INFORMATION TO USERS

This manuscript has been reproduced from the microfilm master. UMI films the text directly from the original or copy submitted. Thus, some thesis and dissertation copies are in typewriter face, while others may be from any type of computer printer.

The quality of this reproduction is dependent upon the quality of the copy submitted. Broken or indistinct print, colored or poor quality illustrations and photographs, print bleedthrough, substandard margins, and improper alignment can adversely affect reproduction.

In the unlikely event that the author did not send UMI a complete manuscript and there are missing pages, these will be noted. Also, if unauthorized copyright material had to be removed, a note will indicate the deletion.

Oversize materials (e.g., maps, drawings, charts) are reproduced by sectioning the original, beginning at the upper left-hand corner and continuing from left to right in equal sections with small overlaps. Each original is also photographed in one exposure and is included in reduced form at the back of the book.

Photographs included in the original manuscript have been reproduced xerographically in this copy. Higher quality 6" x 9" black and white photographic prints are available for any photographs or illustrations appearing in this copy for an additional charge. Contact UMI directly to order.

UMI

A Bell & Howell Information Company
300 North Zeeb Road, Ann Arbor MI 48106-1346 USA
313/761-4700 800/521-0600

Damage Evolution in Ductile Materials at High Rates of Tensile Strain

by

John Paul Fowler
B.Eng. (Aerospace Engineering)

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

Master of Engineering
in Mechanical Engineering

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

Carleton University
Ottawa, Ontario
May 17, 1996

© Copyright
1996, John Fowler

National Library
of Canada

Acquisitions and
Bibliographic Services

395 Wellington Street
Ottawa ON K1A 0N4
Canada

Bibliothèque nationale
du Canada

Acquisitions et
services bibliographiques

395, rue Wellington
Ottawa ON K1A 0N4
Canada

Your file Votre référence

Our file Notre référence

The author has granted a non-exclusive licence allowing the National Library of Canada to reproduce, loan, distribute or sell copies of this thesis in microform, paper or electronic formats.

The author retains ownership of the copyright in this thesis. Neither the thesis nor substantial extracts from it may be printed or otherwise reproduced without the author's permission.

L'auteur a accordé une licence non exclusive permettant à la Bibliothèque nationale du Canada de reproduire, prêter, distribuer ou vendre des copies de cette thèse sous la forme de microfiche/film, de reproduction sur papier ou sur format électronique.

L'auteur conserve la propriété du droit d'auteur qui protège cette thèse. Ni la thèse ni des extraits substantiels de celle-ci ne doivent être imprimés ou autrement reproduits sans son autorisation.

0-612-22111-3

Carleton University
Ottawa, Canada K1S 5J7

1942-1992

Thesis contains black and white and/or
coloured graphs/tables/photographs which when
microfilmed may lose their significance. The
hardcopy of the thesis is available upon
request from Carleton University Library.

The undersigned hereby recommend to
the Faculty of Graduate Studies and Research
acceptance of the thesis

**Damage Evolution in Ductile Materials
at High Rates of Tensile Strain**

submitted by

John Paul Fowler
B.Eng. (Aerospace Engineering)

in partial fulfilment of the requirements
for the degree of Master of Engineering
in Mechanical Engineering

Thesis Supervisor

Chair, Department of Mechanical and Aerospace Engineering

Carleton University
May 17, 1996

Abstract

The study of ductile fracture at high rates of strain is important for the analysis of many impact events and metal forming operations. Ductile fracture occurs through the nucleation, growth and eventual coalescence of voids. Previously, the evolution of void damage with deformation has been determined at quasi-static loading rates. Until recently, no method has existed to perform similar tests at higher strain rates.

For this research, leaded brass was used as a model material to investigate high strain rate damage evolution. A novel momentum trapping technique, recently applied to Carleton's tensile split Hopkinson bar, allowed tests of constant strain rate to be interrupted at various levels of strain. This modified Hopkinson bar was used to test uniaxial and notched specimens. Image processing techniques were applied with quantitative stereology methods to evaluate the resulting damage in the test specimens. In this manner the damage was determined as a function of applied strain and stress-state at strain rates in excess of 10^3 s^{-1} . Plate impact tests were also performed at the Ernst Mach Institute in Freiburg, Germany. Plates which did not spall were quantitatively investigated to evaluate the distribution of damage over their cross-section.

It is concluded that interrupted split Hopkinson bar testing provides a viable means of evaluating damage evolution in ductile materials in the 10^3 to 10^4 s^{-1} strain rate regime. The damage study indicated that: (i) the rate of void evolution increases with triaxiality; (ii) the evolution of void aspect ratio is inversely proportional to triaxiality; and (iii) void nucleation strain is a function of triaxiality. Plate impact tests near spall exhibited damage levels similar to those of the Hopkinson bar tests, supporting a critical void volume fraction criterion for void coalescence.

ACKNOWLEDGEMENTS

I would like to express sincere thanks to my advisor, Dr. M.J. Worswick, for his patience, support and encouragement.

I also am indebted to Dr. H. Nahme and the Ernst Mach Institute in Freiburg, Germany, for performing the experimental plate impact testing presented here.

The technical assistance of Bob Ellis, Fred Barret, Ian Jefferson and Bruce Johnston of the Mechanical and Aerospace Engineering Department as well as John Batky and the ETC Mechanical machinists was invaluable.

Special thanks to Keith Pilkey, colleague, mentor and friend for his help on metallography and materials.

I sincerely thank Pierre Pelletier for letting me get in his way and teaching me the finer side of Carleton's tensile split Hopkinson bar.

My gratitude extends to Christie Egbert for her excellent typing and formatting skills.

Finally, I thank my father Al, mother Sheila and sister Catherine for their love and understanding.

TABLE OF CONTENTS

1	INTRODUCTION	1
1.1	Ductile Fracture	3
1.1.1	Nucleation	4
1.1.1	Growth	9
1.1.3	Coalescence	13
1.2	High Strain Rate Effects on Ductile Fracture	19
1.2.1	Inertial Effects	19
1.2.2	Rate Dependant Material Properties	22
1.2.3	Adiabatic Heating	24
1.3	Current Research	24
2	IMAGE ANALYSIS TECHNIQUES AND QUANTITATIVE MICROSCOPY	27
2.1	Image Acquisition of Metallographic Fields	28
2.1.1	Imaging Equipment	28
2.1.2	Pre-Processing of Image Data	30
2.2	Image Segmentation	32
2.3	Image Processing Methods for Feature Extraction	34

2.3.1	Characterizing Particle Size	34
2.3.2	Evaluation of Particle Shape and Orientation	35
2.3.3	Stereological Relations	36
2.3.4	Sources of Error and Bias in Field Measurements	37
3	MODEL MATERIAL	40
3.1	Heat Treatment	41
3.2	Specimen Preparation	42
3.3	Stock Material Heat Treatment and Particle Populations	43
3.4	Mechanical Properties of Annealed Brass	46
4	TENSILE SPLIT HOPKINSON BAR EXPERIMENTS	47
4.1	Historical Perspectives	48
4.2	Theoretical Basis of the Tensile Hopkinson Bar	48
4.3	Interrupted Tests for the Characterization of Damage Evolution . . .	57
4.4	TSHB Specimen Selection	59
4.5	Carleton's TSHB Apparatus	61
4.6	Firing Control System	63
4.7	Hopkinson Bar Instrumentation	64
4.8	Calibration of Elastic Wave Measurement in the Bars	66
4.8.1	Relating Strain Rate to Striker Velocity	66
4.8.2	Selection of Striker Velocities	68

4.9	Data Reduction Methods	69
4.9.1	Conversion of Engineering Strain to True Strain	69
5	PLATE IMPACT	73
5.1	Theory of Plate Impact Experiments	73
5.2	Apparatus	81
5.3	Spallation of Plates	83
6	EXPERIMENTAL RESULTS	87
6.1	Hopkinson Bar Data	87
6.1.1	Mechanical Properties of Annealed Brass at High Strain Rates	87
6.1.2	Damage Distribution Within Hopkinson Bar Specimens . . .	94
6.1.3	Damage Evolution Within the Hopkinson Bar Specimens . .	97
6.1.4	Fractography of Hopkinson Bar Specimens	101
6.2	Flyer Plate Data	102
6.2.1	Free Surface Velocity Profiles	102
6.2.2	Image Analysis Data	104
6.2.3	Fractography of Plate Specimens	105
7	DISCUSSION, CONCLUSIONS, AND RECOMMENDATIONS	106
7.1	Discussion of Experimental Results	106

7.1.1	Mechanical Properties of Brass	106
7.1.2	Distribution of Damage	107
7.1.3	Evolution of Damage Parameters with Strain	108
7.1.4	Damage at Fracture	109
7.1.5	Applications of Reported Data	109
7.2	Conclusions	110
7.3	Recommendations	111

LIST OF TABLES

Table 3.1	Nominal composition of grains and phase at grain boundaries
Table 4.1	Pulse durations for available striker lengths
Table 4.2	Specimen triaxiality
Table 6.1	C-notch specimen data for UNS-C36000 brass (Triaxiality = 1.026)
Table 6.2	E-notch specimen data for UNS-C36000 brass (Triaxiality = 0.612)
Table 6.3	E-notch specimen data for MS-58 brass (Triaxiality = 0.612)
Table 6.4	Uniaxial tensile specimen data for UNS C36000 brass (Triaxiality = 0.333)

LIST OF FIGURES

- Figure 1.1 The Processes of ductile fracture
- Figure 1.2 Schematic of void nucleation front progressing through the particle size distribution (Embury [21])
- Figure 1.3 Prolate void aligned with the direction of tensile strain
- Figure 1.4 The effect of tensile deformation on porosity, for iron compacts of various initial porosities (Spitzig et al [46])
- Figure 1.5 The evolution of porosity, normalized by initial porosity, for several iron compacts (Spitzig et al. [46])
- Figure 1.6 Geometric representation of Thomason's model for coalescence (Thomason [52])
- Figure 1.7 Void growth at (a) low, (b) moderate, and (c) high triaxialities, as described by Sun Jun [54]
- Figure 1.8 Oriented imperfection plane used to model localization (Yamamoto [65])
- Figure 1.9 Eulerian simulation of stress waves from the top of the page propagating through a copper matrix, resulting in void growth (Benson [79])

Figure 1.10	Eulerian simulation of stress waves from the top of the page propagating through a 4340 steel matrix, resulting in void growth (Benson [79])
Figure 2.1	Steps of quantitative Metallography
Figure 2.2	Bimodal grey-scale histogram
Figure 3.1	Cu-Zn phase diagram
Figure 3.2	Time-temperature-transformation (TTT) diagram for MS-58 brass
Figure 3.3	600°C HT 8.89 cm (3 1/2") brass rod (as polished, 100x)
Figure 3.4	850°C HT 8.89 cm (3 1/2") brass rod (as polished, 100x)
Figure 3.5	600°C HT 8.89 cm (3 1/2") brass rod (as etched, 200x)
Figure 3.6	850°C HT 8.89 cm (3 1/2") brass rod (as etched, 200x)
Figure 3.7	850°C HT 1.27 cm (1/2") brass rod (as etched, 100x)
Figure 3.8	Particle area histogram for 850°C HT brass
Figure 3.9	Particle aspect ratio histogram for 850°C HT brass
Figure 3.10	Particle area histogram for 600°C HT brass
Figure 3.11	Particle aspect ratio histogram for 600°C HT brass
Figure 3.12	Quasi-static engineering stress-strain curves
Figure 3.13	Dynamic engineering stress-strain response
Figure 4.1	Schematic of the Carleton tensile split Hopkinson bar and stress wave propagation
Figure 4.2	Incident bar momentum trap (From Pelletier[110])
Figure 4.3	Transmitter bar momentum trap (From Pelletier[110])

Figure 4.4	Notched specimen geometries
Figure 4.5	Carleton tensile split Hopkinson bar (From Pelletier[110])
Figure 4.6	Gas gun schematic (From Pelletier[110])
Figure 4.7	"Incident bar end" of Hopkinson bar apparatus (From Pelletier[110])
Figure 4.8	Strain rate vs. striker velocity (C-notch data)
Figure 4.9	Strain rate vs. striker velocity (E-notch data)
Figure 4.10	Diametral strain vs. engineering strain (E-notch data)
Figure 4.11	Diametral strain vs. engineering strain (C-notch data)
Figure 5.1	Flyer plate impact geometry and stress wave propagation
Figure 5.2	Wave propagation in a plate impact test (Karnes [120])
Figure 5.3	Standard Hugoniot for a metal (Rolsten [127])
Figure 5.4	Typical free-surface velocity data from VISAR
Figure 5.5	Schematic of Ernst-Mach-Institute light gas gun test section
Figure 6.1	Typical incident and transmitter bar waves for interrupted test
Figure 6.2	Complete Hopkinson bar data set for E-notch specimen
Figure 6.3	Complete Hopkinson bar data set for a C-notch specimen
Figure 6.4	Complete Hopkinson bar data set for a uniaxial specimen
Figure 6.5	True stress and true strain time histories for E-notch
Figure 6.6	Uniaxial stress-strain response for the (a) 850 HT, and (b) 600 HT UNS-C36000
Figure 6.8	True stress-strain data for MS-58, E-notch specimens for (a) 850 HT and (b) 600 HT

Figure 6.9	True stress-strain response for a C-notch of (a) 850 HT, and (b) 600 HT UNS-C36000
Figure 6.10	Comparison of stress-strain response between UNS-C36000 and MS-58 brass for (a) 850 HT, and (b) 600 HT
Figure 6.11	True stress-strain response for notches of various triaxialities with 850 HT for (a) 1.27 cm (0.5 inch), and (b) 8.89 cm (3.5 inch) brass rod
Figure 6.12	True stress-strain response for notches of various triaxialities with 600 HT for (a) 1.27 cm (0.5 inch), and (b) 8.89 cm (3.5 inch) brass rod
Figure 6.13	Ductility of brass as a function of triaxiality
Figure 6.14	Contour plots of (a) initial porosity distribution, (b) approximate location of measured fields with respect to notch geometry, (c-e) contours of porosity in uniaxial specimens with increasing strain
Figure 6.15	Contour plots of aspect ratio evolution with strain in uniaxial specimens
Figure 6.16	Contour plots of porosity in E-notch specimens
Figure 6.17	Contour plots of aspect ratio in E-notch specimens
Figure 6.18	Contour plots of porosity evolution in C-notch specimens
Figure 6.19	Contour plots of mean aspect ratio evolution in C-notch
Figure 6.20	Porosity evolution with strain for the uniaxial specimens
Figure 6.21	Porosity evolution with strain for the E-notch specimens
Figure 6.22	Porosity evolution with strain for the C-notch specimens
Figure 6.23	Comparison of the porosity evolution with strain between various specimen geometries

Figure 6.24	Void growth rate as a function of triaxiality at specimen centres
Figure 6.25	Mean void area evolution with strain for the uniaxial specimens
Figure 6.26	Mean void area evolution with strain for the E-notch specimens
Figure 6.27	Mean void area evolution with strain for the C-notch specimens
Figure 6.28	Mean void aspect ratio evolution with strain for the uniaxial specimens
Figure 6.29	Mean void aspect ratio evolution with strain for the E-notch specimens
Figure 6.30	Mean void aspect ratio evolution with strain for the C-notch specimens
Figure 6.31	Comparison of mean void aspect ratio for various specimen geometries
Figure 6.32	Micrograph of central crack in a E-notch specimen
Figure 6.33	Classic cup-cone failure in brass E-notch
Figure 6.34	Micrograph of crack tip in E-notch
Figure 6.35	SEM micrograph of 850 HT E-notch fracture surface
Figure 6.36	SEM micrograph of 600 HT E-notch fracture surface
Figure 6.37	SEM micrograph of 850 HT C-notch fracture surface
Figure 6.38	SEM micrograph of 600 HT C-notch fracture surface
Figure 6.39	VISAR velocity profiles for 850 HT plates
Figure 6.40	VISAR velocity profiles for 600 HT plates
Figure 6.41	VISAR velocity profiles for "as-received" MS-58 plates
Figure 6.42	VISAR velocity profiles for various material stocks of similar flyer plate velocity
Figure 6.43	Contour plots of void parameters at the centre of the 93 m/s plate impact specimen

Figure 6.44 **Radial distribution of damage in the spall layer of impacted plates near their outer edge**

Figure 6.45 **Axial distribution of damage in the spall layer of impacted plates near their outer surface**

Figure 6.46 **Micrographs of the incipient spall layer in plate impact specimens**

CHAPTER 1

INTRODUCTION

The plastic deformation of ductile materials may lead to ductile fracture through the nucleation, growth and eventual coalescence of microvoids to form cracks. Ductile fracture has been shown to be a prevalent mode of failure for many impact events which include high rates of loading. An improved understanding of the ductile fracture process under high rate conditions would be beneficial in areas including aircraft engine blade and rotor containment, bird strike, automotive crash worthiness, metal forming, and several defence applications. To this end the focus of this research is the quantitative evaluation of void damage evolution under high strain rate conditions.

High strain rate loading is achieved, in the current research, using plate impact and tensile split Hopkinson bar experiments. Plate impact tests allow circular plate specimens to be loaded under conditions approximating uniaxial strain at strain rates of 10^4 to 10^5 s^{-1} . The tensile split Hopkinson bar apparatus is capable of loading tensile specimens at rates of 10^3 to 10^4 s^{-1} . While damage evolution studies have been performed under quasi-static loading rates, in the past, the inability to stop dynamic tests to evaluate the evolution of damage has hindered similar work at high rates. Momentum trapping techniques have recently been developed [1] which allow interrupted Hopkinson bar tests to be performed. With the

availability of such techniques, Nemat-Nasser [1] has noted that the need exists for experimental work to characterize damage evolution at high strain rates.

Quantitative damage measurements have been hampered in the past by the difficulty in obtaining statistically relevant sample sizes using manual quantitative metallographic techniques. The application of digital image processing to quantitative metallography has made this task much easier by automating most of these procedures.

For this research, interrupted Hopkinson bar tests have been performed on uniaxial and notched tensile specimens. Quantitative image analysis was then used to characterize the damage within these samples and the evolution of damage with strain at various triaxialities was obtained. These results are compared to damage measurements from the higher rate plate impact tests. To the knowledge of this investigator, no published data has been produced on void evolution based on interrupted Hopkinson bar tests.

Free-cutting brass (UNS-C36000) will be used for this investigation and its suitability as a model material for damage studies will be evaluated. This brass has a lead phase which may be manipulated through heat treatment to produce various distributions of near-spherical particles. These lead particles nucleate large voids at low strains allowing void growth and coalescence to be studied.

The following sections of this chapter present a review of previous work on ductile fracture (Section 1.1) followed by research considering the effect of strain rate on this process

(Section 1.2). The last section gives an outline of this thesis and the research presented herein.

1.1 Ductile Fracture

Rogers [2] first noted that voids nucleate, grow and eventually coalesce to form cracks in plastically deforming axi-symmetric tensile specimens of ductile materials (Figure 1.1). The coalescence or agglomeration of voids was found to form a central crack which propagated by coalescing with adjacent voids within the plastic zone at the crack tip, a mechanism first described by Tipper [3]. The central crack grows in this manner until a global instability occurs and the specimen is sheared apart at its outer edge resulting in the classic cup-cone failure. The fracture surface exhibits a characteristic "dimpled" appearance where each dimple represents one half of a void. This research will focus on damage accumulation which occurs as a result of bulk material deformation as opposed to that of blunting crack tips, which involve high stress gradients with respect to inter-void spacing. It is assumed that each of the ductile fracture processes, that is, nucleation, growth and coalescence, may be described independently to make the analysis tractable. It is however widely acknowledged [4] that the processes are to a certain extent inter-dependent and that they all occur concurrently to different degrees. This overlap of damage processes makes experimental assessment of models quite difficult.

Analytic models of the ductile fracture process are often incorporated into finite element or finite difference codes for which theories of plastic deformation have been well

characterized. The transient loading by stress waves at high strain rates negates the use of purely analytical models due to the complex, geometry dependent, interaction of stress waves in test specimens. Many researchers investigating these processes at high rates apply quasi-static descriptions of damage evolution to non-linear dynamic codes capable of capturing the dynamic propagation of deformation by stress waves. The assumption behind this approach being that the damage mechanisms do not change to any great extent with strain rate.

The various parameters affecting the three component processes of ductile fracture at quasi-static loading rates will be discussed in the balance of this section. Section 1.2 will address the effect of strain rate on these processes.

1.1.1 Nucleation

The nucleation of voids may be classified as homogeneous or heterogeneous [4]. Homogeneous nucleation occurs within grains in regions of high dislocation density not associated with precipitates or inclusions. Heterogeneous nucleation, the more common form, occurs at discontinuities in the material microstructure. These discontinuities may exist at grain boundary triple points, but in commercial alloys, sites typically include second phase particles and inclusions. Nucleation occurs at a second phase particle by decohesion or cracking of the particle depending on the particle strength, shape, and size, as well as the characteristics of the particle-matrix interface. Argon [5] has shown that sub-micron voids require nucleation criteria based on dislocation theory while the nucleation of larger

voids is driven by macroscopic deformation processes and may be described by a continuum approach. Analytical nucleation models have been developed based on energy criteria as well as local stress and strain criteria. As the current research will focus on a material containing void-nucleating particles of size greater than one micron, this review will be limited to continuum models of void nucleation, growth and coalescence.

Gurland [6] developed an energy criteria which assumed that decohesion of the particle from the matrix occurs when the elastic strain energy released during decohesion is greater than the work of adhesion, W_a , of the particle. The work of adhesion may be defined as,

$$W_a = \gamma_m + \gamma_p - \gamma_{mp} \quad (1.1)$$

for which γ_m and γ_p are the surface energies associated with the matrix and particle respectively, and γ_{mp} represents the interface energy. While equation (1.1) is a required condition for nucleation, it has been shown [6] that it is not a sufficient condition as the particle-matrix interface strength must also be exceeded. Care must be taken in defining what characterizes a nucleation event. Models which assume that nucleation occurs upon particle-matrix decohesion at the poles aligned with the tensile stress direction provide much more accurate results [4] than those which assume complete separation of the entire particle is required [7]. Tanka et al. [8] has suggested that this criterion is easily satisfied for particles larger than 0.1 μm . As nucleation is already energetically favourable for such particles, the condition for decohesion reduces to a local stress criterion [9]. While the exact size of particle which denotes this transition is in dispute [9], it is on the order of one

micron. The analytic model developed by Tanka [8] and similar work by Fisher [9] have identified a particle size effect on the critical nucleation strain for which nucleation strain varied inversely with the square of the particle radius. Gurland [10] noted this effect experimentally and suggested that it occurred due to the increased probability of critically sized flaws in larger particles. Experimental work to investigate the energy criteria described by equation (1.1) has been performed by Fishmeister [11] and Hippsley [12] in which metallurgical techniques were used to decrease the work of adhesion in various steels. These studies indicated that a reduction in work of adhesion lowered the nucleation and fracture strain of these materials.

Criteria for nucleation based on critical stress or strain have also been developed. Argon [5,13] developed a nucleation model based on the critical radial stress at the particle-matrix boundary required to overcome the interfacial strength. The stress at the boundary was related to the local stress represented by the matrix flow stress, $\bar{\sigma}$, summed with the superimposed hydrostatic stress, Σ_{hyd} . Thomson [14] recognized that a stress concentration existed at the void and included a constant, C , to account for it. The condition for nucleation is then

$$\sigma_i \geq C \bar{\sigma} + \Sigma_{hyd} \quad (1.2)$$

where σ_i is the interfacial strength. Goods and Brown [4] reasoned that the local stress would be proportional to the root of the local plastic strain, ϵ_p , which would correspond to the nucleation strain, ϵ_N , when the nucleation criteria was met. Combined with equation

1.2, this argument suggests that

$$(\epsilon_N)^{1/2} = H(\sigma_I - \Sigma_{hyd}) \quad (1.3)$$

in which H is a constant. This relationship between the strain for void nucleation and hydrostatic stress has been confirmed experimentally by Argon [13] and French [15] in studies of iron carbide particles in iron.

The concept of a single critical value of stress or strain to nucleate voids is unrealistic since interface characteristics vary between particles such that decohesion progresses gradually over the entire particle population [16]. The nucleation strain will be altered by both the local microstructure as well as the stress-state. Local microstructural features affecting nucleation include particle, size, shape, orientation, spacing and interfacial strength. In most materials, statistical variations exist in these parameters which suggests nucleation is a continuous process over a range of strains. Large iron carbide particles in iron have been noted to nucleate at lower strains than smaller ones [17]. Nucleation has also been shown to occur preferentially along the longest axis of particles parallel to the tensile direction [18,19,20].

Embury [21] has suggested that a wave front may be defined which passes through the particle area histogram (Figure 1.2). The front delineates particles which have nucleated from those which have not. The movement of the front is defined by the strain and hydrostatic stress in the matrix. The nucleation strain is then defined accounting for the

statistical variations in microstructure. While this approach complicates the analysis and increases the number of experimental values required to define nucleation, it results in a more physically-based model.

The majority of analytic nucleation models are derived for isolated, rigid, spherical particles in an elastic-perfectly plastic or linearly hardening matrix. While this may approximate hard particles at low void volume fractions, it does not describe all systems. Fisher [20] has suggested that nucleation occurs preferentially in clustered regions of particles as the plastic flow between the particles is constrained [5,9]. These interaction effects become more prevalent as spacing decreases with increasing porosity [4].

In experimental work, the existence of bimodal void nucleation has been used to account for the difference between the bulk porosity at fracture and the much higher levels at the fracture surface [19,22,23]. Close examination of fracture surfaces often shows large voids connected with much smaller and more uniform voids. This observation is accounted for by suggesting that large particles nucleate early during deformation while a second phase of much smaller and more strongly bonded inclusions nucleate voids late in the deformation process at much higher strains. These voids form in the inter-void ligaments of the large particles and may be promoted by void sheeting mechanisms proposed by Rogers [2].

1.1.2 Growth

The growth of voids in ductile materials has been widely studied both analytically and, to a lesser extent, experimentally. Plastic deformation of the matrix material induces volumetric and shape change in voids. The local stress and strain values around the void are higher due to concentration factors resulting from the requirement for a stress free surface of the void. Typically, analytic models assume that the dilational (volume changing) and extensional (shape changing) cavity growth may be determined independently. Much of the information on void growth has been determined from analytic models. These models are based on continuum-plasticity models of the growth of through-holes in plates and embedded voids in infinite blocks or periodic unit cells.

McClintock [24] derived an analytic approximation of the growth of a through-hole in a plastically deformed plate in terms of the applied far field stress and plastic strain. Using a development from Berg [25], the eccentricity of the hole was also determined to describe the approximate rate of shape change or extensional growth. The resulting model was shown to predict the eccentricity of through-holes in copper plates by Perra [26] and agreed with subsequent finite element simulations by Becker [27]. Both Perra and Becker noted that the calculated void growth was higher than was predicted using McClintock's model. Nemat-Nasser and Taya [28] used finite element methods (FEM) to consider the growth of a periodic array of holes and found good agreement in both eccentricity and hole growth. Despite differences in geometry, the McClintock model has also been compared to embedded voids. Experimental testing by Thomson [23] on iron carbide particles in iron

and by Melander [29] on cuprous oxide inclusions in copper both showed reasonable agreement with the model. A review of the topic by Sowerby [30] indicates the model also matched the results of LeRoy [19] and Brownrigg [31] on spherodized iron carbide particles in steel. Work by Floreen [32] and Cox [33] suggests that the growth is over predicted using McClintock's model compared to measured void growth in maraging steels.

Analytic models based on isolated spherical voids embedded within an infinite block subjected to remote stress and plastic strain have been developed by Rice and Tracey [34] as well as Budiansky [35]. The Rice-Tracey model assumes a rigid-perfectly plastic matrix behaviour and predicts void growth from the initial void geometry,

$$\frac{\dot{R}_o}{R_o} = \frac{2}{3} (\dot{\epsilon}_{ij}^{\text{rem}} \dot{\epsilon}_{ij}^{\text{rem}})^{1/2} D \quad (1.4)$$

in which R_o is the initial radius of the void, $\dot{R}_o$ is the average radial velocity of the void surface and $\dot{\epsilon}_{ij}^{\text{rem}}$ is the remote tensile strain rate. The dilational growth factor, D , is given by

$$D = 0.558 \sinh \left(\frac{3}{2} \frac{\Sigma_{hyd}}{\bar{\sigma}} \right) + 0.008 v \cosh \left(\frac{3}{2} \frac{\Sigma_{hyd}}{\bar{\sigma}} \right) \quad (1.5)$$

in which v is the Lode parameter describing the strain state,

$$v = \frac{-3 \dot{\epsilon}_2}{\dot{\epsilon}_{11} - \dot{\epsilon}_{33}} \quad (1.6)$$

and the ratio $(\Sigma_{hyd} / \bar{\sigma})$ is the triaxiality.

It has been suggested that the Rice-Tracey model over predicts void growth at high triaxialities [36]. Barnby [37] compared the Rice-Tracey predictions to measured void growth in uniaxial tensile specimens and three point bend tests of C-Mn steel. Good agreement was noted for the low triaxiality tensile test but not for the higher triaxiality crack. The exponential dependence of dilational growth rate, D , with triaxiality was confirmed for A508 steel by Bermin [38] and Marini [39], however, the void growth rate was under predicted for triaxialities ranging from .33 to 1.50. Bourcier [40] and Worswick [36] both performed finite element analyses which showed that void shape or aspect ratio has a significant effect on growth rate. Voids that are prolate, that is elongated in the tensile straining direction (Figure 1.3), tend to grow at a slower rate than spherical ones. LeRoy [19] modified the Rice-Tracey model to account for void shape change with deformation. Budiansky [35] developed a model similar to that of Rice and Tracey with a rigid-viscoplastic rate hardening matrix. The initial and final void growth rates were both considered to account for the change in void shape with deformation. The model of Budiansky predicted that void growth rate decreases with increasing hardening exponent. Budiansky's model captured the interesting phenomenon in which at high triaxialities deforming voids do not elongate in the tensile direction but flatten or become oblate. This

effect has been confirmed by numerical studies [36,41] and is attributed to the incompressibility of the matrix and the increase in transverse restraint on plastic flow at high triaxialities. This flattening effect was not captured by Rice and Tracey [34] who assumed that voids would always elongate along the tensile straining direction.

Worswick [42] measured the void growth rate in uniaxial and notched tension specimens of leaded brass. The lead particles burst or tore early in the deformation process resulting in the nucleation of quasi-spherical voids. The measured dilational growth rate was found to compare well with the models of both Rice-Tracey and Budiansky [42]. The extensional growth rate of voids was found to decrease with increasing triaxiality.

A second aspect of the ductile fracture process is the degradation in material load carrying capacity due to the evolving damage. This phenomena is termed constitutive softening and results in a lower remote stress. The most cited analytic model which accounts for constitutive softening due to porosity was developed by Gurson [43,44,45]. This model describes the reduction in strength, described by the far field equivalent stress, Σ_{eq} , as a function of bulk porosity, f , and the hydrostatic stress, Σ_{hyd} :

$$\phi = \left(\frac{\Sigma_{eq}}{\bar{\sigma}} \right)^2 + 2fq_1 \cosh \left(q_2 \frac{3}{2} \frac{\Sigma_{hyd}}{\bar{\sigma}} \right) - 1 - q_3 f^2 \quad (1.7)$$

in which $\bar{\sigma}$ is the matrix flow stress and q_1 , q_2 and q_3 are "calibration coefficients".

Utilizing this yield function as a flow potential, Gurson [43,44,45] then derived a set of

constitutive relations for porous-plastic materials. This model predicts dilational growth but not the void shape change.

The assumption of isolated voids holds for low volume fractions but in highly clustered regions or in materials with higher void volume fractions, void interactions will alter void growth rates. The dependence of dilational growth on the initial volume fraction of nucleating particles was shown by Marini et al. [39] for low carbon steels with MnS inclusions. Interestingly, a dependence of void growth rate on initial porosity has not been observed in unit cell calculations performed by Worswick [36]. Quasi-static tensile tests performed on iron compacts with initial porosity ranging from 0.3% to 11% have been performed by Spitzig et al. [46]. The iron compacts of higher porosity were found to have progressively larger voids. The larger voids in the higher porosity compacts were found to grow faster than the smaller voids in compacts of lower porosity (Figure 1.4). The rate of void volume fraction evolution with plastic strain in each compact, when normalized by the initial void volume fraction, was found to follow a common trend (Figure 1.5).

1.1.3 Coalescence

The coalescence process is typically assumed to initiate once voids begin to interact. Coalescence of voids may occur by the agglomeration of neighbouring voids through direct impingement as well as by necking or shear failure of the inter-void ligament.

Initially, coalescence was considered a catastrophic event. Upon reaching some critical void volume fraction the material was assumed to fail [47]. Two other approaches have been taken to model coalescence. The first defines a geometric criteria for the joining of neighbouring voids based on the local void morphology [4,48]. The second approach is based on flow localization within intervoid ligaments. Such processes represent a bifurcation from stable void growth conditions [51]. The true nature of coalescence involves both local void geometry and is effected by stress state and material work or rate hardening effects. Coalescence is the least studied stage of the ductile fracture process due to the complex nature of experimental work.

The assumption that coalescence is a catastrophic event lead researchers to define a critical void volume fraction to cause fracture. This method is still used in practice and is based on the argument that residual strength diminishes once porosity becomes high within a material. It was noted by Fyfe [47] that this value of critical void volume fraction was not a material property but varied with triaxiality for specimens of similar material. Fyfe has suggested that a critical void volume fraction may be used to define failure for loading cases in which the mean stress is quite high, but that it does not work well for lower mean stress loading.

Two forms of geometric criteria have been developed. Originally coalescence was modelled to occur as a result of impingement of voids [33] while later models define criteria for shear localization in the ligament between two neighbouring voids [2,49,50]. McClintock [24] developed a model for coalescence based on impingement which has been

shown to predict the behaviour of spherodized steel under tension [30]. It has been noted [51] that this model is questionable as it assumes stable void growth to failure. Thomason [52] developed a model to predict coalescence through shear localization between neighbouring voids. Coalescence was found to occur when the height of neighbouring voids equalled their inter-void spacing (Figure 1.6). This model captured the decrease in fracture strain with increasing initial porosity which has been identified experimentally [53]. LeRoy [19] determined the ratio of void length to spacing experimentally for spherodized steel and reported a ratio of 1.23 which is greater than the value of 1.0 assumed for most models. Sun Jun [54] performed metallographic studies on uniaxial and notched tensile specimens of spherodized carbides in 37Mn5 steel to examine the effects of triaxiality on coalescence. The examination of sections parallel to the tensile direction allowed measurement of the dimple size and aspect ratio at the fracture surface. Their results suggest that, at low triaxialities, voids experience greater elongation and that coalescence occurs through necking or shear of the inter-void ligament. Higher triaxiality tests in notched specimens suggested that voids coalesced through lateral growth and impingement (Figure 1.7). Sun Jun [54] developed a criteria for the strain at fracture based on void spacing which accounts for triaxiality by incorporating the Rice-Tracey void growth model. While these geometric criteria capture the effects of local void morphology on ductile fracture, they neglect effects related to localization such as matrix work hardening.

In considering the effects of local microstructure on coalescence it has been recognized that irregular void spacing or clustering of voids can dominate the process. Dubensky [55] has performed experimental studies on randomly spaced through-holes in axially loaded plates.

It was found that closely spaced holes grow at a slightly higher rate early in the deformation. Shear bands eventually develop and lead to coalescence or linking of closely spaced neighbouring holes. If a third hole was favourably located it would immediately link with the cluster resulting in a defect large enough to initiate fracture of the entire plate. The absence of a third hole in a suitable location resulted in the deformation progressing and other hole pairs linking until three holes joined to produce a large enough defect to fracture the plate. It has been suggested by Teirlinck [17,56] that damage accumulation should be predicted on two levels to account for the bulk composition as well as the clustered regions. The existence of clusters accelerates the bulk damage and eventually leads to failure as damage "percolates" through the coalesced regions as described by Tonks [57]. The difficulties associated with defining and measuring clustering [58] as well as capturing the coalescence process prior to fracture make experimental evaluation of such models very complex.

The constitutive softening which occurs as a result of void growth destabilizes deformation and accelerates the growth rate [59,60]. Unfortunately, models based on stable void growth alone, such as Gurson [43,44,45], tend to over predict ductility since they do not consider the onset of coalescence and neglect local void morphology. Tvergaard and Needleman [61,62,63] modified the Gurson constitutive model to artificially accelerate the void growth rate as the constitutive softening develops in an effort to capture coalescence behaviour. Their modified version of equation 1.7 is given as,

$$\phi = \left(\frac{\Sigma_{eq}}{\bar{\sigma}} \right)^2 + 2f^* q_1 \cosh \left(q_2 \frac{3}{2} \frac{\Sigma_{hyd}}{\bar{\sigma}} \right) - 1 - q_{31} f^{*2} \quad (1.8)$$

in which the parameter f^* is introduced to model coalescence and is defined as,

$$\begin{aligned} f^* &= f \quad \text{for} \quad f \leq f_c \\ f^* &= f_c + \left(\frac{f_u - f_c}{f_f - f_c} \right) (f - f_c) \quad \text{for} \quad f \geq f_c \end{aligned} \quad (1.9)$$

where

$$f_u = \frac{1}{q_1}$$

defines point where material strength vanishes and f_f is the void volume fraction at fracture. The term f_c identifies the critical void volume fraction at which the interaction of voids or coalescence commences.

The effect of thin disks of higher void volume fraction on localization has been investigated by Ohno and Hutchinson [64]. The size of the disk or local inhomogeneity on the localization strain was determined. It was found that the critical disk size which resulted in localization decreased with increasing triaxiality. Yamamoto [65] accounted for inhomogeneities in the matrix by performing a M-K localization type analysis [66] of tensile tests with oriented planes of higher initial porosity (Figure 1.8). Their results

indicated that lower initial porosity and increased work hardening delayed the onset of localization. It was also noted that the fracture strain under plane strain conditions was lower than that under uniaxial stress as shown experimentally by Hancock and Brown [67].

Thomason [68] has proposed a two part criterion requiring the occurrence of instability as well as attainment of the limit load across the inter-void ligament. His upper-bound rigid-plastic solution accounts for the evolution of void shape and spacing on the conditions for internal necking or plastic limit load failure. This limit is defined as the point where deformation is localized in the inter-void matrix over a single sheet of voids which forms the fracture surface. A recent modification to Gurson's constitutive model by Zhang [69], incorporates Thomason's model as a failure criteria. This eliminates the need to define f_c in equation 1.9. Zhang shows that this value, where void coalescence is assumed to initiate, is a function of triaxiality.

As mentioned earlier, failure in ductile materials often occurs by the second stage nucleation, growth, and coalescence of small voids in the ligaments between larger voids. Broek [70] has suggested that large particles, which nucleate at small strain, determine the location and moment of failure while the set of smaller voids initiates after the conditions for fracture have been met. This mechanism was supported by metallographic studies in which small voids are not visible in deformed specimens prior to fracture. When this second stage occurs it is often referred to as void sheeting mechanism [33]. The inclusion of this mechanism in two-dimensional [71,72] and three-dimensional [73] numerical work has shown that significant shear banding develops between voids, resulting in a lower strain

to fracture.

1.2 High Strain Rate Effects on Ductile Fracture

High rates of loading may result in the same processes of nucleation, growth and coalescence as occur during quasi-static loading. While the parameters and trends indicated in the preceding sections still hold, they are affected by the additional considerations of rate dependent material properties and localized adiabatic heating, as well as the introduction of inertial effects.

1.2.1 Inertial Effects

The inertia of an impacted body results in load propagation in the form of stress waves, as the applied load may not be transmitted to every point in the body instantaneously. The intensity and duration of the loading is defined by the propagation and interaction of these transient stress waves through the material. Strain rates above 10^3 s^{-1} are typically defined as being part of the dynamic range [74] in which high loading rate effects become important.

The strain experienced by the material in the initial stages of loading is not homogeneous as the elastic and plastic deformation are propagated by waves of different velocity. Depending on the impacted test geometry, the material may be reloaded by reflected stress waves and the interaction of these stress waves may generate zones of tensile stress state within which void nucleation and growth occur. Alternatively, if these waves are

compressive and their magnitude is large enough, void collapse may result [75,76]. The inertia of accelerated plastic flow can also promote uniform deformation, delaying the onset of localization and thus fracture. This effect occurs both globally as well as on the microstructural level within inter-void ligaments. Regazzani [77] has described this effect for the development of a neck in a uniaxial tensile test. Upon meeting the conditions for mechanical instability, a finite amount of time is required for the neck to develop. At high rates of strain, a considerable amount of uniform strain occurs over this time increasing the strain to failure. Once the neck has formed, the region outside of the neck is unloaded by a wave propagated from the outside of the neck and plastic flow is localized to within the neck. At the microstructural level, a similar process is assumed to occur as strain localizes within the inter-void ligaments. The effect of inertia is assumed to be more pronounced at this level as the local strain rates will be far greater than the nominal values due to the stress and strain concentration associated with the void geometry.

It has been noted by Shockey [78] that the effects of microstructure on microfailure and coalescence under dynamic loading are still not well understood. The dynamic processes of nucleation, growth and coalescence under quasi-static loading have been shown to be dependent on the existence of clusters of nucleating particles [5,9,20]. It has also been established that clusters affect localization [79] and fracture of materials. Tonks [57] has created two- and three-dimensional numerical simulations to investigate the effects of clusters at various loading rates. Tonks noted that when clusters of voids or damage are formed at high rates of loading, there is not enough time for the stress and strain fields to develop at the boundaries of clusters. This inhibits the larger clustered regions from

growing faster than smaller ones as noted in quasi-static simulations [80]. This delay results in a wider distribution of damage at higher rates, as smaller randomly spaced clustered regions continue to form until the conditions for fracture are met, as opposed to one, well positioned cluster, dominating fracture.

Benson [79] has investigated the effect of wave propagation on ductile fracture by considering various arrangements of cylindrical voids under the action of a tensile stress wave using an Eulerian numerical formulation. This model compared the coalescence of small clusters of six cylindrical voids under both static and dynamic loading conditions. Materials having a void volume fraction of 0.0184 were considered, utilizing both copper and 4340 steel matrix material models. Under dynamic conditions, the coalescence of voids in copper followed a percolation path through the first voids that the stress wave interacted with, regardless of their spacing (Figure 1.9). The voids "downwind" from this region were found to have much lower growth rates than those within the percolation path. This result differed from lower rate simulations in which coalescence occurred between nearest neighbours and all of the voids grew at approximately the same rate. Changing the matrix material to 4340 steel resulted in a higher stress amplitude being required to force localization to occur within first voids interacting with the stress wave. The steel matrix model also exhibited void sheeting between large voids at approximately 45° to the incident wave (Figure 1.10) as noted experimentally by Cox [81]. Benson concluded that the direction of the stress wave must be considered in coalescence models at high rates and that the regions of shielding created by rapidly growing voids are important. The magnitude of this shielding effect is related to the applied stress since higher void growth rates within

the percolation path will result in reduced downstream void growth.

1.2.2 Rate Dependant Material Properties

Many materials, particularly those of body centred cubic structure, exhibit strain rate sensitivity. This sensitivity leads to a change in stress-strain response with increasing strain rate. Typically, the stress associated with a given strain will increase with strain rate, as described by the equation,

$$\sigma = C(\dot{\epsilon})^m \quad (1.10)$$

where C is a constant and the strain rate sensitivity exponent, m , may be defined as,

$$m = \frac{\partial \ln(\sigma)}{\partial \ln(\dot{\epsilon})} \quad (1.11)$$

Strain rate sensitivity delays the onset of mechanical instability. As the neck forms, the strain rate in the neck increases. The increase in strain rate results in a higher stress being required to maintain flow, delaying localization. This higher stress may also lead to more extensive void nucleation in the later stages of fracture. Magnusen [82] performed tests on nickel and titanium compacts with initial void volume fractions of 0.1 to 9.4% for strain rates up to 10^2 s^{-1} . The influence of strain rate on fracture was found to be small for low strain rate sensitivity materials. The strain at fracture was found to be sensitive to the strain rate sensitivity coefficient.

It has already been noted, in Section 1.1.2, that void growth is exponentially sensitive to triaxiality. The formation of necking raises the triaxiality as described by Bridgman [83] and results in accelerated void growth. Preventing localization through elevated material rate sensitivities will allow more time for the voids to grow at a stable rate delaying the onset of accelerated growth which accompanies necking and fracture.

The behaviour of materials under very high strain rates after localization or under the action of shock loading may be defined by viscous effects. The viscosity of a material describes the relation between stress and strain rate in which the stress will tend to increase in proportion to strain rate. The increase in stress will alter the processes of nucleation, growth and coalescence in much the same manner as the strain rate sensitivity coefficient. Seaman et al. [75], Johnson and Addessio [85], Johnson [82,83] have developed void growth laws describing void volume fraction in terms of viscosity and mean stress. These models were developed for predicting spall failure through ductile fracture in impacted plates under conditions of uniaxial strain. Spall fracture will be discussed further in the review of plate impact techniques in Chapter 5. Johnson's analytical model for viscous void growth [84] was based on a void collapse model developed by Carrol and Holt [76]. Seaman and co-workers [75] developed an empirical model which included viscosity effects.

As discussed earlier, the nucleation of voids from particles larger than micron size is governed by a critical stress or strain criteria. It has been noted that the strain for void

nucleation is rate dependent suggesting that nucleation is a stress controlled process. Cortes [86] has shown that for very high deviatoric strain rates, in excess of 10^6 s^{-1} , viscous effects slow void growth to a point where fracture is controlled by nucleation alone. While the nominal strain rates considered in this research are lower than the rates in [86], nucleation may be important in defining fracture in plate impact experiments.

1.2.3 Adiabatic Heating

At high rates of strain, the heat generated in regions of large plastic deformation does not have sufficient time to conduct away. This retention of heat results in thermal softening of the matrix material, particularly within inter-void ligaments and at void walls where shear strains as high as 1000% [84] may be achieved. Thermal softening promotes instability thereby lowering the strain to fracture. Thermal softening may also lead to adiabatic shear band formation on a microstructural level promoting coalescence by shearing of inter-void ligaments.

1.3 Current Research

From the above review of the literature, it is evident that a need exists for experimental quantification of the rate of damage accumulation under high rate loading conditions. In particular, there is a lack of data concerning the development of damage as a function of applied strain, stress state, and strain rate.

To address this need, the current research has been undertaken to measure damage development under high strain rate conditions. Two apparatus have been used to perform mechanical testing, the Tensile Split Hopkinson Bar (TSHB), available at Carleton University, and the plate impact facilities at the German Ernst Mach Institute. The TSHB permits tensile testing at strain rates of 10^3 - 10^4 s⁻¹ while the plate impact facility generates spall through ductile fracture at rates of 10^4 - 10^5 s⁻¹. In addition, the TSHB apparatus at Carleton incorporates momentum trapping fixtures which allow interruption of deformation after which damage development can be determined. Image processing techniques were coupled with quantitative stereology methods to obtain measures of void volume fraction, void aspect ratio, and other particle/void fractures at various stages of loading.

The material adopted in this investigation was UNS-C36000 leaded brass. This "model" material contains a globular lead phase which introduces well-defined void nucleation sites. Using this material and the two high strain rate facilities permits void damage to be determined for two levels of strain rate and a variety of stress/strain states and partially addresses the need for experimental data under high rate conditions.

The balance of this thesis is organised as follows. Chapter 2 presents the image processing and stereology techniques. The heat treatment performed on the brass and the resulting material properties are discussed in Chapter 3 and the characterization of the lead particle population morphology is given. Descriptions of the TSHB and plate impact experiments are given in Chapters 4 and 5, respectively. The results of the Hopkinson bar and plate impact tests are presented in Chapter 6. A discussion of the results and conclusions from

this research are provided in Chapter 7.

CHAPTER 2

IMAGE ANALYSIS TECHNIQUES AND QUANTITATIVE MICROSCOPY

Digital image analysis was used to quantitatively evaluate the morphology of lead particles and voids within the brass matrix investigated for this thesis. This evaluation relies on mathematical relations that exist within the field of quantitative stereology to characterize three-dimensional structures from two-dimensional measurements. These methods [87,88,89] tend to be tedious as well as time consuming, however, automated systems based on image processing techniques provide a viable means by which to apply these methods. Imaging techniques limit operator fatigue by reducing data acquisition time and allowing many types of measurements to be made simultaneously. The automation of feature measurement also leads to more accurate and reproducible results as the measurements are less dependent on operator judgement and larger samples may be obtained. This research utilized such techniques to evaluate void and particle area, aspect ratio and volume fraction.

An overview of the major components comprising an image analysis system is shown in Figure 2.1 and will be discussed in the following sections. The fundamentals of digital image processing are well documented. Three texts of note have been authored by Pratt [90], Russ [91] and Castleman [92].

2.1 Image Acquisition of Metallographic Fields

The first step in image analysis is to acquire the best possible image. Carefully prepared specimens were placed on an optical microscope. An image of the microscope field was captured by a camera mounted on the microscope. The particles and voids were identified by a contrast technique and their outlines or borders are stored in a format utilizing data compression.

2.1.1 Imaging Equipment

Images were taken from an Olympus PME-3 inverted stage optical microscope with bright field illumination. A mounting bracket was designed which allowed a precisely adjustable micrometer stage from a Reichert 261885 microscope to be used as a platform. The microscope objective was a 20x MSPlan20 with a numerical aperture of 0.46. A green filter was placed in the light path to enhance the resolution of the image. The microscope was also fitted with an Olympus U-PMTVC trinocular head with a 2.5x optic. The trinocular head provides separate light paths to the eyepieces and a video camera.

A COHU monochrome charged coupled device (CCD) camera model 4915-2001/2000 was mounted on the trinocular head to image the micrographs. The electric iris and auto gain compensation of the camera were both disabled to allow greater control of field intensity and assure better repeatability of results. The camera hosts a 1/2" (diagonal) CCD array of light sensitive sensors. The image from the CCD array is transferred as an

analog RS-272 video signal through a coaxial cable to a frame grabber card in the data acquisition computer.

The frame grabber was a PCVISIONplus (PFGPLUS-640-3-60) card manufactured by Imaging Technologies Inc. The function of the frame grabber is to acquire a digital image from the video signal. This board was housed in an IPC 486 personal computer, model 486-33 DT. The system works with two displays. The first display was a high resolution NEC Multisync II RGB monitor (model JC-1402HMA) and displays the digital image obtained by the frame grabber card. The second monitor operates as a standard VGA monitor (TVM Low Radiation Colour Monitor, model MD-14II U08LR) for the personal computer hosting the frame grabber. The microscope field is digitized by quantizing the light intensity at each discrete point in the image. In this way each picture element or pixel is assigned a grey scale value ranging from black (0) to white (255). The spatial resolution of the digitized image is 640 by 480 pixels.

The resolution of the microscope objective, d , may be defined as,

$$d = 0.61 \frac{\lambda}{NA} \quad (2.1)$$

where λ is the wavelength of the light being used and NA is the numerical aperture of the objective. Substituting the numerical aperture of the MSPlan 20 objective of 0.46 and the

wavelength of green light as 546 nm, the resolution of the objective may be found to be 0.724 μm . The digital image is separated into 640 by 480 square pixels with each pixel having an edge length of 0.625 μm for the magnification used in this work. The resolution of the camera is thus higher than that of the objective and, as such, no loss of resolution occurs during digitization. It is important to note that for small objects, the accuracy of length-based measurements is limited by the system resolution. As the resolution of the system is on the order of one pixel, objects of ten pixels in length can at best be measured with an accuracy of 10%. The majority of objects of interest in this investigation were larger than ten pixels in length or 3.9 μm .

The image acquisition was controlled by a modified version of the ENCODE [93] software developed at Carleton University. The modifications were made as part of this research and allowed for more automated image acquisition which was important due to the large number of images required. Images were given numerically sequential file names automatically with a preset prefix or series designator. Image acquisition was then performed with a single key stroke.

2.1.2 Pre-Processing of Image Data

Effective image analysis requires that the best possible microscope image be obtained. Poor sample preparation, field illumination and video noise all bias the final results. The image acquisition software ENCODE performed a multiple frame average of the grey scale image to average-out any random noise in the image improving the signal to noise

ratio. A light source correction was also applied to account for uneven illumination across the image.

Multiple Frame Average

The image from the camera is fed into the frame grabber memory. Noise from random sources, introduced primarily in the analogue video signal, is present in any single image. In an attempt to limit the noise, successive images of the field of interest were acquired and their resulting grey scales for each pixel were averaged. In this manner the random noise was averaged thus improving the signal-to-noise ratio.

Light Source Correction

Despite careful alignment of the light source, the distribution of light across a microscope field is rarely even. While not visible when viewing the image by eye, variations do exist and alter the acquired image as was evident after segmentation. To reduce this variation, field illumination compensation was applied to the averaged grey scale image. Prior to each data acquisition sequence a specimen of cartridge brass was acquired. This material does not contain lead particles and provided a monotone target of similar reflectivity to the brass matrix material of the leaded brass samples. If the illumination is constant across the field, the corresponding digitized image will have a uniform or a "relatively clean" constant grey-level. Uneven illumination will lead to differences between local pixel grey levels and the average grey level for the image. To correct for uneven illumination, multiplicative scale factors can be calculated as the ratio of the mean or

average grey level across the entire image divided by the actual grey level for each pixel. These factors are stored and applied to subsequent images containing particles or voids, thereby correcting for the variation in field illumination. In the current work, correction factors were not applied on a pixel-to-pixel basis but were obtained for four-by-four blocks of pixels thus smoothing any inconsistency due to local noise or polishing artifacts.

2.2 Image Segmentation

Having generated a "relatively clean" grey scale image, it was next segmented. That is to say, the lead particles or voids were identified and separated from the background. Any image processing requires the structure exhibit adequate contrast to allow the various components to be distinguished. The leaded brass used in this research provides excellent contrast and all of the voids or particles were identified as objects of interest. The segmentation was performed using a grey scale thresholding technique. A threshold value of grey scale was determined for which all pixels of that grey scale level or lower (darker) were identified as comprising particles or voids. The pixels of higher grey level were identified as the matrix material. Ideally a wide range of grey scale levels will exist between the particles and background, clearly delineating the objects of interest. Using this threshold, the grey scale image was transformed to a binary image. This binary image was then stored in a run-length encoded format. For each row of the acquired image only the column locations in which the pixel colour changed between particle and matrix or vice versa were recorded. The use of this format allowed a data compression ratio of approximately 30 to be achieved.

The choice of the optimum threshold grey scale value is a complex problem which must be carefully considered for the class of objects in the images. Several methods have been employed in previous studies. These range from a flicker method [94,95] to adaptive feature segmentation [91]. With the flicker method, the threshold is set by alternating between the live and the thresholded image and observing the correlation between the two images by eye. Feature dependant adaptive thresholding analyzes each individual object and assigns a unique threshold value to that object. For each method, assumptions must be made about the objects of interest. The majority of methods assume that the objects of interest are monotonically increasing spots of grey scale of nearly equivalent grey scale distribution. The centre of the spot is the darkest region and pixels progressively become lighter moving towards the edge and, finally, the background. The choice of the final threshold value may alter the measured quantities significantly based on the clarity or resolution of the edges of the objects of interest.

The method used in this study to select the threshold value of grey scale was based on the image having a bimodal grey-scale frequency histogram. This histogram plots the frequency of each of the 256 grey levels for the entire image. A schematic of a bimodal grey-scale histogram is shown in Figure 2.2. It is characterized by a lower and upper peak as well as a constant background level. This background arises from the imperfections in sample preparation such as polishing scratches as well as the border pixels surrounding each object. The lower peak is assumed to comprise the particle pixels while the upper peak corresponds to matrix or background pixels. The threshold value is normally taken as the mid-point between the two peaks.

2.3 Image Processing Methods for Feature Extraction

Having acquired the binary image, the objects of interest must then be extracted and the feature measurements obtained. For this work, the particle size, shape and orientation were evaluated. This was done using unix-based particle analysis software PAS [96] which was developed over the course of this work from an earlier code IMAGE [93] based on an algorithm developed by Agin [97].

2.3.1 Characterizing Particle Size

The acquired image in run-length encode format was analyzed using PAS. The connectivity of pixels is first defined. This operation establishes which pixels are attached to each other in such a manner that individual particles or voids are identified and separated from the background of the image. The algorithm stores the border pixels defining the outline of each particle. Features such as the area and aspect ratio of individual particles are also determined concurrently.

As a result of the metallographic techniques employed for this investigation, the larger voids often had interior regions which were lighter than the rest of the object. This resulted in some interior regions being identified as background. These regions were automatically merged within the surrounding objects in the evaluation of area and other measurements.

Calibration of the measurements is required to relate particle size in pixels to the actual size in microns. This calibration was done using the stage micrometers to evaluate the screen size in physical units and then dividing by the number of pixels along each edge. The pixels were determined to be square with a width and height of $0.625\text{ }\mu\text{m}$ and an area of $0.391\text{ }\mu\text{m}^2$.

2.3.2 Evaluation of Particle Shape and Orientation

The particles and voids observed were primarily ellipsoidal in form and, as such, an ellipse was fitted to these objects. This fit allowed the aspect ratio and orientation of each object to be determined.

Many feature specific measurements, such as aspect ratio, are a function of the object's orientation with respect to the screen and the method of measurement. The principal axes of an object may be defined based on its second moments of area. These axes may be used to determine the orientation of the object, as well as yielding values of principal length and width which are independent of the original image axis.

The ellipse fit was performed using two methods. The first was based on the second moment of area calculated during connectivity analysis. An ellipse is defined having an equivalent second moments of area to the object being modelled. Following this construction process, the dimensions and orientation of the principal axes may be determined for each ellipse. The second method was based on a least square fit to the

pixels at the edge of the object. The second method tended to fail for small or non-elliptical objects since the resulting aspect ratio and orientation of smaller objects tended to be biased due to the limited number of possible orientations and aspect ratios. As such, a minimum particle size was defined to obtain more representative average measures of these quantities and particles below this size were not used in subsequent aspect ratio statistics.

For this study, the "least squares fit to perimeter" method was primarily used with the "area moment method" being included on occasions when the least squares fit failed for a given object.

2.3.3 Stereological Relations

The fundamental goal of stereology is to determine three-dimensional information from two-dimensional measurements. The relation,

$$f = A_A \tag{2.2}$$

describing the equivalence of volume fraction and the measured area fraction is a classical example of such a relation and the one used in this work. The derivation of this and similar relations may be found in several texts [87,88,89].

The three-dimensional porosity or volume fraction is inferred from the two-dimensional area fraction with the assumption of a randomly oriented sampling plane with respect to the structure being measured. Several of these randomly oriented planes must then be measured to obtain an accurate measure of mean volume fraction. This was impractical for the measurements made here for several reasons. The most obvious problem is the large number of test specimens at each state which would be required to accurately determine the structure at that state while assuming random planes of measurement. The second problem arises from the localized nature of damage evolution. Neither the initial morphology nor the strain distribution over the microscopic field was homogeneous. In general the voids and particles of the axi-symmetric tensile specimens and plates used in this study were originally randomly oriented. As deformation proceeds, particles and voids elongate in the axial direction at different rates depending on the local strain field. For this research, the mid-plane of the specimens were analyzed. While this introduces some bias into the results, particularly in regions of high deformation, it was assumed that the trends of interest could still be obtained from this data. To apply a more complex analysis would be beyond the scope of this research.

2.3.4 Sources of Error and Bias in Field Measurements

Every step of the quantitative metallography process outlined in Figure 2.1 and discussed above adds uncertainty and bias to the final measured quantities. The first step in limiting these errors was to define the sources from which they arise. The majority of these points have been addressed in preceding sections.

Sample preparation is the first and perhaps the most important step in assuring an accurate measurement. Great care was taken in this task, as described in Section 3.2, to assure both accurate representation of the particles or voids as well as limiting any defects in the polishing which would contribute to poor measurements as described in Section 3.2.

The selection of magnification must be done carefully to assure adequate resolution of the objects of interest while at the same time assuring a large sample area may be measured in a reasonable amount of time. The resolution of both the microscope objective and the imaging equipment must be analyzed to assure the proper limiting factor is determined and defined as the resolution of the system. The largest possible field size which maintains adequate resolution must be determined to ensure both accuracy and precision of measurements. Also, the magnification must be selected so as to provide a reasonable number of pixels per particle in order to accurately capture the particle shape.

Noise in the image signal and the fluctuations in illumination across the image field must both be considered for the grey scale image. Care must also be taken in focusing each individual field to assure the best possible measurements are obtained.

The thresholds must be carefully chosen to assure proper segmentation of the data as this will have a significant effect on both accuracy and repeatability of measurements. This is of particular importance for fields of small particles.

Problems arise when particles intersect the edge of the field of view. The inclusion of partial objects tends to bias feature measurements. To deal with this problem a guard zone is defined within the image describing the area which will be analyzed. Objects outside the guard zone are ignored.

The assumptions made for stereological relations which are neglected will bias the results. This was discussed in some detail in Section 2.3.3.

Uncertainty arising from sampling errors must also be considered. For low volume fractions, small fields exhibit a high variability in measurements and a relatively large area must be investigated to improve precision. Measurement fields were chosen to follow a systematic pattern. Variance in field to field measures also occurs due to the existence of field-sized clusters of particles or voids [98] further complicates the analysis.

The average and standard deviation of measured quantities were taken as the starting point for statistical analysis. This analysis was complicated by the non-gaussian distribution of data, the large natural variations in object parameters and the complex methods of measurement.

CHAPTER 3

MODEL MATERIAL

The majority of models describing the nucleation, growth and coalescence of microvoids assume an array of spherical or cylindrical voids. It has been shown by Worswick [42], that free-cutting brass (UNS-C36000) provides a useful model material for experimental investigations paralleling the development of such theoretical work. The brass matrix has a nominal composition by weight of 61.5% copper and 35.5% zinc with 3% lead included to improve machinability. The lead phase is essentially insoluble in the brass matrix and forms globular particles which aid machining by providing lubrication as well as promoting chip formation.

The tensile strength of lead is roughly 20 MPa which is very low when compared to the tensile strength of 600 MPa exhibited by the brass. This difference indicates that the lead particles will tear or debond to nucleate voids at a relatively low strain. The brass matrix is ductile allowing a large amount of strain and thus cavity growth to occur prior to fracture. The combination of these two features suggests that free-cutting brass would be a good model material for studying the mechanisms of ductile fracture as voids are nucleated early in the deformation history allowing the growth of these voids to be studied with evolving strain.

3.1 Heat Treatment

The lead particles in the brass matrix were spherodized by heat treating the material. Two annealing temperatures of 600°C and 850°C were applied. The heat treatment lasted for two hours and was followed by an air cool to room temperature. These schedules were optimized based on a trade-off study performed by Worswick [42] which considered the resulting area and aspect ratio distributions of the lead particles. The 850°C treatment resulted in larger and more evenly spaced particles than the 600°C treatment. The final result was two batches of brass having similar matrix properties and volume fractions of lead but different lead particle morphologies.

While spheroidization was promoted by these heat treatments, both the spatial and size distribution of particles remained non-uniform. It was deemed impractical to obtain a commercial alloy could meet a criterion of uniform distribution. Thus growth of individual voids could not be ascertained and statistical measurement of void area or volume fraction and average aspect ratio were used to determine void growth. This approach assumed that the mean void volume would be representative of the average growth of an individual void.

The brass rod purchased for this work was received in the half-hard condition. All of the cold work associated with this condition is relieved during the heat treatment. The anneal also promotes recrystallization and grain growth, particularly for the higher temperature 850°C anneal. Due to the difference in annealing temperature, grain growth occurs at

different rates. Relatively little grain growth occurs at 600°C and the resulting grain size is much smaller than that of the 850°C treatment.

3.2 Specimen Preparation

The accurate representation of small voids or soft particles in metal has been recognized as one of the most difficult problems in metallography [99]. In general, it is difficult to polish soft metals without promoting rounding of the void edges which obscures the actual void size and shape. Several techniques were employed in an attempt to achieve optimum results. These ranged from skidding techniques to vibrational polishing sequences. The final sequence is summarized below:

1. The specimens were milled or sectioned with an abrasive cut off wheel approximately two mm from the final metallographic plane.
2. Specimens were mounted in 1.25 inch bakelite mounts and roughly ground on 240 grit silicon carbide paper to approximately 20 μm from the plane of interest.

This rough grinding process was followed by the following fine grinding sequence on a Buehler ECONOMET III automatic polishing machine fitted with a WHIRLMET polishing attachment. The wheel was set to rotate at 125 rpm and a 507 g weight was used with water as a lubricant:

3. 15 seconds on 240 grit silicon carbide paper
4. 30 seconds on 400 grit silicon carbide paper

5. 30 seconds on 600 grit silicon carbide paper

For the rough polishing stage, weight was then increased to 1421 g and the wheel speed was increased to 250 rpm:

6. 2 minutes on Struers TexMet cloth charged with 3 μm diamond paste
7. 4 minutes on TexMet charged with 1 μm metadi oil-based diamond suspension
8. 1 minute on TexMet charged with 1/4 μm diamond paste

The samples were then mounted in 410.0 g pucks for final polishing on a Buehler Vibromet I vibrational polishing machine:

9. 15 minutes on Micromet cloth with a Struers OPS magnesium oxide-water slurry with a rotational rate of 5 revolutions per minute

This final stage was performed to improve the sharpness of the particle edges as well as removing or lightening any polishing artifacts. The final product was viewed at high magnification to assure that the edges of the voids were not rounded and that all major polishing artifacts from previous polishing stages had been removed.

3.3 Stock Material Heat Treatment and Particle Populations

The plate specimens investigated in this research were manufactured from annealed 8.89 cm (3.5 inch) diameter UNS-C36000 brass rod. The same 8.89 cm (3.5 inch) rod was also used to manufacture the majority of the E-Notch tensile specimens. Additional plate specimens machined by the Ernst Mach Institute in Freiburg Germany were made of "as-

received" 3 inch diameter MS-58 brass rod. The majority of the uniaxial tensile specimens and C-Notch specimens were manufactured from annealed 1.28 cm (0.5 inch) diameter UNS-C36000 brass rod.

The difference in the rod diameter of the UNS-C36000 brass used for the specimens manufactured at Carleton resulted in different cooling rates during air cooling after annealing. This variation resulted in slightly different lead particle populations and microstructure.

The MS-58 brass had a nominal composition of 58% copper - 40% zinc - 2% lead. The higher zinc concentration places the equilibrium structure in the beta phase of the phase diagram (Figure 3.1). However the material goes through a bainitic transformation as shown in the TTT diagram given as Figure 3.2 which does not occur with the UNS-C36000 brass. This increases the complexity of the microstructure and results in a stronger, more brittle material. Due to the large change in microstructure in comparison to the UNS-C36000 brass, the results of tensile specimens manufactured from this material will not be discussed to simplify the analysis of the void growth data.

Micrographs of the UNS-C36000 brass 8.89 cm (3.5 inch) rod heat treated at 600°C and 850°C are shown in Figures 3.3 and 3.4 (as polished) and Figures 3.5 and 3.6 (after etching). A micrograph of the 850°C 1.28 cm (0.5 inch) rod after etching is also given as Figure 3.7.

The as-polished micrographs (Figures 3.3 and 3.4) reveal the difference in the particle size between the 600°C and 850°C heat treatments. Both heat treatments have resulted in approximately spherical lead particles. These particle fields have been characterized using the image processing system. The resulting particle area and aspect ratio histograms are in Figures 3.8 through 3.11. The mean particle area of the 600°C brass of $11.3 \mu\text{m}^2$ was considerably lower than that of the 850°C brass, $24.3 \mu\text{m}^2$. The number of particles was much higher for the 600°C anneal since the lead fraction is essentially constant while the particle size is smaller.

The etched microstructure revealed two important features. The average grain size of the 600°C brass was smaller than that of the 850°C brass and the 8.89 cm (3.5 inch) rod had a slightly higher proportion of retained beta phase and larger grain size than the 1.28 cm (0.5 inch) rod.

The composition of the UNS-C36000 brass was confirmed by microprobe (EDX) analysis performed by Carleton's Electron Microscope Research Facility. The results of this analysis are summarized in Table 3.1. The nominal composition of grains was determined to be 65% copper-35% zinc. This was compared to the nominal composition of the phase at the grain boundaries of 55% copper - 45% zinc. If equilibrium structures are assumed to exist, the phase at the grain boundaries will be ordered beta phase, as shown on the phase diagram in Figure 3.1, while the grains are primarily alpha brass.

3.4 Mechanical Properties of Annealed Brass

Both static and dynamic tensile stress-strain curves were determined. The static data was obtained using an Instron tensile testing apparatus available in Carleton's Civil Engineering department. The results are given in Figure 3.12 where it is shown that the 600°C brass had a slightly lower tensile strength and slightly greater ductility than the 850°C brass. It should be noted that the elastic modulus and yield strength of the material were not well defined in these tests.

Dynamic tensile testing using the tensile split Hopkinson bar (discussed in the next chapter) was also performed to determine the dynamic stress-strain response. The resulting stress-strain curves are given in Figure 3.13. The 600°C brass is again shown to have a slightly lower tensile strength as compared to the 850°C brass. The stress-strain curve presented does not represent the complete response of the material as fracture did not occur.

The strength of the annealed brass was found to be comparable with the published static tensile strength of 400 MPa [99] under both quasi-static as well as dynamic loading conditions.

CHAPTER 4

TENSILE SPLIT HOPKINSON BAR EXPERIMENTS

This research will investigate damage accumulation in the high strain rate regime using Hopkinson bar and flyer plate impact techniques. Hopkinson bars are capable of achieving nominal strain rates of 10^3 s^{-1} for a range of durations while plate impacts can generate strain rates as high as 10^5 s^{-1} for shorter durations. At these high loading rates stress waves are propagated and their interaction produces dynamic deformation and fracture of the test specimens. This chapter describes the Hopkinson bar apparatus used in this research while Chapter 5 describes the plate impact experiments.

The major benefit of the split Hopkinson bar apparatus is that it allows the stress and strain in a specimen to be determined without directly measuring the gauge length or the load within the specimen. Load cells and most extensometers have complicated frequency responses and display inertial effects at high loading rates which would obscure the data being measured. These problems are avoided when using the tensile split Hopkinson bar as the load and displacement of the specimen are inferred from the elastic stress waves which propagate in the bars.

4.1 Historical Perspectives

Bertram Hopkinson [101] first proposed using the elastic response of a cylindrical bar to characterize an impact event. In his tests, a pressure load was applied to one end of a bar and its magnitude was determined by the momentum transferred to the bar. Kolsky [102] showed that by placing a specimen between two bars, the dynamic response of the material could be determined. It was Kolsky's developments that led to the first split Hopkinson or Kolsky bar. Various forms of Kolsky's apparatus have been developed to perform high strain rate tests in compression, shear, tension and torsion at strain rates between 10^2 and 10^4 s^{-1} . Harding, Wood and Campbell [103] are generally attributed with producing the first tensile split Hopkinson bar.

Pochhammer [104] and Chree [105] independently solved the equations of motion describing sinusoidal wave propagation in an infinite cylindrical bar. These equations form the theoretical foundation of the wave propagation in Hopkinson bars and have been used by Davies [106] to describe three-dimensional effects due to dispersion and inertia in the bars that are neglected by Kolsky's analysis. The theoretical analysis of Davies establishes the accuracy of the one dimensional assumptions used in analyzing the Hopkinson bar data.

4.2 Theoretical Basis of the Tensile Hopkinson Bar

The fundamental operation of any Hopkinson bar is quite similar. An impact event

causes the propagation of an elastic wave in the incident bar which travels at the velocity of sound in the bar material. In the case of Carleton's tensile bar, an initially compressive pulse generated by the impact of a striker, shown in Figure 4.1, reflects from the free end of the incident bar as a tensile pulse maintaining the stress free boundary condition at the end of the bar. This tensile wave travels along the incident bar (Figure 4.1a) and impinges on the test specimen which is of smaller cross-section, A , than the bar. Due to the change in impedance (ρc) and linear mass (ρA) at the specimen-bar interface, the specimen is loaded and deformed plastically. A portion of the incident loading wave is propagated through the specimen into the transmitter bar and a portion is reflected back along the incident bar as shown in Figure 4.1(b). The waves are recorded using piezoresistive strain gauges. These gauges are typically located far enough away from the specimen that interactions between the incident and reflected waves in the incident bar are avoided (Figure 4.1(c)).

The recorded strain gauge data may then be used to determine the stress-strain response of the specimen. The magnitude and duration of the reflected pulse define the strain rate history. The transmitted pulse represents the portion of the incident stress wave which is propagated through the specimen and is used to determine the stress experienced by the specimen over the duration of the test. The strain pulses in the bars are converted to the engineering stress-strain curve of the specimen using the so-called Hopkinson bar equations derived below.

The Hopkinson bar equations were derived by Kolsky [102] assuming:

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

Based on these assumptions, one-dimensional theory for elastic wave propagation may be used to describe the displacement of the bars. The development makes use of the fact that stress is not instantaneously applied to every particle in the bar, but travels as a wave with a finite velocity. The velocity for elastic waves, c , may be found by solving the one dimensional wave equation for a longitudinal disturbance and is given here as,

$$c = (E/\rho)^{1/2}$$

in which E is Young's modulus and ρ is the density of the bars. By applying Newton's second law of motion to the problem of wave propagation, the elastic stress in the bar, S_b may also be shown to be,

$$S_b = \rho c \Delta v_p$$

in which Δv_p is the change in particle velocity due to the stress wave. It may also be shown that a one dimensional longitudinal wave, when striking a free surface, will be

reflected as a pulse of identical magnitude but opposite sense. This reflection occurs such that the stress free condition of the free surface is maintained, as noted earlier.

The strain rate, $\dot{e}$, is defined as the change in engineering strain with time,

$$\dot{e} = \Delta e / \Delta t$$

Engineering strain is defined as

$$e = \Delta l / l_0$$

in which Δl is the change in length of the specimen, given by the relative displacement of the ends of the incident and transmitter bar, u_1 and u_2 , respectively, and l_0 is the initial specimen gauge length.

Thus the engineering strain rate of the specimen may be expressed as,

$$\dot{e} = (\dot{u}_1 - \dot{u}_2) / L_s \quad (4.1)$$

From elementary wave propagation theory [107], the velocity of the bar ends may be shown to be

$$\dot{u}_1 = c (e_i - e_R)$$

and

$$\dot{u}_2 = c (e_T)$$

in which e_I , e_R and e_T are the strains within the incident, reflected and transmitted pulses, respectively.

Substituting these results into equation 4.1, one obtains

$$\dot{\epsilon} = (c/L_o)(e_I - e_R - e_T) \quad (4.2)$$

The average stress, S_s , in the specimen may be determined as the average of the forces acting at the ends of the bars divided by the area, A_s , of the specimen.

$$S_s = \frac{(P_1 + P_2)/2}{A_s} \quad (4.3)$$

in which P_1 and P_2 are the forces acting at the incident and transmitter bar ends of the specimen, respectively.

The force at the ends of the bars may be determined using

$$P = S_b A_b$$

in which S_b and A_b are the stress and area of the bar, respectively. Applying Hooke's law one obtains:

$$P = (E_b e_b) A_b$$

such that,

$$P_1 = E_b (e_l - e_R) (A_b)$$

$$P_2 = E_b (e_T) (A_b)$$

Substituting this result into equation 4.3 yields

$$S_s = \frac{E_b A_b}{2A_s} (e_l - e_R + e_T) \quad (4.4)$$

By assuming uniform deformation, the strain at the specimen ends of the bars may be

assumed to be compatible and as such

$$e_T = e_I + e_R \quad (4.5)$$

This allows the further simplification of equations 4.2 and 4.4 to

$$\dot{\epsilon} = 2(c/L_0) e_R \quad (4.6)$$

and

$$S_s = (E_b A_b / A_s) e_T \quad (4.7)$$

Thus the specimen stress and strain rate histories during the experiment may be calculated directly from the recorded transmitted and reflected pulses, respectively.

Given that the length of the striker (L_{ST}) and the velocity at which it is travelling prior to impact (V_{ST}) are known, the magnitude and duration of the stress pulse in the bar may be determined *a priori*. Since the change in particle velocity which occurs in a symmetric

impact is known to be one half of the impact velocity [107],

$$S_b = \frac{\rho_b c_o V_{st}}{2} \quad (4.8)$$

and

$$\Delta t = \frac{2 L_{st}}{c_o} \quad (4.9)$$

are the stress and period of the incident pulse generated by the striker.

Davies [106] demonstrated the over-simplification by Kolsky of neglecting wave dispersion and established the accuracy of the one dimensional assumption. Analysis of error due to dispersion in the Hopkinson bars identifies a difference between the surface and the centre line strain. It was shown that this error is less than 5% for an apparatus with bar radius to pulse length ratios less than 0.1. Thus the assumptions which were made regarding one dimensional wave propagation are generally good for bars with long stress pulses and small radii. While impact typically generates numerous forms of waves they tend to be quickly damped and may be neglected for bar lengths greater than ten bar diameters according to Follansbee [108].

Uniform deformation, which is also assumed in Hopkinson bar equation development, is altered by longitudinal and radial inertia effects as well as wave transient time through the specimen. Davies and Hunter [109] developed an inequality to express the validity of the assumption of uniform deformation in the specimen. They showed that inertia effects would not greatly influence the results of the experiment after π reverberations of the stress wave in the specimen. They argued that during the time required for stress to reach equilibrium, deformation was not uniform and the Hopkinson bar equations were not valid. The time to uniform deformation was defined by:

$$t_E \geq (\pi L_s \rho) / (d\sigma/d\varepsilon) \quad (4.10)$$

in which L_s is the specimen length.

Davies and Hunter [109] have suggested a correction to the stress calculation which accounts for the inertial variation after uniform deformation is achieved. The expression,

$$S = S_{\text{measured}} + \rho_s [(L_s^2/6 - v(D_s/8))] d^2\varepsilon/dt^2 \quad (4.11)$$

in which D_s is the specimen diameter, suggests that the specimen should be designed such that the length to diameter ratio is

$$L_s/D_s \approx (3v/4)^{1/2}$$

This ratio would minimize the error due to inertia; unfortunately this constraint was found to be impractical for the specimens used in this investigation due to the test geometry and the desire to be able to fracture specimens. Equation 4.11 also indicates that the error in the stress measurement is minimized through the use of constant strain rate tests such as those performed in this investigation.

As mentioned earlier, the strain determined from these equations represents the engineering strain across the ends of the bars or the gauge length of the specimen. Following localization in a tensile test, force-displacement data can no longer be converted to true stress-strain data. The reduction in area of the specimen as a function of the axial strain must be determined to calculate the true diametral strain and true stress experienced by the specimen.

4.3 Interrupted Tests for the Characterization of Damage Evolution

During conventional Hopkinson bar testing in which the specimen does not fail, the transmitted and reflected waves would continue to propagate along the bars and their successive reflections would reload the specimen numerous times. This is not normally a problem as the goal of standard Hopkinson bar testing is to fracture the specimen on the first pulse to obtain the complete stress-strain response of the material. For damage accumulation studies it is desired that the specimen be loaded only once without fracture by a single well-characterized stress wave.

A wave trapping mechanism was developed for the Carleton tensile Hopkinson bar by Pelletier [110], utilizing the momentum trapping techniques due to Nemat-Nasser [1]. These methods utilize the simple fact that tensile waves can not travel across a free interface.

On the incident bar side of the apparatus, a second bar long enough to trap the entire incident pulse of the longest striker rests on two bushings collinear with the incident bar as shown in Figure 4.2 (from Pelletier[110]). At the start of the test a precision gap is set between the incident bar end cap and the incident bar momentum trap. This gap is sized based on the anticipated striker velocity such that the gap will just close once the incident pulse has completely entered and displaced the incident bar. The reflected pulse travels as a compressive wave to the trap where it passes through the now closed gap. This wave reflects as a tensile pulse in the momentum trap rod but is now unable to travel back through the free interface into the incident bar. Thus the reflected wave is caught in the momentum trap and unable to reload the specimen.

Capturing the transmitted pulse is less critical as the magnitude tends to be substantially lower than the reflected wave for the bar and specimen geometries used in this research. The transmitted pulse is trapped by a momentum trap cylinder similar in form to the striker located at the transmitter bar end cap, as shown in Figure 4.3 (from Pelletier). The wave first reflects off the end of the transmitter bar as a compressive pulse and crosses the free interface between the bar and the cylinder. Once in the momentum trap cylinder,

the wave is reflected as a tensile pulse and is trapped in the cylinder since it is unable to travel back into the transmitter bar.

The use of various striker lengths allowed constant strain rate and stress level tests to be performed and interrupted at different strains, allowing the damage evolution to be investigated and mapped as a function of strain. The strikers used in this study and the duration of the pulses they produce are presented in Table 4.1.

4.4 TSHB Specimen Selection

Three specimen configurations were utilized for the testing discussed in this thesis to provide a range of notch severity and triaxiality. These were the uniaxial tensile specimen, the E-notch and the C-notch. The geometry and physical dimensions of these specimens are given in Figure 4.4.

Specimen design for Carleton's TSHB was covered in detail by Pelletier [110] and the notch profiles used in this work were similar to those adapted by Pelletier based on the designs used by Hancock and Mackenzie [111]. Hancock and Mackenzie performed quasi-static tensile tests on several specimens of varying notch severity to determine the effect of stress state on ductility. Triaxiality, $\Sigma_{\text{hyd}}/\bar{\sigma}$, is defined as the ratio of hydrostatic stress to flow stress and serves as an indicator of the severity of loading. For the notched specimens, triaxiality was determined using Bridgman's analysis [83] for each geometry. Rajendran and Bless [112] have confirmed the validity of Bridgman's analysis for high

strain rates which states,

$$\Sigma_{hyd} / \bar{\sigma} = 1/3 + \ln(r_s/2R_n + 1) \quad (4.12)$$

where r_s is the specimen radius = $D_s/2$ and R_n is the notch radius. The triaxiality of the tensile specimens used in this study, based on Bridgman's equation (4.12) are given in Table 4.2.

The notch serves not only to elevate hydrostatic stress but to localize strain as well. While the Hopkinson equations are based on the gauge length of the entire specimen, the actual region which is severely strained is much smaller. This results in strain rates at the centre of the notches which are higher than that based on specimen length. While typical strain rates of 1000 s^{-1} have been achieved within uniaxial specimens, values as high as 10^4 s^{-1} have been realized at the minimum section of the C-notch [113].

The specimen ends have one left-hand thread and one right-hand thread. This allows for easy installation as the bars do not have to rotate to thread the specimen which avoids tangling the strain gauge wires. It should be noted that the loading of the specimen will depend on the tightness of fit. Any lag or slack in the specimen will alter the strain response. The fit will depend on whether or not the specimen is bottoming at the base of the threads or on its shoulders. The E-notch specimens in this investigation had large

shoulders. It was shown by Pelletier [110] that this shoulder design would not greatly influence the results.

4.5 Carleton's TSHB Apparatus

The TSHB facility at Carleton, pictured in Figures 4.2 through 4.7, has been well described by preceding graduate students Pelletier [110] and Clarke [114]. This section will serve as an overview of the apparatus to familiarize the reader with its operation. Readers who wish a more detailed description of specific components and their design methodologies are directed to the aforementioned theses.

Carleton's facility is capable of nominal strain rates of 10^3 s^{-1} for durations up to $160 \mu\text{s}$. Values of strain rate as high as 10^4 s^{-1} have been achieved in the centre of some notched specimens. The maximum uniaxial stress in the bar is defined by the yield stress of annealed AISI 4340 with a safety margin of 25%. Hopkinson bar tests on the bar material itself indicate a somewhat higher dynamic yield strength. The maximum striker velocity is now limited to 34 m/s rather than the 27.3 m/s reported by Clarke [114]. This value is high enough that the actual limit has been found to be imposed by the lateral inertia of the bar rather than yielding.

The schematic shown in Figure 4.5 (from Pelletier[110]) illustrates all of the major components of the TSHB. The gas gun illustrated in Figure 4.6 (from Pelletier[110]) consists of two annular thick walled pressure vessels which sit concentrically about the

incident bar. The incident bar itself rests on two brass bushings with the incident bar end cap resting in a teflon cradle. The incident and transmitter bars are both machined from fully annealed AISI 4340 steel characterized by high yield stress and toughness. The aluminum pusher, labelled in Figure 4.6, runs on PVC bushings along the incident bar and serves to accelerate the hollow cylindrical striker which runs on nylon or teflon bushings. The striker rests in the pusher with a sliding fit. When the pusher/striker assembly is loaded into the rear of the gas gun bore, the O-rings seal the openings between the inner and outer cylinders. Nitrogen is then used to pressurize the outer cylinder of the gas gun while the inner cylinder remains at ambient pressure. When the data collection system has been correctly set and the bar is ready to fire, nitrogen is bled in behind the pusher moving it forward. As the pusher O-rings slide past the holes separating the inner and outer cylinder; the gas in the main cylinder flows into the central cylinder accelerating the pusher and striker out of the gas gun. The pusher is then stripped off of the striker by a fixture placed at the exit of the gas gun.

The striker travels along the bar and the leading edge passes a pair of infrared sensors used by a velocity detector to measure impact speed. The striker then impacts the integrally machined end cap of the incident bar. Upon impact a compressive stress wave is generated which reflects off the end of the incident bar as a tensile wave and progresses towards the specimen. The end of the incident bar is displaced as the wave enters and this closes the gap to the momentum trap. When the reflected pulse returns from the sample it is trapped in the momentum bar. The momentum bar is accelerated

to the right where it is eventually stopped by the heavy momentum brake shown in Figure 4.7 (from Pelletier). The brake consists of an aluminum pin pressed between two rubber blocks. This component dissipates the energy within the momentum trapping bar.

The stress wave is measured by a piezoresistive strain gauge located 80 cm from the end of the incident bar, shown in Figure 4.5. The incident wave then impinges on the test specimen. The reflected wave is measured by the incident bar strain gauge and the transmitted wave is recorded by another strain gauge located at the same distance from the end of the transmitter bar.

4.6 Firing Control System

As part of the current research, a modification was made to the firing control system in order to improve control of the striker velocity. These changes were made necessary by the momentum trapping scheme in which the size of the gap between the incident bar and the momentum trapping rod is directly a function of impact velocity.

In work by Pelletier [110] and Clarke [114], fire control was achieved by a manual valve on the nitrogen feed line located well away from the gas gun. In the current work, a solenoid valve was installed at the gas gun manifold, allowing the bleed line to be fully charged prior to firing. In this manner a larger initial impulse is imparted to the pusher thereby reducing the variability in firing velocity due to friction of the pusher. This modification makes the tests more repeatable as friction of the system was variable over

a series of tests.

The solenoid valve used to fire the gas gun was found to generate sufficient electromagnetic noise as to interfere with the triggering mechanism of the oscilloscope. To combat this problem, a solid state relay was added between the solenoid and its control switch. This modification eliminated the noise problem.

Two electronic pressure gauges were also added. One gauge is used to measure the bleed line pressure behind the pusher. The second pressure gauge replaced a mechanical gauge which is used to monitor the pressure in the cylinder. These changes increased both the accuracy as well as the precision with which the striker velocity could be controlled.

4.7 Hopkinson Bar Instrumentation

As mentioned previously, strain gauges are mounted on the transmitter bar and incident bars, located 80 cm from the specimen end. Two gauges are mounted at each location, installed on opposite sides of the bar. The gauges are wired to opposite arms of a wheatstone bridge which doubles the voltage output of the bridge resulting from axial elongation as well as cancelling any bending which may occur during the test. Kulite semiconductor strain gauges(VPG-1000-060) were used, which are piezoresistive elements characterized by high gauge factors on the order of 155. The two wheatstone bridges are both housed separately to avoid electromagnetic interference or cross-talk. Each bridge has its own Hewlett Packard 6213A power supply adjusted to provide 1400 ± 0.1 mV

excitation to the bridges. The bridges are balanced using a potentiometer located on one of the bridge arms at the beginning of every test and the excitation voltage was also confirmed. The output of these bridges are fed into a two channel Nicolet Pro30 Digital Oscilloscope. Data is recorded using the oscilloscope with 12 bit resolution and a sampling period of 200 ns. This system has a temporal accuracy of 1 ns. Over the 2 ms duration of a test, 10,000 points are stored.

The measured voltage versus time data was converted to strain values using the full non-linear form of the wheatstone bridge equation,

$$(\Delta E_o / a V_e) = \frac{(GF_2 \epsilon + GF_1 \epsilon + GF_2 GF_1 \epsilon^2)}{(1 + a)^2 + (1 + a)(GF_2 + a GF_1) \epsilon + GF_1 GF_2 \epsilon^2} \quad (4.13)$$

where V_E is the bridge excitation voltage and ΔE_o is the measured change in voltage across the bridge. Factor a is the bridge ratio which is defined as,

$$a = R_F / R_{g1}$$

given the fixed resistance, R_F , and the resistance of the first strain gauge of the bridge, R_{g1} .

4.8 Calibration of Elastic Wave Measurement in the Bars

To ensure accurate measurement of the stress waves propagating within the bars, the piezoresistive strain gauges must be calibrated. Static calibration of the strain gauges was performed on a regular basis during testing using the procedure described by Pelletier [110].

Concerns have arisen as to the validity of using this form of quasi-static calibration in the measurement of waves generated by dynamic loading (Safford [115]). While no attempt was made to dynamically calibrate the strain gauges, the validity of the calibration was checked under the dynamic test conditions. The strain in the incident bar was recorded using the gauges and then compared to the strain predicted based on the measured striker velocity as suggested in [107]. The measured incident pulse magnitude from the strain gauges were typically within 1% of the magnitude calculated based on the measured striker impact velocity.

4.8.1 Relating Strain Rate to Striker Velocity

The effect of triaxiality on the rate of damage accumulation is of particular interest in this study. While previous studies have established the low strain rate sensitivity of brass [116], an effort was made to test all of the various notch geometries at approximately the same strain rate. Controlling the strain rate would limit any rate dependant material or inertial effects which could alter the mechanical properties and affect void growth rate.

The strain rate within each specimen varies over its geometry. It was decided to match the diametral strain rate between specimens as this is a bulk measure of the plastic deformation that is comparable between the notches. To this end, finite element studies were undertaken to relate the axial and diametral strain rates to the striker velocity for the C and E notches. Regression fits to the results of these simulations yielded the following relations,

$$\text{E-Notch:} \quad \dot{\epsilon}=329.5V_{ST} - 752.4 \quad [s^{-1}] \quad (4.14)$$

$$\dot{e}=148.4V_{ST} - 56.73 \quad [s^{-1}] \quad (4.15)$$

$$\text{C-Notch} \quad \dot{\epsilon}=507.8V_{ST} - 1730.8 \quad [s^{-1}] \quad (4.16)$$

$$\dot{e}=468.2V_{ST} - 328.5 \quad [s^{-1}] \quad (4.17)$$

in which ϵ is the diametral (logarithmic) strain across the minimum section, e is the axial (engineering) strain across the notch opening, and V_{ST} , the striker velocity, is given in m/s.

The dependence of engineering strain rate on striker velocity can also be derived from one dimensional wave theory and the Hopkinson Bar equations. The resulting equations for the notched specimens used in this study are:

$$\text{E-Notch:} \quad \dot{e}=143V_{ST} \quad [s^{-1}] \quad (4.18)$$

$$\text{C-Notch: } \dot{\epsilon}=437V_{ST} \quad [s^{-1}] \quad (4.19)$$

again assuming V_{ST} to be measured in m/s.

These numerical and theoretical relations are plotted as Figures 4.8 and 4.9 for the E and C notches respectively. Note the close agreement between the theoretical and numerical values of engineering strain rate.

4.8.2 Selection of Striker Velocities

The diametral strain at failure for the C-notch was determined to be 0.37 m/m from tests performed to fracture. A slightly lower value of strain, 0.35 m/m, was targeted as the maximum value of strain for interrupted testing of the C-notch. The strain rate which would generate this strain in the C-notch over the 119 μ s pulse duration associated with the 12" striker was found to be

$$\dot{\epsilon} = \frac{\Delta e}{\Delta t} = \frac{0.35}{119 \times 10^{-6}} = 2940 \quad [s^{-1}]$$

This strain rate was then selected as the nominal rate for all of the specimens tested. Equations 4.14 and 4.16 were then used to determine the striker velocities required to match the diametral strain rates between the two notch geometries. These velocities were found to be $V_C=9.2$ m/s, and, $V_E=11.2$ m/s to obtain a strain rate of $\dot{\epsilon}=2940 \text{ s}^{-1}$ in the C

and E-notch specimens respectively. The C-notch specimens were tested using the 4", 8" and 12" strikers as failure would occur using the 16" striker. For the E-notch specimens, tests were performed using the 8", 12" and 16" strikers. Use of the 16" striker for the E-notch was possible due to the higher strain to failure (ductility) of this less sharply notched geometry.

4.9 Data Reduction Methods

The strain gauge data recorded by the oscilloscope is converted from a binary to an ASCII file using Viewpoint 4.0 analysis software. The voltages are then converted to strain measurements using the full non-linear bridge equations given earlier as equation 4.13 and the gauge factors determined during calibration. The Hopkinson bar equations are then applied to the strain data to determine the engineering stress and strain time histories of the test specimens. The true stress-strain histories are also calculated based on the relations between axial and diametral strain, described in the following section.

4.9.1 Conversion of Engineering Strain to True Strain

The useful limit of gauge length-based axial strain measurement in uniaxial tensile test is the onset of localized necking. At this point in the deformation, force-displacement data can no longer be converted directly to true stress-strain data. In order to determine this data, a relation is required between axial elongation and reduction in area or between the engineering and true strain. Under high rate testing conditions, this expression may

be determined using high speed photography of a test for a given notch geometry and material but such testing is very expensive. Numerical simulations may also be used, however, the reduction in area will be sensitive to the constitutive data. An alternative approach, made possible through the use of momentum trapping, is to directly measure axial and diametral strains from recovered interrupted specimens. Care must be taken in defining the axial gauge length as this will alter these results.

From numerical simulation of the experiments, the conversion of axial elongation to the reduction of cross-sectional area at the minimum section was found. The relation expressed in terms of the true diametral or log strain as a function of engineering strain was found to be fit well using a cubic equation as suggested by Pelletier [110]. For the E-notch geometry the relation was

$$\varepsilon = 0.0007355 + 2.0814e - 3.9212e^2 + 12.2071e^3 \quad (R^2=0.999) \quad (4.20)$$

while for the C-notch, the relation was

$$\varepsilon = 0.004289 + 0.9418e - 1.0470e^2 + 1.2060e^3 \quad (R^2=0.997) \quad (4.21)$$

The R^2 terms are the regression coefficients from fitting a cubic to the numerical data. The gauge length for the engineering strain, e , was taken as the notch opening or twice the notch radius.

These relations, as well as values measured from interrupted specimens from pulse trapping experiments, are given as Figures 4.10 and 4.12 for the C and E notches respectively.

It should be noted that there is a significant amount of error associated with the measurement of the experimental strain values. Due to the narrow gauge length of the C-notch specimen, the diameter had to be measured using a shadow graph device. The specimen geometry made it difficult to make this measurement accurately. Deformation of the notch shoulders during specimen installation also lead to errors in the measurement of final gauge length after testing.

Having determined the relation for the diametral strain in terms of the engineering strain the method of stress conversion was derived as follows.

True stress is defined as

$$\sigma = F/A \quad (4.22)$$

where F is the applied force and A is the instantaneous cross-sectional area of the specimen.

Engineering stress is given as

$$S = F/A_0 \quad (4.23)$$

where A_0 is the original cross-sectional area.

Combining equation 4.22 and 4.23 yields

$$\begin{aligned}\sigma &= (F/A)(A_0/A_0) \\ &= (F/A_0)(A_0/A) \\ &= S(A_0/A)\end{aligned}\tag{4.24}$$

Recalling that the diametral strain is defined as

$$\epsilon \equiv \ln(A_0/A)$$

then,

$$(A_0/A) = \exp(\epsilon)\tag{4.25}$$

combining equation 4.24 with 4.25 yields

$$\sigma = S \exp(\epsilon)\tag{4.26}$$

Thus equation 4.26 provides the conversion from engineering to true stress. It also shows the conversion to be exponentially sensitive to error in the conversion of strain.

CHAPTER 5

PLATE IMPACT

The planar plate impact or flyer plate test allows dynamic material properties to be determined for strain rates in excess of 10^5s^{-1} . These strain rates are only achievable under conditions of uniaxial strain for short durations. The plate tests presented in this work were performed by Dr. H. Nahme at the Ernst-Mach-Institute in Freiburg, Germany under a collaborative arrangement with Carleton University. The particle velocity-time history and the dynamic properties for the brass used in this thesis have been determined and published by Nahme [117]. Numerical simulations showing reasonable agreement with these results have also been reported by Worswick [118, 119]. Metallographic studies were performed as part of the current research in order to examine damage development at strain rates in excess of those obtained in the TSHB and under a different stress state.

5.1 Theory of Plate Impact Experiments

Plate impact was first studied by Hopkinson [101]. It had long been known that the impact of a high velocity projectile onto a plate may not lead to penetration or produce any significant damage on the impacted surface but could cause significant damage to the rear of the plate. Tests have shown that, given the proper impact velocity, a portion of

the back surface could separate and fly off. This phenomenon was termed scabbing or spallation. It was Hopkinson who first identified that spallation occurred as a result of a tensile stress in the impacted plate created by the interaction of a propagating stress wave with its rarefaction.

Excellent reviews exist on the topic of spall and dynamic fracture [120, 121, 122, 123], of particular note is a treatise by Meyers and Aime [124]. A large amount of shock data has been gathered from plate impact testing and reported by McQueen and Marsh [125].

Today, planar plate impact tests are the primary method for determining Hugoniot data, that is, the equation of state of the material, as well as gathering information on dynamic tensile fracture by spall. The standard planar plate impact test has been described by Karnes [121]. In general, a thin flyer plate is accelerated and impacts a thicker stationary target. To assure that only uniaxial waves are generated at the plate centre, it is important to minimize the tilt between the two plates. This is especially important during lower velocity tests [126].

The simplified wave propagation occurring during a planar plate impact test is presented in Figure 5.1. The flyer plate, moving at some initial velocity, strikes the target (Figure 5.1(a) generating strong compressive waves which propagate into each plate at some velocity $\dot{U}_s$, (Figure 5.1(b)). The wave in the thinner flyer plate reaches its free

surface first and reflects as a tensile wave of equal magnitude which unloads the flyer plate as seen from the net stress in Figure 5.1(c). Once this release wave has travelled through the flyer plate, Figure 5.1(d), a tensile release wave is generated in the thicker target plate. Thus a pulse equal to twice the flyer plate thickness propagates through the target plate Figure 5.1(e). The leading edge of this pulse reaches the free-surface of the target plate and reflects as a tensile release wave to maintain the stress free boundary condition at the free surface, as shown in Figure 5.1(f). The net stress in the target is a compressive stress which appears to be rapidly decreasing in size as the two release waves propagate towards each other. At a distance of one half of the pulse length (which is equal to the flyer plate thickness) the net stress in the target drops to zero (Figure 5.1(g)) as the two release waves meet. As the release waves continue to propagate, a net tensile stress appears to "grow" from this point located at a distance equal to the flyer thickness from the rear surface of the target (Figure 5.1(h)). Note that this point is subjected to the tensile pulse for the longest duration. The net tensile stress pulse length eventually equals twice the flyer plate thickness (Figure 5.1(i)). If no fracture or damage has occurred, the pulses will then travel to and reflect off of the free surfaces as a compressive wave and will continue to propagate in the manner outlined above within the target plate. If the magnitude of the incident compressive pulse is sufficiently large, its reflected tensile pulse will overwhelm the material strength and fracture will occur in the target plate at a distance from its free surface equal to the width of the flyer plate.

These tests provide a state of uniaxial strain for a short amount of time as the material

is unable to deform other than axially due to the plate geometry and radial inertia. This time is limited for real loading conditions, since a portion of the wave generated on impact is not axial but propagates from the plate edge as a spherical unloading wave (Figure 5.2). As such the state of uniaxial strain at the plate centre persists only until the radial release waves from the outer edge of the plate arrive. The geometry of the plates is designed such that the uniaxial wave will reverberate several times at the centre of the plate prior to the edge relief waves altering the strain state. Typically the target is supported by a spall ring which serves to delay and suppress the tensile radial release waves in a manner similar to the momentum trapping techniques described earlier for the Hopkinson bar. For damage studies involving plate impact, it is assumed that the majority of the damage takes place early in the deformation process.

Another advantage of plate impact testing in damage studies is that given the flyer plate velocity and the shock velocity of the specimen material, the peak tensile stress and its duration are known *a priori* for a given test geometry. Symmetric impact is defined as the impact between two materials of similar impedance (ρc). In this case it can be shown, using simple one dimensional wave theory, that the magnitude of the stress pulse is,

$$\sigma = \rho \dot{U}_s \Delta \dot{u}_p \quad (5.1)$$

and this stress is maintained for a duration of,

$$\Delta t = 2d_f / \dot{U}_s \quad (5.2)$$

in which ρ and $\dot{U}_s$ are the density and wave velocity and d_f is the flyer plate thickness.

For the case of symmetric impact, in which the rear surface of the plates are free, the particle velocity, $\Delta \dot{u}_p$, may be defined as one half of the flyer plate velocity, V_o , such that,

$$\Delta \dot{u}_p = V_o / 2 \quad (5.3)$$

so the initial compressive pulse is,

$$\sigma = \rho \dot{U}_s V_o / 2 \quad (5.4)$$

which reflects from the rear surface of the target as a tensile wave of the same magnitude.

The impact of the flyer and the target plate may actually produce more than one wave type depending on the magnitude of the generated stress. If the impact velocity is low,

an elastic wave will propagate through the material at a distinct velocity, c_E , defined by the speed of sound in the plate material:

$$c_E = (E/\rho)^{1/2} \quad (5.5)$$

If the plates are impacted at moderate velocities, elastic and plastic waves are created and travel through the material at two distinct velocities. The elastic wave remains at c_E while the plastic wave moves more slowly with a velocity defined by the constitutive response of the material. The velocity of plastic wave propagation, c_p , has been derived by von Karman [127] as

$$c_p = [(d\sigma/d\varepsilon)/(\rho)]^{1/2} \quad (5.6)$$

where $d\sigma/d\varepsilon$ is the instantaneous slope of the stress strain curve.

If the material is impacted at still higher velocities, a shock wave is created and propagates through the material as a single disturbance with a unique velocity defined by the equation of state or Hugoniot of the material.

The Hugoniot of a solid may be expressed in terms of its stress and strain response as shown in Figure 5.3 after Rolsten [128]. It should be noted that the Hugoniot is not a load path but a locus of equilibrium states obtained under uniaxial strain shock wave

conditions. The curve can be represented by a relation between any two of the particle velocity, shock velocity, pressure, volume, stress or strain and is typically termed the equation of state for the metal. For values of stress below the Hugoniot elastic limit, the material responds elastically and only a single wave is propagated at velocity c_E . For values of stress greater than this limit but less than the point marked as C in Figure 5.3, both elastic and plastic waves propagate. Note that the plastic shock velocity in this case is defined by the slope of a Rayleigh line for the given stress and strain values. Beyond point C, elastic and plastic shocks travel at the same velocity and a true shock wave will result.

The free-surface velocity of the target plate, $\dot{u}_{fs}$, is measured to determine the mechanical response of the material. The free surface velocity will equal the sum of the particle velocity and the velocity of the reflected wave due to the requirement for velocity compatibility at the surface.

$$\dot{u}_{fs} = \dot{u}_p + \dot{u}_r \quad (5.7)$$

It has been shown by Walsh [129] that the reflected wave velocity, $\dot{u}_r$, will approximately equal the particle velocity

$$\dot{u}_r \approx \dot{u}_p$$

such that

$$\dot{u}_{fs} = \dot{u}_p + \dot{u}_r = \dot{u}_p + \dot{u}_p = 2\dot{u}_p \quad (5.8)$$

or,

$$\dot{u}_p = \dot{u}_{fs} / 2$$

Given the free-surface or particle velocity history, several properties may be inferred. The Hugoniot Elastic Limit (HEL) shown in Figure 5.3 represents the highest stress which will result in only an elastic wave propagating. A typical free surface velocity time history is shown in Figure 5.4 for the case where both elastic and plastic waves propagate. The fast moving elastic wave arrives first and increases the rear surface velocity to the velocity associated with the HEL. The magnitude of this wave may be defined as,

$$\sigma_{HEL} = \rho c_L \dot{u}_{HEL} / 2 \quad (5.9)$$

The uniaxial dynamic yield stress, Y , may also be determined from this data by assuming a strain hardening material and von Mises or Tresca yield condition, as has been shown by Zukas [107] to be

$$Y = [(2G)/(K+4G/3)] \sigma_{HEL} \quad (5.10)$$

in which G and K are the shear and bulk moduli, respectively.

The signal persists at the HEL until the slower but stronger plastic wave arrives at E' and accelerates the free surface to the initial flyer plate velocity. This higher level is maintained until the elastic tensile release wave from the impact surface arrives and lowers the velocity by an amount almost equal to u_{HEL} . This is typically followed by a high strain rate tensile release of the plastic wave at R' to NS in cases where spall does not occur. If the tensile stress is large enough and applied for a long enough duration, the target plate will spall creating the spall cusp or ringing in spall signal shown starting with the pull back at S. This analysis of the spall signal will be discussed with more depth in Section 5.3. The slopes and peaks of the velocity history tend to be defined by the equation of state while the constitutive relation for the material defines the elastic limit. The failure process defines the spall signal.

5.2 Apparatus

The Ernst-Mach-Institute plate impact facility utilizes a 70 mm single stage light gas gun to accelerate flyer plates mounted on plastic sabots. A schematic of the test section as it was set-up for this work is shown in Figure 5.5. For the data presented in this work, 3 mm flyer plates, 58 mm in diameter were accelerated to impact 8 mm thick target plates which were 50 mm in diameter. The plastic sabot is used to assure that a planar impact occurs. The void shown in the sabot behind the flyer plate in Figure 5.5 provides the free surface for wave reflection in the flyer. A precision adjustable target sample

holder aids in minimizing the tilt between the plates. It is very important that the impact be planar and tilts of less than 1 mrad are typically recorded for the Ernst-Mach test facility [117]. Impact velocities between 53 and 256 m/s were investigated. The impact velocities of the plates were measured from self shorting trigger pins inside the gunbarrel and near the target surface. These were compared with the velocity of the free surface in light of equation 5.8,

$$\dot{u}_{fs} = 2 \dot{u}_p$$

and recalling that for the configuration used here,

$$\dot{u}_p = V_o / 2$$

so,

$$\dot{u}_{fs} = V_o \tag{5.11}$$

and found to match quite closely. On exiting the bore of the gas gun, the flyer plate triggers a Velocity Interferometer System for Any Reflector (VISAR) data collection system which records the free surface velocity. The flyer then impacts the target plate and generates an elastic and plastic shock in the material. The rarefaction of these waves off the rear surface, causes a region of intense tensile stress at a distance of the flyer plate thickness from the rear surface, as discussed earlier. Specimens are then "softly"

recovered such that their post-loading microstructure may be investigated.

The particle velocity induced by the waves interacting with the free surface of the target plate is determined by the VISAR system developed by Barker and Hollenbach [130]. The system measures surface velocity based on the doppler shift of an incident beam reflected off the free-surface. The initial beam, is split and part of it is reflected off the free-surface while the remainder is reflected from a mirror at an equal distance. The two beams are recombined and the resulting interference may be used to determine the velocity and acceleration or deceleration of the surface. The VISAR data is digitized and stored with a sample interval of 5ns using a Tektronics DSA602 transient recorder. The raw data is transformed to particle velocity data for the rear face of the specimen by in-house software at the Ernst-Mach-Institute.

5.3 Spallation of Plates

Spall is a form of dynamic tensile failure which occurs under high strain rate loading conditions as a result of a spherical stress tensor. Johnson [84] has defined spall as the planar separation of material parallel to the wave front as a result of a tensile stress occurring normal to its plane. The conditions which bring about spall are usually defined in terms of a stress and the duration. As such, spall studies typically define the impact velocity required for a given flyer plate thickness which result in spallation.

The spall cusp or pull back shown (point S in Figure 5.4) results from the arrival at the

free surface of the target of the stress release wave generated when the damage develops in the material [131]. As mentioned earlier, the VISAR data allows the tensile stress which brings about spall to be determined. The process may then be summarized as follows, a compressive pulse generated by impact interacts with its own tensile rarefaction wave to cause spall. The sudden relaxation of the tensile stress at the spall layer creates its own stress wave which propagates back to the free surface.

In their review of spallation Seaman [75] describes three principal characteristics of damage. The first was that a range of damage occurs in the test specimens. The undeformed material does not instantaneously separate. The second point was that this damage evolves as a function of time and applied stress. For this reason the dynamic fracture process may not be described by a single stress or strain but must involve some time integral of these quantities. Finally it was noted that damage weakens the structure and decrease the stiffness which alters the wave propagation characteristics during the test.

Traditionally, threshold values have been set to define spall failure. Fracture is said to occur at some critical value of stress for a given pulse duration. Tuler and Butcher [132] recognized the fact that damage alters the mechanical properties of the material. To properly model fracture, the nucleation and growth of damage must be included. This creates a time dependant failure in which spall is said to occur when some critical value of damage is reached. Spallation may occur by the nucleation and growth of ductile

voids or brittle cracks. When the process occurs by ductile void growth, plastic deformation is localized around small voids nucleated at the spall plane. Davison [131] showed that even small amount of damage have a profound effect on the stress field in the spall plane.

Experimental evidence [84, 131] suggests that the nucleation of damage in highly triaxial stress states is controlled by the mean stress. This is different from the case of uniaxial stress where nucleation is controlled by plastic strain [4]. Curran [133] quantitatively evaluated the microstructural damage distribution of several materials under various test geometries. The statistical distributions of void area and aspect ratio were evaluated and this data was used to generate models for the nucleation and growth of damage. It is assumed that spallation in ductile materials is controlled by the localization of plastic deformation around voids. These voids grow and coalesce to form the spall plane. While coalescence is a result of local instabilities in the flow field around individual voids, it is a very rapid process. For this reason, coalescence is generally assumed to be an instantaneous process which occurs at some critical value of damage.

It should be noted that the target plate is subjected to a repeated compression-tension loading cycle. If complete separation of the spall plane does not occur on the first pulse the damage accumulated is subjected to a compressive pulse prior to the following tensile loading. This compressive reload may collapse voids and lower the apparent damage level. Davison [131] argued that the collapse of voids is an important consideration

which must be included to properly capture the dynamic fracture process.

CHAPTER 6

EXPERIMENTAL RESULTS

This chapter presents the results obtained from the current research. Two sections follow: Section 6.1 presents the results from the Hopkinson bar testing while Section 6.2 deals with the flyer plate data.

6.1 Hopkinson Bar Data

Interrupted split Hopkinson bar testing was performed on uniaxial, E-notch and C-notch specimens. A summary of these tests is presented in Tables 6.1 for the C-notch specimens, 6.2 and 6.3 for the E-notch specimens, and 6.4 for the Uniaxial specimens. Tables 6.2 and 6.3 provide the E-notch results for the UNS-C36000 brass and the MS-58 brass respectively.

6.1.1 Mechanical Properties of Annealed Brass at High Strain Rates

The experimental stress and strain rate data presented in this chapter were determined from the Hopkinson bar equations discussed in Chapter 4. The two equations describing the stress and strain rate of the specimen were given as:

$$(4.7) \quad S_s = \frac{E_b A_b}{A_s} e_T$$

$$(4.6) \quad \dot{e}_s = \frac{-2C_o}{L_s} e_R$$

where e_R and e_T are the measured reflected and transmitted waves respectively. An example of a typical incident and transmitter bar wave are presented in Figure 6.1. It is seen in the figure that the application of pulse trapping has essentially eliminated subsequent wave reflections after the initial loading.

The stress and strain data derived from equations 4.7 and 4.8 were then converted to true stress-strain data in the manner described in Section 4.9.1. An experimental fit was used to define the relation between the diametral or log strain and the engineering strain. This allowed the true stress to be defined as,

$$(4.26) \quad \sigma = S \exp(\epsilon)$$

Utilizing the relations given above, the true stress and strain can be found from the engineering values. Complete data sets for typical E-notch and uniaxial specimens tested with the 16 inch striker are shown in Figures 6.2 and 6.3 respectively. Similar data for the C-notch tested with the 12 inch striker is provided in Figure 6.4. The mechanical data presented includes the strain rate and stress time histories as well as the engineering and the true stress-strain response of the specimen. The C-notch test with the 12 inch striker

is shorter in duration than those with the 16 inch striker, as discussed in Chapter 4. It should be noted that the engineering strain rate is presented. The difference in the evolution of diametral strain with respect to engineering strain varies with specimen geometry. To match the diametral strain rate, and thus the diametral strain, between each of the specimens, the E-notch tests must be performed at a higher engineering strain rate than the C-notch. The uniaxial specimen must be tested at a much higher engineering strain rate again.

Figure 6.5 plots the true-stress(6.5(a)) and true-strain(6.5(b)) time histories for the E-notch data presented in Figure 6.2. The strain, determined by integrating the strain rate data, is quite smooth (Figure 6.5(b)) while the true stress history (Figure 6.5(a)) shows an oscillation after the initial rise of the wave. This results in an oscillation in the stress-strain data. This oscillation occurs due to two effects. The first is the oscillation of the loading wave travelling in the bar due to dispersion. This effect was shown by Clarke [113] to be small for the geometry of bar and strain rates used in this study. The second effect is the oscillation of the stress wave in the specimen itself. The stress requires a finite amount of time to come to equilibrium within the specimen. This suggests that the initial values of stress at lower strains may be suspect and limits the ability to define the dynamic yield stress from these tests.

Material Effects on Stress-Strain Data

Two different heat treatments were used in this investigation which provided two different particle morphologies of equal volume fraction and similar metallurgical properties as discussed in Chapter 3. For ease of referral, the 2 hour 850°C heat treatment will be referred to as 850HT. Similarly, the 2 hour 600°C anneal will be referred to as 600HT. Two types of brass were also investigated: free-cutting UNS-C36000 brass, as well as its German equivalent, MS-58. Specimens were machined from two stocks of material. The majority of the C-notches and uniaxial specimens were manufactured from 1.27 cm (0.5 inch), UNS-C36000, brass rod. The E-notches were manufactured from a 8.89 cm (3.5 inch) rod of UNS-C36000 brass as well as a 7.62 cm (3.0 inch) rod of MS-58 brass.

The stress-strain curves for tests of similar geometry or triaxiality levels are presented in Figures 6.6 for the uniaxial specimens, Figure 6.7 and 6.8 for the E-notch specimens machined from the UNS-C36000 and MS-58 brass respectively, and finally Figure 6.9 for the C-notch. Two plots are presented in each of these figures, one for the 850HT material and a second with the 600HT material. Note that in these plots, the specimens which were tested to fracture displayed an unloading characterized by a negative slope at large strains. The specimens which display elastic unloading were interrupted using momentum trapping. The majority of the uniaxial specimens tested were machined from the 1.27 cm (0.5 inch) rod. For comparison, a similar specimen was manufactured from the 8.89 cm (3.5 inch) rod and is indicated in both the 600HT and 850HT data. These specimens from the 8.89 cm (3.5 inch) rod were of some what lower strength, particularly

for the 850HT specimen. This result was not seen for the 600HT specimen.

The E-notch data for the UNS-C36000 brass is given as (Figure 6.7), while similar tests with the MS-58 brass are presented in Figure 6.8. The UNS-C36000 results in Figure 6.7 include specimens manufactured from both the 1.27 cm (0.5 inch) and the 8.89 cm (3.5 inch) rod. These plots indicate that the 1.27 cm (0.5 inch) rod is stronger than the 8.89 cm (3.5 inch) rod used for most of these specimens. This effect is most notable in the 850HT brass as compared to the 600HT brass.

Figure 6.9 presents the C-notch stress-strain response. The C-notches were manufactured primarily of 1.27 cm (0.5 inch) rod. C-notch specimens manufactured from the 8.89 cm (3.5 inch) rod were again plotted along with this data. A result similar to the uniaxial and E-notch specimens was obtained in that the 8.89 cm (3.5 inch) rod material exhibited a lower strength than its 1.27 cm (0.5 inch) counterpart for the 850HT and not in the 600HT data.

A series of stress-strain curves for these various E-Notches is presented in Figure 6.10(a) and (b) for the 850HT and 600HT respectively, which includes both the UNS-C36000 and MS-58 brass for comparison. The UNS-C36000 material is shown to have a considerably lower hardening rate than that of the MS-58. The tensile stress at a given strain is also lower.

The amount of scatter evident in the stress-strain curves may occur as a result of the differences in calibration of the strain gauges. The mechanical alignment of the tensile split Hopkinson bar apparatus itself may also play a role. Each striker was mounted on its own incident bar. These bars were changed to perform tests on specimens with various striker lengths. While the strain gauges on each bar were frequently recalibrated over the course of testing, errors in this process could result in the observed variation of measured stress and strain values. When frequently removing the incident bar to perform tests with various striker lengths it was quite difficult to keep the Hopkinson bars aligned. None the less, the 600HT and 850HT materials for each given notch compared well within the range of scatter in the data.

Triaxiality Effects on Stress-Strain Data

By comparing the stress-strain data for the uniaxial, E-notch and C-notch geometries, an appreciation for the effects of triaxiality can be obtained. The stress-strain responses of typical uniaxial, E-notch and C-notch specimens for the 8.89 cm (3.5 inch) and 1.27 cm (0.5 inch) rod materials are presented in Figures 6.11 and 6.12. Figure 6.11 provides this data for the 850HT material, while 6.12 provides the 600HT data. In each case the stress at a given strain increases with increasing triaxiality (notch severity). This elevation in stress results from the generation of transverse components of stress due to the constraint of the notch.

Note that the uniaxial specimens in Figures 6.11 and 6.12 were not tested to failure. For

the notched specimens, the sharper C-notch displays lower ductility, or strain to fracture, than the E-notch.

Ductility Measurements

Measured values of ductility for all of the failed specimens and are presented as diametral strain in Tables 6.1 to 6.4. The limitations of Carleton's split Hopkinson bar would not allow uniaxial specimens to be tested to failure. Ductility measurements are therefore limited to those for the E-notch and C-notch. The two C-notch specimens manufactured from the 8.89 cm (3.5 inch) rod of the UNS-C36000 material also did not fail. As such, the only data available for the ductility of the C-notches is for the 1.27 cm (0.5 inch) rod. From the ductility measurements and strain rates presented in Tables 6.1 through 6.4, no dependence of fracture strain on strain rate is noted for the range of strain rates used in this investigation.

The effect of triaxiality on the strain to failure is plotted in Figure 6.13. The ductility of the MS-58 E-notch specimens have also been included in 6.13 for comparison. By comparing the C-notch and E-notch results, the ductility is noted to increase as the triaxiality drops from the C-notch(1.026) to the E-notch(0.613). A very strong material effect is noted between the various E-notch results.

For the E-notches in which incipient or partial fracture was found during metallographic examination, the mean strain was calculated to be 0.641 m/m. The measured ductility

from the fractured E-notch specimen was 0.731 m/m. This comparison suggests a significant amount of deformation occurs after the formation of the central crack in the E-notch specimens. The largest strain obtained for a C-notch without fracture was 0.318 m/m as compared to the mean strain at fracture for the E-notch of 0.342m/m. This suggests that not only does fracture initiate earlier, in the deformation history, but that it propagates more rapidly with increasing triaxiality.

6.1.2 Damage Distribution Within the Hopkinson Bar Specimens

Hopkinson bar specimens were sectioned, milled and polished to their mid-plane. They were then examined under a light microscope and quantitatively analyzed with the image processing procedure outlined in Chapter 2. Microscope fields were examined on a grid pattern covering the notched region of each specimen. The microstructural damage was determined from these field measurements. Damage measurements were limited to specimens annealed at 850°C. Since the 600°C treatment produced smaller particles and much greater scatter in the metallographic results.

Distribution of Damage over Specimen Geometry

Colour contour plots of the measured porosity and void aspect ratio at various strain levels have been created. These plots (Figures 6.14 through 6.19) indicate the relative magnitude and the distribution of the measured quantities over the mid-plane of the specimen notch region. It should be noted that the actual gradients within the material may be higher than those indicated in the colour contour plots which are based on field

averaged values.

The distribution of damage within an undeformed sample is given in Figure 6.14(a). The observed nature of the initial porosity variation is an important consideration in damage studies. Figure 6.14(b) shows the approximate location of the measured fields with respect to the notch geometry. All of the contours have their axial directions aligned horizontally to the page.

The stress and strain state within each deformed specimen varies significantly over its unique geometry and evolves with time. Strains are largest near the notch surface due to the stress and strain concentrating effects of the notch. At the notched specimen centres, the triaxiality will be higher due to the plastic constraint imposed by the notch geometry.

The distribution of porosity for the uniaxial specimens are presented in Figures 6.14(c,d,e). Very little void expansion is indicated with strain for the uniaxial specimens. The aspect ratio increases uniformly with increasing strain as shown in Figures 6.15(b,c,d).

The porosity and aspect ratio distribution for the E-notch are presented in Figures 6.16 and 6.17, respectively. No significant growth has occurred at the 0.200 m/m strain level shown in Figure 6.16(b). As the strain increases, voids grow and the porosity level

increases and localizes across the minimum section of the notch. Figure 6.16(d) represents the measured porosity levels for a specimen exhibiting incipient fracture. A central crack formed in this specimen but did not have time to propagate to the outer edge to bring about failure. Note that in Figure 6.16(d), each microscope field which included the crack was separated into several smaller images surrounding the crack. The area fraction of these smaller fields was then used to define the porosity of this field without the crack. It is these values which are indicated both on the contour plot, as well as in subsequent data discussed in the proceeding section. The aspect ratio increased with plastic strain, evolving first at the notch surface and remains highest there (Figure 6.17).

The C-notch porosity data is presented in Figure 6.18, and the mean aspect ratio data is presented in Figure 6.19. In Figure 6.18(b), little void growth has occurred at this relatively low strain level. With an increment in strain to Figure 6.18(c), the porosity has increased substantially from the nominal levels shown in 6.14(b). The porosity data in Figure 6.18(c,d) demonstrates that void growth is relatively uniform across the minimum section. The high level of porosity noted at the bottom surface of the C-notch, in Figure 6.14(c), is artificially high due to the presence of an edge crack. This crack was not visible on the opposite surface of the C-notch.

The evolution of aspect ratio with strain (Figure 6.19) was quite limited. No significant change was noted at the lowest strain (Figure 6.19(a)). As the aspect ratio began to

slowly evolve, it started at the notch surface and moved inwards, as shown in Figures 6.19(b) and (c).

To summarize, the porosity and aspect ratio are shown to increase with evolving strain. This rate of increase in damage is dependent on the notch geometry and resultant triaxiality. The porosity increase at 0.352 m/m strain in the uniaxial specimen, Figure 6.14(c), is smaller than that of the E-notch and particularly that of the C-notch at similar strains. The aspect ratio increases first at the notch surface, and progresses towards the specimen centre. The distribution of damage or porosity in the C-notch appears to be more uniform across its diameter than in the uniaxial specimen or the E-notch.

6.1.3 Damage Evolution Within the Hopkinson Bar Specimens

In order to construct histories of damage evolution, only two regions of the specimens were considered, namely the specimen centre (lower strain, high triaxiality region) and the region near the notch surface (higher strain, low triaxiality region).

The damage levels from the central and edge fields were considered as well as averages from the three by three region of fields about the centre of the notch and a two by three region at the specimen edge. Values of the lead volume fraction measured in the undeformed material are given in each plot as well as the strain at fracture where available.

Porosity Growth

The increase in porosity with strain is presented in Figure 16.20 for the uniaxial specimens. The error bars indicated in these plots represent the square root of the variance in field measurements for the multiple field averages. The size of these bars is quite large due to the natural statistical variation in the measured quantities. The porosity appears to undergo stable growth at a rate of about, 1.80, at the specimen centre and, 2.95, at the specimen edge. These values were calculated from a linear fit to the three by three and two by three field measurements and represent the rate of increase in porosity with strain during stable void growth. Note that these rates were based on only a few data points and should be viewed as approximate.

The evolution of porosity with strain for the E-notch is given in Figure 16.21. The rate of increase in porosity at the notch centre, 7.78, is much higher than that at the notch surface, 3.33. The porosity of the high triaxiality C-notch, (Figure 6.22), shows close agreement in the rate of increase in porosity evolution between the centre and surface of the notch. The rate of porosity increase with strain was, 13.31, at the centre of the specimen and, 8.80, at the surface of the notch. These rates are considerably higher than for the E-notch and uniaxial geometries. The higher growth rates at the surface of the C-notch are due to the large strain concentration in this region. Thus higher strains and low triaxialities at the surface produced net void growth comparable to the centre where lower strains but high triaxialities exist.

While only a limited amount of data is available, the strain at which a central crack develops in the E-notch is larger than for the C-notch. It is interesting to note that the growth of porosity does not initiate until a strain of roughly 30% for the E-notch and 10-15% for the C-notch.

The evolution of porosity, based on the multiple field averages of each specimen geometry, is compared in Figures 6.23(a) and 6.23(b) for the notch centre and surface respectively. The E-notch data is noted to fall below both the uniaxial and C-notch specimens due to the lower initial porosity of the undeformed material from which it was manufactured. The growth rate, or the slope of the porosity curve as a function of the strain curve, is clearly increasing with triaxiality. This result is summarized in Figure 6.24 which plots the measured growth rates as functions of triaxiality at the specimen centre.

Mean Void Area

The mean area trends, presented in Figures 6.25, 6.26, and 6.27, follow similar trends to the porosity as expected. The void growth for the uniaxial specimen was found to be very low over the range of strains studied as shown in Figure 6.25. This was particularly noticeable when compared to the near doubling which occurred at the centre of both the E-Notch (Figure 6.26) and the C-notch (Figure 6.27).

The similarity between the mean area and porosity trends indicate that it is in fact void

growth rather than nucleation which is defining failure. The abundance of the lead particles and the relative ease with which they nucleate voids suggests that the increase in porosity should be primarily a function of void growth.

The increase in the variance of particle area and porosity with strain, noted in the C-notch and E-notch measurements, could be an indication of clusters evolving as discussed by VanderVoort [97].

Mean Void Aspect Ratio

The evolution of mean void aspect ratio with strain is given in Figures 6.28, 6.29 and 6.30, for the uniaxial, E-notch and C-notch respectively, at both the notch centre and surface.

The uniaxial data, Figure 6.28, suggests that the aspect ratio at the edges and surface of the notch centre are identical as expected. This is not the case for the C-notch (Figure 6.29) or E-notch (Figure 6.30), for which the aspect ratio is highest at the notch surface in both cases. The aspect ratio for the central region of the C-notch remains essentially constant. Analytical and numerical studies by previous researchers[35,36,40] suggest that the rate of void elongation will decrease with increase triaxiality. This trend would account for the low aspect ratios in the high triaxiality region of the C-notch centre. Thus, the greatest void elongation occurs at the notch surfaces under conditions of large strain and low triaxiality.

The void aspect ratio versus diametral strain data from the notch centre, is plotted in Figure 6.31. The decrease in void aspect ratio with triaxiality is evident with lower void extension with strain for the notched geometries.

6.1.4 Fractography of the Hopkinson Bar Specimens

Optical micrographs of specimens of the E-notch specimens exhibiting incipient failure indicate that failure initiates at the notch centre (Figure 6.32). A central crack is formed and propagates radially until plastic instability brings about shearing at the outer edge of the central crack along the planes of maximum shear (approximately 45° to the tensile direction). This generates the classical cup-cone failure (Figure 6.33) often exhibited in ductile failure of tensile specimens.

Micrographs taken transverse to the fracture surface using a light microscope are given in Figure 6.34 for the E-notch. The E-notch central crack follows a meandering path through the central region of the notch. It appears as though fracture occurs through localized regions of coalesced voids. Figure 6.34 shows a region of coalesced voids which may have become part of the evolving fracture surface had the deformation been permitted to continue.

Scanning electron microscope was used to examine the fracture surfaces of both 600HT and 850HT, E-notches and C-notches. The fracture surface of the E-notch for the 850HT is presented in Figure 6.35, and for the 600HT in Figure 6.36. Overall the E-notch

exhibited a very tortuous fracture path evidenced by the surface relief. The fracture surface consists of the characteristic dimples associated with ductile fracture. The C-notch results are presented in Figure 6.37 for the 850HT and Figure 6.38 for the 600HT. The fracture surfaces in both C-notch cases are much flatter than the E-notch surfaces. Comparison of the 600HT and 850HT for similar notches, indicates that the characteristic dimple size was smaller for the 600HT brass. The dimple size on the fracture surface of the higher triaxiality C-notch was also noted to be larger than that of the E-notch for comparable materials.

6.2 Flyer Plate Data

As mentioned previously, all of the plate impact testing was performed by the Dr. H. Nahme at the Ernst-Mach Institute in Freiburg, Germany. The mechanical response of the plates has been reported in [117]. In this section the VISAR velocity profiles will be presented, as well as the measured damage.

6.2.1 Free Surface Velocity Profiles

The free surface velocity profiles measured using the VISAR system are presented in Figures 6.39, 6.40, and 6.41 for each of the materials investigated. The variation between these materials may be determined by examining Figure 6.42 which shows typical curves at similar impact velocities from spalled tests.

As discussed in Chapter 5, the VISAR system records the free surface velocity at the back of the plate. The arrival of the elastic wave at the rear surface in all of the VISAR plots (Figures 6.39 through 6.41) is marked by the initial rise of the pulse. The increase in free-surface velocity is slowed at the Hugoniot elastic limit (HEL) until the stronger but slower plastic wave arrives. The particle velocity at the rear surface then increases to a level defined by the Hugoniot state for that particular test. This level is maintained for the duration of the pulse until a moderate strain rate release of about the same magnitude of the HEL occurs as a result of the elastic unloading wave of the flyer plate arriving. This initial release is not shown very clearly in these plots. The higher strain rate release which follows is a substantial reduction of particle velocity indicating the initial arrival of the plastic wave to the rear of the plate. This is followed by a third region of very low strain rate release. The low strain rate release is sustained until the release wave reflected from the impact face or the spall plane reaches the rear surface. The drop in the velocity is related to the magnitude of the tensile pulse reflected from the rear of the target plate.

For the MS-58 plate tested at 119 m/s (Figure 6.42), the conditions for spall were just met. This created an incipient spall surface. The amount of time from the initial arrival of the tensile wave to the evolution of a strong spall signal is shown in the dotted box of Figure 6.42 and indicates the fracture process is at most a 4 microsecond process. Data of this type offers interesting possibilities in terms of defining the coalescence process as the wave may be decomposed and related to the evolution of the fracture surface.

6.2.2 Image Analysis Data

Note on Loading

While a great deal of control over the stress state and the applied strain is available in Hopkinson bar tests, the same is not the case for the plate impact specimens. The particle velocity plots shown earlier indicate that the material is repeatedly loaded by a tension compression cycle until the waves in the plate dissipate. The majority of the damage is believed to occur on the primary pulse which for a short period of time produces a state of uniaxial strain at the plate centre. This strain state is supported until radial release waves arrive. For the geometry of test used in this study, this state is disturbed during the initial pulse. The addition of a radial spall ring would suppress the radial release waves which disturb the uniaxial strain state. As a result, the damage within the plate was not greatest at the plate centre as expected but was largest at the outer edge due to the radial release waves from the edge of the plate convergence with the release waves originating from the flyer and target plate free surfaces.

Distribution of Damage in Plates

The plate impact specimens of 850HT material will be investigated. Only the plates which did not spall will be considered. The distribution of damage at the centre of the plate is given in Figure 6.43 for the plate impacted at 93 m/s. The porosity, mean area, aspect ratio, and mean particle counts are presented. The porosity and mean area contour plot show discrete elliptical regions of higher porosity that are distributed along the spall layer. The measured porosity values were quite low at the centre of the plates compared

to the near edge region.

In light of the radial release waves, measurements were made closer to the edge of the plate. A strip, three microscope fields (0.84 mm) in width, was sampled from the region near the outer edge of the plate along a line moving radially inwards along the incipient spall layer. The resulting distributions are presented in Figure 6.44 for each of the three plates considered. The porosity measured in these regions at the edge of the plates is much higher than that indicated at the plate centre. This suggests that failure of the plate is initiated in these regions of higher stress and triaxiality, and propagates radially inwards towards the centre of the plate. This crack most likely "percolates" through the previously coalesced regions noted in the central region contour plots (Figure 6.43), within the spall layer.

6.2.3 Fractography of Plate Specimens

A micrograph of the incipient spall surface for the MS-58 plate tested at 119m/s is given in Figure 6.45. This micrograph exhibits regions of locally coalesced particles as noted in the void volume distributions of central region of the plate.

The fracture surface appears to evolve through the progressive joining of these regions of coalesced particles. While final failure is dependent on the joining of these various regions of coalesced particles they are not all included the fracture surface.

CHAPTER 7

DISCUSSION, CONCLUSIONS AND RECOMMENDATIONS

7.1 Discussion of Experimental Results

The focus of this research has been the experimental examination of damage evolution in leaded brass under high rates of tensile loading. Interrupted split Hopkinson bar tests have been successfully performed to capture increases in porosity and void aspect ratio with strain at high rates. Plate impact specimens have also been studied and the damage state in incipient spall layers has been determined. Further work is required to determine damage evolution rates during plate impact experiments.

7.1.1 Mechanical Properties of Brass

The MS-58 brass has been shown to exhibit a higher flow stress at a given strain than the UNS-C36000. The 600HT and 850HT brass have been shown to have an equivalent stress-strain response, within the range of scatter for the tensile split Hopkinson bar apparatus.

Tests on the notched specimens demonstrated the dependence of ductility on triaxiality. High triaxiality C-notch specimens were found to fracture at a lower diametral strain than

the lower triaxiality E-notch specimens. The stress-strain data for the uniaxial, E-notch and C-notch specimens confirmed that stress increased with notch severity or triaxiality.

7.1.2 Distribution of Damage

Colour contour plots of void volume fraction, average area and aspect ratio have been developed and are useful in interpreting the distribution of damage over the specimen geometry. The uniaxial Hopkinson bar specimens exhibited a uniform porosity and aspect ratio across their gauge as expected. The porosity within the E-notch specimens grew first at the surface of the notch but concentrated at the notch centre as deformation progressed. The C-notch specimens showed a much more uniform distribution of porosity than in the E-notch. The largest increase in aspect ratio occurred in the low and moderate triaxiality uniaxial and E-notch specimens, respectively.

The plate impact specimens impacted at velocities just below the threshold for spallation were found to have discrete regions of higher porosity along their incipient spall plane. The measured values of porosity near the centres of the plates were much lower than expected, and further investigation revealed that the greatest extent of the plate's damage had occurred away from the centre line of the impact, near the plate edge. In this location the levels of damage were comparable to those of the Hopkinson bar specimens interrupted just prior to failure. The similarity in porosity between the Hopkinson bar specimens and the impacted plates, both near failure, suggests that this material may be governed by a critical porosity for void coalescence criterion.

7.1.3 Evolution of Damage Parameters with Strain

No change in porosity was noted within E-notch specimens until approximately 30 % strain had been achieved. The C-notch specimens showed growth initiating between 10 and 15 % strain. This indicates that the nucleation of damage is dependent on triaxiality, and suggests a stress based criteria should be used to describe void nucleation.

The evolution of damage has been shown to be related to the triaxiality. Porosity evolved the most quickly for the C-notch, followed by the lower triaxiality E-notch and finally the uniaxial specimen. The evolution of aspect ratio showed the opposite dependence on triaxiality. In the C-notch the aspect ratio showed very little change as compared to the E-notch and the uniaxial specimens. These results may be used to describe the local variations which exist across the specimen gauge, as noted in the previous section.

The strains near the surface of a notched specimen are higher than those at their centre due to the stress and strain concentrations in this region. The triaxiality arises as a result of the plastic constraint to flow and is highest in the centre of the notch. Interestingly, the C-notch specimen showed a uniform extent of damage across its minimum section. This conformity is attributed to the rate of void growth at the notch surface, occurring under conditions of low triaxiality but high strain, being equal to the rate at the centre where the strain is lower but the triaxiality is much greater.

There was an increase in the variance of the void area and porosity measurements with

increasing strain. This trend indicates the development of field size clustering in the material which may be evolving with strain.

7.1.4 Damage at Fracture

Optical micrographs of Hopkinson bar specimens have indicated that small regions of coalesced voids are involved in defining the fracture path. Similar results have been noted for the plate impact specimens, however the regions of coalesced voids are larger. The fracture surface seems to evolve by the joining of these coalesced regions. This behaviour is more noticeable for the plate impact specimens than the Hopkinson bar specimens.

If the effects of damage distribution on fracture are to be studied, alternative model materials should be considered. The leaded brass provide numerous sites for easy nucleation of voids. This limits the size of clusters as any given void has several neighbours to coalesce with which are nearly equally favoured. To study the percolation of damage, materials of lower nucleating particle fraction should be used.

7.1.5 Applications of Reported Data

Due to the difficulties in obtaining void growth data, particularly at high rates, models developed for these processes do not typically use experimentally defined values to describe damage evolution. For example, Gurson's [43,44,45] constitutive model, as modified by Tvergaard and Needleman [62,63], is commonly used to predict damage

accumulation and failure in ductile materials. Tvergaard and Needleman [62,63] have incorporated a model to account for the onset of coalescence, as discussed in Section 1.2. This model utilizes a critical porosity level at which coalescence is assumed to initiate, f_c , and the material is said to fail completely at f_f . Typically, these values are arbitrarily set to 0.15 and 0.25, respectively, which may be valid for some materials, but are clearly much higher than the porosity at coalescence in this study.

The evolution of void volume fraction or void growth, determined over the course of this work, could be used to calibrate such void growth models for problems at higher rates. Such applications were beyond the scope of the current research, but represents the next step in data analysis.

7.2 Conclusions

The tensile split Hopkinson bar, with momentum trapping, has been shown to provide a viable means of determining damage evolution for various ductile metals in the 10^3 to 10^4 s^{-1} strain rate regime.

The results of this study on leaded brass support the following conclusions:

1. Nucleation strain is dependent on triaxiality and not strain;
2. The evolution of aspect ratio was found to be inversely proportional to

triaxiality and strongly dependent on strain;

3. The rate at which porosity evolves has been shown to be a strong function of triaxiality; and,
4. Clustering of voids important in fracture and may also influence growth rates during the deformation process.

While control over the plate impact tests was limited, this experiment provides a unique stress and strain state not normally attainable in Hopkinson bars. The damage in the plates, prior to fracture, was found to reach a similar level to that observed in the tensile split Hopkinson bar specimens, suggesting a critical level of damage at fracture.

7.3 Recommendations for Future Research

The interrupted tensile split Hopkinson bar apparatus should be used to characterize damage evolution in commercial alloys. This would provide much needed experimental data at these higher rates. It would also allow dynamic void growth models, to be properly calibrated for these materials.

Future studies to measure growth rate for the plate impact tests should require the use of a spall ring. Methods of momentum trapping, similar to that used in the Hopkinson bar, may also be useful to interrupt damage.

Cluster analysis, through methods such as Voronoi diagrams and dilational counting techniques, should be performed to evaluate both the extent and evolution of particle clusters during plastic deformation at various triaxialities.

Alternative model materials should be considered to investigate the effects of clustering on damage evolution and fracture. In earlier studies by other researchers, compacts of copper and iron powder have been used for quasi-static void evolution studies. These materials allow the initial porosity of the material to be defined through the pressing process used to form the material, and may be useful for high rate studies.

BIBLIOGRAPHY

- [1] Nemat-Nasser, S., Isaacs, J.B. and Starrett, J.E., Hopkinson Techniques for Dynamic Recovery Experiments, *Proceedings of the Royal Society of London, A*, **435** (1991), pp. 371-391.
- [2] Rogers, H.C., The Tensile Fracture of Ductile Metals, *Transactions of the Metallurgical Society of AIME*, **218** (1960), pp. 498-506.
- [3] Tipper, C.F., The Fracture of Metals, *Metallurgica*, **39** (1948), pp. 133-137.
- [4] Goods, S.H. and Brown, L.M., The Nucleation of Cavities by Plastic Deformation, *Acta Metallurgica*, **27** (1979), pp. 1-15.
- [5] Argon, A.S., Im, J. and Safoglu, R., Cavity Formation from Inclusions in Ductile Fracture, *Metallurgical Transactions A*, **6A** (1975), pp. 825-837.
- [6] Gurland, J. and Plateau, J., The Mechanism of Ductile Rupture of Metals Containing Inclusions, *Transactions of the ASM*, **56** (1963), pp. 443-4454.

- [7] Brown, L.M. and Stobbs, W.M., The Work-Hardening of Copper-Silica V. Equilibrium Plastic Relaxation By Secondary Dislocations, *Philosophical Magazine*, **34** (1976), pp. 351-372.
- [8] Tanaka, K., Mori, T. and Nakamura, T., *Philosophical Magazine*, **21** (1970), pp. 267-279.
- [9] Fisher, J.R. and Gurland, J., Void Nucleation in Spheroidized Carbon Steels, Part 2: Model, *Metal Science*, May 1981, pp. 193-202.
- [10] Gurland, J., Observations on the Fracture of Cementite Particles in a Spheroidized 1.05% Steel Deformed at Room Temperature, *Acta Metallurgica*, **20** (1972), pp. 735-741.
- [11] Fischmeister, H.F., Navara, E. and Easterling, K.E., Effects of Alloying On Structural Stability and Cohesion Between Phases in Oxide/Metal Composites, *Metal Science*, **6** (1972), pp. 211-215.
- [12] Hipsley, C.A. and Druce, S.G., The Influence of Phosphorous Segregation to Particle/Matrix Interfaces on Ductile Fracture in a High Strength Steel, *Acta Metallurgica*, **31** (1983), pp. 1861-1872.

- [13] Argon, A.S. and Im, J., Separation of Second Phase Particles in Spheroidized 1045 Steel, Cu-0.6Pct Cr Alloy, and Marging Steel in Plastic Straining, *Metallurgical Transactions A*, **6A** (1975), pp. 839-851.
- [14] Thomson, R.D. and Hancock, J.W., Local Stress and Strain Fields Near a Spherical Elastic Inclusion in a Plastically Deforming Matrix, *International Journal of Fracture*, **24** (1984), pp. 209-228.
- [15] French, J.A. and Weinrich, P.F., *Metallurgical Transactions A*, **6A** (1975), p. 785.
- [16] Needleman, A., A Continuum Model for Void Nucleation by Inclusion Debonding, *Journal of Applied Mechanics*, **54** (1987), pp. 525-531.
- [17] Teirlinck, D., Zok, F., Embury, J.D. and Ashby, M.F., Fracture Mechanism Maps in Stress Space, *Acta Metallurgica*, **36** (1988), pp. 1213-1228.
- [18] Liu, C.T. and Gurland, J., The Fracture Behaviour of Spheroidized Carbon Steels, *Transactions of the ASM*, **61** (1968), pp. 156-167.

- [19] LeRoy, G., Embury, J.D., Edwards, G. and Ashby, M.F., A Model of Ductile Fracture Based on the Nucleation and Growth of Voids, *Acta Metallurgica*, **29** (1981), pp. 1509-1522.

- [20] Fisher, J.R. and Gurland, J., Void Nucleation in Spheroidized Carbon Steels Part I: Experimental, *Metal Science*, May 1981, pp. 185-192.

- [21] Embury, J.D., Plastic Flow in Dispersion Hardened Materials, *Metallurgical Transactions A*, **16A** (1985), pp. 2191-2200.

- [22] Thomason, P.F., Ductile Fracture by the Growth and Coalescence of Microvoids of Non-Uniform Size and Spacing, *Acta Metallurgica and Materialia*, **41** (1993), pp. 2127-2134.

- [23] Thomson, R.D. and Hancock, J.W., Ductile Failure by Void Nucleation, Growth and Coalescence, *International Journal of Fracture*, **25** (1984), pp. 99-112.

- [24] McClintock, F.A., A Criterion for Ductile Fracture by the Growth of Holes, *Journal of Applied Mechanics*, **35** (1968), pp. 363-371.

- [25] Berg, C.A., The Motion of Cracks in Plane Viscous Deformation, proceedings of the *Fourth Annual U.S. National Congress of Applied Mechanics*, ASME Vol. 2 (1962), pp. 885-892.
- [26] Perra, M. and Finnie, I., Void Growth and Localization of Shear in Plane Strain Tension, *Fracture 1977, Volume 2, ICF4, Waterloo Canada*, (1977), pp. 415-423.
- [27] Becker, R., Smelser, R.E., Richmond, O. and Appleby, E.J., The Effect of Void Shape on Void Growth and Ductility in Axisymmetric Tension Tests, *Metallurgical Transactions A*, 20A (1989), pp. 853-861.
- [28] Nemat-Nasser, S. and Taya, M., Numerical Studies of Void Growth in a Necked Bar, *International Journal of Solids and Structures*, 16 (1980), pp. 483-494.
- [29] Melander, A., Void Growth During Wire Drawing and Tensile Testing of Copper, *Scandinavian Journal of Metallurgy*, 9 (1980), pp. 267-272.

- [30] Sowerby, R. and Chandrasekaran, N., Applications of McClintock's Ductile Fracture Model of the Growth of Holes to the Tensile Straining of Some Spheroidized Steels Under Superimposed Hydrostatic Pressure, *Materials Science and Engineering*, **79** (1986), pp. 15-25.
- [31] Brownrigg, A., Spitzig, W.A., Richmond, O., Teirlinck, D. and Embury, J.D., The Influence of Hydrostatic Pressure on the Flow Stress and Ductility of Spheroidized 1045 Steel, *Acta Metallurgica*, **31** (1983), pp. 1141-1150.
- [32] Floreen, S. and Hayden, H.W., Some Observations of Void Growth During the Tensile Deformation of a High Strength Steel, *Scripta Metallurgica*, **4** (1970), pp. 87-94.
- [33] Cox, T.B. and Low Jr., J.R., An Investigation of the Plastic Fracture of AISI 4340 and 18 Nickel-200 Grade Maraging Steels, *Metallurgical Transactions*, **5** (1974), pp. 1457-1470.
- [34] Rice, J.R. and Tracey, D.M., On the Ductile Enlargement of Voids in Triaxial Stress Fields, *Journal of the Mechanics and Physics of Solids*, **17** (1969), pp. 201-217.

- [35] Budiansky, B., Hutchinson, J.W. and Slutsky, S., Void Growth and Collapse in Viscous Solids, *Mechanics of Solids, The Rodney Hill 60th Anniversary Volume*, eds. H.G. Hopkins and M.J. Sewell, (1982), pp. 13-45, Pergamon Press, Oxford.

- [36] Worswick, M.J. and Pick, R.J., Void Growth and Constitutive Softening in a Periodically Voided Solid, *Journal of the Mechanics and Physics of Solids*, **38** (1990), pp. 601-625.

- [37] Barnby, J.T., Shi, Y.W., and Nadkarni, A.S., On the Void Growth in C-Mn Structural Steel During Plastic Deformation, *International Journal of Fracture*, **25** (1984), pp. 273-283.

- [38] Beremin, F.M., Study of Criteria for Ductile Rupture of A508 Steel, from *Advances in Fracture Research*, ICF5, Vol. 2, (1981), pp. 809-816.

- [39] Marini, B., Mudry, F. and Pineau, A., Experimental Study of Cavity Growth in Ductile Rupture, *Engineering Fracture Mechanics*, **22** (1985), pp. 989-996.

- [40] Bourcier, R.J., Koss, D.A., Smelser, R.E. and Richmond, O., The Influence of Porosity on the Deformation and Fracture of Alloys, *Acta Metallurgica*, **34** (1986), pp. 2443-2453.

- [41] Andersson, H., Analysis of a Model for Void Growth and Coalescence Ahead of a Moving Crack Tip, *Journal of Mechanics and Physics of Solids*, **25** (1977), pp. 217-233.
- [42] Worswick, M.J. and Pick, R.J., Void Growth in Plastically Deformed Free-Cutting Brass, *Journal of Applied Mechanics*, **58** (1991), pp. 631-638.
- [43] Gurson, A.L., Continuum Theory of Ductile Rupture by Void Nucleation and Growth: Part I - Yield Criteria and Flow Rules for Porous Ductile Media, *Transactions of the ASME*, (1977), pp. 2-15.
- [44] Gurson, A.L., Porous Rigid-Plastic Materials Containing Rigid Inclusions-Yield Function, Plastic Potential, and Void Growth, *Fracture 1977*, Vol 2, ICF4, Waterloo, 1977.
- [45] Gurson, A.L., *Plastic Flow and Fracture Behaviour of Ductile Materials Incorporating Void Nucleation, Growth and Interaction*, PhD Thesis, Brown University, Providence, R.I., 1975.
- [46] Spitzig, W.A., Smelser, R.E. and Richmond, O., The Evolution of Damage and Fracture in Iron Compacts with Various Initial Porosities, *Acta Metallurgica*, **5** (1988), pp. 1201-1211.

- [47] Fyfe, I.M., A Damage Criterion for Dynamic Ductile Failure, in Structural Impact and Crashworthiness, Vol. 2, Morton (ed.), Elsevier Applied Science Publishers, July 1984.
- [48] Ashby, M., Gandhi, Taplin, Drm. R. *Acta Metallurgica*, **27** (1979), pp. 669.
- [49] Beachem, C.D., An Electron Fractographic Study of the Influence of Plastic Strain Conditions Upon Ductile Rupture Process in Metals, *Transactions of the ASM*, **58** (1963), pp. 318-326.
- [50] Bluhm, J.I. and Morrissey, R.J., Fracture in a Tensile Specimen, Proceedings of the 1st International Conference of Fracture, Sendai, Vol 3, Sendai, Japan, 1966, pp. 1739-1780.
- [51] Thompson, A.W., Modeling of Local Strains in Ductile Fracture, *Metallurgical Transactions A*, **18A** (1987), pp. 1877-1886.
- [52] Thomason, P.F., A Theory for Ductile Fracture by Internal Necking of Cavities, *Journal of the Institute of Metals*, **96** (1968), pp. 360-365.

- [53] Edelson, B.I. and Baldwin, W.M. Jr., The Effects of Second Phases on the Mechanical Properties of Alloys, *Transactions of the ASM*, **55** (1962), pp. 230-250.

- [54] Sun Jun, Effects of Stress Triaxiality on Micro-Mechanisms of Void Coalescence and Micro-Fracture Ductility of Materials, *Engineering Fracture Mechanics*, **39** (1991), pp. 799-805.

- [55] Dubensky, E.M. and Koss, D.A., Void/Pore Distributions and Ductile Fracture, *Metallurgical Transactions A*, **18A** (1987), pp. 1887-1895.

- [56] Teirlinck, D., Ashby, M.F. and Embury, J.D., Damage Accumulation During Ductile Rupture and the Development of Fracture Maps, *Advances in Fracture Research (Fracture 84)*, ICF6, 1984.

- [57] Tonks, D.L., Percolation Wave Propagation, and void link-up effects in ductile fracture, *Journal de Physique IV*, **C8**(1994), pp. 665-670.

- [58] Worswick, M.J., Pilkey, K.A., Fowler, J.P., Effects of Microstructural Defect Distribution on Failure and Localization of Aluminum Sheet - Second Progress Report: Preliminary Testing and Image Analysis, Contract Report for ALCAN Kingston Research and Development Centre, Carleton University, March 29, 1994.

- [59] Needleman, A. and Tvergaard, V., Analysis of Plastic Flow Localization in Metals, *Applied Mechanics Reviews*, **45** (1992), pp. S3-S18.

- [60] Nemat-Nasser, S., Phenomenological Theories of Elastoplasticity and Strain Localization at High Strain Rates, *Applied Mechanics Reviews*, **45** (1992), pp.S19-S45.

- [61] Tvergaard, V., Material Failure by Void Coalescence in Localized Shear Bands, *International Journal of Solids and Structures*, **18** (1982), pp. 659-672.

- [62] Tvergaard, V. and Needleman, A., Analysis of the Cup-Cone Fracture in a Round Tensile Bar, *Acta Metallurgica*, **32** (1984), pp. 157-169.

- [63] Needleman, A. and Tvergaard, V., An Analysis of Ductile Rupture in Notched Bars, *Journal for the Mechanics of Physics of Solids*, **32** (1984), pp. 461-490.

- [64] Ohno, N. and Hutchinson, J.W., Plastic Flow Localization Due to Non-Uniform Void Distribution, *Journal for the Mechanics and Physics of Solids*, **32** (1984), pp. 63-85.
- [65] Yamamoto, H., Conditions for Shear Localization in the Ductile Fracture of Void-Containing Materials, *International Journal of Fracture*, **14** (1978), pp. 347-365.
- [66] Marciniak, A. and Kuczynski, K., *International Journal for Mechanical Science*, **9** (1967), pp. 609.
- [67] Hancock, J.W. and Brown, D.K., On the Role of Strain and Stress State in Ductile Failure, *Journal for the Mechanics and Physics of Solids*, **31** (1983), pp. 1-24.
- [68] Thomason, P.F., A Three-Dimensional Model for Ductile Fracture by the Growth and Coalescence of Microvoids, *Acta Metallurgica*, **33** (1985), pp. 1087-1095.
- [69] Zhang, Z.L. and Niemi, E., A New Failure Criterion for the Gurson-Tvergaard Dilational Constitutive Model, *International Journal of Fracture*, **70** (1995), pp. 321-334.

- [70] Broek, D., The Role of Inclusions in Ductile Fracture and Fracture Toughness, *Engineering Fracture Mechanics*, **5** (1973), pp. 55-66.
- [71] Oh, S.I. and Kobayashi, S., *Deformation Model of Void-Growth and Coalescence in the Process of Ductile Fracture*, Air Force Material Laboratory Report AFML-TR-75-95, July 1976.
- [72] Tvergaard, V., Ductile Fracture by Cavity Nucleation Between Larger Voids, *Journal for the Mechanics and Physics of Solids*, **30** (1982), pp. 265-286.
- [73] Tvergaard, V., A Numerical Investigation of 3-D Localization Failure by a Void-Sheet Mechanism, *Engineering Fracture Mechanics*, **41** (1992), pp. 787-803.
- [74] Regazzoni, G. and Montheillet, F., High Strain Rate Ductility in Uniaxial Tension : A Review, *Journal de Physique IV*, **C5** (1985), pp. 435-444.
- [75] Seaman, L., Curran, D.R. and Shockey, D.A., Computational Models for Ductile and Brittle Fracture, *Journal of Applied Physics*, **47** (1976), pp. 4814-4826.

- [76] Carroll, M.M. and Holt, A.C., Static and Dynamic Pore-Collapse Relations for Ductile Porous Materials, *Journal of Applied Physics*, **43** (1972), pp. 1626-1636.
- [77] Regazzoni, G., Follansbee, P.S. and Johnson, J.N., Inertia, Stability and Ductility in Uniaxial Tension at High Strain Rate, Los Alamos National Laboratory Report LA-UR--85-632, (1985).
- [78] Shockey, D.A., Understanding the Dynamic Fracture Phenomenology with Fractography, in Metallurgical and Materials Applications of Shock Wave and High-Strain-Rate Phenomena, Staudhammer, K.P. and Meyers, M.A.(eds.), Elsevier Science B.V., (1995), pp. 151-162.
- [79] Benson, D.J., An Analysis of Void Distribution Effects on the Dynamic Growth and Coalescence of Voids in Ductile Metals, *Journal for the Mechanics and Physics of Solids*, **41**(1993), pp. 1285-1308.
- [80] Tonks, D.L., Zurek, W.R. and Thissell, W.R., Ductile Damage Modelling Based on Void Coalescence and Percolation Theories, Los Alamos National Laboratory Report LA-UR--95-2920, (1995).

- [81] Cox, T.B. and Low Jr., J.R., An Investigation of the Plastic Fracture of AISI 4340 and 18 Nickel-200 Grade Maraging Steels, *Metallurgical Transactions*, **5** (1974), pp. 1457-1470.
- [82] Magnusen, P.E., Follansbee, P.S. and Koss, D.A., The Influence of Strain Rate and Porosity on the Deformation and Fracture of Titanium and Nickel, *Metallurgical Transactions A*, **16A** (1985), pp. 2273-2281.
- [83] Bridgman, P.W., Studies in Large Plastic Flow and Fracture, Chapter 1. Harvard University Press, Cambridge Massachusetts, (1964).
- [84] Johnson, J.N., Dynamic Fracture and Spallation in Ductile Solids, *Journal of Applied Physics*, **52** (1981), pp. 2812-2825.
- [85] Johnson, J.N. and Addessio, F.L., Tensile Plasticity and Ductile Fracture, *Journal of Applied Physics*, **64** (1988), pp. 6699-6712.
- [86] Cortes, R., Dynamic Growth of Microvoids Under Combined Hydrostatic and Deviatoric Stresses, *International Journal of Solids and Structures*, **29** (1992), pp. 1637-1645.

- [87] DeHoff, R.T. and Rhines, F.N., Quantitative Microscopy, McGraw-Hill Book Company, New York, New York, 1968.
- [88] Underwood, E.E., Quantitative Stereology, Addison-Wesley Publishing Co., Reading, Mass., 1970.
- [89] Russ, C.R., Practical Stereology, Plenum Press, New York, New York, 1986.
- [90] Pratt, WK., Digital Image Processing, 2nd ed., Wiley, New York, 1991.
- [91] Russ, JC., Computer-Assisted Microscopy; The Measurement and Analysis of Images, Plenum, New York, 1990.
- [92] Castleman, KR., Digital Image Processing, Prentice-Hall, Englewood Cliffs, New Jersey, 1979.
- [93] Worswick, M.J., *An Image Processing-Based Particle Analysis System - User Manual*, Department of Mechanical Engineering, University of Waterloo.
- [94] ASTM E1245-89, "Standard Practice for Determining Inclusion Content in Steel and Other Metals by Automatic Image Analysis", in Annual Book of ASTM Standards, Vol 03.01.

- [95] Cole, GS., "Practical Examples of Useful Algorithms in Computer-Aided Microscopy", Micon 90: Advances in Video Technology for Microstructural Control, ASTM STP 1094, George F. VanderVoort, Ed., ASTM, Philadelphia, 1990. pp.5-28.
- [96] Fowler, J.P., *Particle Image Analysis System for Low Area Fraction Measurements - User Manual*, Department of Mechanical and Aerospace Engineering, Carleton University, in preparation.
- [97] Agin, G.J., in Handbook for Industrial Robotics, preprint of *Section on Vision*, 1983, published by Wiley.
- [98] VanderVoort, GF., "Evaluating Clustering of Second-Phase Particles", Micon 90: Advances in Video Technology for Microstructural Control, ASTM STP 1094, George F. VanderVoort, Ed., ASTM, Philadelphia, 1990. pp.242-265.
- [99] Samuels, L.E., *Metagraphic Polishing by Mechanical Methods*, 3rd Edition, SME, 1992.
- [100] "Properties and Selection - Nonferrous alloys and pure metals," Metals Handbook, Ninth Edition, Volume 2, ASM, 1985.

- [101] Hopkinson, B., *A Method of Measuring the pressure produced in the Detonation of High Explosives or by the Impact of Bullets*, Phil. Trans. A, **213**, 1914, p. 437-456.
- [102] Kolsky, H., *An Investigation of the Mechanical Properties of Materials at Very High Rates of Loading*, Proc. Royal Soc. A, **62**, 1949, p. 676-700.
- [103] Harding, J., Wood, E.D. and Campbell, J.D., *Tensile Testing of Material at Impact Rates of Strain*, J.Mech.Eng., Sci., **2**, 1960, p. 88-96.
- [104] Pochhammer, L., *On the Propagation Velocities of Small Oscillations in an Unlimited Isotropic Circular cylinder*, J. Reine Angewandte Math., **81**, 1876, p. 324-336.
- [105] Chree, C., *The Equations of an Isotropic Elastic in Polar and Cylindrical Coordinates, Their solutions and Applications*, Cambridge Phil. Soc. Trans., **14**, 1889, p. 250.
- [106] Davies, R.M., *A Critical Study of the Hopkinson Pressure Bar*, Phil. Trans. A, **240**, 1948, p. 375-457.

- [107] Zukas, J.A., *Stress Waves in Solids*, Chapter 1 of *Impact Dynamics* by J.A. Zukas, T. Nicholas, H.F. Swift, L.B. Grezschuck and D.R. Curran, Krieger Publishing Corp., Malabar, FL, 1992, p. 1.
- [108] Follansbee, P.S., *The Hopkinson Bar* in *The Metal handbook*, 9th edn., 8, ASM, Metals parks, OH, 1985, p. 198-203.
- [109] Davies, E.D.H. and Hunter, S.C., *The Dynamic Compression Testing of Solids by the Method of the Split Hopkinson Pressure Bar*, J. Mech. Phys. Solids, **11**, 1963, p. 155-179.
- [110] Pelletier, P., *Mechanical Behaviour of Armco Iron at High Strain Rates and Elevated Temperatures*, M.Eng. Thesis, Carleton University, May, 1995.
- [111] Hancock, J.W. and Mackenzie, A.C., *On the Mechanisms of Ductile Fracture in High-Strength Steels Subjected to Multi-Axial Stress-States*, J. Mech. Phys. Sol., **24**, 1976, p. 147-169.
- [112] Rajendran, A.M., Bless, S.J., *Plastic Flow and Failure Modeling under High Strain Rate Loading*, AF Wright Aeronautical laboratories Report AFWAL-TR-87-4134, 1988.

- [113] Worswick, M.J., Pelletier, P., Fowler, J., Gallagher, P.J. and Weickert, C.A., Ductile Fracture during High Rate Forming of Armco Iron, *Proceedings EXPLOMET 95*, eds. L. Murr, El Paso, Texas, 1995.
- [114] Clarke, J.A.M., *High Strain Rate Tensile Testing*, M.Eng. Thesis, Department of Mechanical and Aerospace Engineering, Carleton University, 1993.
- [115] Safford, N.A., *High Strain Rate Studies*, Ph.D. Thesis, Cambridge University, 1988.
- [116] Wong, B., *Void Growth Modelling in a dynamically Loaded Ductile Material*, Ph.D. Thesis, University of Waterloo, 1990.
- [117] Nahme, H. and Worswick, M.J., Dynamic Properties and Spall Plane Formation of Brass, *DYMAT Journal de Physique IV*, C8, 4, 1994, pp. 707-712.
- [118] Worswick, M.J., Nahme, H. and Fowler, J., Spall through Void Nucleation, Growth and Coalescence, *Dymat Journal de Physique IV*, C8, 4, 1994, pp. 623-628.

- [119] Worswick, M.J., Nahme, H., Clarke, J. and Fowler, J., Ductile Fracture during Tensile Hopkinson Bar and Plate Impact Testing, *DYMAT Journal de Physique*, C, 1993.
- [120] Lindholm, V.S., Review of Dynamic Testing Techniques and Material Behaviour, *Proceedings of the Conference on Mechanical Properties of Materials at High Rates of Strain*, Oxford, 1974, pp. 3-21.
- [121] Karnes, C.H., The Plate Impact configuration for Determining Mechanical Properties of Materials at High Strain Rates, *Proceedings of the Conference on Mechanical Properties of Materials at High Rates of Strain*, Oxford, 1974, pp. 270-293.
- [122] Curran, D.R., Seaman, L. and Shockey, D.A., Dynamic Failure of Solids, *Physics Reports*, 147, 1987, pp. 253-288.
- [123] Dremin, A.N. and Molodets, A.M., in *Shock Compression of Condensed Matter*, 1989, edited by S.C. Schmidt, J.N. Johnson and L.W. Davison, North-Holland, Amsterdam, 1990, pps. 415-420.
- [124] Meyers, M.A. and Aimone, C.T., Dynamic Fracture of Metals, *Progress in Materials Science*, 28, 1983, ppg. 1-96.

- [125] Marsh, S.P., LASL Shock Hugoniot Data, Los Alamos Scientific Laboratory, University of California Press, Berkeley, 1980.
- [126] Metals Handbook, Mechanical Testing, 8, ed. 9, ASM, 1985.
- [127] Von Karman, T. and Duwez, P., *Journal of Applied Physics*, **21**, 1950, p. 987.
- [128] Rolsten, R.F., A study of the shock loading of materials, *Transactions New York Academy of Sciences* **2**, 1974, pp. 416-480.
- [129] Walsh, J.M., Christian, R.H., Equation of State of Metals from Shock Wave Measurements *Physics Reviews* **97/6**, 1955.
- [130] Barker, L.M. and Hollenbach, R.E., Laser Interferometer for Measuring High Velocities of any Reflecting Surface, *Journal of Applied Physics*, **43**, 11, 1972.
- [131] Davison, L., Stevens, A.L., Kipp, M.E., Theory of Spall Damage Accululation in Ductile Metals, *J.Mech.Phys.Solids*, **25**, 1977, pp. 11-28.
- [132] Romenchenko, V.I. and Stepanov, G.V., Dependence of the Critical Stresses on the Loading Time Parameters during Spall in Copper, Aluminum, and Steel, *Journal of Applied Mechanics and Technical Physics*, **21**, 1980, pp. 555-561.

- [133] Curran, D.R., Shockey, D.A. and Seaman, L., Dynamic Failure in Solids, *Journal of Applied Physics*, **44**, 1973, pp. 4025-4038.

TABLES

Table 3.1 : Nominal Composition of grains and phase at grain boundaries

	5505	555	405	525	55	45
CV	45.82	35.45	35.13	35.74	34.44	34.44
Si	55.13	66.35	66.21	55.86	57.21	55.55
FE	1.00	1.00	1.00	1.25	1.13	1.13
SE	1.00	1.00	1.00	1.00	1.00	1.00
TOTAL	100.95	102.80	101.33	99.60	101.63	100.15
BASIS : 1 ATOM%						
CV	1.000	1.040	1.041	1.031	1.002	1.007
Si	1.000	1.066	1.059	1.005	1.006	1.000
FE	1.000	1.000	1.000	1.175	1.102	1.113
SE	1.000	1.000	1.000	1.000	1.000	1.000
	1.000	1.000	1.000	1.000	1.000	1.000

Table 4.1 : Pulse durations for available striker lengths

Striker Length mm (inches)	Pulse Duration (microseconds)
101.6 (4)	39.7
203.2 (8)	79.4
304.8 (12)	119.0
406.4 (16)	158.7

Table 4.2 : Specimen triaxiality

Geometry	Triaxiality
C-Notch	1.026
E-notch	0.612
Uniaxial	0.333 (prior to necking)

Table 6.1 : C-notch specimen data for UNS-C36000 brass
(Triaxiality = 1.026)

Specimen Number	Rod Diameter (inches)	Diametral Strain (m/m)	Failed (No) Trap (F/T/NT)	Test Velocity (m/s)	Nominal Strain Rate (1/s)
850 Anneal					
S75	0.5	0.103	T	8.38	2590
S72	0.5	0.106	T	8.16	2594
S73	0.5	0.130	NT	8.08	2205
S77	0.5	0.244	T	7.55	2341
S76	0.5	0.292	T	8.55	2630
S54	0.5	0.318	T	8.79	3377
S45	0.5	0.325	F	9.02	3582
S85	0.5	0.350	F	9.64	3644
S91	0.5	0.351	F	8.32	2532
S97	3.5	0.431	T	10.47	3334
600 Anneal					
S74	0.5	0.077	T	7.96	2437
S71	0.5	0.093	T	8.91	4218
S60	0.5	0.271	T	9.16	3380
S78	0.5	0.277	T	7.78	2610
S42	0.5	0.311	T	9.20	3424
S53	0.5	0.355	F	9.19	3514
S41	0.5	0.371	F	9.82	3779
S40	0.5	0.391	F	11.09	4347
S90	0.5	0.393	F	9.64	2958
S84	0.5	0.402	F	8.10	2885
S96	3.5	0.473	T	10.61	3257

Table 6.2 : E-notch specimen data for UNS-C36000 brass
(Triaxiality = 0.612)

Specimen Number	Rod Diameter (inches)	Diametral Strain (m/m)	Failed or (No) Trap (F/T/NT)	Test Velocity (m/s)	Nominal Strain Rate (1/s)
850 Anneal					
S50	3.5	0.200	T	11.04	1578
S47	3.5	0.308	NT	11.15	1527
S80	3.5	0.370	T	13.00	1637
S58	3.5	0.454	T	11.85	1672
S34	3.5	0.636	Incipient F	15.15	2096
S66	3.5	0.646	Incipient F	12.93	1826
S87	3.5	0.731	F	15.12	2199
S92	0.5	0.466	F	14.18	2069
S95	0.5	0.506	F	14.48	2040
600 Anneal					
S43	3.5	0.191	T	9.38	1256
S49	3.5	0.201	T	10.96	1541
S44	3.5	0.327	T	11.01	1499
S79	3.5	0.397	T	13.00	1598
S55	3.5	0.449	T	11.48	1611
S65	3.5	0.490	T	12.18	1628
S86	3.5	0.723	F	14.71	2149
S93	0.5	0.497	F	14.33	2113
S94	0.5	0.510	F	14.45	2038

Table 6.3 : E-notch specimen data for MS-58 brass
(Triaxiality = 0.612)

Specimen Number	Rod Diameter (inches)	Diametral Strain (m/m)	Failed or (No) Trap (F/T/NT)	Test Velocity (m/s)	Nominal Strain Rate (1/s)
850 Anneal					
S52	3.0	0.194	T	11.38	1511
S48	3.0	0.336	T	11.22	1478
S64	3.0	0.404	T	13.17	1518
S59	3.0	0.423	T	10.44	1354
S57	3.0	0.628	F	12.06	1673
S32	3.0	0.633	F	15.08	2148
S68	3.0	0.628	F	14.90	2004
600 Anneal					
S51	3.0	0.194	T	10.89	1477
S46	3.0	0.307	T	11.29	1507
S63	3.0	0.393	T	13.00	1545
S56	3.0	0.476	T	10.63	1447
S31	3.0	0.559	F	14.35	1991
S88	3.0	0.601	F	15.18	2188
S67	3.0	0.623	F	14.45	1940

Table 6.4 : Uniaxial tensile specimen data for UNS C36000 brass
(Triaxiality = 0.333)

Specimen Number	Rod Diameter	Diametral Strain	Failed or (No) Trap	Striker Velocity	Nominal Strain Rate
	(inches)	(m/m)	(F/T/NT)	(m/s)	(1/s)
850 Anneal					
S104	0.5	0.151	T	13.44	1512
S82	0.5	0.211	T	19.66	2225
S101	0.5	0.246	T	18.98	2875
S100	3.5	0.297	T	18.87	2867
S70	0.5	0.352	T	20.18	2840
600 Anneal					
S103	0.5	0.151	T	13.47	1544
S81	0.5	0.224	T	18.63	2223
S98	3.5	0.242	T	19.19	2923
S99	0.5	0.266	T	18.88	2751
S69	0.5	0.366	T	20.41	2910

FIGURES

Figure 1.1 : The Processes of ductile fracture

Figure 1.2 : Schematic of void nucleation front progressing through the particle size distribution (Embury [21])

Figure 1.3 : Prolate void aligned with the direction of tensile strain

Figure 1.4 : The effect of tensile deformation on porosity, for iron compacts of various initial porosities (Spitzig et al [46])

Figure 1.5 : The evolution of porosity, normalized by initial porosity, for several iron compacts (Spitzig et al. [46])

Figure 1.6 : Geometric representation of Thomason's model for coalescence
(Thomason [52])

Figure 1.7 : Void growth at (a) low, (b) moderate, and (c) high triaxialities, as described by Sun Jun [54]

Figure 1.8 : Oriented imperfection plane used to model localization (Yamamoto [65])

Initial Distribution

Final Geometry

Figure 1.9 : Eulerian simulation of stress waves from the top of the page propagating through a copper matrix, resulting in void growth (Benson [79])

Figure 1.10 : Eulerian simulation of stress waves from the top of the page propagating through a 4340 steel matrix, resulting in void growth (Benson [79])

Figure 2.1 : Steps of quantitative Metallography

Figure 2.2 : Bimodal grey-scale histogram

Figure 3.1 : Cu-Zn phase diagram

Figure 3.2 : Time-temperature-transformation (TTT) diagram for MS-58 brass

Figure 3.3 : 600°C HT 8.89 cm (3 1/2") brass rod (as polished, 100x)

Figure 3.4 : 850°C HT 8.89 cm (3 1/2") brass rod (as polished, 100x)

Figure 3.5 : 600°C HT 8.89 cm (3 1/2") brass rod (as etched, 200 x)

Figure 3.6 : 850°C HT 8.89 cm (3 1/2") brass rod (as etched, 200x)

Figure 3.7 : 850°C HT 1.27 cm (1/2") brass rod (as etched, 200x)

Figure 3.8 : Particle area histogram for 850°C HT brass

Figure 3.9 : Particle aspect ratio histogram for 850°C HT brass

Figure 3.10 : Particle area histogram for 600°C HT brass

Figure 3.11 : Particle aspect ratio histogram for 600°C HT brass

Figure 3.12 : Quasi-static engineering stress-strain curves

Dynamic Engineering Stress Strain Response

Figure 3.13 : Dynamic engineering stress-strain response

Figure 4.1 : Schematic of the Carleton tensile split Hopkinson bar and stress wave propagation

a) Schematic view

b) Photograph

Figure 4.2 : Incident bar momentum trap

a) Schematic view

b) Photograph

Figure 4.3 : Transmitter bar momentum trap

Specimen Geometry	Minimum Diameter (ϕ)	Shoulder Diameter (Φ)	Shoulder Width (W_s)	Notch Radius (R_n)	Gauge Length (G_o)	Specimen Length (L_o)
C-Notch	0.161	0.375	0.125	0.041	0.082	1.232
E-Notch	0.161	0.500	0.081	0.125	0.250	1.311
Uniaxial	0.125	0.375	0.125	N/A	0.500	1.650

(All Dimensions in Inches)

Figure 4.4 : Notched specimen geometries

Figure 4.5 : Carleton tensile split Hopkinson bar

Figure 4.6 : Gas gun schematic

Figure 4.7 : "Incident bar end" of Hopkinson bar apparatus

Figure 4.8 : Strain rate vs. striker velocity (C-notch data)

Figure 4.9 : Strain rate vs. striker velocity (E-notch data)

Diametral Strain vs. Engineering Strain

(Experimental Data and Nicolet Conversion Curve)

Figure 4.10 : Diametral strain vs. engineering strain (E-notch data)

Diametral Strain vs. Engineering Strain

(Experimental Data and Nicolet Conversion Curve)

Figure 4.11 : Diametral strain vs. engineering strain (C-notch data)

Figure 5.1 (a) : Flyer plate impact geometry and stress wave propagation

Figure 5.1 (b) : Flyer plate impact geometry and stress wave propagation

Figure 5.2 : Wave propagation in a plate impact test (Karnes [120])

Figure 5.3 : Standard Hugoniot for a metal (Rolsten [127])

Figure 5.4 : Typical free-surface velocity data from VISAR

Figure 5.5 : Schematic of Ernst-Mach-Institute light gas gun test section

Incident and Transmitter Bar Waves

Figure 6.1 : Typical incident and transmitter bar waves for interrupted test

Figure 6.2 : Complete Hopkinson bar data set for E-notch specimen

Figure 6.3 : Complete Hopkinson bar data set for a C-notch specimen

Figure 6.4 : Complete Hopkinson bar data set for a uniaxial specimen

Figure 6.5 : True stress and true strain time histories for E-notch

Figure 6.6 : Uniaxial stress-strain response for the (a) 850 HT, and (b) 600 HT UNS-C36000

Figure 6.8 : True stress-strain data for MS-58, E-notch specimens for (a) 850 HT and (b) 600 HT

Figure 6.9 : True stress-strain response for a C-notch of (a) 850 HT, and (b) 600 HT
UNS-C36000

Figure 6.10 : Comparison of stress-strain response between UNS-C36000 and MS-58 brass for (a) 850 HT, and (b) 600 HT

Figure 6.11 : True stress-strain response for notches of various triaxialities with 850 HT for (a) 1.27 cm (0.5 inch), and (b) 8.89 cm (3.5 inch) brass rod

Figure 6.12 : True stress-strain response for notches of various triaxialities with 600 HT for (a) 1.27 cm (0.5 inch), and (b) 8.89 cm (3.5 inch) brass rod

Failure Strain vs. Triaxiality

Figure 6.13 : Ductility of brass as a function of triaxiality

Figure 6.14 : Contour plots of (a) initial porosity distribution, (b) approximate location of measured fields with respect to notch geometry, (c-e) contours of porosity in uniaxial specimens with increasing strain

Figure 6.15 : Contour plots of aspect ratio evolution with strain in uniaxial specimens

(a) $e = 0.200$ [m/m]

(b) $e = 0.308$ [m/m]

(c) $e = 0.454$ [m/m]

(d) $e = 0.636$ [m/m]

Figure 6.16 : Contour plots of porosity in E-notch specimens

Figure 6.17 : Contour plots of aspect ratio in E-notch specimens

Figure 6.18 : Contour plots of porosity evolution in C-notch specimens

Figure 6.19 : Contour plots of mean aspect ratio evolution in C-notch

Figure 6.20 : Porosity evolution with strain for the uniaxial specimens

Figure 6.21 : Porosity evolution with strain for the E-notch specimens

Figure 6.22 : Porosity evolution with strain for the C-notch specimens

Figure 6.23 : Comparison of the porosity evolution with strain between various specimen geometries

Figure 6.24 : Void growth rate as a function of triaxiality at specimen centres

Figure 6.25 : Mean void area evolution with strain for the uniaxial specimens

Figure 6.26 : Mean void area evolution with strain for the E-notch specimens

Figure 6.27 : Mean void area evolution with strain for the C-notch specimens

Figure 6.28 : Mean void aspect ratio evolution with strain for the uniaxial specimens

Mean Void Aspect Ratio vs. Diametral Strain
(E-Notch Central Region)

Mean Void Aspect Ratio vs. Diametral Strain
(E-Notch Edge Region)

Figure 6.29 : Mean void aspect ratio evolution with strain for the E-notch specimens

Mean Void Aspect Ratio vs. Diametral Strain
(C-Notch Central Region)

Mean Void Aspect Ratio vs. Diametral Strain
(C-Notch Edge Region)

Figure 6.30 : Mean void aspect ratio evolution with strain for the C-notch specimens

Figure 6.31 : Comparison of mean void aspect ratio for various specimen geometries

Figure 6.32 : Micrograph of central crack in a E-notch specimen
(as polished, 200x)

Figure 6.33 : Classic cup-cone failure in brass E-notch

Figure 6.34 : Micrograph of crack tip in E-notch
(as polished, 200x)

Figure 6.35 : SEM micrograph of 850 HT E-notch fracture surface

Figure 6.36 : SEM micrograph of 600 HT E-notch fracture surface

Figure 6.37 : SEM micrograph of 850 HT C-notch fracture surface

Figure 6.38 : SEM micrograph of 600 HT C-notch fracture surface

Figure 6.39 : VISAR velocity profiles for 850 HT plates

Figure 6.40 : VISAR velocity profiles for 600 HT plates

Figure 6.41 : VISAR velocity profiles for "as-received" MS-58 plates

Figure 6.42 : VISAR velocity profiles for various material stocks of similar flyer plate velocity

Figure 6.43 : Contour plot of void volume fraction
at the centre of the 93 m/s plate

Figure 6.44 : Distribution of damage in the spall layer of impacted plates near their outer edge

Figure 6.45 : Micrographs of the incipient spall layer in plate impact specimens
(as polished, 200x)