other nonmetallic materials.

Deformation-induced metallurgical effects can be generally well documented to be the result of stress or strain-induced microstructures, or microstructural changes in crystalline (polycrystalline) metals and alloys [114]. Materials having low stacking-fault energies such as 304 stainless steel (21 mJ/m²) or brass (14 mJ/m²) develop planar dislocation arrays, stacking faults, microtwins or twin-faults [2]. Materials having high stacking-fault energies such as copper (70 mJ/m²) or nickel (128 mJ/m²) develop dislocation cells because dislocations can cross-slip easily [114]. In BCC metals, stacking faults have not been observed experimentally, and the ease of cross slip suggested that faults are at best of very high energy [115]. The relationship between the residual stress and residual microstructure can be expressed by the Hall-Petch relation:

$$\sigma_y = \sigma_o + K_1 \rho^{1/2} + K_2 V_c d^{-1} + K_3 V_T \Delta^{-1/2} + K_4 D^{-1/2} \quad (1.30)$$

where V_c and V_T denote the volume fractions of dislocation cells (c) and twin-faults (T), respectively, d is the dislocation cell diameter, D is the grain size, Δ is the spacing between twins or twin-fault bundles, and K_1 , K_2 , K_3 , and K_4 are material constants [114].

When materials are deformed under high strain rates, the dislocation velocity becomes

an important factor. The rate of inelastic strain due to the movement of dislocations is commonly expressed by Orowan's equation:

$$\dot{\epsilon}^p = b\rho_m(\epsilon^p, \sigma_s)\bar{v}(\epsilon^p, \sigma_s) \quad (1.31)$$

where b is the Burgers vector, ρ_m is the mobile dislocation density, and $\bar{v}$ is the average dislocation velocity. Both ρ_m and $\bar{v}$ are indicated by (ϵ^p, σ_s) to be functions of stress (σ_s) and plastic strain (ϵ^p). Considering the distinctly different behavior of BCC and FCC metals deformed under high strain rates, the relations between flow stress, strain, strain rates, and thermal effects for FCC and BCC metals have been incorporated into the Zerilli-Armstrong relations [116] as,
for the FCC polycrystalline materials:

$$\sigma = \Delta\sigma_G + c_2\dot{\epsilon}^{1/2} \exp(-c_3T + c_4T \ln \dot{\epsilon}) + kl^{-1/2} \quad (1.32)$$

and for the BCC metals as:

$$\sigma = \Delta\sigma_G + c_1 \exp(-c_3T + c_4T \ln \dot{\epsilon}) + c_5\dot{\epsilon}^n + kl^{-1/2}. \quad (1.33)$$

In equations 1.32 and 1.33, $\Delta\sigma_G$ is an additional component of stress accounting for the influence of solutes and the original dislocation density on the yield stress, σ is the plastic flow stresses of the BCC or FCC metals, ϵ is the strain, $\dot{\epsilon}$ is the strain rate, l is the average grain diameter, and c_i and n are material constants.

Another successful constitutive model of high strain-rate deformation was developed by Johnson and Cook [117, 118]. They have successfully described cylinder impact (Taylor) test results for a variety of materials by using a Lagrangian finite element code. A single type of

numerical constitutive relationship was employed in the code, giving the von Mises yield stress as follows:

$$\sigma = (A + B\epsilon^n) \left(1 + C \ln \dot{\epsilon} \right) (1 - T^{*m}) \quad (1.34)$$

where $\dot{\epsilon}$ is the strain rate, A, B, n, C, and m are material constants, T^* is the ratio $(T - T_{\text{room}})/(T_{\text{melt}} - T_{\text{room}})$, and T is the absolute temperature. The material constants are determined from tests done in tension or torsion. Strain hardening, strain-rate hardening, and thermal softening are taken into account.

1.4 Current Research

The main objective of the current research is to characterize and quantify the damage and failure processes occurring in several ductile materials under high strain rate conditions. A secondary objective is to assess or validate the numerical predictions of damage obtained using a so-called Gurson-based constitutive model applied to high rate conditions. In previous research, Worswick *et al.* [97, 100-102] applied this model to predict spall strengths in leaded brass under plate impact conditions with strain rates in excess of 10^4 s^{-1} . In that work, it was speculated that the predicted damage rates were too high. In other work, simulations of ductile damage in Armco iron were undertaken by Worswick and Pelletier [119]. One drawback of their study was that the nucleation parameters were not validated experimentally. Hence, the current research was undertaken to obtain accurate measures of damage development for these two materials under high rate conditions.

Four materials were considered in the experiments: Armco iron, leaded brass, pure tantalum and tantalum-10%tungsten alloy. The iron and brass materials contain second phase particles or inclusions which lead to ductile damage. The tantalum and tantalum-tungsten were extremely pure metals and were studied to determine whether ductile damage would form under high rate conditions. For each material, uniaxial and two types of notched tensile sample geometries were tested using a tensile split Hopkinson bar (TSHB) at strain rates on the order of $10^3 - 10^4 \text{ s}^{-1}$. The notched geometries were used to elevate triaxiality levels and accelerate damage rates.

Momentum trapping was used in the experiments to arrest deformation, thereby allowing studies of damage using a combination of metallographic techniques coupled with quantitative stereology methods developed by Pilkey *et al.* [103-104] and previously used to study Al-Si and Al-Mg alloys [103-110].

The measures of damage versus applied strain and triaxiality have been used to assess predicted damage rates from a Gurson based constitutive model. This model was implemented in the LS-DYNA finite element code user material subroutine written by Worswick and Pelletier [119].

The balance of this thesis is organized as follows. Chapter 2 presents the materials and as-received microstructures. The experimental methods are described in Chapter 3. The experimental results are presented in terms of the mechanical response and damage characterization in Chapter 4. In Chapter 5, the numerical models of the experiments are described along with a comparison of the predicted and measured damage. Further discussion of the results is in Chapter 6, while conclusions and recommendations for further research are given in Chapter 7.

Chapter 2

Materials

Four kinds of materials, Armco iron, leaded brass, tantalum and a tantalum alloy, were used in this research. Armco iron is a commercial high-purity material of BCC structure. Leaded brass is considered a good model material for the study of ductile fracture due to void nucleation, growth, and coalescence. The tantalum and tungsten-tantalum alloy used in this study are currently used in defence application because of their extremely high costs. Research on these materials will examine the effects of different crystal structure, material purities, alloying elements, second phase, and second phase particles on material behavior under high-strain, high strain-rate deformation. The metallographic study will be used to validate the numerical analysis, and comments on further research will be given based on the results.

In this chapter, the theoretical analysis of materials behavior under high strain-rate is discussed briefly. The materials used in this research are also discussed based on the survey on the previous research.

2.1 Armco Iron

Armco iron is a commercial high-purity alpha iron of BCC crystal structure. When Armco iron is deformed at large strains, the elevating dislocation densities are concentrated in dislocation cell walls. The residual microstructures developed in polycrystalline iron after deformation at high rates of strain were characterized by Khalid *et al.* [120] using transmission electron microscopy (TEM). Long, screw-type dislocations were observed in Hopkinson bar and impacted specimens, but no deformation twinning was found. Localized deformation bands are observed in impacted specimens. Furthermore, deformation twinning and dislocation cell structures were found in shocked specimens. The grain size of Armco iron not only affected the flow stress (Hall-Petch relation), but also affected the dynamic fracture toughness [121, 122], when the grain size was in the range 38 -1050 μm [121].

Typical tensile stress-strain plots associated with grain sizes in the range of 38 μm -1050 μm from Srinivas *et al.* [123] are shown in Figure 2.1. The phenomenon of pop-in, or drop in load, has been observed when the grain sizes smaller than 1050. It is clearly seen that the difference between the upper and lower yield points is largest for finest grain size (38 μm) and decreases markedly with increasing grain size, eventually disappearing at 1050 μm .

Microcracks or microvoids in Armco iron can be nucleated by either dislocation interaction or at the sites of inclusion. Cottrell [124] proposed that in BCC materials, i.e. iron at low temperatures, dislocations could interact on intersecting slip planes and that this might account for cleavage failures. In iron, the primary slip systems are $\{100\} \langle 100 \rangle$ and the resulting cleavage planes are $\{001\}$. This method of crack nucleation in iron has been confirmed by Honda under microscopic observations [125]. In contrast, Armco iron tensile tests performed by Qiang *et al.* [126] found that fracture initiated at inclusions.

The Armco iron used in this research project has been produced by ARMCO GmbH, a subsidiary of Thyssen Stahl (situated in Köln, Germany) and obtained through Koëinig industries in Switzerland. The samples were from material used by SNC Industrial Technologies Inc. to produce explosively formed projectiles (EFP) liners. The specimens were from 25 mm plate purchased from ARMCO GmbH. Observation of the as-received Armco iron using the Scanning Electron Microscope (SEM) revealed elongated slag inclusions distributed along the rolling direction (see Figures 2.2 (a) and (b)). The EDX analysis indicated that the slag contains Al, Si and Mn, as shown in Figure 2.2 (c). Before machining the specimens, the Armco iron was annealed at 550 °C for 2 hrs. After heat treatment, the slag distribution was examined using light microscopy, as shown in Figure 2.3. The samples of Armco iron have been etched and the grain size was measured to be approximately 30 μm (Figure 2.4). The material properties and constitutive relations of this material have been determined in previous work [9, 101] and that data was used in this research.

2.2 Leaded Brass

The brass used in this study was UNSC 3600 free machining or “leaded brass”. The matrix is a copper-zinc alloy in which the zinc content ranges up to 39%. There are two phases in a brass, that is, alpha phase (α) and beta phase (β). Alpha phase is face-centered cubic and shows a maximum zinc content of 39% at 454 °C falling to about 33% at 150 °C. Beta phase is a body-centered cubic phase resulting from the peritectic reaction between the alpha phase and the melt. It is of the disordered type above 460 °C and ordered below this temperature. The beta phase is relatively brittle and high in hardness compared to the alpha phase. The mechanical properties of brasses vary widely with the strength and hardness depending on alloying and/or cold work. Free-cutting brass has 3.25% lead which provides optimum machinability.

Leaded brass is used as a model material for the study of ductile fracture. The lead particles in a leaded brass, being considerably weaker than the metal matrix, burst or tear easily during deformation introducing well-defined void nucleation sites. Through the application of suitable annealing heat treatments, spheroidization and coarsening of the lead particles can be achieved. Such heat treatments allow model materials with different particle morphologies (size, shape, and spacing) to be tested. Once nucleated, the voids grow in a stable manner with continued deformation until a global or local instability causes the voids to coalesce or link, through localized intervoid shearing, which eventually leads to the formation of a catastrophic crack.

The leaded brass used in this research was UNS C3600 free-cutting brass which is predominantly an alpha brass and small regions of beta brass containing 2.5 % dispersed lead phase. The as-received brass was annealed at 850 °C for 2 hrs in order to promote spheroidization of the lead particles, as shown in Figure 2.5. This heat treatment [10] resulted in an average particle size of roughly 3–4 μm with an aspect ratio near unity. The grain size is typically 50-55 μm while the average lead particle size is roughly 5 μm . The in-plane mean nearest-neighbor lead particle spacing is approximately 15 μm and the volume fraction of lead particles is 0.022.

The material stress-strain behavior was described using a polynomial fit to quasi-static uniaxial tensile data by Worswick and Pick [97]. The flow stress obtained using this fit was then scaled using a log-strain-rate relationship:

$$\bar{\sigma} = (94.4 + 1594\bar{\epsilon}^p - 1693(\bar{\epsilon}^p)^2 + 645(\bar{\epsilon}^p)^3)(1 + 0.0009 \ln(\dot{\epsilon})) \quad (2.1)$$

where $\bar{\sigma}$ is the material flow stress, $\bar{\epsilon}^p$ is the effective plastic strain, and $\dot{\epsilon}$ is the effective strain rate.

2.3 Tantalum and Tungsten Tantalum Alloy

Tantalum is a transition element in Group VB of the Periodic Table with an atomic number of 73 and atomic weight of 180.95. The crystal structure of tantalum is BCC and the rate of strain hardening is low. Bechtold [127] determined that the work-hardening exponents are rather small for annealed tantalum, especially below room temperature, and are observed to decrease with decreasing testing temperature. It must be emphasized that the mechanical properties of tantalum are dependent on the purity and the metallurgical state of the metals, i.e., on their mechanical and thermal history [128]. Therefore, property data are most meaningful when accompanied by a precise description of the metal.

Tungsten and tantalum alloys are binary isomorphous alloy, i.e. tungsten (BCC) and tantalum (BCC) have complete liquid solid solubility. With increased levels of tungsten, the temperature of both liquidus and solidus lines increases from 3020 °C (the melt temperature of tantalum) to 3387 °C (the melt temperature of tungsten). Ta-W forms a substitutional alloy. The binary addition of tungsten is found to be effective in raising the recrystallization temperature of pure tantalum. The highest recrystallization temperature measured was 1600 °C for the Ta-20%W alloy. While Ta-15% W and Ta-10% W-C (630 ppm C) had recrystallization temperatures of 1500 °C.

The mechanical properties of a large number of binary and ternary tantalum alloys have been surveyed by Schmidt *et al.* [129-131] at both room and elevated temperature. Their data showed that tungsten is a good strengthener at the 5% addition level. Meanwhile, the addition of tungsten noticeably reduces the strain-rate dependence of the flow stress of tantalum near yield. It also subtly alters the strain-rate behavior of the work hardening, making it more like that of copper, an FCC metal. Tungsten alloys may therefore be preferable to unalloyed tantalum in

applications that demand substantial ductility at high rates of strain [132].

The effects of strain rate, temperature, and tungsten alloying on the yield stress and strain-hardening behavior of tantalum were investigated by Chen *et al.* [133]. Their results showed that the yield and flow stresses of unalloyed Ta and tantalum-tungsten alloys exhibit very high rate sensitivities. The hardening rates in Ta and Ta-W alloys were found to be insensitive to strain rate and temperature at lower temperatures or at higher strain rates. The dependence of yield stress on temperature and strain rate was found to decrease, while the strain-hardening rate increased with tungsten alloying content.

TEM observations of the substructures of tantalum [134-138] under high-strain, high-strain-rate deformation showed well-defined dislocation arrays, grouped dislocations, elongated dislocation cells, subgrains, and dynamically recrystallized micron-sized grains. The mechanism of the evolution of the microstructure progresses from highly dislocated grains, to elongated dislocation cells, to subgrains, and to micrograins as the shear strain increase. Dynamic recovery and recrystallization have been observed in high plastic deformation zones due to the high estimated adiabatic temperatures expected in those regions [139].

Specimens of the tantalum and tungsten-tantalum alloy studied in this research were fabricated from rod stock and used in the as-received state.

Chapter 3

Experimental Details

The experimental component of this research includes high strain rate tensile split Hopkinson bar (TSHB) tests followed by characterization of the test specimen microstructures. The constitutive behavior of the different materials under high strain-rates has been determined. In previous Mechanical Engineering theses at Carleton University, momentum trapping techniques have been developed for the TSHB system [9,10]. For the current research, this technique has made it possible to examine the microstructures of each sample corresponding at various levels of deformation. In addition, three specimen geometries were considered, a uniaxial sample and two kinds of notched geometries, in order to vary the level of stress triaxiality and resultant damage rate. Metallographic examination under an optical microscope included image acquisition from as-polished specimens for porosity measurement as well as microstructural observation of etched specimens. Fractographs of all the materials used in this research have been analyzed using Scanning Electron Microscopy (SEM). In this chapter, the principles of TSHB testing are discussed briefly and the experimental conditions are discussed in detail.

3.1 Mechanical Testing

In this section, the principles behind TSHB testing are discussed, after while sample design and geometry are presented.

3.1.1 TSHB Testing

The split Hopkinson Bar technique has been widely used in studying high strain rate deformation. A split Hopkinson bar generally consists of two long elastic cylindrical bars, the incident and transmitter bars, with a specimen positioned between them. A striker is accelerated to a high speed and impacts the end of the incident bar causing a stress wave to be transmitted along the incident bar through the specimen to the transmitter bar. Figure 3.1 [140] illustrates the components of a tensile split Hopkinson bar. The striker, incident bar and transmitter bar must remain elastic and are made from a high-strength material. The diameter of the bars is chosen depending on the desired strain rate. Higher strain rates necessitate a smaller bar diameter in order to avoid lateral inertia effects. The theoretical analysis is carried out based on the following assumptions [141]:

1. The Hopkinson bars remain elastic;
2. Wave propagation may be idealized as one-dimensional;
3. A uniform strain condition exists within the specimen.

The stress wave passing through the bars can be treated as a one-dimensional longitudinal plane wave. A longitudinal plane wave is characterized by the condition that the displacements of the particles of the material all take place in the direction of a single Cartesian coordinate and that the wave also propagates in this same direction. The equation of motion of an infinitesimal

element of the material for this case is

$$\rho \frac{\partial^2 u}{\partial t^2} = \frac{\partial \sigma_x}{\partial x} \quad (3.1)$$

where x is the coordinate; u is the particle displacement; t is the time; σ_x is the direct stress in the x direction (with tension stress taken to be positive), and ρ is the density of the material. For the longitudinal motion of long, thin, elastic rods [142], the stress-strain relation can be expressed by Hooke's law,

$$\sigma_x = E \varepsilon_x \quad (3.2)$$

where E is Young's modulus, and

$$\varepsilon_x = \frac{\partial u}{\partial x}. \quad (3.3)$$

Equation 3.1 can then be rewritten as:

$$\frac{\partial^2 u}{\partial t^2} = \frac{E}{\rho} \frac{\partial^2 u}{\partial x^2}. \quad (3.4)$$

It can be shown by substitution that any function $f(x + \sqrt{E/\rho}t)$ is a solution to equation 3.1 as is any function $f_1(x - \sqrt{E/\rho}t)$. Therefore, the general solution for the differential equation of motion given in equation 3.1 may be written as:

$$u(x, t) = f(x + c_0 t) + f_1(x - c_0 t) \quad (3.5)$$

with:

$$c_0 = \sqrt{\frac{E}{\rho}}. \quad (3.6)$$

The function f_1 represents a wave propagating in the $+x$ direction without change in form while f similarly represents a wave propagating in the $-x$ direction.

The propagation of plastic strain waves in long thin rods can be expressed using von Karman theory [143]. Assuming that the plastic stress-strain relationship is independent of strain rate:

$$\sigma_x = \sigma(\varepsilon) \quad . \quad (3.7)$$

Equation 3.1 can be applied to the case in which plastic as well as elastic deformation is involved for which the velocity of wave propagation becomes:

$$c = \sqrt{\frac{1}{\rho} \frac{d\sigma}{d\varepsilon}} \quad (3.8)$$

When an elastic stress wave travels through an interface in TSHB, it partially transmits through the interface and partially reflects back. As a result, the strain rate in the sample can be expressed as:

$$\dot{\varepsilon} = \frac{v_1 - v_2}{L_0} \quad (3.9)$$

where L_0 is the initial gauge length of the specimen. The parameters v_1 and v_2 are the velocities in the incident bar-specimen and specimen-transmitter bar interface, respectively, and can be expressed as:

$$v_1 = c_0(\varepsilon_I - \varepsilon_R) \quad (3.10a)$$

and

$$v_2 = c_0 \varepsilon_T \quad (3.10b)$$

where ε_I , ε_R , and ε_T are the strains due to the stress pulse. The subscripts I, R, and T indicate the incident, reflected and transmitted pulses, respectively. In the TSHB apparatus, the incident and transmitted pulses generate tensile stresses and are defined as positive, while the reflected one is compressive. So the strain rate $\dot{\varepsilon}_{(t)}$ and the average stress $\sigma_{AVG(t)}$ experienced by the specimen can be expressed as:

$$\dot{\varepsilon}_{(t)} = \frac{c_0}{L_0} [\varepsilon_{I(t)} - \varepsilon_{R(t)} - \varepsilon_{T(t)}] \quad (3.11)$$

$$\sigma_{AVG(t)} = \frac{F_{I(t)} + F_{T(t)}}{2A} \quad (3.12)$$

where A is the instantaneous cross-sectional area of the specimen. $F_{I(t)}$ and $F_{T(t)}$ are the forces at the incident bar-specimen and transmitter bar-specimen interfaces. These forces are given by:

$$F_{I(T)} = E[\varepsilon_{I(t)} + \varepsilon_{R(t)}]A_b \quad (3.13a)$$

and

$$F_{T(t)} = EA_b[\varepsilon_{T(t)}] \quad (3.13b)$$

where E is the Young's modulus and A_b is the cross-sectional area of the bar. Since the forces in both bar sections at the interface must be equal, then:

$$\epsilon_{I(t)} + \epsilon_{R(t)} = \epsilon_{T(t)}. \quad (3.14)$$

All told, the final form of the Hopkinson bar equations are:

$$\dot{\epsilon}_{(t)} = -\frac{2c_0}{L_0} \epsilon_{R(t)} \quad (3.15)$$

and

$$\sigma_{AVG(t)} = E \frac{A_b}{A} \epsilon_{T(t)} \quad (3.16)$$

where A is the instantaneous cross-sectional area of the specimen, A_b the cross-sectional area of the bar.

At Carleton University, the strain gauges were bonded onto both incident bar and transmitter bar. The Wheatstone bridge principle was used to convert load applied to the specimen into an electronic signal. Calibration of the strain gauges has been done by Pelletier [9] and the detailed information regarding this process can be found in [9]. A series of strikers with lengths ranging from 4 inches to 20 inches have been designed so that the duration of the stress pulse can be varied from 40 μ s up to 200 μ s. A novel technique used in the Tensile Split Hopkinson Bar at Carleton University is the use of momentum trapping systems on both the incident and transmitter bars, as illustrated in Figure 3.2 [9]. The advantage of this technique is its ability to interrupt testing and avoid reloading of the specimen by reflected pulses in a manner similar to that used by Nemat-Nasser [144]. This allows recovery of the interrupted sample for metallographic examination after any given amount of deformation. Detailed information concerning this technique is presented by Pelletier [9] and Fowler [10].

3.1.2 Specimen Design

In this research, uniaxial and notched specimen configurations have been employed based on the study by Pelletier [9]. The uniaxial configurations were used to obtain the stress-strain curves under uniaxial stress conditions. The purpose of the notched configurations was to obtain tensile data under the higher triaxiality induced by the notch and to focus deformation within the smaller notched region, thus obtaining higher strains for a given incident strain pulse. Notched geometries were also employed following the approach of Hancock and MacKenzie [147-148]. The effect of the notch is to elevate the stress triaxiality at the notch center, thereby elevating damage rates [145]. Notched specimens also provide changes in gauge diameter, causing the local strain rate to rise to nearly 10^5 s^{-1} . Figures 3.3, 3.4, and 3.5 illustrate the three kinds of specimen configurations adopted for this research, i.e. uniaxial, c-notch, and e-notch. The triaxiality, defined as the ratio of the hydrostatic or mean stress over the flow or effective stress, has been calculated for the three different configurations. The maximum values of triaxiality, based on a Bridgman [145] analysis, were given in [9, 10] as 0.333, 0.612, and 1.026 for uniaxial, e-notched, and c-notched specimens, respectively. Both notched and uniaxial specimens were machined with left-hand threads on one end of the specimen and right-hand threads on the other end so that they could be screwed onto the bars without turning the bars. Tables 3.1 to 3.3 summarize all of the TSHB experiments performed in this research, noting the test conditions and the degree of deformation produced.

Prior to an experiment, the dimensions of the specimen were measured with a Vernier caliper and an optical comparator to check their precision and to obtain an accurate measure of the gauge length and the gauge diameter for subsequent stress, strain, and strain rate calculations. The level of precision obtained for the gauge length and diameter, which are the critical dimensions, was within $\pm 0.125 \text{ mm}$ ($\pm 0.005''$). The precision for the other specimen dimensions was within $\pm 0.51 \text{ mm}$ ($\pm 0.020''$).

3.2 Metallography

All the interrupted test specimens obtained using the TSHB momentum trapping techniques were examined using light microscopy. These examinations included microstructural observations, image acquisition and porosity determination. Samples strained to various levels were sectioned, ground, and polished for porosity measurements. Images were acquired under light microscopy and the void volume fractions were measured using an image processing code programmed in C++. The polished samples were then etched to examine the microstructural change of the matrix with strain. Fractographs were obtained using Scanning Electron Microscopy and used to characterize the final failure modes for each material.

3.2.1 Sample Preparation

All light microscopy specimens were carefully prepared to obtain precise metallographic results. The procedure of sample preparation consisted of the following:

1. All momentum trapped specimens were sectioned on a milling machine to within roughly 1mm of the middle axial plane.
2. In order to fit within the diameter of the bakelite mounts, the threaded end on one side of the notched specimens and both ends of the uniaxial specimens were cut off.
3. The specimens were thermally mounted in 24 mm bakelite mounts for 40 minutes.

4. The mounted specimens were rough ground on 120 grit silicon carbide paper, followed by 240 grit silicon carbide paper to within 0.5 mm of the middle axial plane, using water as a lubricant.
5. The specimens were fine ground on a Buehler ECONOMET III automatic polishing machine fitted with a WHIRLMET polishing attachment, using 400 and 600 grit silicon carbide paper for the Armco iron, tantalum, and tungsten-tantalum alloy samples and using 400, 600, and 1200 grit silicon carbide paper for the brass samples. The wheel was set to rotate at 125 rpm and a 1422 g weight was used with water as a lubricant.
6. The specimens were roughly polished for 6 minutes on Buchler METCLOTH cloth charged with 6 μm and then 1 μm oil-based diamond suspensions in order for the Armco iron, tantalum, and tungsten-tantalum alloy specimens. Only the 1 μm oil-based diamond suspension was used for brass. A 1422 g weight was used in all cases.
7. The specimens were fine polished on METCLOTH cloth charged with 0.25 μm oil-based diamond suspension for 6 minutes, employing a 1422 g weight.
8. The specimens were finally polished on a Buehler Vibromet I vibrational polishing machine for 4 hours on a Buchler MASTERTEX polish cloth at a rotational rate of 6 revolutions per minute, using a 600 g weight. The Struers OP-S collordal SiO_2 Suspension was used as a polishing agent for the brass, while the Struers OP-U SiO_2 -Suspension was employed for the Armco iron, tantalum, and tantalum-tungsten alloy samples.

The as-polished specimens were examined with the aid of an image data acquisition system, from which the porosity of the materials was measured. The polished Armco iron specimens were etched using 3% nitric acid. After etching, all of the samples were cleaned by water and alcohol and dried by an air blast. Further observations of the microstructure were carried out on etched specimens.

3.2.2 Optical Microscopy

The samples were examined optically using an Olympus PME-3 inverted stage optical microscope under bright field illumination. For image analysis, the microscope field was digitized by quantifying the light intensity with a grey scale value ranging from black (0) to white (255) at each discrete point in the image. The spatial resolution of the digitized image is 640 by 480 pixels [10]. The image acquisition was controlled by a modified version of the ENCODE [146] software developed at Carleton University (refer to [10] for the details).

All of the images were converted to particle bit map (pbm) files using code developed by Pilkey [75]. The pbm files were then passed through an image processing program, written in C++ code and running on a PC, to calculate the porosity values.

3.2.3 SEM Observation

The fracture features in all of the samples tested to failure were observed utilizing a scanning electron microscope (SEM). The type of SEM is a JSM6400 with an EDX Link Analytical System exL. The working voltage used in this experiment was 20 kV for all the materials. Energy-dispersive X-ray Spectrometry (EDS) was used to analyze inclusions or second phase particles. A Digital Image Acquiring System was employed for the image processing of SEM samples.

Chapter 4

Experimental Results

The experimental results include both the mechanical behavior and the damage evolution within the various materials under high strain rates. Using the tensile split Hopkinson bar, the dynamic stress-strain response was determined from the measured strain gauge signals. By taking the advantage of the interrupted TSHB technique, physical measurements of axial strain (engineering strain) and diametral strain (true strain) were obtained from the interrupted specimens. From these measurements, the engineering stress-strain data was converted to true stress-strain data.

The damage evolution of materials under high strain-rates was investigated on the interrupted specimens using light microscopy. Scanning electron microscopy (SEM) was used to observe the fracture surfaces of these materials for the different specimen configurations (notch geometries). The ductile fracture mechanism of void nucleation, growth, coalescence was observed in the Armco iron and brass. Ductile rupture through gross necking was observed in Ta and TaW.

4.1 Mechanical Test Results

4.1.1 TSHB Test Results

The various striker velocities for the TSHB tests were established for each specimen geometry (i. e. uniaxial, c-notch, and e-notch) and for the various materials used in this research. For tantalum and tantalum-tungsten specimens, the strain rates are roughly from 1800 to 8800 s⁻¹ with the highest rates in the notched samples. Table 3.3 lists the striker velocities used for the tantalum and tantalum-tungsten samples. For Armco iron, the uniaxial specimens were tested at striker velocities of 18.5 m/s, while the c-notch and e-notch used 7 m/s and 10 m/s, respectively. These striker velocities produced strain rates in the range of 1200 - 2500 s⁻¹. The strikers velocities for the brass tests were 17.4 m/s, 9.5 m/s and 12.5 m/s for the uniaxial, c-notch and e-notch specimens, respectively, producing strain rates in the range 1700 - 4200 s⁻¹. Figures 4.1 - 4.4 show the histories of strain rate and engineering stress versus engineering strain for Armco iron, brass, tantalum, and tantalum-tungsten. The strain rates during the tests are roughly constant, apart from the initial and final transients, resulting in a roughly square shape. The stress history shows oscillation of stress with time associated with the reflection of stress waves within the sample length. The notched specimens exhibit less oscillation due to the shorter gauge length.

The specimen configurations cause different mechanical behavior in the TSHB test. For Armco iron, the c-notched configuration is of highest triaxiality, resulting in a peak stress of about 800 MPa as shown in Figure 4.1 (b). Since the e-notched configuration is of intermediate triaxiality, its peak stress is about 650 MPa, while the uniaxial yield strength is roughly 500 MPa. Similar trends have also been observed in the other materials as shown in Figures 4.2 (b), 4.3 (b), and 4.4 (b) for brass, tantalum, and tantalum-tungsten, respectively. The yield stress of brass, as shown in Figure 4.2 (b), is roughly 150 MPa based on the uniaxial configuration. The peak

stresses of e-notched, and c-notched configurations are 210 MPa, and 320 MPa, respectively. In Figure 4.3 (b), the peak stresses of the pure tantalum are seen to be 450 MPa (yield value) for the uniaxial specimen, 550 MPa for the e-notched specimen, and 750 MPa for the c-notched specimen. Figure 4.4 (b) shows the tantalum-tungsten peak values of the uniaxial, e-notched, and c-notched configurations to be roughly 650 MPa, 900 MPa, and 1200 MPa respectively.

The Armco iron, tantalum, and tantalum-tungsten materials display considerable strain softening at large strains as necking and failure progress. In contrast, the lower strength brass shows considerable strain hardening. Since the uniaxial tensile test provides the general constitutive relationships for materials, the uniaxial responses for each material are compared in Figure 4.5. The other notch configurations are compared in Figures 4.6 and 4.7, respectively. The ranking of strengths is consistent with the uniaxial data.

The strong softening response seen in the tantalum and tantalum-tungsten is due to a low work hardening response combined with a tendency to neck to a chisel point. This behavior was observed in all three specimen configurations.

4.1.2 True Stress Versus True Strain

The relationship between true strain (diametral strain) and engineering strain (axial strain) was determined based on a physical measurement of the changes in both diameter and length of specimens interrupted at different levels of strain. This relationship is needed to obtain true stress-strain curves from the measured engineering stress-strain data. The data was fit to analytical equations using MATLAB. A convenient power-law expression was used to fit the data

$$\varepsilon_T = C\varepsilon_E^B \quad (4.1)$$

where ε_T is the true strain, ε_E the engineering strain, and B and C are material-dependent constants. The results of these regressions for the Armco iron and brass notched samples are given in Table 4.1. The uniaxial samples were not fit nor were the tantalum and tantalum-tungsten data since the samples necked substantially to a chisel point.

Using the analytical expression in equation 4.1, true stress can be obtained using

$$\sigma_T = \frac{F}{A} \quad (4.2)$$

and

$$\sigma_T = \frac{F}{A_0} \frac{A_0}{A} = \sigma_E \exp(\varepsilon_T) \quad (4.3)$$

where A_0 is the initial cross-sectional area of the specimen and F is the axial load acting on the specimen.

The fits to equation (4.1) and the resulting true stress-strain curves are plotted in Figures 4.8 - 4.15. In general, these regression fits appear quite good. The Armco iron samples exhibit relatively less-hardening response. The strain softening in the Armco engineering stress-strain curves in Figure 4.1 can therefore be attributed to a gross reduction in specimen cross-section as well as thermal softening due to adiabatic heating.

4.2 Damage Processes

4.2.1 Damage Development - Armco Iron

The evolution of damage within Armco iron samples with c-notched configuration has been recorded using optical microscopy, as shown in Figures 4.16 to 4.20. Figure 4.16 shows the undeformed specimen, whereas, Figures 4.17 to 4.20 show the progression of void nucleation, growth, and coalescence associated with different levels of deformation corresponding to the samples labeled 1, 2, 3, and 4 in Figure 4.9, respectively. Void nucleation was observed after a strain of 10%. The growth did not progress uniformly within the sample. Some nuclei grew quickly and formed deep cavities (Figure 4.18), while void coalescence was observed after 32 % true strain. After the onset of coalescence, the crack developed rapidly, which is seen in Figures 4.19 and 4.20.

The damage sequence for the e-notched Armco iron specimens is shown in Figures 4.21 to 4.25. Note that the stress-strain curves labeled 1-5 in Figure 4.11 correspond to the specimens in Figures 4.21 to 4.25 respectively. Void nucleation and growth is evident at a strain of 27% (Figure 4.22). Local coalescence within void clusters is observed at a strain of 29% in Figure 4.23, leading to the formation of larger voids. The coalescence of these larger voids is seen at a strain of 40% (Figure 4.24) resulting in large crack formation. This process is quite variable as a large stable crack is also evident at a strain of 64% in Figure 4.25. For the e-notch geometry, the voids were elongated along the tensile stress direction, producing an elliptic shape, as seen in Figure 4.24, for example. This contrasts the more spherical void shapes seen in the higher triaxiality c-notch samples (see Figure 4.18).

4.2.2 Damage Development - Brass

Figure 4.26 shows the initial microstructure of the leaded brass. In general, the lead particles are quasi-spherical and display variable clustering. Figures 4.27 to 4.29 show the void evolution in samples tested to strains of 13%, 25%, and 36% (corresponding to curves 1, 2, and 3 in Figure 4.13, respectively). Void growth becomes apparent at a strain of 13% as shown in Figure 4.28. Void coalescence is observed at strain of 25%, associated with the formation of larger voids. There were no large cracks observed before 39.5% strain.

The evolution of damage for the e-notched brass samples is seen in Figures 4.30 to 4.33. Again, the stress-strain curves for the corresponding samples are shown in Figure 4.15. Void growth is noticeable when the strain is around 23%, as seen in Figure 4.30. Void growth was characterized by void elongation along the tensile stress direction, resulting in elliptical shapes. Local void coalescence by void impingement or localized shearing is evident at strain around 56%, as shown in Figure 4.31. Widespread coalescence is seen at a strain of 60.5% (see Figure 4.32) and several cracks have formed by 65.5% strain (see Figure 4.33). When the strain reached about 75%, the e-notched brass specimens fracture.

4.3 Final Fracture

4.3.1 Fracture Surfaces - Armco Iron

The Armco iron fracture surfaces are representative of ductile fracture through void nucleation, growth, and coalescence. Failure initiates and propagates internally at inclusions or voids and ultimately fails by shear on 45° planes at the surface, that is, forming shear lips and a typical cup-cone fracture. Figure 4.34 includes SEM images of a failed c-notch specimen. Relatively large and deep cavities are observed on the fracture surface in Figure 4.34 (a) and (b). These large cavities correspond to the primary growing nuclei. Inclusions are readily located in these cavities, as shown in Figure 4.34 (b), (e) and (f), with an average size of roughly $10\ \mu\text{m}$. This indicates that the void are nucleated by decohesion of the interface between the inclusion and matrix. Detailed fracture profiles can be seen in Figure 4.34 (c) and (d). The average size of shallow cell, was equal to the grain size of Armco iron (approximately $30\ \mu\text{m}$), as shown in Figure 4.34 (c) and (d). The formation of shallow cells indicates that voids grow mainly in the lateral directions, causing them to remain shallow [149]. Tiny inclusions deposited on grain boundaries, as shown in Figure 4.34 (f), promoted grain boundary cracking.

Fractographs for an e-notched Armco iron specimens are shown in Figure 4.35. Void nucleation by decohesion of the interface between the inclusions and matrix can be observed in Figure 4.35 (b), (c), (d), (e), and (f). Deeper dimples around the inclusion, compared to the c-notch specimen, indicate a larger local degree of deformation around the inclusions and a tendency for greater void elongation, as expected for lower triaxiality levels. The inclusions on the grain boundaries made the grain boundaries weaker and easier to separate. This can be confirmed from Figure 4.35 (d) where the deep narrow dimples were on the grain boundaries.

4.3.2 Fracture Surfaces - Brass

The brass c-notched specimens failed under high strain-rates at a strain of around 55%. The fractographs for a c-notched specimen are shown in Figure 4.36. Although the pre-existing nuclei, the lead particles, were homogeneously distributed in the material, there were still some relatively larger cavities on the fracture surface as shown in Figure 4.36 (a) and (b). Careful observations of the bottom of these larger cavities were carried out and the results were given in Figure 4.36 (c) and (e). The larger cavities comprised several dimples and formed as a result of coalescence of local voids. Since the depth of each dimple on the bottom of large cavities was same, the micro-mechanism of coalescence can be deduced as the result of local necking of the matrix material. Shallow equiaxed dimples away from the larger voids, as shown in Figure 4.36 (d) and (f), indicated that the fracture occurred under high stress triaxiality.

Deep dimples were the main characteristic of the fracture profile of the e-notched brass specimens, since voids experienced large elongation. SEM observation provided good evidence of large deformation along the tensile stress direction, as shown in Figure 4.37. Unequal depths of neighboring voids in Figure 4.37 (c) indicated that void coalescence was accomplished through necking of matrix material between voids.

4.3.3 Fracture Surfaces - Ta and TaW

The tantalum and tantalum-tungsten alloy used in this research were extremely pure metals, virtually free of second phase particles. Plastic deformation under high strain-rates continued on conjugate slip planes until the specimens necked down to an arrow point with 100 per cent reduction of area. Such a failure is a geometric consequence of gross slip deformation. The fracture mechanism for the uniaxial specimens was studied under the SEM and is presented

in Figure 4.38 (a) and (b). Fracture is characterized as necking down to a point. The shear deformation process is reflected by the layered structure on the fracture surface as is evident in Figure 4.38 (c) and (d). Since the uneven shear deformation caused an irregularly shaped neck, the relationship between true strain (diametral strain) and engineering strain (axial strain) could not be accurately obtained through physical measurements of the interrupted specimens.

SEM observation of the fracture surfaces of c-notched tantalum specimens indicated that necking down to a chisel point was also the fracture mechanism under high triaxiality. Figure 4.39 (a) shows an overview of the fracture surface. A closer look of the chisel point is given in Figure 4.39 (b). The layered structure in Figure 4.39 (c) records the shear bands formed during plastic deformation. Tiny inclusions were observed, as shown in Figure 4.39 (d) and (e). EDX analysis indicated that these inclusions mainly consisted of S, Cl, Ca, Fe, Zn, P, and K, as illustrated in Figure 4.39 (f). From the nature of these inclusions, it was hard to determine whether they contributed to the fracture process.

The definitive characteristics observed in the uniaxial and c-notched tantalum fracture surfaces are also observed in the fracture profile of the e-notched specimens, that is, chisel point features and shear deformation bands, as depicted in Figure 4.40.

The tantalum-tungsten alloy used in this research was free of second phase particles. SEM observation of the fracture surface of a uniaxial tantalum-tungsten specimen revealed ductile fracture characteristics, as shown in Figure 4.41. Necking associated with large plastic deformation has been observed in Figure 4.41 (a) and (b). The flat feature in Figure 4.41 was due to rebound and impact of the two surfaces during testing.

SEM observations of the fracture surface of one c-notched sample are shown in Figure 4.42. Microvoids have been seen on the edge of the flat final fracture surface, as shown in Figure 4.42 (b). There were no obvious microcracks and plastic deformation in the form of shear deformation lead to final fracture. Three observations can be seen in Figure 4.42 (d), (e), and (f).

Cleavage was also observed on several c-notched tantalum-tungsten samples, as shown in Figure 4.43 (a) and (b). Since cleavage is associated with separation along crystallographic planes, its microscopic appearance is very flat. The presence of cleavage cracks indicates a brittle failure which was surprising. From the river patterns, the crack source seems to exist at the center of sample where several inclusions were observed, as illustrated in Figure 4.43 (e).

The fracture profiles of an e-notched tantalum-tungsten sample are shown in Figure 4.44 and were of same the nature as the ductile uniaxial samples. Figure 4.44 (a) gives a general overview of the fracture surfaces, while Figure 4.44 (b) provides a closer look at the final fracture. Some inclusions around the edge of the final flat fracture surface have been shown more closely in Figure 4.44 (c), (d), and (e). The composition of these inclusions has been analyzed using EDX and the results are given in Figure 4.44 (f), which indicates that the main components are Si, S, Cl, K, Ca, Zn, Fe, and Cu.

4.4 Measured Damage Rates

Measurements of damage within the interrupted specimens were obtained using the metallographic/image analysis techniques described in Chapter 3. These consisted of measured particle and void areal fraction taken at the center of each notched sample. Attention was focussed on the second-phase particle containing brass and Armco iron since the tantalum and tantalum-tungsten samples did not exhibit significant void damage.

Figures 4.45 and 4.46 show the increase in porosity with strain for the c-notch and e-notch Armco iron samples. The data from the smaller sample area of 0.3x0.22 mm corresponds to the single field at the center of each sample. The larger field corresponds to the 0.6x0.45 mm fields at the center and allows an estimate of scatter which is large. The very high porosity values at

larger strains are from samples interrupted after significant void coalescence resulting in large cracks. The smaller strain plots in each figure (Figure 4.45 (b) and Figure 4.46 (b)) encompass the “stable void growth” stage of ductile damage development. Comparing the data for the two samples, Figure 4.47 shows higher void growth rates in the c-notched samples due to higher levels of triaxiality.

Figure 4.48 shows the damage history data for the brass samples. Note that the “initial porosity” level is around 2.2% which corresponds to the lead particle volume fraction. As in the iron results, the higher triaxiality c-notch samples exhibit higher rates of damage. Interestingly, large regions of coalescence or cracking were not observed in the brass compared to the iron samples. The higher numbers of particles in the brass provides an easier path for crack propagation through void link-up once cracking initiates. Thus, large arrested cracks were not observed in interrupted brass test samples.

The spatial void distributions within the Armco iron c-notch and e-notch configurations are presented as contour plots in Figures 4.49 and 4.50, respectively. This image data was acquired such that the specimen gauge area was divided into smaller sub-areas of 0.3 mm X 0.22 mm. The average void fractions within these sub-area areas were calculated and the contours were obtained using the TechPlot graphics software. It is evident from Figures 4.49 and 4.50 that the higher void fraction is at the center of specimen gauge. Similar results are also observed for the brass specimens, as shown in Figure 4.51.

Chapter 5

Damage Model

Although the main focus of this thesis is on the experimental investigation of ductile fracture, the experimental results were also used to assess damage predictions produced by an available numerical model. The Gurson model [82-84] was used to describe the fracture process within the Armco iron and brass specimens. The numerical simulations were done with LS-DYNA, which is a lumped mass, explicit, two-dimensional, axi-symmetric finite element code used for analyzing large deformation, dynamic responses in inelastic solids. An existing "user subroutine" [119] was used to introduce the Gurson constitutive model into LS-DYNA. In this chapter, the meshes and the finite element material model are described briefly, after which the numerical simulation results for Armco iron and brass are presented.

5.1 Mesh and Boundary Conditions

Only the specimen and the ends of the Hopkinson bars were modeled using a so-called specimen-only mesh; that is, the incident and transmitter bars which surround the threaded ends of the specimen are included in the model. Quadrilateral elements were used for spatial discretization in this research. These meshes are shown in Figures 5.1 and 5.2 for the e-notched configuration and 5.3 and 5.4 for the c-notched configuration. Prior research by Pelletier [9] has indicated that similar results are achieved for both simulations considering the full length of the Hopkinson bars, or, as in the current research, specimen-only simulations.

Since all of the components of the TSHB considered in the model are axi-symmetric about their common axis, only one-half of their cross-section needed to be discretized. For both the e- and c-notched configurations, the specimens failed within the gauge regions. Thus, the meshes need to be refined locally in the gauge region to optimize accuracy and computing time. The details of the refined regions of the e- and c-notched configurations are seen in Figures 5.2 and 5.4, respectively. For the simulations, the following boundary conditions were adopted:

- 1) longitudinal axis was constrained to impede radial displacements;
- 2) the nodes along the bottom of the specimen-only mesh, as shown in the mesh plots, were constrained from axial displacement; and,

3) the nodes along the top of the specimen-only mesh were prescribed a velocity-time history consistent with the measured strain rates in the experiments.

The loading curve corresponding to this prescribed velocity boundary condition was determined based on the TSHB test results. The histories of strain-rate for the various specimens used in this research, as presented in Chapter 4, were idealized as a constant velocity boundary condition because the amplitude of the prescribed velocity (V_{spec}) is related to the nominal strain rate of the specimen by

$$V_{spec} = V_1 - V_2 = \dot{\epsilon}_s L_0 \quad (5.1)$$

where V_1 and V_2 are the velocities at the incident bar-specimen interface and the specimen-transmitter bar interface, respectively, $\dot{\epsilon}_s$ is the nominal strain rate in the specimen and L_0 is the initial gauge length of the specimen. Since the transmitter end of the specimen is constrained axially ($V_2 = 0$) in the model, equation 5.1 becomes

$$V_{spec} = V_1 \quad (5.2)$$

The time duration of the prescribed velocity is selected to be the same as the experimental duration of the incident pulse, such that:

$$t_l = \frac{2L_s}{C_0} \quad (5.3)$$

where L_s is the length of the striker and C_0 is the longitudinal sound velocity in the bar material.

5.2 Material Model

The incident and transmitter bars were made from a high-strength material, AISI 4340 steel. The bars were modeled as elastic-perfectly plastic with material properties corresponding to AISI 4340, that is, a density of 7.8869 g/cm³, with values for Young's modulus, Poisson's ratio and yield stress being 206.9 GPa, 0.3, and 689.5 MPa, respectively.

The constitutive response of the Armco iron was modeled using the Johnson-Cook [117, 118] equation (1.34) which incorporates the separate effects of strain hardening, strain rate hardening and thermal softening as described in Chapter 1. The parameters of A, B, m, and n were determined from the true stress-strain curves by Worswick and Pelletier [119]. In that

research, due to very limited range of strain rates in their high strain rate data, they adopted a value for parameter C equal to 0.06, given by Johnson and Cook [117]. Table 5.1 lists the various material parameters adopted for the Armco iron simulations. The nucleation parameters were determined by Pelletier and Worswick [119] to match the predicted fracture behaviour of Armco iron. In that work, they assumed an initial particle fraction of 0.5% and an average nucleation strain of 5%. No quantitative damage assessments have been performed on these parameters to-date.

The brass material data was also taken from Worswick and Pelletier [97]. Table 5.1 summarizes relevant parameters governing the yield and fracture behaviour of the brass. The material stress-strain behavior was described using a polynomial fit to quasi-static uniaxial tensile data by Worswick and Pick [97]. The flow stress obtained from this fit was then scaled using a log-strain-rate relationship:

$$\bar{\sigma} = (94.4 + 1594\bar{\epsilon}^p - 1693(\bar{\epsilon}^p)^2 + 645(\bar{\epsilon}^p)^3)(1 + 0.0009\ln(\dot{\epsilon})) \quad (5.4)$$

where $\bar{\sigma}$ is the material flow stress, $\bar{\epsilon}^p$ is the effective plastic strain, and $\dot{\epsilon}$ is the effective strain-rate. The 0.0009 constant prior to the logarithmic term represents the strain rate sensitivity of the material determined by Wong [150], which is recognized as being quite small. The nucleation parameters for the brass material were derived by Worswick and Pelletier [97] to match measured spall strengths from plate impact experiments. Of interest in the current work is the validation of these parameters based on measured porosity data.

5.3 Damage Predictions

The predicted damage development is presented as contours of porosity in Figures 5.5 and 5.6 for the Armco iron c-notched and e-notched samples, respectively. Figures 5.7 and 5.8 plot the respective porosities for the brass c-notched and e-notched samples. It is evident that larger porosity can be found at the center of the specimens, in general agreement with the experimental results. Hence, only the center void porosity was examined further in the following sections.

Figures 5.9 and 5.10 show the predicted and measured porosity levels for the Armco iron c-notched and e-notched specimens, respectively. In general, the predicted values lie well above the measured data. The reason for these discrepancies is attributed to the nucleation parameters for this material. It is evident from the figures that the predicted rate of nucleation is too high and that larger strains to nucleate damage should be considered. Consequently, simulations were performed with higher nucleation strain levels which demonstrated somewhat better agreement with the measured data. Comparison of the damage rate for the brass experiments and models are given in Figures 5.11 and 5.12. The predicted damage rate for this material are in much better agreement than for the iron. The nucleation parameters for the brass were determined by Worswick and Pelletier [119] based on plate impact experiments on this material. It is

encouraging to note that the predicted damage rates for the rather different experiments in the current research are in relatively good accord with the measured data.

Chapter 6

Discussion

The manner in which materials fail under high strain rate deformation will depend on a number of factors, such as, material purity, second phase distribution, second phase properties, and stress state, for example. In this chapter, the failure mechanisms of the different materials considered in this research will be discussed based on the microstructural and fractographic analysis. Also, the effect of stress state (i. e. triaxiality), on the fracture characteristics is discussed based on the experimental results.

6.1 Material Behavior Under High Strain Rate Deformation

The purity of a ductile material is the main factor determining its ductile failure mechanism. The tantalum used in this research was of high purity. The engineering stress versus strain curve displayed a large negative slope compared to the other materials used in this research. Since there were no inclusions large enough to be observed under the microscope, fracture by the nucleation and growth of voids was not a primary failure mode. Thus, the most common failure mechanism was shearing off through ductile rupture to form a chisel point.

When tungsten is alloyed with tantalum, it forms a substitutional solid solution and the stress versus strain curve of tantalum-tungsten shows a solution hardening effect in comparison to pure tantalum. Since tungsten increases the strength of the matrix without an increase in potential void nucleation sites, the fracture of tungsten-tantalum alloy under high strain rates was characterized as plastic deformation by shear. However, if there were small particles/impurities sitting on the grain boundary, they can serve as failure nucleation sites. Cleavage failure were observed in some c-notch cases, where the separation along crystallographic planes caused flat fractographic features. It was evident that, with the help of high triaxiality, initial cracks created around the inclusions in some of the c-notched tungsten-tantalum alloy specimens triggered the

cleavage fractures. Cleavage was not observed in any of the low triaxiality e-notch specimens and most of the c-notch specimens.

Brass, on the other hand, is a zinc and copper alloy. They form a two phase alloy system, that is, FCC alpha phase (α) and BCC beta phase (β). The interfaces of the alpha/alpha, beta/beta, and alpha/beta phases increase the potential for void nucleation under high triaxiality. Besides these interfaces, lead particles existed in the matrix as void nuclei. These factors determine that the fracture of the brass is through void nucleation, growth, and coalescence.

The Armco iron used in this research was a relatively pure metal of BCC structure with a certain amount of inclusions. Those inclusions during plastic deformation become void nucleation sites through splitting of the interfaces between inclusions and matrix. The failure mechanism of Armco iron is through void nucleation, growth, and coalescence. The void nucleation mechanism is dependent on the inclusion location, that is, within grain or on grain boundary.

6.2 Void Nucleation

The Armco iron and brass used in this research have second phase particles which cause material failure under high strain rate deformation through void nucleation, growth, and coalescence. The void nucleation mechanism at inclusions depended on the location of

inclusions. The micro-process of plastic deformation is through the motion of dislocations. Dislocations slipping against the inclusions and/or grain boundary results in dislocation pile-ups. Piled-up dislocations against the grain boundary caused linear defects in the grains of Armco iron as shown in Figure 6.1. Piled-up dislocations against the grain boundary cause large concentrated stresses on the interface of the inclusion and/or grain boundary resulting in that the grain boundaries split and form “island” voids with the size of the original grain. This nucleation mechanism is shown in Figures 6.2 and 6.3.

Void nucleation in the brass is mainly through interface decohesion at the matrix and lead particles. Meanwhile, nucleation also occurs at the interfaces of the alpha/alpha, beta/beta, and alpha/beta phases, as illustrated in Figure 6.4. Nucleation on the interfaces of alpha/alpha, beta/beta, and alpha/beta phases is through splitting of grain boundaries as shown in Figures 6.4 (b) and (c).

6.3 Void Growth and Coalescence

The void growth mechanism is either through local plastic deformation or local grain boundary decohesion depending on the void nucleation position. Such damage processes leading to ductile fracture are complex and exhibit a strong coupling with the particle morphology.

When the voids nucleate on the grain boundary, the growth is mainly through the process of grain boundary splitting since the grain boundary is weak (lower surface energy [2]) as shown in Figures 6.3 and 6.4. Hence, voids nucleate at inclusions within a grain, they can be considered as isolated.

Upon nucleation, voids grow while new nuclei form. During void growth, local zones of high stress concentration emanate from large voids and spread across the material, thereby raising the stresses at nearby microvoids. As a result, the hydrostatic stress surrounding one or more microvoids is raised to a level that activates an unstable deformation mode in which the stored elastic energy drives the plastic expansion of the microvoid. Although the overall stress decreases rapidly, small zones of high stress concentration are generated near growing voids, causing even smaller nearby microvoids to grow rapidly. This process continues until the submicron inter-void ligament fails either by microcleavage or by shearing along crystallographic planes [161].

When the nominal strain reaches a certain amount, that is, about 30 % for Armco iron and 40 % and 56 % for c-notched and e-notched brass samples respectively, coalescence happens locally through necking and shear. Bigger voids form and void volume increases dramatically, eventually resulting in material failure. The mechanisms of void coalescence can be categorized into three types: (i) local necking (see Figure 6.5 (a)); (ii) shear off (see Figure 6.5 (c)); and (iii) splitting void linkage with the aid of either small void formation (Figure 6.6 (b)) or a grain boundary (Figures 6.5 (c), 6.6 (e) and 6.7) functioning as a bridge. Plastic deformation between two voids causing elongated grain as shown in Figure 6.5 (a). A primary feature of the

coalescence phase is an unstable deformation mode whereby a minute, benign void rapidly enlarges, reaching a size set by the characteristic length of the locally elevated stress field [161].

The mechanisms of void coalescence depend upon the void shape, size, and location relative to the local stress state and neighbors. When neighboring voids are of the same size and are oriented parallel to each other, a local neck forms between two voids, as shown in Figures 6.5 (a), 6.6 (d), 6.6 (e), 6.6 (f), 6.7, and 6.8 (a). In this case, the matrix between the two voids was also of high triaxiality, in which the level of triaxiality is determined by the radii of the two voids. The coalescence of two voids with a larger spacing is usually through void nucleation and growth in the inter-void ligament. Such voids act as a bridge to connect the two primary voids. When the voids are arranged with an angular offset, void coalescence can result from the shearing off of the local matrix material. Once two voids coalesce, they are treated as a single larger void. Such big voids repeat the process of growth and coalescence. Propagation of damage can then proceed at three levels: (i) further coalescence of isolated voids (see Figure 6.7 (a)); (ii) coalescence of isolated voids with existing cracks (see Figure 6.6 (f)); and (iii) coalescence of two existing cracks (see Figure 6.9). All of these processes can be expected to occur concurrently with continued deformation.

6.4 The Effect of Triaxiality on Void Nucleation, Growth, and Coalescence

Triaxiality affects the mechanism of ductile fracture. The stress state to which a component is subjected influences the fracture appearance. A tensile test is uniaxial; however, when necking occurs the stress state becomes triaxial. This causes internal fissuring which leads to failure. A typical fracture process for a tensile test on a ductile material involves necking which in turn produces higher triaxiality at the center of neck and local void formation. These events indicate that triaxiality plays a very important role in the failure process. In the current work, the notched configurations increase the triaxiality in the center of the gauge region of the specimens. The effect of triaxiality on ductile fracture can then be easily studied from such sample configurations.

The triaxiality controls the shape of void growth and accelerates the rate of void growth and coalescence. The voids in the three kinds of specimens were elliptical in cross-section with axes parallel to the principal directions of the applied stress. The uniaxial specimens had the largest aspect ratio while the c-notched specimens had the smallest. In the c-notch specimens, the voids were more likely to be spherical due to the higher triaxiality at the center of the gauge region. These observations indicated that the nature of the void growth was dependent on the stress state, i.e. the dilational growth is the main characteristic of the c-notch specimen, while

both shape change and dilational growth are characteristic of the e-notched and uniaxial specimens. Comparing the depths of the dimples in the c-notch and e-notch specimens for all the materials considered here, the e-notch specimens were found to have deeper dimples than the c-notch specimens due to the lower stress triaxiality.

Comparing the failure strains of the c-notched, e-notched and uniaxial specimens, the c-notched specimens exhibited the lowest ductility while the uniaxial specimens failed at higher strains. This trend is due to higher damage rates in the higher triaxiality c-notch samples.

The effects of triaxiality on nucleation depended on the matrix materials. Any weak structures which can be potential nucleation sites are more likely to become nuclei as triaxiality increases. At higher triaxiality, the interfaces of alpha and beta phases in brass can become nucleation positions. However, if there were no or few void nuclei, as in the tantalum and tantalum-tungsten specimens, the material deforms to large strains and fails through ductile rupture.

The rate of damage is non-uniform at the microstructural scale. The contours of void volume fraction corresponding to different level of plastic deformation, as shown in Figures 4.49 to 4.51, provide evidence of the uneven distribution of void damage in the sectioned Armco iron and brass samples. This effect was strongest in the iron samples and even greater variation may be expected as the size of the image samples is reduced.

Chapter 7

Conclusions and Recommendations

7.1 Conclusions

The failure mechanisms of four metallic materials under high strain rate deformation have been investigated and can be categorized as either fracture by ductile rupture to a chisel point, cleavage, or ductile fracture through void nucleation, growth, and coalescence. Shear-off to a

chisel point occurred in tantalum and a tantalum-tungsten alloy of high purity. Cleavage fracture was produced in the tantalum-tungsten alloy under high triaxiality conditions, with the aid of very small inclusions. Void nucleation, growth and coalescence is the failure mechanism of both Armco iron and brass due to the extensive pre-existing void nucleation sites in these materials. Based on the discussion of the experimental results, the following conclusions can be made:

- The mechanisms of ductile failure under high strain rate deformation is mainly dependent upon the purity and microstructure of a metallic material, as well as the stress state.
- Elevated stress triaxiality accelerates void nucleation rates as reflected by lower strains to nucleate damage.
- When voids nucleate on the grain boundaries, grain boundary splitting becomes the micro-mechanism of void growth and coalescence.
- Use of the Gurson damage model to simulate ductile fracture of Armco iron under high strain rate conditions requires a 35% true strain as an average nucleation strain in order to match the experimented results.
- The Gurson model has captured the damage rates in the brass quite well. Extensive scatter exists in the iron damage rate data.

7.2 Recommendations

1. In order to accurately describe the process of void nucleation, growth, and coalescence, more samples are needed, especially in the stages of growth and coalescence for brass. Further research on void nucleation behaviour should be done for both the Armco iron and brass materials.
2. The damage percolation modelling approach of Worswick et al [105] should be applied to the current image data.

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Table 3.1 Armco iron TSHB test conditions and final deformations

Striker	Geometry	specimen # (Measured Strain %)[Velocities m/s]	Note
4 inches	C-Notch	AC12 (8.0) [7.3]; AC13 (10.7) [8.3]	A - Armco iron
	E-notch	AE12 (11.2) [9.5]	
5 inches	C-Notch	AC14 (9.9) [7.45]	C - C-Notch E - E-Notch
	E-notch	AE14 (13.6) [9.64]	
7 inches	C-Notch	AC15 (15.6) [7.43]	
	E-notch	AE15 (28.2) [9.3]; AE16 (20) [9.16]	
8 inches	C-Notch	AC10 (26.5) [7.9]; AC11 (26.8) [7.8]	
	E-notch	AE10 (28.2) [9.9]; AE11 (28.6) [9.8]	
9 inches	C-Notch	AC17 (23.4) [7.5]	
	E-notch	AE17 (39) [9.67]	
10 inches	C-Notch	AC6 (28.8) [7.3]; AC7 (19.4) [5.88]	U - Uniaxial
	E-notch	AE6 (36) [9.7]; AE7 (32.4) [9.2]	
	Uniaxial	AU3 (63.3) [18.7]	
11 inches	E-notch	AE18 (34.5) [9.34]	
12 inches	C-Notch	AC8 (41.0) [7.8]; AC9 (36.8) [7.2]	
	E-notch	AE8 (46.3) [9.8]; AE9 (48.1) [9.9]	
	Uniaxial	AU4 (58.5) [18.8]	
13 inches	C-Notch	AC18 (34) [7]	
14 inches	C-Notch	AC4 (66.1) [7.1] "F"; AC5 (45.4) [6.5]	
	E-notch	AE4 (54.6) [9.4]; AE5 (35.5) [9.22]	
16 inches	C-Notch	AC3 (71.6) [7.3] "F"; AC2 (30.7) [5.4]; AC1 (21.7) [4.9]	"F" designates a fractured specimen.
	E-notch	AE1 (98.1) [10.2] "F"; AE2 (57.9) [9.2]; AE3 (64.4) [9.8]; AE13 (66.4) [9.09]	
	Uniaxial	AU1 (108) [18.5] "F"	

Table 3.2 Brass TSHB test conditions and final deformations

Striker	Geometry	specimen # (Measured Strain %)[Velocities m/s]	Note
4 inches	C-Notch	BC14 (16.1) [9.5]; BC15 (16.3) [9.9]	B - Brass C - C-Notch E - E-Notch
	E-notch		
8 inches	C-Notch	BC12 (27.7) [9.8]; BC13 (29.9) [10.0]	
	E-notch	BE14 (24.9) [12.6]	
10 inches	C-Notch	BC8 (31.6) [9.5]; BC9 (32.7) [9.8]	
	E-notch	BE11 (33.5) [12.7]; BE12 (33.8) [12.6]	
	Uniaxial	BU4 (19.6) [18]	
12 inches	C-Notch	BC1 (31.5) [8.5]; BC2 (32.3) [8.3]; BC10 (40.6) [9.5]; BC11 (40.4) [10.0]; BC18 (36.5) [9.48]	
	E-notch	BE1 (34.6) [11.7]; BE2 (34.3) [11.7]; BE13 (42.5) [12.7]	
	Uniaxial	BU1 (27.2) [18.3]	
14 inches	C-Notch	BC6 (34.3) [8.9]; BC7 (38.5) [8.9]	
	E-notch	BE9 (51.4) [12.8]; BE10 (49.5) [12.9]	
16 inches	C-Notch	BC3 (37.0) [8.1]; BC4 (38.0) [7.4]; BC5 (59.2) [9.8] "F"	"F" designates a fractured specimen. U - Uniaxial
	E-notch	BE3 (48.5) [12.3]; BE4 (71.0) [16.0] "F"; BE5 (70.9) [14.2] "F"; BE7 (58.9) [12.7]; BE8 F (66.6) [12.9]	
	Uniaxial	BU2 (27.2) [17.4]	
18 inches	C-Notch		
	E-notch	BE15 (52.3) [11.8]; BE16 (60.2) [12.3]	

Table 3.3 Ta and TaW TSHB test processed data

<u>Specimen</u>	<u>Striker Length</u>	<u>Velocity</u>	<u>Fail or Not</u>	<u>Trapping? Y/N/R/SR</u>	<u>Comment</u>
TaC1	16	11.93	N	Y	
TaC2	18	11.84	N	Y	
TaC3	20	11.87	N	Y	
TaC4	20	14.01	Y	Y	
TaC5	20	16.88	Y	Y	
TaC6	20	19.29	Y	Y	
TaC7	16	14.15	N	Y	
TaC8	14	14.12	N	Y	
TaC9	12	13.98	N	Y	
WTaC1	16	14.18	N	Y	
WTaC2	18	14.13	Y	Y	
WTaC3	14	14.06	N	Y	
WTaC4	12	14.04	Y	Y	Intergranular Failure
WTaC5	12	13.73	N	Y	
TaE1	16	16.45	Y	Y	
TaE2	18	16.51	Y	Y	
TaE3	14	16.31	N	Y	
TaE4	15	16.01	N	Y	
TaE5	16	15.91	N	Y	Reflected Pulse Cut off
TaE6	16	15.91	N	Y	
WTaE1	20	13.78	Y	R	Lost Data files
WTaE2	18	16.09	Y	Y	
WTaE3	18	16.06	Y	Y	
WTaE4	14	16.31	N	Y	
WTaE5	15	15.98	N	Y	
WTaE6	16	15.79	N	Y	
TaU1	16	18.42	Y	N	Incident Pulse Cut off
TaU2	18	18.39	Y	Y	
TaU3	18	18.15	Y	Y	
TaU4	20	9.99	Y	SR	
TaU5	20	10.07	N	Y	
TaU6	20	10.05	N	Y	
WTaU1	18	18.47	Y	Y	
WTaU2	18	18.44	Y	Y	
WTaU3	20	10.12	N	Y	
WTaU4	20	10.04	N	Y	
WTaU5	20	10.07	N	Y	
Note: "R" is Reload and "SR" is Slight Reload					

Table 4.1 The regression results of Armco iron and brass using equation 4.1 *.

Material	Specimen Configuration	log (A)	B	R ² **
Armco iron	C-notch	0.2199	0.9662	0.9712
Armco iron	E-notch	0.2403	1.258	0.9818
Brass	C-notch	0.1028	0.8966	0.9323
Brass	E-notch	0.1145	1.298	0.9612

Note:

* The equation 4.1 can be rewritten as

$$\log(\varepsilon_T) = \log(A) + B \log(\varepsilon_E)$$

* R^2 R-squared value is the proportion of the variance in Y attributable to the variance in X and can be expressed as

$$R^2 = \frac{\left[n(\sum XY) - (\sum X)(\sum Y) \right]^2}{\left[n\sum X^2 - (\sum X)^2 \right] \left[n\sum Y^2 - (\sum Y)^2 \right]}$$

Table 5.1 Parameters used to model the Armco iron and brass materials

	Armco Iron	Brass
K (Pa)	174	100.4
G (GPa)	80.4	40.1
ρ (g/cm ³)	7.89	8.54
q_1	1.25	1.25
q_2	0.95	0.95
q_3	q_1^2	q_1^2
f_N	0.005	0.025
ε_N	0.05/0.15/0.25/0.35/0.45	N/A
S_N	0.025	N/A
f_c	0.05	0.15
f_u	0.06	0.25

BROAD CLASSES OF FRACTURE MECHANISM

Figure 1.1 Classification of fracture mechanisms [12].

Figure 1.2 Various dislocation models for cleavage crack nucleation and growth [14].

Figure 1.3 (a) Formation of a grain boundary ledge due to slip. (b) The opening of a cavity due to the combined action of the bulk deformation and grain boundary sliding. (c) Additional cavity formation by the repetition of the process. (d) Fractographic evidence of the type of cavity formation [30].

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Figure 1.4 The Griffith criterion for an infinite cracked plate of unit thickness with a central transverse crack of length $2a$.

Figure 1.5 Void nucleation after Ashby [55].

Figure 1.6 A pattern of cylindrical holes visualized by F. A. McClintock [59].

Figure 1.7 An isolated spherical void within a matrix which is subjected to uniaxial tension [60].

Figure 1.8 Void growth at (a) low, (b) moderate, and (c) high triaxialities, as described by Sun Jun [73].

Figure 1.9 The two-dimensional plane-strain plasticity model for nucleation-controlled ductile fracture showing: (a) strongly bonded particles; (b) particle-sized voids at the nucleation strain; (c) the local slip-line field at incipient limite-load failure; (d) the complete void-coalescence process by internal necking of the intervold matrix [44].

Figure 1.10 A two-dimensional plane-strain plastic model of a unit cell, with a central circular void. at the point of incipient limit-load failure of the intervoid matrix, with frictionless constraints on the sides and the plastic limit-load stress on the ends [44].

Figure 1.11 The two-dimensional plane-strain plasticity model for ductile fracture by nucleation, growth and coalescence of microvoids showing: (a) strongly bonded particles; (b) particle-sized voids at the void nucleation strain; (c) the extensional- and dilational-plastic growth of voids; (d) the incipient limit-load failure of the intervoid matrix when critical stress σ_{lc} has reduced to the σ_1 and internal necking can begin; (e) internal necking; (f) the completion of the internal necking process to give 'knife-edge' separation of the fracture surfaces [44].

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Figure 1.12 Long Circular cylindrical void in a matrix of rigid-perfectly plastic von Mises material [82].

Figure 1.13 Spherical void model, fully plastic flow [82].

Figure 1.14 Yield surface dependence on the hydrostatic tension for various values of the void porosity [88].

Figure 2.1 Yield phenomenon in Armco iron as a function of grain size [123].

Figure 2.2 The distribution and composites of inclusions in Armco iron as received. a) string of inclusion along rolling direction. b) magnified inclusion for EDX analysis, c) EDX analysis of the inclusion in b).

Figure 2.3 Particle distribution in Armco iron after heat treated at 550 °C 2 hrs.

Figure 2.4 The microstructure of Armco iron after heat treated at 550 °C 2 hrs and slow cooling. The grain size is about 30 μm .

Figure 2.5 The distribution of lead particles in brass after heat treated at 850 °C 2 hrs.

Figure 3.1 Schematic view of the TSHB configuration [140].

① Strain gauges	⑦ Gas gun
② Specimen	⑧ Striker
③ Incident bar endcap	⑨ Transmitter bar screwed-on endcap
④ Incident bar momentum trap brake	⑩ Transmitter bar momentum trap brake
⑤ Transmitter bar	⑪ Transmitter bar momentum trap cylinder
⑥ Incident bar	⑫ Incident bar momentum trap rod

Figure 3.2 Schematic view of the novel technique used in TSHB [9].

Figure 3.3 Uniaxial specimen configuration (units used in inches).

Figure 3.4 C-notched specimen configuration (units used in inches).

Figure 3.5 E-notched specimen configuration (units used in inches).

a)

b)

Figure 4.1 The TSHB test results showing (a) the history of strain rate and (b) engineering stress versus engineering strain for Armco iron.

a)

b)

Figure 4.2 The TSHB test results showing (a) the history of strain rate and (b) engineering stress versus engineering strain for leaded brass.

a)

b)

Figure 4.3 The TSHB test results showing (a) the history of strain rate and (b) engineering stress versus engineering strain for tantalum.

a)

b)

Figure 4.4 The TSHB test results showing (a) the history of strain rate and (b) engineering stress versus engineering strain for tantalum-tungsten.

Figure 4.5 The engineering stress-strain relations of uniaxial specimens for different materials under TSHB testing.

Figure 4.6 The engineering stress-strain relations of the e-notched specimens for different materials under TSHB testing.

Figure 4.7 The engineering stress-strain relations of the c-notched specimens for different materials under TSHB testing.

Figure 4.8 True strain versus engineering strain relation for c-notched Armco iron.

Figure 4.9 The true stress-strain relationship of c-notched Armco iron failed sample and interrupted samples with different levels of deformation, labeled 1, 2, 3, and 4 which represent 10%, 31.5%, 33.5%, and 37% strain, respectively.

Figure 4.10 True strain versus engineering strain relation for e-notched Armco iron.

Figure 4.11 The true stress-strain relationship of e-notched Armco iron failed samples and interrupted samples with different level of deformation, labeled 1, 2, 3, 4 which represent 11%, 27%, 29%, 40%, and 64% strain, respectively.

Figure 4.12 True strain versus engineering strain relationship for c-notched brass.

Figure 4.13 The true stress-strain relationship of c-notched brass failed sample and interrupted samples with different level of deformation, labeled 1, 2, 3 which represent 13%, 28%, and 40% strain, respectively.

Figure 4.14 True strain versus engineering strain relationship for e-notched brass.

Figure 4.15 The true stress versus true strain relationship of e-notched brass failed sample and interrupted samples with different level of deformation labeled 1, 2, 3, and 4 which represent 23%, 56%, 60.5%, and 65.5% strain, respectively.

Figure 4.16 Light microscope image (scale 9 $\mu\text{m}/\text{mm}$) showing the distribution of inclusions in the Armco iron after annealing at 550 °C for 2 hrs (as-polished).

Figure 4.17 Light microscope image (scale 18 $\mu\text{m}/\text{mm}$) of an as-polished Armco iron specimen showing the damage evolution in the c-notched region at a true strain of 10 %. Tensile direction is vertical.

Figure 4.18 Light microscope image (scale 18 $\mu\text{m}/\text{mm}$) of an as-polished Armco iron specimen showing the damage evolution in the c-notched region at a true strain of 32 %. Tensile direction is vertical.

Figure 4.19 Light microscope image (scale 18 $\mu\text{m}/\text{mm}$) of an as-polished Armco iron specimen showing the damage evolution in the c-notched region at a true strain of 34%. Tensile direction is vertical.

Figure 4.20 Light microscope image (scale $18\text{ }\mu\text{m/mm}$) of an as-polished Armco iron specimen showing the damage evolution in the c-notched region at a true strain of 37%. Tensile direction is vertical.

Figure 4.21 Light microscope image (scale 18 $\mu\text{m}/\text{mm}$) of an as-polished Armco iron specimen showing the damage evolution in the c-notched region at a true strain of 11%. Tensile direction is vertical.

Figure 4.22 Light microscope image (scale $18\text{ }\mu\text{m/mm}$) of an as-polished Armco iron specimen showing the damage evolution in the e-notched region at a true strain of 27 %. Tensile direction is vertical.

Figure 4.23 Light microscope image (scale 18 $\mu\text{m}/\text{mm}$) of an as-polished Armco iron specimen showing the damage evolution in the c-notched region at a true strain of 29%. Tensile direction is vertical.

Figure 4.24 Light microscope image (scale 18 $\mu\text{m}/\text{mm}$) of an as-polished Armco iron specimen showing the damage evolution in the e-notched region at a true strain of 40%. Tensile direction is vertical.

Figure 4.25 Light microscope image (scale $18\text{ }\mu\text{m/mm}$) of an as-polished Armco iron specimen showing the damage evolution in the e-notched region at a true strain 64%. Tensile direction is vertical.

Figure 4.26 Light microscope image (scale 9 $\mu\text{m}/\text{mm}$) of lead particle distribution in brass after being annealed at 850 $^{\circ}\text{C}$ for 2 hrs (as-polished).

Figure 4.27 (a) Partial (left half) light microscope image (scale $9\text{ }\mu\text{m/mm}$) of c-notched region of as-polished brass specimen at a true strain of 13%. Tensile direction is vertical.

Figure 4.27 (b) Partial (right half) light microscope image (scale $9\text{ }\mu\text{m/mm}$) of c-notched region of as-polished brass specimen at a true strain of 13%. Tensile direction is vertical.

Figure 4.28 (a) Partial (left half) light microscope image (scale $9\text{ }\mu\text{m/mm}$) of c-notched region of as-polished brass specimen at a true strain of 28%. Tensile direction is vertical.

Figure 4.28 (b) Partial (right half) light microscope image (scale 9 $\mu\text{m}/\text{mm}$) of c-notched region of as-polished brass specimen at a true strain of 28% under TSHB test for 80 μs . Tensile direction is vertical.

Figure 4.29 (a) Partial (left half) light microscope image (scale $9\text{ }\mu\text{m/mm}$) of c-notched region of as-polished brass specimen at a true strain of 39.5%. Tensile direction is vertical.

Figure 4.29 (b) Partial (right half) light microscope image (scale $9\text{ }\mu\text{m/mm}$) of c-notched brass specimen at a true strain of 39.5%. Tensile direction is vertical.

Figure 4.30 (a) Partial (left half) light microscope image (scale $9\text{ }\mu\text{m/mm}$) of e-notched brass specimen at a true strain of 23%. Tensile direction is vertical.

Figure 4.30 (b) Partial (right half) light microscope image (scale $9\text{ }\mu\text{m/mm}$) of etched region of as-polished brass specimen at a true strain of 23%. Tensile direction is vertical.

Figure 4.31 (a) Partial (left half) light microscope image (scale $9\text{ }\mu\text{m/mm}$) of e-notched region of as-polished brass specimen at a true strain of 56%. Tensile direction is vertical.

Figure 4.31 (b) Partial (right half) light microscope image (scale $9\text{ }\mu\text{m/mm}$) of e-notched region of as-polished brass specimen at a true strain of 56%. Tensile direction is vertical.

Figure 4.32 (a) Partial (left half) light microscope image (scale $9\text{ }\mu\text{m/mm}$) of e-notched brass specimen at a true strain of 60.5%. Tensile direction is vertical.

Figure 4.32 (b) Partial (right half) light microscope image (scale $9\text{ }\mu\text{m/mm}$) of c-notched region of as-polished brass specimen at a true strain of 60.5%. Tensile direction is vertical.

Figure 4.33 (a) Partial (left half) light microscope image (scale $9\text{ }\mu\text{m/mm}$) of e-notched region of as-polished brass specimen at a true strain of 65.5%. Tensile direction is vertical.

Figure 4.33 (b) Partial (right half) light microscope image (scale 9 $\mu\text{m}/\text{mm}$) of e-notched region of as-polished brass specimen at a true strain of 65.5%. Tensile direction is vertical.

Figure 4.34 SEM observation of the fracture surface of c-notched Armco iron specimens showing: (a) overview of fracture surface from which large cavities can be observed at about one third of radii of notched range from center; (b) local magnified image of center in (a) where void nucleation at inclusions can be observed; (c) high magnification of center area showing smaller voids nucleated at grain boundary through grain boundaries splitting, associated with second phase particles on the grain boundary; (d) SEM image at 2/3 radii from center illustrated that the void nucleated on grain boundary; (e) nucleation at second phase particle, the result of EDX analysis on the particle indicated that the composition of the particles are Al, Si, and Mn; (f) nucleation at inclusion and on grain boundary.

Figure 4.35 SME observation of the fracture surface of e-notched Armco iron specimen showing: (a) the profile of broken surface; (b) void nucleation on grain boundary with inclusions on the grain boundary; (c) the center area in (a); (d) higher magnification of (c); (e) local image at one thirds of notch radii from center; (f) local image at two third of notch radii from center. the result of EDX analysis indicated that the composition of particles are Al, Si, and Mn.

Figure 4.36 SEM observation of the fracture surface of c-notched specimen of brass showing: (a) profile of broken surface; (b) voids coalescing with their neighboring voids to form larger cavities; (c) large plastic deformation around the voids associated with grain boundary splitting; (d), (e), and (f), typical ductile dimpled fracture.

Figure 4.37 SEM observation of e-notched brass specimen showing: (a) profile of broken surface; (b) dimples and cavities at the center corresponding to the nuclei (lead particles); (c) and (d) growth and coalescence.

Figure 4.38 Fractographs of the tantalum uniaxial specimen illustrating the typical ductile rupture fracture with necking to a point in (a) and (b) with large magnification and shear bands in (c) and (d).

Figure 4.39 SEM observation of the fracture surface of c-notched tantalum specimen showing: (a) overview of fracture surface; (b) enlarged center area of (a); (c) shear deformation resulting in a layered structure; (d) and (e) illustrating the inclusion; and (f) EDX analysis on the inclusion.

Figure 4.40 SEM observation of the fracture surface of e-notched tantalum specimen showing: (a) and (b) ductile rupture fracture with necking to a point; (c) river track due to different orientation of shear deformation; and (d) shear bands associated with micro-voids and micro-cracks.

a)

b)

Figure 4.41 SEM observation of the fracture surface of uniaxial tantalum tungsten showing an overview fracture surface (a) and (b).

Figure 4.42 Fractographs of tungsten tantalum alloy c-notched specimen showing: (a) necking; (b) micro-voids; (c) flat surface on the final broken surface; and (d) to (f) profile of plastic deformation before final fracture.

Figure 4.43 Cleavage fractographs of c-notched TaW specimen under SEM: (a) originating from inclusions: (b) and (c) by tracing quasi-cleavage source: (d) of river pattern (e) and (f).

Figure 4.44 SEM observation and EDX analysis of the fracture surface of tungsten-tantalum alloy e-notched specimen. (a) and (b) showing ductile rupture with different magnifications, (c), (d), and (e) showing final broken surface, and (f) illustrating the composition of inclusions in (c), (d), and (e) using EDX analysis.

a)

b)

Figure 4.45 Measured void volume fraction as a function of true strain for the interrupted c-notched Armco iron specimens.

a)

b)

Figure 4.46 Measured void volume fraction as a function of true strain for the interrupted e-notched Armco iron specimens.

Figure 4.47 Comparison of damage rates in the Armco iron c- and e- notch specimens.

a)

b)

Figure 4.48 Measured void volume fraction as a function of true strain for the brass c-notched (a) and e-notched (b) specimens.

Figure 4.49 Void fraction contours of c-notched Armco iron specimens at (a) 27%, and (b) 41% true strain. Vertical mid-points are tangent to the notch. Tensile direction is vertical. Units used in this figure are mm.

Figure 4.50 Void fraction contour of e-notched Armco iron specimen at 48% true strain. Vertical mid-points are tangent to the notch. Tensile direction is vertical. Units used in this figure are mm.

Figure 4.51 Void fraction contours for (a) c-notched brass specimen at 36% true strain and (b) e-notch at 65% true strain. Vertical mid-points are tangent to the notch. Tensile direction is vertical. Units used in this figure are mm.

Figure 5.1 Mesh of c-notched specimen.

Figure 5.2 The partial mesh of c-notched specimen as show Figure 5.1 present the gauge area.

Figure 5.3 Mesh of e-notched specimen

Figure 5.4 The partial mesh of e-notched specimen as show Figure 5.2 present the gauge area.

a)

b)

c-notch specimen only - s.r. = 1000/s
time = 7.9999E+01 fringes of history variable # 1
dsf = 0.10000E+01

minval = 0.000E+00
maxval = 5.931E-02
fringe levels

c)

d)

Figure 5.5 Contour of simulation results of the Armco iron with c-notch configuration at (a) 20 μ s, (b) 46 μ s, (c) 80 μ s, and (d) 90 μ s.

a)

```

minval= 0.000E+00
maxval= 3.238E-02
fringe levels

```

c)

d)

Figure 5.6 Contours of simulation result of Armco iron with e-notch configuration at (a) 60 μs, (b) 68 μs, (c) 72 μs, and (d) 74 μs.

c-notch specimen only - s.r. = 1000/s
time = 2.0000E+01 fringes of history variable # 1
dsf = 0.10000E+01

minval = 0.000E+00
maxval = 9.315E-03
fringe levels

a)

c-notch specimen only - s.r. = 1000/s
time = 4.0000E+01 fringes of history variable # 1
dsf = 0.10000E+01

minval = 0.000E+00
maxval = 2.659E-02
fringe levels

b)

c-notch specimen only - s.r. = 1000/s
time = 5.9997E+01 fringes of history variable # 1
dsf = 0.10000E+01

minval = 0.000E+00
maxval = 3.446E-02
fringe levels

c)

Figure 5.7 Contour of simulation results of the brass with c-notch configuration at (a) 20 μ s, (b) 40 μ s, (c) 60 μ s, and (d) 80 μ s.

a)

b)

```

minval= 0.000E+00
maxval= 5.539E-02
fringe levels

```

c)

d)

Figure 5.8 Contours of simulation result of the brass with e-notch configuration at (a) 60 μs, (b) 120 μs, (c) 140 μs, and (d) 158 μs.

Figure 5.9 Comparison of predicted porosity using 5%, 15%, 25%, 35%, and 45% nucleation strain, and measured porosity for the c-notched Armco iron specimens.

Figure 5.10 Comparison of predicted porosity using 5%, 15%, 25%, 35%, and 45% nucleation strain, and measured porosity for the e-notched Armco iron specimens.

Figure 5.11 Comparison of predicted and measured porosity for the c-notched brass specimens.

Figure 5.12 Comparison of predicted and measured porosity for the e-notched brass specimens.

Figure 6.1 Defects formed at grain boundary triple points in Armco iron (see arrow).

Figure 6.2 Light microscopy observation on Armco iron showing the void nucleation and growth at grain boundaries (arrow a) and larger sized inclusions (arrow b) (a), (b), and (c) in c-notched samples: (d), (e), and (f) in e-notched samples.

Figure 6.3 Void nucleated along grain boundary with the aid of inclusions on the grain boundary (see arrow).

Figure 6.3 Void nucleated along grain boundary with the aid of inclusions on the grain boundary (see arrow).

Figure 6.4 Void nucleation in brass showing (a) lead particle distribution without deformation, (b) and (c) void nucleation by either interface decohesion at inclusion (arrow a) or grain boundary split (arrow b) in c-notch and e-notched specimens, respectively. Tensile deformation was along vertical direction.

Figure 6.5 Three types of micro-processes of void coalescence in Armco iron, that is, (a) necking in matrix material between voids; (b) grain boundary split, and (c) shear deformation between the two neighbor voids.

Figure 6.6 Void coalescence in the brass of (a), (c), and (e) c-notched configuration, and (b), (d), and (f) e-notched configuration.

Figure 6.7 Void coalescence through grain boundary split (see arrow) in the e-notched brass specimen. Tensile direction is horizontal.

Figure 6.8 Void coalescence in c-notched Armco iron with tensile deformation in vertical direction.

Figure 6.9 Typical void coalescence and crack development in c-notched samples. No elongated grains have been observed.