DEVELOPMENT OF HIGH STRAIN RATE MECHANICAL TESTING FOR METALLIC MATERIALS

Michael John Cox

August 2013

Mechanics of Materials
Department of Mechanical Engineering
Imperial College London

A thesis submitted for the Diploma of Imperial College and the Degree of Doctor of Philosophy from Imperial College
Copyright Declaration
The copyright of this thesis rests with the author and is made available under a Creative Commons Attribution Non-Commercial No Derivatives licence. Researchers are free to copy, distribute or transmit the thesis on the condition that they attribute it, that they do not use it for commercial purposes and that they do not alter, transform or build upon it. For any reuse or redistribution, researchers must make clear to others the licence terms of this work.

Declaration of Originality
The work in this thesis is the own work of Michael John Cox and that all else is appropriately referenced.
Abstract

Recent developments of high strain rate servo-hydraulic systems and high speed video imaging equipment have made the simple tensile test covering the quasi-static to dynamic range possible. However, obtaining reliable material data from the raw data produced from such tests requires a good understanding of the unique set of problems this testing technique can present.

Tensile tests covering the quasi-static to 760 s⁻¹ strain rate range were performed on Oxygen Free Electrolytic (OFE) copper, Al 6061-T6 and a Ta-2.5%W alloy. Modified standard sized tensile specimens were used in all tests and evaluations of the higher strain rate tests were carried out to understand the specimen dynamics at these high strain rates. Digital Image Correlation was used to measure strain at the higher strain rates and was ideal as a non-contact extensometer and could provide an indication if dynamic equilibrium is maintained throughout the test. The work strongly suggests that each material and specimen geometry will have its own strain rate threshold at which stress equilibrium is maintained. Appropriate methods were also necessary in processing the raw dynamic output to extract meaningful material data from the tests.

Data obtained from the tests were successful in evaluating the materials behaviour over the quasi-static to dynamic strain rate range. The materials responded in a typical manner to that expected of their crystal structure and stacking fault energy, agreeing with results available from open literature. The tests performed in tension were compared with tests carried out in compression and showed the strain rate sensitivity in tension did not differ substantially to that in compression.

Three constitutive material models were assessed, the Johnson-Cook (J-C) model was found to represent the experimental results of the OFE Cu and Al 6061-T6 materials well, but did not give such a good fit to the Ta-2.5%W material. The Zerilli-Armstrong (Z-A) model provided a good fit to Ta-2.5%W but not the OFE Cu and Al 6061-T6 materials. No satisfactory fit was achieved using the Mechanical Threshold Stress (MTS) model.
Acknowledgements

I am in debt to my academic supervisor Professor John Dear (Imperial College) for giving me the opportunity to undertake this research and his support with all aspects of my work at Imperial College. My sincerest thanks also go to Mr. Alex Worley (Imperial College), Dr. Paul Hooper (Imperial College), Dr. Ruth Brooker (Imperial College), for their help and assistance in using the high strain rate equipment at Imperial College. I would also like to thank Dr. Joe Punni (AWE), Dr. Paul Ryan (AWE), and the late Mr. Mike Parsley (AWE) for assisting in some of the metallography, Mr. Phil Mallard (AWE) for help with the Finite Element Analysis, and Tom Dunnet (Imperial College) for some DIC measurements. Dr. Garry Burnell (AWE), Dr James Talbot and Dr. Giles Aldrich-Smith (AWE) for proof reading this thesis. Finally, I would like to acknowledge AWE plc in financing this work.
# Contents

<table>
<thead>
<tr>
<th>Section</th>
<th>Page</th>
</tr>
</thead>
<tbody>
<tr>
<td>ABSTRACT</td>
<td>3</td>
</tr>
<tr>
<td>ACKNOWLEDGEMENTS</td>
<td>4</td>
</tr>
<tr>
<td>CONTENTS</td>
<td>5</td>
</tr>
<tr>
<td>LIST OF FIGURES</td>
<td>9</td>
</tr>
<tr>
<td>LIST OF TABLES</td>
<td>17</td>
</tr>
<tr>
<td>NOMENCLATURE</td>
<td>20</td>
</tr>
<tr>
<td>ACRONYMS AND ABBREVIATIONS</td>
<td>23</td>
</tr>
<tr>
<td>1. INTRODUCTION</td>
<td>25</td>
</tr>
<tr>
<td>1.1 Aims and Objectives</td>
<td>25</td>
</tr>
<tr>
<td>1.2 Thesis Outline</td>
<td>26</td>
</tr>
<tr>
<td>2. LITERATURE REVIEW</td>
<td>29</td>
</tr>
<tr>
<td>2.1 Mechanical Testing Techniques</td>
<td>29</td>
</tr>
<tr>
<td>2.1.1 Tensile Testing</td>
<td>30</td>
</tr>
<tr>
<td>2.1.2 Split Hopkinson Pressure Bar Test</td>
<td>33</td>
</tr>
<tr>
<td>2.1.3 Dynamic Tensile Testing</td>
<td>35</td>
</tr>
<tr>
<td>2.2 Mechanical Behaviour of Metallic Materials</td>
<td>41</td>
</tr>
<tr>
<td>2.2.1 Effect of Microstructure</td>
<td>41</td>
</tr>
<tr>
<td>2.2.2 Measurement of Strain Rate Sensitivity in a Material</td>
<td>45</td>
</tr>
<tr>
<td>2.2.3 Influence of Stress State or Strain State</td>
<td>46</td>
</tr>
<tr>
<td>2.3 Constitutive Modelling</td>
<td>47</td>
</tr>
<tr>
<td>2.3.1 Development of Constitutive Models</td>
<td>47</td>
</tr>
<tr>
<td>2.3.2 Johnson-Cook Equation</td>
<td>48</td>
</tr>
<tr>
<td>2.3.3 Zerilli-Armstong Equation</td>
<td>49</td>
</tr>
<tr>
<td>2.3.4 Mechanical Threshold Stress Model</td>
<td>50</td>
</tr>
<tr>
<td>2.4 Conclusions</td>
<td>56</td>
</tr>
<tr>
<td>3. EVALUATION OF DYNAMIC TENSILE TESTING</td>
<td>59</td>
</tr>
<tr>
<td>3.1 Introduction</td>
<td>59</td>
</tr>
<tr>
<td>3.2 Experimental Procedure</td>
<td>59</td>
</tr>
<tr>
<td>3.2.1 Tensile Test Sample</td>
<td>59</td>
</tr>
</tbody>
</table>
3.2.2 Test Equipment
3.2.3 Load Measurement
3.2.4 Strain Measurement
3.2.5 Determination of Dynamic Stress Equilibrium
3.2.6 Assessment of Load Measurement and Data Processing Methods
3.2.7 Evaluation of DIC as a Measurement and Analytical Technique
3.3 Results and Discussion
3.3.1 Load Signal and Dynamic Equilibrium
3.3.2 Processing and Evaluating the Stress-Strain Curve
3.3.3 Assessment of Strain Measurement by High Speed Imaging
3.3.4 Analysis of the DIC Technique
3.4 Conclusions
4. EFFECT OF STRAIN RATE AND TEMPERATURE
4.1 Introduction
4.2 Experimental Procedure
4.2.1 Test Materials
4.2.2 Tensile Test Sample Preparation
4.2.3 Tension Testing Machines
4.2.4 Load Measurement
4.2.5 Strain Measurement
4.2.6 Environmental Chamber
4.2.7 Experimental Test Conditions
4.2.8 Metallography/Fractography
4.3 Results
4.3.1 Aluminium 6061-T6 Alloy
4.3.2 OFE Copper
4.3.3 Ta-2.5% wt % W Wrought Alloy
4.3.4 Ta-2.5 wt % HIP Alloy
4.4 Discussion
4.4.1 Effect of Crystal Structure
4.4.2 Effect of Tantalum-Tungsten Manufacture Route
4.5 Conclusions
5. COMPARISON OF TENSILE AND COMPRESSION TESTS

5.1 Introduction

5.2 Experimental Procedure

5.2.1 Compression Test Sample Preparation
5.2.2 Quasi-Static compression Testing Method
5.2.3 High Rate Compression Test Method

5.3 Results

5.3.1 Aluminium 6061-T6 Alloy
5.3.2 Tantalum-2.5 wt % Tungsten Wrought Alloy
5.3.3 Tantalum-2.5 wt % Tungsten HIP Alloy

5.4 Discussion

5.5 Conclusions

6. CONSTITUTIVE MODELLING

6.1 Introduction

6.2 Experimental Procedure

6.2.1 Determination of Material Model Parameters
6.2.2 Degree of Model Fit

6.3 Results

6.3.1 Johnson-Cook Calculated Curves
6.3.2 Zerilli-Armstrong Calculated Curves
6.3.3 Mechanical Threshold Stress Calculated Curves

6.4 Discussion

6.4.1 Evaluation of the J-C Model for FCC Metals/Alloys
6.4.2 Evaluation of the J-C Model for BCC Metals/Alloys
6.4.3 Evaluation of the Z-A Model for FCC Metals/Alloys
6.4.4 Evaluation of the Z-A Model for BCC Metals/Alloys
6.4.5 Evaluation of the MTS Model

6.5 Conclusions

7. CONCLUSIONS AND FURTHER WORK

7.1 Introduction

7.2 Evaluation of dynamic tensile testing
7.3 Effect of strain rate and temperature
7.4 Comparison of tensile and compression tests
<table>
<thead>
<tr>
<th>Section</th>
<th>Title</th>
<th>Page</th>
</tr>
</thead>
<tbody>
<tr>
<td>7.5</td>
<td>Constitutive modelling</td>
<td>178</td>
</tr>
<tr>
<td>7.6</td>
<td>Conclusions and further work</td>
<td>179</td>
</tr>
<tr>
<td></td>
<td>REFERENCES</td>
<td>181</td>
</tr>
<tr>
<td></td>
<td>APPENDIX</td>
<td>187</td>
</tr>
</tbody>
</table>
List of Figures

Figure 2-1  Typical strain rates covered by the various testing systems.  29
Figure 2-2  Determination of 0.2% proof strength.  31
Figure 2-3  SHPB equipment set-up.  33
Figure 2-4  Examples of system ringing in dynamic tensile experiments.  38
Figure 2-5  Face centred cubic crystal lattice, showing the \{111\} plane of highest atomic packing density.  42
Figure 2-6  Body centred cubic crystal lattice showing the lack of any truly close-packed plane, although the \{110\} planes have the highest packing atomic density in the BCC structure.  44
Figure 2-7  Hexagonal close packed crystal lattice showing the \{0001\} planes of greatest atomic packing density which are few in number in a unit cell.  45
Figure 2-8  Hall-Petch strengthening is limited by the size of dislocations. Once the grain size reaches about 10 nm, grain boundaries start to slide.  45
Figure 2-9  Description of the $\hat{\sigma}_\alpha$ and $\theta_0$ parameters from the strain curve.  52
Figure 2-10 Fisher plot based on equation (2.36).  55
Figure 2-11 Fisher plot based on equation (2.40).  56
Figure 3-1  Tensile sample geometry.  60
Figure 3-2  Loss motion device for intermediate and high strain rate testing.  61
Figure 3-3  General overview of the experimental configuration for high strain rate testing using the DIC method.  63
Figure 3-4  Detail of the test sample area for high strain rate testing using the DIC method.  64
Figure 3-5  Load history trace of the metallic material tested at 760 s$^{-1}$.  66
Figure 3-6  DIC images of double necking.  68
Figure 3-7  Load-time curve, showing a double peak.  69
Figure 3-8  FEA images of double necking.  71
Figure 3-9  A typical example of stress as a function of strain curves provided by a load cell and strain gauge for OFE copper.  72
Figure 3-10 Example of the determination of stress at 0.2% off-set strain for OFE copper.

Figure 3-11 Example of engineering stress as a function of engineering plastic strain for data acquired and processed by four different methods for OFE copper.

Figure 3-12 Example of engineering stress as a function of engineering strain for data acquired using an extensometer and by the imaging of the specimen by two synchronised high speed cameras (strain rate 0.675 s⁻¹).

Figure 3-13 A series of DIC images of an aluminium 6061 alloy tensile specimen over the duration of a tensile test.

Figure 3-14 True stress-true strain curves showing the projection of the curves beyond the instability point using the Bridgeman correction factor.

Figure 4-1 True stress-true plastic strain curves for Al 6061-T6 alloy at a strain rate of 0.001 s⁻¹.

Figure 4-2 True stress-true plastic strain curves for Al 6061-T6 alloy at a strain rate of 0.675 s⁻¹.

Figure 4-3 True stress-true plastic strain curves for Al 6061-T6 alloy at a strain rate of 760 s⁻¹.

Figure 4-4 Average true stress-true plastic strain curves for Al 6061-T6 alloy at different strain rate at a test temperature of 294 K.

Figure 4-5 Average true stress-true plastic strain curves for Al 6061-T6 alloy at different strain rate at a test temperature of 203 K.

Figure 4-6 True stress as a function of temperature for Al 6061-T6 alloy at strain rates of 0.001 s⁻¹ and 280 s⁻¹.

Figure 4-7 True strain as a function of temperature for Al 6061-T6 at strain rates of 0.001 s⁻¹ and 280 s⁻¹.

Figure 4-8 True stress as a function of strain rate for Al 6061-T6 alloy at a temperature of 294 K.

Figure 4-9 True plastic strain as a function of strain rate for Al 6061-T6.
Figure 4-10 Fractography of a typical Al 6061 tensile specimen showing the deformation that takes place in the neck and fracture area (strain rate of 760 s\(^{-1}\)).

Figure 4-11 FEG SEM EDX map of a typical Al 6061-T6 tensile specimen fracture surface showing the cup and cone structure characteristic of ductile fracture.

Figure 4-12 True stress-true plastic strain curve of OFE copper at various strain rates and a test temperature of 294 K.

Figure 4-13 True stress-true plastic strain curve of OFE copper at various strain rates and a test temperature of 203 K.

Figure 4-14 Average true stress-true plastic strain curves for OFE copper at different test strain rates and temperatures.

Figure 4-15 True stress as a function of temperature for OFE copper.

Figure 4-16 True plastic strain as a function of temperature for OFE copper.

Figure 4-17 True stress as a function of strain rate for OFE copper.

Figure 4-18 True plastic strain as a function of strain rate for OFE copper.

Figure 4-19 Fractography of a typical OFE copper tensile specimen showing the deformation that takes place in the neck and fracture area (strain rate of 760 s\(^{-1}\)).

Figure 4-20 True stress-true plastic strain curves for Ta-2.5% W wrought alloy at a strain rate of 0.001 s\(^{-1}\).

Figure 4-21 True stress-true plastic strain curves for Ta-2.5% W wrought alloy at a strain rate of 0.675 s\(^{-1}\).

Figure 4-22 True stress-true plastic strain curves for Ta-2.5% W wrought alloy at a strain rate of 760 s\(^{-1}\).

Figure 4-23 Average true stress-true plastic strain curves for Ta-2.5% W wrought alloy at various strain rates and test temperatures.

Figure 4-24 EBSD images of the Ta-2.5% W wrought alloy plate material showing variation in the grain morphology with respect to spatial direction.

Figure 4-25 True stress as a function of temperature for Ta-2.5% W wrought alloy.
Figure 4-26  True plastic strain as a function of temperature for Ta-2.5% W wrought alloy.

Figure 4-27  True stress as a function of strain rate for Ta-2.5% W wrought alloy tested at 294 K.

Figure 4-28  True plastic strain as a function of strain rate for Ta-2.5% W wrought alloy.

Figure 4-29  Fractography of a Ta-2.5% W wrought alloy tensile specimen showing the degree of anisotropic deformation that takes place in the neck and fracture area.

Figure 4-30  Image of the Ta-2.5% W wrought alloy tensile specimen neck region showing deformation elongation of grains (Strain rate 760/s).

Figure 4-31  Image of the Ta-2.5% W wrought alloy tensile specimen neck region near to the fracture surface showing extreme grain deformation and recrystallisation (Strain rate 760 s⁻¹).

Figure 4-32  True stress-true plastic strain curves for Ta-2.5% W HIP alloy at different strain rates and temperatures.

Figure 4-33  Average true stress-true plastic strain curves for Ta-2.5 wt % W HIP alloy at different strain rates and temperatures.

Figure 4-34  True stress as a function of temperature for Ta-2.5 wt % W HIP alloy at a strain rate of 0.001 s⁻¹.

Figure 4-35  True plastic strain as a function of temperature for Ta-2.5% W HIP alloy at a strain rate of 0.001 s⁻¹.

Figure 4-36  True stress as a function of strain rate for Ta-2.5% W HIP alloy at a test temperature of 294 K.

Figure 4-37  True plastic strain as a function of strain rate for Ta-2.9% W HIP alloy at a test temperature of 294 K.

Figure 4-38  EBSD images of the Ta-2.5% W HIP alloy material showing the equiaxed grain structure.

Figure 4-39  Fractography of a Ta-2.5% HIP alloy tensile specimen showing the isotropic deformation of the neck and fracture area.
Figure 4-40  Difference in true stress-true plastic strain curves for copper, Al 6061-T6 and Ta-2.5% W wrought alloys with respect to strain rate.

Figure 4-41  Difference in true stress-true plastic strain curves for copper, Al 6061-T6 and Ta-2.5% W wrought alloys with respect to test temperature.

Figure 4-42  Difference in true stress-true plastic strain curves for wrought and HIP Ta-2.5% W alloys

Figure 5-1  Experimental set-up for quasi-static compression tests.

Figure 5-2  SHPB equipment test set-up.

Figure 5-3  Cooling curve for the Ta-2.5% W plate SHPB test piece.

Figure 5-4  Quasi-static compression true stress-true strain curves for Al 6061-T6.

Figure 5-5  True stress-true plastic strain curves for Al 6061-T6 for quasi-static compression tests.

Figure 5-6  True stress as a function of temperature and true strain for Al 6061-T6 at a 0.001 s$^{-1}$ strain rate.

Figure 5-7  True stress as a function of true strain for Al 6061-T6 at a 2450 s$^{-1}$ strain rate on SHPB.

Figure 5-8  Example of the determination of 0.2% off-set stress and polynomial curve fit for Al 6061-T6 at a 2450 s$^{-1}$ strain rate on SHPB.

Figure 5-9  True stress as a function of true plastic strain for Al 6061-T6 at a 2450 s$^{-1}$ strain rate on SHPB.

Figure 5-10 True stress as a function of temperature and true strain for Al 6061-T6 at a 2450 s$^{-1}$ strain rate.

Figure 5-11 True yield stress as a function of strain rate for Al 6061-T6.

Figure 5-12 Quasi-static compression true stress-true strain curves for Ta-2.5% W wrought alloy.

Figure 5-13 Quasi-static compression true stress-true plastic strain curves for Ta-2.5% W wrought alloy.
Figure 5-14  True stress-true as a function of temperature and true plastic strain for Ta-2.5% W wrought alloy at a 0.001 s\(^{-1}\) strain rate.

Figure 5-15  True stress-true as a function of true strain for Ta-2.5% W wrought alloy at a 2200 s\(^{-1}\) strain rate.

Figure 5-16  Average true stress-true as a function of true plastic strain curves for Ta-2.5% W wrought alloy at a 2200 s\(^{-1}\) strain rate.

Figure 5-17  True stress-true as a function of temperature and true plastic strain for Ta-2.5% W wrought alloy at a 2200 s\(^{-1}\) strain rate.

Figure 5-18  True yield stress as a function of strain rate for Ta-2.5% W wrought alloy.

Figure 5-19  Quasi-static compression true stress-true strain curves for Ta-2.5% W HIP alloy.

Figure 5-20  Quasi-static compression true stress-true plastic strain curves for Ta-2.5% W HIP alloy.

Figure 5-21  True stress a function of temperature and true strain for Ta-2.5% W HIP alloy at a 0.001 s\(^{-1}\) strain rate.

Figure 5-22  True stress a function of temperature and true strain for Ta-2.5% W HIP alloy at a 2090 s\(^{-1}\) strain rate on SHPB.

Figure 5-23  True stress a function of true plastic strain for Ta-2.5% W HIP alloy at a 2090 s\(^{-1}\) strain rate on SHPB.

Figure 5-24  True stress a function of temperature and true plastic for Ta-2.5% W HIP alloy at a 2090 s\(^{-1}\) strain rate on SHPB.

Figure 5-25  True yield stress as a function of strain rate for Ta-2.5% W HIP alloy.

Figure 5-26  Comparison of true stress-true plastic strain curves at quasi-static and dynamic strain rates in tension and compression for the Al 6061-T6 alloy.

Figure 5-27  Comparison of true stress-true plastic strain curves at quasi-static and dynamic strain rates in tension and compression for the Ta-2.5%W wrought alloy.

Figure 5-28  Comparison of true stress-true plastic strain curves at quasi-static and dynamic strain rates in tension and compression for the Ta-2.5%W HIP alloy.
Figure 6-1  Comparison of true stress-true strain curves as calculated using the original J-C equation with the experimentally determined average true stress-true strain tensile curves for OFE copper.

Figure 6-2  Comparison of true stress-true plastic strain curves as calculated using a reference temperature in the J-C equation with the experimentally determined true stress-true tensile strain curves for OFE copper.

Figure 6-3  Comparison of true stress-true plastic strain curves as calculated using the original J-C equation with the experimentally determined true stress-true plastic tensile strain curves for Al 6061-T6.

Figure 6-4  Comparison of true stress-true plastic strain curves as calculated using a reference temperature in the J-C equation with the experimentally determined true stress-true plastic tensile strain curves for Al 6061-T6.

Figure 6-5  Comparison of true stress-true plastic strain curves as calculated using the original J-C equation with the experimentally determined true stress-true plastic tensile strain curves for Ta-2.5%W wrought alloy.

Figure 6-6  Comparison of true stress-true plastic strain curves as calculated using a reference temperature in the J-C equation with the experimentally determined true stress-true plastic tensile strain curves for Ta-2.5%W wrought alloy.

Figure 6-7  Comparison of true stress-true plastic strain curves as calculated using a reference temperature in the J-C equation with the experimentally determined true stress-true plastic tensile strain curves for Ta-2.5%W HIP alloy.

Figure 6-8  Comparison of true stress-true plastic strain curves as calculated using the original Zerilli-Armstrong equation with the experimentally determined true stress-true tensile plastic strain curves for OFE copper.
Figure 6-9  Comparison of true stress-true plastic strain curves as calculated using the original Zerilli-Armstrong equation with the experimentally determined true stress-true tensile plastic strain curves for Al 6061-T6.

Figure 6-10  Comparison of true stress-true plastic strain curves as calculated using the original Zerilli-Armstrong equation with the experimentally determined true tensile stress-true plastic strain curves for Al 6061-T6 (C₀ pinned at experimental value for 0.001 s⁻¹).

Figure 6-11  Comparison of true stress-true plastic strain curves as calculated using the Zerilli-Armstrong equation with the experimentally determined true stress-true plastic tensile strain curves for Ta-W wrought alloy.

Figure 6-12  Comparison of true stress-true plastic strain curves as calculated using the Zerilli-Armstrong equation with the experimentally determined true stress-true plastic tensile strain curves for Ta-W HIP alloy.

Figure 6-13  Comparison of true stress-true plastic strain curves as calculated using the Zerilli-Armstrong equation with the experimentally determined true stress-true plastic compression strain curves for Ta-W wrought alloy.

Figure 6-14  Comparison of true stress-true plastic strain curves as calculated using the Zerilli-Armstrong equation with the experimentally determined true stress-true plastic compression strain curves for Al 6061-T6.

Figure 6-15  Fisher plot used to calculate the structure evolution parameters.

Figure 6-16  Fisher plot used to calculate the thermal activation parameters.
List of Tables

<table>
<thead>
<tr>
<th>Table</th>
<th>Description</th>
<th>Page</th>
</tr>
</thead>
<tbody>
<tr>
<td>Table 3-1</td>
<td>Load/time trace from load cell readings.</td>
<td>66</td>
</tr>
<tr>
<td>Table 3-2</td>
<td>Comparison of stress data by four different methods.</td>
<td>74</td>
</tr>
<tr>
<td>Table 3-3</td>
<td>Comparison of strain readings obtained from extensometer and DIC.</td>
<td>76</td>
</tr>
<tr>
<td>Table 3-4</td>
<td>Standard deviation of the DIC strain readings with those obtained by an extensometer (Ta-2.5% W alloy at $1 \times 10^{-3}$ s$^{-1}$).</td>
<td>76</td>
</tr>
<tr>
<td>Table 4-1</td>
<td>Typical mechanical properties of supplied materials.</td>
<td>84</td>
</tr>
<tr>
<td>Table 4-2</td>
<td>Surface roughness measurements from tensile specimens prepared from Ta-2.5%W, Al 6061-T6 wrought plates and OFE Cu rod.</td>
<td>86</td>
</tr>
<tr>
<td>Table 4-3</td>
<td>Mechanical properties of Al 6061-T6.</td>
<td>92</td>
</tr>
<tr>
<td>Table 4-4</td>
<td>Strain rate sensitivity of true yield stress and true UTS for Al 6061-T6 at 294 K.</td>
<td>95</td>
</tr>
<tr>
<td>Table 4-5</td>
<td>Mechanical properties of OFE copper.</td>
<td>99</td>
</tr>
<tr>
<td>Table 4-6</td>
<td>Strain rate sensitivity of true yield stress and true UTS for OFE copper at 294 K.</td>
<td>101</td>
</tr>
<tr>
<td>Table 4-7</td>
<td>Grain structure of the Ta-2.5%W wrought alloy.</td>
<td>107</td>
</tr>
<tr>
<td>Table 4-8</td>
<td>Mechanical properties of Ta-2.5%W wrought alloy.</td>
<td>107</td>
</tr>
<tr>
<td>Table 4-9</td>
<td>Strain rate sensitivity of true yield stress and true UTS for Ta-2.5%W wrought alloy at 294 K.</td>
<td>108</td>
</tr>
<tr>
<td>Table 4-10</td>
<td>Mechanical properties of Ta-2.5%W HIP alloy.</td>
<td>113</td>
</tr>
<tr>
<td>Table 4-11</td>
<td>Strain rate sensitivity of true yield stress and true UTS for Ta-2.5%W HIP alloy at 294 K.</td>
<td>114</td>
</tr>
<tr>
<td>Table 4-12</td>
<td>Grain structure of the Ta-2.5%W HIP alloy.</td>
<td>117</td>
</tr>
<tr>
<td>Table 4-13</td>
<td>Comparison of strain rate sensitivity values for the wrought metals tested over the 0.001 s$^{-1}$ to 760 s$^{-1}$ strain rate range.</td>
<td>120</td>
</tr>
<tr>
<td>Table 4-14</td>
<td>Comparison of strain rate sensitivity values for the wrought and HIP Ta-2.4%W processed materials.</td>
<td>122</td>
</tr>
<tr>
<td>Table 5-1</td>
<td>Average true stress as a function strain for Al 6061-T6 alloy at a strain rate of 0.001 s$^{-1}$.</td>
<td>131</td>
</tr>
<tr>
<td>Table 5-2</td>
<td>Average true stress as a function strain for Al 6061-T6 alloy at a strain rate of 2540 s$^{-1}$.</td>
<td>134</td>
</tr>
</tbody>
</table>
Table 5-3  Strain rate sensitivity of the true yield stress for Al 6061-T6 alloy.
Table 5-4  Average true stress as a function strain for Ta-2.5%W wrought alloy at a strain rate of 0.001 s\(^{-1}\).
Table 5-5  Average true stress as a function strain for Ta-2.5%W wrought alloy at a strain rate of 2200 s\(^{-1}\).
Table 5-6  Strain rate sensitivity of the true yield stress for Ta-2.5%W wrought alloy.
Table 5-7  Average true stress as a function strain for Ta-2.5%W HIP alloy at a strain rate of 0.001 s\(^{-1}\).
Table 5-8  Average true stress as a function strain for Ta-2.5%W HIP alloy at a strain rate of 2090 s\(^{-1}\).
Table 5-9  Strain rate sensitivity of the true yield stress for Ta-2.5%W HIP alloy.
Table 5-10 Summary of Al 6061-T6 and Ta-2.5%W alloys strain rate sensitivity in tension and compression.
Table 6-1  Johnson-Cook parameters determined from experimental tension data.
Table 6-2  Degree of fit for the true stress true plastic strain curves calculated using the J-C equation from experimental tension data.
Table 6-3  Zerilli-Armstrong parameters determined from experimental data.
Table 6-4  Degree of fit for the true stress true plastic strain curves calculated using the Z-A equation from experimental tension data.
Table 6-5  Degree of fit for the true stress true plastic strain curves calculated using the Z-A equation from experimental compression data.
Table 6-6  Comparison of J-C model constants determined from experimental tension data for Al 6061-T6 with those obtained from literature.
| Table 6-7 | Comparison of J-C model constants determined from experimental tension data for OFE copper with those obtained from literature. |
| Table 6-8 | Comparison of J-C model constants determined from experimental tension data for wrought Ta-2.5%W with those obtained from literature. |
| Table 6-9 | Z-A constants established for OFE copper with those in open literature. |
| Table 6-10 | Z-A constants established for Ta-2.5%W with those in open literature. |
Nomenclature

**English Alphabet**

\( A_b \) Pressure bar diameter (SHPB)
\( A_F \) Percentage reduction in area on failure
\( A_f \) Tensile sample’s final gauge area
\( A_0 \) Initial cross sectional area of a sample’s gauge length
\( b \) Magnitude of the Burgers vector
\( C_m \) Elastic wave velocity in a material
\( C_p \) Specific heat capacity of the material
\( C_{p1} \) and \( C_{p2} \) Constants
\( C_{p0} \) Constant
\( C_{0i} \) Constants in the Z-A model
\( C_1 \) Constants in the Z-A model
\( C_2 \) Constants in the Z-A Model
\( C_3 \) Constants in the Z-A model
\( C_4 \) Constants in the Z-A model
\( C_5 \) Constants in the Z-A model
\( D_0 \) Material constant
\( d \) Mean grain size diameter
\( d\sigma \) Incremental change in stress
\( d\varepsilon \) Incremental change in strain
\( E \) Young’s modulus
\( g_{0i} \) Normalised activation energies
\( g_{0c} \) Normalised activation energies
\( K \) Constant
\( k \) Boltzmann constant
\( L \) Specimen Length
\( l_f \) Tensile sample’s gauge length at failure
\( l_{\text{min}} \) Minimum initial parallel gauge length of the specimen
\( l_0 \) Initial gauge length of the tensile specimen
\( M \) Thermal softening fraction
\( m \) Strain rate sensitivity

\( N \) Number of round-trips a stress wave makes in a sample during a test

\( n \) Constant in the Z-A model

\( P \) Applied load

\( p_i \) Constants describing the dislocation glide resistance profile (MTS model)

\( q_i \) Constants describing the dislocation glide resistance profile (MTS model)

\( R \) Radius of curvature of the neck in a tensile sample

\( r_n \) Radius of the thinnest part of the neck in a tensile sample

\( S_i \) and \( S_\varepsilon \) Arrhenius form for a temperature scaling factors, specifying the ratio between the applied stress and the mechanical threshold stress (MTS model).

\( T \) Test temperature

\( T^* \) Homologous temperature

\( T_m \) Melting point of the metal/alloy being tested

\( T_0 \) Material constant

\( T_r \) Reference temperature

\( t_r \) Load rising time

\( V \) Velocity of the tensile machine’s actuator

\( V_{\text{max}} \) Maximum speed of the tensile machine’s actuator

**Greek alphabet**

\( \Delta l \) Change in the tensile sample gauge length

\( \varepsilon_e \) Engineering strain

\( \varepsilon_F \) Percentage elongation to failure

\( \varepsilon_i \) Incident strain (SHPB)

\( \varepsilon_t \) True strain

\( \varepsilon_r \) Reflected strain (SHPB)

\( \varepsilon_T \) Transmitted strain (SHPB)

\( \dot{\varepsilon} \) Strain rate

\( \dot{\varepsilon}_{\text{max}} \) Maximum strain rate

\( \dot{\varepsilon}_0 \) Reference strain rate

\( \varepsilon^* \) Dimensionless plastic strain term
\( \dot{\varepsilon}_{0i} \) Reference strain rate (MTS model)
\( \dot{\varepsilon}_{0c} \) Reference strain rate (MTS model)
\( \dot{\varepsilon}_{0ss} \) (MTS model)
\( \theta_0 \) Initial slope of the stress-strain curve at the moment of yielding (MTS model)
\( \kappa \) Non-dimensional parameter that characterises the shape of the hardening law (MTS model)
\( \mu \) Shear modulus
\( \mu_0 \) Shear modulus at 0 K
\( \rho \) Sample material density
\( \sigma_a \) Athermal component of the flow stress (MTS model)
\( \sigma_{av} \) Nominal stress after the instability point
\( \sigma_{calculated} \) Calculated stress
\( \sigma_e \) Engineering stress
\( \sigma_{experimental} \) Experimentally measured stress
\( \hat{\sigma}_i \) Rate-dependent portion of the yield stress (MTS model)
\( \sigma_t \) True stress
\( \sigma_{UTS} \) Ultimate tensile strength
\( \sigma_y \) Yield strength
\( \sigma_0 \) Zero plastic stress
\( \sigma_1, \sigma_2 \) Stress at two respective strain rates \( \dot{\varepsilon}_1 \) and \( \dot{\varepsilon}_2 \)
\( \hat{\sigma}_{\varepsilon} \) Strain hardening component of flow stress (MTS model)
\( \hat{\sigma}_{as} \) Saturation stress (MTS model)
\( \psi \) Conversion factor from mechanical work to heating
<table>
<thead>
<tr>
<th>Acronym</th>
<th>Description</th>
</tr>
</thead>
<tbody>
<tr>
<td>BCC</td>
<td>Body Centred Cubic</td>
</tr>
<tr>
<td>DIC</td>
<td>Digital Image Correlation</td>
</tr>
<tr>
<td>EBSD</td>
<td>Electron Backscattered Diffraction</td>
</tr>
<tr>
<td>EDM</td>
<td>Electric Discharge Machining</td>
</tr>
<tr>
<td>EDX</td>
<td>Electron Diffraction Spectroscopy</td>
</tr>
<tr>
<td>FCC</td>
<td>Face Centred Cubic</td>
</tr>
<tr>
<td>FEA</td>
<td>Finite Element Analysis</td>
</tr>
<tr>
<td>HCP</td>
<td>Hexagonal Close Packed</td>
</tr>
<tr>
<td>HIP</td>
<td>Hot Iso-Statically Pressed/Hot Iso-Static Pressed/Hot Iso-Static Pressings.</td>
</tr>
<tr>
<td>J-C</td>
<td>Johnson-Cook</td>
</tr>
<tr>
<td>MTS</td>
<td>Mechanical Threshold Stress</td>
</tr>
<tr>
<td>OFE</td>
<td>Oxygen Free Electrolytic</td>
</tr>
<tr>
<td>PREP</td>
<td>Plasma Rotating Electrode Process</td>
</tr>
<tr>
<td>SHPB</td>
<td>Split Hopkinson Pressure Bar</td>
</tr>
<tr>
<td>UTS</td>
<td>Ultimate Tensile Stress</td>
</tr>
<tr>
<td>Z-A</td>
<td>Zerilli-Armstrong</td>
</tr>
</tbody>
</table>
CHAPTER 1

Introduction

1.1 Aims and Objectives

The behaviour of metals at high strain rates has been the subject of an increasing number of studies since the 1950s; nevertheless, the comparison of the results obtained over a strain rate regime can sometimes be difficult, especially if different test equipment and specimen geometries are used in the low and high strain rate tests. With high strain rate servo-hydraulic systems and high speed video imaging equipment capable of strain measurement becoming available, the simple tensile test covering the quasi-static strain rate of $10^{-3}\,\text{s}^{-1}$ to $10^{3}\,\text{s}^{-1}$ often referred to as the “dynamic range” has over the last decade become practically possible. However, the direct measurement of stress and strain in the dynamic strain rate range has brought with it, its own set of unique problems and the lack of reliable guidelines available on the testing methods, specimen dimensions, measurement devices, signal damping and curve smoothing techniques necessary to assure the quality of the test results has greatly hampered comparison of data between different testing laboratories.

In the absence of reliable guidelines for testing and data analysis the aim of this research is to improve the understanding of high strain rate tensile testing in metallic materials so that a greater appreciation of the specific problems encountered during high strain rate testing can be understood and how the interpretation of the results can be improved.

The aims and objectives of the work described in this thesis are:

1. To develop an understanding of the practical difficulties in acquiring stress and strain data in a high strain rate tensile test, covering the problems with stress wave propagation within the test specimen and the measuring techniques necessary to record the output data.

2. To explore the problems of interpreting and acquiring meaningful material information from the raw data provided by the test equipment.
(3) To evaluate the material tensile data acquired covering the quasi-static to dynamic strain rate regime and to compare it with those acquired by other mechanical testing techniques such as those acquired under compression.

(4) To apply the data acquired over the quasi-static to dynamic tensile strain rate testing regime in constitutive material models.

1.2 Thesis Outline

This thesis is divided into seven chapters. Chapter 1 introduces the work described in the thesis, and outlines the aims and objectives. Chapter 2 is a review of the literature relating to the most commonly used mechanical testing techniques as a function of strain rate. It outlines the specific problems encountered with dynamic tensile testing, and the factors that influence the mechanical behaviour of a metallic material when subjected to high strain rate conditions, such as its microstructure, grain size and adiabatic heating. Chapter 2 also gives an overview of the most commonly used constitutive material models which can help to predict the mechanical behaviour of a material over a wide range of test parameters, outside the range of test conditions usually available in the standard testing laboratory.

Chapter 3 investigates the practical requirements for tensile testing of metallic materials at high strain rates, such as tensile specimen design, test equipment requirements, the methods used to determine the validity of a test, and methods and techniques for interpreting and processing data in order to provide meaningful results.

Chapter 4 investigates the tensile behaviour of a number of metals and alloys over a range of strain rates using the methods established in Chapter 3. It evaluates the mechanical behaviour of these materials over the quasi-static to dynamic range and at test temperatures of 203 K to 373 K.

Chapter 5 details the results from experiments carried out on the same metallic materials as in Chapter 4, but this time in compression over the quasi-static to dynamic strain rate range and at test temperatures between 204 K and 373 K. These results are compared with their tensile behaviour established in Chapter 4.
Chapter 6 uses the true stress-true plastic strain curves determined experimentally in Chapters 4 and 5 to assess the practicalities and validity of the most common constitutive material models in describing material behaviour in tension and compression over the quasi-static to dynamic strain rate regime. The model constants established are compared with those available in the literature. Chapter 7 summarises the findings of this work.
CHAPTER 2

Literature Survey

2.1 Mechanical Testing Techniques

There are numerous examples of engineering materials being subjected to dynamic or impact types of loading, which subject the material to high strain rates. Most metals and alloys show a significant change in their mechanical response under such increased rates of strain [1], and in order to design or analyse dynamically loaded structures more accurately, it is necessary to know the mechanical properties of the materials involved at the strain rates to which the material is to be subjected to.

Several types of testing techniques have been used to generate rate dependent material data; each serves a specific range of strain rates and provides a specific type of information. The conventional screw driven tensile tests can provide strain rate dependent data up to 1 s⁻¹, while the high rate servo-hydraulic and the drop weight systems can provide strain rates up to 10³ s⁻¹, the Split Hopkinson Pressure Bar (SHPB), a widely used high strain rate testing technique, may be used to generate data at the high end of 10² s⁻¹ to 10⁵ s⁻¹. Figure 2-1 presents a schematic representation of the strain rate coverage range of the most commonly used testing systems.

![Diagram showing strain rate coverage range of testing systems](image)

*Figure 2-1. Typical strain rates covered by the various testing systems [2].*
2.1.1 Tensile Testing

One of the simplest and most effective laboratory tests for obtaining material data is the tension test carried out on a screw or servohydraulic machine. A specimen, usually a round rod or a flat bar with a reduced section gauge length is pulled with a controlled force in tension. As the force is increased, the elongation of the specimen gauge length is recorded. The relationship between the applied force and resulting deformation (elastic-plastic) is measured and recorded; this data is then plotted graphically in the form of a stress-strain curve taking into account the specimen dimensions. The engineering stress, \( \sigma_e \), can be obtained from:

\[
\sigma_e = \frac{P}{A_0}
\]  

(2.1)

where, \( P \) is the load measurement and, \( A_0 \) is the initial area of the specimen gauge length cross section. Engineering strain, \( \varepsilon_e \), is obtained from:

\[
\varepsilon_e = \frac{\Delta l}{l_0}
\]  

(2.2)

where, \( l_0 \) is the initial gauge length of the specimen and, \( \Delta l \) is the change in the sample gauge length. The Young’s modulus, \( E \) is defined as the gradient of the elastic region of the engineering stress-strain curve and is given by:

\[
E = \frac{d\sigma}{d\varepsilon}
\]  

(2.3)

where, \( d\sigma \) is the change in stress, \( d\varepsilon \) is the change in strain. The yield strength, \( \sigma_y \), is defined as the stress corresponding to an offset in strain of 0.2% from a line described by equation (2.3) as shown in Figure 2-2.
The ultimate tensile strength, $\sigma_{UTS}$, is the maximum recorded stress value from the stress-strain curve. Other values recorded are the percentage elongation to failure, $\varepsilon_F$, and the percentage reduction in area on failure, $A_F$.

\[
\varepsilon_F = \left( \frac{l_f - l_0}{l_0} \right) \times 100 \tag{2.4}
\]

and

\[
A_F = \left( \frac{A_0 - A_f}{A_0} \right) \times 100 \tag{2.5}
\]

where, $l_f$ is the sample’s gauge length at failure.

The engineering stress, $\sigma_e$, and engineering strain, $\varepsilon_e$, are converted to true stress, $\sigma_t$, and true strain, $\varepsilon_t$, (based on the actual cross sectional area of the specimen gauge length) using the formulae:

\[
\sigma_t = (1 + \varepsilon_e)\sigma_e \tag{2.6}
\]

and

\[
\varepsilon_t = \ln(1 + \varepsilon_e) \tag{2.7}
\]

Equations (2.6) and (2.7) are only suitable for converting stresses and strains up to the ultimate tensile strength at which point the specimen starts to neck. At approximately this point the reduction in sample’s gauge cross section makes the true stress value invalid. The onset of necking is accompanied by the establishment of a tri-axial state of stress in the neck; the uni-axial stress distribution is disrupted by the geometrical
irregularity. The flow stress of a material is strongly dependent on the state of stress; hence a correction has to be introduced to covert the tri-axial flow stress into uni-axial stress as the external boundaries of the neck generate tensile components perpendicular to the axis of the specimen. The magnitude of the transverse tensile stresses depends on the geometry of the neck, and Bridgmen [3] introduced a correction factor to cylindrical specimens after performing a stress analysis in the neck. The equation that expresses the correction stress is:

\[
\sigma_i = \sigma_{av} \left(1 + \frac{2R}{r_n} \right) \ln \left(1 + \frac{r_n}{2R} \right)
\]

where, \( R \) is the radius of curvature of the neck and, \( r_n \) is the radius of the cross section in the thinnest part of the neck.

The strain rate expressed by the test specimen can be roughly calculated by the test machine velocity of the actuator and the gauge length of the specimen:

\[
\dot{\varepsilon} = \frac{V}{l_0}
\]

where, \( \dot{\varepsilon} \) is the engineering strain rate, \( V \) is the velocity of the actuator and, \( l_0 \) is the initial parallel length of the specimen gauge length, which is the reduced section in the specimen with a constant width. Thus the maximum engineering strain rate achievable for a system can be calculated as:

\[
\dot{\varepsilon}_{\text{max}} = \frac{V_{\text{max}}}{l_{\text{min}}}
\]

where, \( V_{\text{max}} \) is the maximum speed of the actuator that is limited by the machine capability. \( l_{\text{min}} \) is the minimum initial parallel length of the specimen, which is controlled by the requirement to achieve the uni-axial stress condition throughout the parallel length of the specimen. The smaller the sample’s parallel gauge length, the higher the maximum engineering strain rate. When the radius of the shoulder is very small, as is the case for many specimens designed for testing at high strain rates, the engineering strain rate can be roughly estimated using the total length, the distance including the shoulders.
2.1.2 Split Hopkinson Pressure Bar Test

The Split Hopkinson Pressure Bar (SHPB) test in its original form is a test developed to measure the compressive mechanical behaviour of a material at high strain rates. The technique was first suggested by Bertram Hopkinson in 1914 [4] as a way to measure stress pulse propagation in a metal bar and later refined by R.M. Davies in 1948 [5,6] and H. Kolsky [7] in 1949, applies uniaxial compression to samples at strain rates in order of $10^3$ s$^{-1}$. The underlying principles for the test and measurement [8] is that a cylindrical specimen is placed between the ends of two parallel sided cylindrical bars, called the incident bar and the transmitted bar as shown in Figure 2-3.

![Figure 2-3. Split Hopkinson pressure bar (SHPB) equipment set-up][2].

A third cylindrical bar, called the striker bar, is impacted at the free end of the incident bar to create a compressive stress pulse (referred to as the incident wave) that propagates in the incident bar towards the specimen. This compressive stress pulse on reaching the incident bar-specimen interface is partly reflected back as a wave of tension (referred to as the reflected wave) and partly transmitted through the specimen into the transmitted bar (referred to as the transmitted wave). As the wave travels through the specimen, the specimen undergoes dynamic elastoplastic deformation at a high rate. Strain gauges bonded on the surface of the incident and transmitted bars connected in a Wheatstone bridge record the incident, reflected, and transmitted waveforms on an oscilloscope. From these recordings, the strain, strain rate, and stress in the specimen can be calculated.

The engineering stress, $\sigma_e(t)$ in the sample can be calculated from the transmitted waveform only using the following equation:
\[ \sigma_e(t) = \frac{E \cdot A_b \cdot \varepsilon_t(t)}{A_0} \]  
(2.11)

where, \( E \) is the elastic modulus of the pressure bar, \( A_b \) is the pressure bar diameter, \( \varepsilon_t(t) \) is the transmitted strain, and \( A_0 \) is the initial sample cross-sectional area. and is known as 1-wave analysis. Conversely, the sample’s engineering stress, \( \sigma_e(t) \), can be calculated by the summation of the incident and reflected waveforms at this interface, termed 2-wave analysis using equation (2.12):

\[ \sigma_e(t) = \frac{E \cdot A_b}{A_0} \cdot (\varepsilon_i(t) + \varepsilon_r(t)) \]  
(2.12)

where, \( \varepsilon_i(t) \) is the incident strain and \( \varepsilon_r(t) \) is the reflected strain. Both these equations assume a uniform distribution of stress and strain over the entire sample’s length as a function of time. However, at the early stages of any test the loading starts at the incident bar-specimen interface while the sample’s other face remains at rest. Time is therefore required for stress equilibrium to be achieved in the sample. The different forces on the ends of the specimen may be taken into account by averaging the stress in the sample at the specimen-transmitted bar interface (back stress), and combining incident and reflected pulses to calculate the stress at the incident bar-specimen interface (front stress). This is termed 3-wave analysis and can provide additional information on how soon the actual stress state in a sample achieves homogeneity. The engineering stress, \( \sigma_e(t) \), using 3-wave analysis may be derived simply by taking the average of the two principal forces divided by the combined interface areas:

\[ \sigma_e(t) = \frac{E \cdot A_b}{2 \cdot A_0} \cdot (\varepsilon_i(t) + \varepsilon_r(t) + \varepsilon_t(t)) \]  
(2.13)

where, \( \varepsilon_i(t) \) is the transmitted strain. The strain rate, \( \dot{\varepsilon}(t) \), over the period of the test can be derived from equation (2.14):

\[ \dot{\varepsilon}(t) = \frac{2 \cdot C_b \cdot \varepsilon_r(t)}{l_0} \]  
(2.14)

where, \( C_b \) is the wave speed in the pressure bar, \( \varepsilon_r \) is the reflected strain, and \( l_0 \) is the initial sample length. The wave speed in the bar, \( C_b \), can be calculated from equation (2.15):
\[ C_b = \frac{E}{\rho} \quad (2.15) \]

where, \( \rho \) is the bars density. Equation (2.14) can be integrated to give the engineering strain, \( \varepsilon_e(t) \):

\[ \varepsilon_e(t) = \frac{2 \cdot C_b}{l_0} \int_0^t \varepsilon_r(t) dt \quad (2.16) \]

These engineering stresses and strains can be converted to true stresses and strains using equations (2.6) and (2.7).

### 2.1.3 Dynamic Tensile Testing

The SHPB test was originally developed to assess materials in compression. The fact that a material may respond differently at high strain rates under tension to that found under compression resulted in the SHPB system being modified for tensile testing. However, while this technique has provided invaluable data, there are a number of drawbacks, including the fact that the stress-strain data obtained is valid only after some degree of stress and strain rate uniformity is achieved, and even small variations in the strain rate thereafter do not allow the rate to be quoted to an accuracy of better than the nearest hundred. The SHPB tension test can also only test materials under high strain rate conditions, making it difficult to compare with data obtained from quasi-static and intermediate strain rates conducted on a conventional load frame owing to the different test conditions in terms of equipment, geometry of specimen and application of the loading force.

High strain rate servo-hydraulic systems capable of testing at rates of up to \( 10^3 \) s\(^{-1} \) have been available for some time. However, measurement of the loading and strains at these rates is challenging. The test methods and standards established for quasi-static test conditions are not automatically valid in dynamic material testing. Although a number of societies and institutes have published guidelines for dynamic tensile testing [9-12], there is no official standard available, and often the guidelines lack detailed information as to the testing method, specimen dimensions, measuring devices, signal damping, and curve smoothing techniques, all of which are critical to the quality of testing results. As a result, data from different laboratories are often not
comparable. Some of the sources of variability in the reported data are discussed below.

**Inertia**

To eliminate the inertia effect in a dynamic tensile test during the acceleration phase, use is made of a slack adapter, which consists of a hollow tube and a sliding bar. When the machine is actuated, the hollow tube travels freely with the hydraulic head over a distance needed to reach a predefined speed before coming into contact with the surface of the sliding bar. The initial position of the hydraulic head together with the tube in respect to the specimen is also an experimental variable. In open loop, a servo-hydraulic machine maintains nearly constant speed over a given distance. The hydraulic head needs to be placed at a position such that the dynamic tensile test is performed within the constant speed window. If the initial position of the hydraulic head is too low, the loading of the specimen may start before a constant speed has been reached. On the other hand, if the initial position of the hydraulic head is too high, the machine may run out of actuation (travel at desired speed) before the specimen fractures.

**System Ringing and Damping**

In a dynamic tensile test, the loading of the specimen starts when the sliding bar comes into contact with the hollow tube in the slack adapter. The sudden engagement with the upper portion of the load train introduces a shock wave in the test system generating a high amplitude stress wave, causing the test system to oscillate, a phenomenon known as “system ring.” The test system will ring (oscillate) at the system’s natural frequency, which is dependent on the load train arrangement and includes, the slack adapter, the size of the test specimen, and the yield strength of the test material. All these factors that influence system ringing will determine the maximum nominal strain rate that might possibly be achieved with a particular test system setup.

System ringing under high rate testing conditions is inevitable and can be quite excessive, unless a “damping joint” is introduced at the sliding-bar and hollow-tube contact face. The damper is a necessary requirement in the dynamic testing system, as it not only reduces the system ring and its rate of decay, but also provides a way to
regulate the load rising time and therefore reduces the magnitude of the stress oscillation. A system with higher natural frequencies results in oscillations of lower amplitudes and would respond to damping in a time interval of the same order as its natural period and therefore its ringing can be suppressed effectively by damping. Theoretically, the system ringing can be damped out regardless of the natural frequency of a testing system. In the extreme case an isolator may be used, however, the rate of decay in such cases would be unrealistically slow for high rate testing. If the test system is damped the actual elastic strain rate will depend on the modified load rising curve. Therefore, the maximum strain rate at which a testing system can generate acceptable data is limited by its natural frequency.

The draft SAE standard [11] recommends that damping is needed for strain rates greater than 10 s\(^{-1}\) and the damping-related effect should be finished by the time the applied load is at 25% of the yield. Therefore not all testing systems can be effectively damped for use in high rate testing.

**Dynamic Stress Equilibrium**

To obtain valid stress-strain data in a material test, the specimen should be in state of stress equilibrium, undergoing homogeneous deformation in the specimen gauge section. Under quasi-static loading conditions, the stress wave has time to travel back and forth many times inside the specimen allowing the sample to remain in a quasi-equilibrium state providing deformation homogeneity throughout the test. Under dynamic loading conditions the time available for a stress-wave to travel back and forth inside the specimen is extremely small, which can result in a large variation in the stress distribution across the specimen length and in extreme cases the stress at one end of the specimen can exceed the strength of the material, causing fracture, while other areas experience little deformation. It is recognised therefore that to ensure the validity of a dynamic material test the number of stress waves and their amplitude should be controlled so that the specimen remains in a condition of stress equilibrium ensuring homogeneous deformation at a constant rate.

In a dynamic material test, it is impossible to reach the state of stress equilibrium as in a quasi-static test. Instead, a dynamic stress equilibrium condition is sought. A qualitative determination may be achieved by examining the system ringing in the
load-time curves produced from the experiment, such as that shown in Figure 2-4. However, the quality of high rate testing varies greatly among labs, and a quantitative definition or criterion on what constitutes a valid dynamic test is required. While the analysis of dynamic tensile testing data is still under development and reports are relatively scarce there is a large amount of work published on the knowledge acquired using SHPB. Xiao [13] proposed applying the SHPB criterion for dynamic equilibrium to the dynamic tensile test.

Figure 2-4. Examples of system ringing in dynamic tensile experiments [10].

In SHPB, the criterion for the condition of dynamic stress equilibrium is rather simple. Research on SHPB has shown that to achieve dynamic stress equilibrium the load rising time should be sufficient to allow a certain number of round trips of the stress waves inside the specimen and this has been determined by experiment to be a minimum of three [8]. Therefore a quantitative method may be defined by determining the number of round-trips \( N \) a stress wave makes during a test with a criterion for a specimen being under dynamic equilibrium conditions if \( N \geq 3 \), which may be determined from:

\[
N = \frac{t_r C_m}{2L}
\]  

(2.17)

where \( t_r \) is the load rising time (defined as the time between the onset of loading to the peak stress), \( L \) is the length of the specimen and \( C_m \) is the elastic stress wave velocity of the material. The elastic stress wave velocity in its simplest form assuming a one-
dimensional longitudinal elastic stress wave velocity in an isotropic material may be given as described in equation (2.15) [14]. Equation (2.17) shows that for the same rising time a larger specimen size results in a smaller $N$ value. It also shows that the load rising time varies with the stress wave velocity of the material. Therefore, whether a test meets the acceptable condition of dynamic stress equilibrium needs to be determined for each material, either by the qualitative method of visually inspecting the stress history trace or by the quantitative method proposed above.

**Load Measurement**

In high rate testing, the response of the load cell is a further concern. At quasi-static strain rates, a load cell generally deforms homogeneously and the loading force is measured accurately by the strain gauge arrangement attached to it. At high strain rates the homogeneity of elastic deformation within the load cell is lost and propagation waves occur within the load cell [9]. Piezoelectric load washers are recommended for dynamic tests as they have much faster response times, but the load reading will still be affected by propagation waves generated at the very high rates. The load can also be measured by attaching a strain gauge to a specimen which incorporates a dynamometer section in its design so that the load can be measured directly from the specimen. Experimental work has demonstrated that measuring the load signal from such an arrangement reduces the amplitude of the oscillations in the load trace compared to that obtained from a piezoelectric load washer [15].

**Strain Measurement**

The direct measurement of strain by extensometer as used in tests carried out at quasi-static rates are not suitable for tests carried out under dynamic conditions, as they do not have the necessary response characteristics and are normally physically too fragile at these higher strain rates. Non-contact measurement techniques are thus required at these dynamic strain rates and the development of high-speed digital cameras capable of imaging rates as high as several million frames per second and at relatively high spatial resolutions has allowed the measurement of surface displacement on a tensile specimen by Digital Image Correlation (DIC).

The DIC technique tracks the position of a random speckle pattern on the surface of a test specimen. In the analysis, images are divided up into facets, the speckled pattern
in each facet being unique so they can be tracked from one image to the next. This allows local displacements to be quantified as the sample deforms, from which a strain field can be calculated. DIC can be carried out in two dimensions (2-D), or in three dimensions (3-D) with two cameras. In 2-D DIC, displacements are assumed to occur in a single plane, such as a flat plate under-in-plane stress. The results from this method will be accurate if there are no significant out-of-plane movements.

In 3-D DIC, out-of-plane displacements are measured using a second camera place at an angle to the first. 3-D DIC enables the measurement of diametrical contraction in round tensile specimens, allowing materials displaying deformation anisotropy to be readily quantified for a fuller understanding of a material’s mechanical behaviour and subsequent modelling.

**Adiabatic Heating**

Adiabatic heating is another issue to consider during dynamic testing. A metal that is subjected to deformation will undergo irreversible plastic deformation, which proceeds by dislocation motion and displacive mechanisms such as slip mechanical twinning (in BCC or HCP metals) or martensite transformation. During this process, the associated deformation energy is predominately dissipated by the motion of defects, leading to a distinct thermomechanical response which can result in local temperature increases of up to several hundred degrees centigrade. Chen et al. [16] measured overall temperature increases due to the Portevin-Le Chatelier effect\(^1\) in Fe-Mn-C steels of about 100°C while work on Ti alloys by Ranc et al. [17] observed maximum local temperatures of 1000°C inside shear bands that formed just before failure during dynamic torsion tests.

The thermally activated processes associated with plastic flow such as dislocation motion, interface motion, nucleation and strain hardening depend exponentially on the temperature, so local temperature rises can significantly change the material flow characteristics of a deforming metal compared with those expected at the nominal temperature. The friction stresses generated in metals are rate dependent, and high

---

1 Portevin-Le Chatelier effect describes a serrated stress-strain curve or jerky flow, which some metallic materials exhibit as they undergo plastic deformation. This effect has long been associated with the competition between diffusing solutes pinning dislocations and dislocations breaking free of these obstructions.
shear rates lead to a high energy release rate, and hence, higher frictional heating when compared to low velocities. The instantaneous dissipative heating that occurs during plastic deformation is adiabatic, meaning that the heat generation rate is much larger than the heat transfer rate. Hence, the deformation energy released as heat remains in the strained zone for a short time interval before dissipating by heat conduction. These very localised temperature rises mean that the plastically deforming metal becomes less resistant to further deformation, which becomes even more pronounced as necking develops. As necking occurs the deformation is localised further and the temperature rises mainly within the necking region. Since the flow stress decreases with increasing temperature, further deformation is localised preferentially in this zone and this “autocatalytic” process continues until failure occurs. For many metals, adiabatic heating has a dual effect on the flow-forming characteristics; increasing the ductility (failure limit) and decreasing the strength or flow stress of the metal.

The adiabatic heating effect will often be apparent in the microstructure of the metal in the form “adiabatic shear bands.” Adiabatic shear bands are usually very narrow; typically 5-500 μm and consist of very highly sheared material and tend to show the signs typical of a metal which has undergone a dynamic re-crystallisation process. Usually the centre of a shear band shows a significant increase in hardness which might occur following a cycle of work hardening and rapid quenching [18].

2.2 Mechanical Behaviour of Metallic Materials

2.2.1 Effect of Microstructure

One of the underlying principles of metallurgy is that properties can be deduced from a knowledge of a metal’s or alloy’s microstructure. By the term microstructure we can mean the crystal structure, and all imperfections present including their shape, orientation, composition, and spatial distribution, etc. Clearly with so many factors influencing a material’s mechanical response, defining a material structure-property relationship is complex. However, much work has been conducted in the past and some important relationships have been found [19]. Dislocations being one of the prime movers in plastic deformation respond to an applied traction by their ability to overcome lattice obstacles. These lattice obstacles may be short range barriers such as
the other atoms (Peierls-Nabarro Barriers), interstitial atoms, substitutional atoms and forest dislocations, or long range barriers such as grain boundaries, twin boundaries, phase boundaries, inclusions, precipitates, micro-cracks, and voids. The influence of the key microstructural factors on mechanical behaviour is discussed in more detail below.

**Crystal Structure**

The crystal structure of a metal has a major effect on its mechanical behaviour, for example, the significant difference in temperature and strain rate response seen in typical Face Centred Cubic (FCC) and Body Centred Cubic (BCC) metals [20]. In a FCC metal dislocations move at a very low stress known as the Peierls stress\(^2\), but they also rapidly work harden. In the FCC crystal lattice (Figure 2-5) there are 12 different close packed slip systems, four planes \{111\} and three directions <110>, that are well distributed in space. Slip can occur much more easily on a close packed plane rather than on a plane of lower atomic packing density and therefore the flow stress of a typical FCC metal such as copper, gold or aluminium will have a lower temperature and strain rate dependence, as the activation energy associated with flow over such an obstacle (other atoms) is vanishingly small [21].

![Figure 2-5. Face Centred Cubic crystal lattice, showing the {111} plane of highest atomic packing density.](image)

The large number of slip systems present in the FCC lattice makes it practically impossible that a slip plane will not be favourably orientated to the applied stress so

\(^2\) The Peierls stress is the shear stress needed to move a dislocation through a crystal lattice in a particular direction.
that the critical resolved shear stress\(^3\) is low. However, it also means that slip in FCC metals will take place on more than one of the close packed planes. When slip occurs at the same time on several intersecting slip planes, the stress required to produce additional deformation rises rapidly, as the flow stress is controlled by dislocation-obstacle interaction, in this case other dislocations (forest dislocations) [21]. The activation energy associated with flow over such obstacles is large and work hardening increases with strain rate owing to an increase in dislocation generation (increase in dislocation density) and their subsequent interactions, while work hardening reduces with temperature as greater numbers of dislocations are annihilated by thermal diffusion. This, therefore, makes the yield stress of FCC metals insensitive to strain rate and temperature, but work hardening increases with strain rate and declines with increasing temperature.

In contrast, the Body Centred Cubic (BCC) lattice (Figure 2-6) is characterised by four close-packed directions <111> and has no truly close-packed planes, and in comparison to the FCC metals, requires a much higher stress to initiate slip.

![Figure 2-6. Body Centred Cubic crystal lattice showing the lack of any truly close-packed plane, although the {110} planes have the highest packing atomic density in the BCC structure.](image)

The Peierls barrier will therefore control the yield and initial flow stress in a BCC metal and the yield stress in these crystals will increase as the temperature is lowered, the reduction in temperature reducing the thermal energy assistance to move the dislocation and thereby requiring a greater force to overcome the barrier. A rise in the strain rate will also increase the yield stress in these BCC metals as less time is

---

\(^3\) Critical resolved shear stress is the yield stress that must be exceeded in order to produce plastic deformation in a favourably orientated slip plane and direction.
available for a dislocation to overcome these Peierls barriers. Additionally, the structural evolution (change in the flow stress with deformation) in a BCC crystal will be weak compared to FCC metals as dislocation interaction on intersecting slip planes will be considerably smaller.

Hexagonal Close Packed (HCP) metals, have very few slip systems (Figure 2-7). Slip takes place predominately on the basal \{0001\} plane, being the plane of highest atomic packing density, and is therefore only possible in the directions that lie in that basal plane <1120> orientation. This makes this lattice structure highly dependent on the orientation of this plane resulting in an anisotropic response to an applied stress.

![Figure 2-7. Hexagonal Close Packed crystal lattice showing the \{0001\} planes of greatest atomic packing density which are few in number in a unit cell.](image)

**Grain Boundaries**

The most basic difference between single crystals and polycrystalline materials are the presence of grain boundaries and the possible randomisation of the orientation of individual grains to the applied stress. While microscopically these grains are isotropic due to their preferred slip direction, the randomisation of their orientation often makes them anisotropic in a polycrystalline material. The influence of the grain boundaries on the properties can also be significant, but will depend on the exact conditions of deformation and on the particular material [23, 24]. At low temperatures grain boundaries act as barriers to dislocation motion, thus strengthening the material. At elevated temperatures the opposite is true, grain boundaries can lower strength by providing an alternative path for both diffusion and dislocation motion e.g. creep.
Also, impurities tend to segregate to grain boundaries, which can alter the properties of the grain boundaries and hence the material in a number of ways. Work published by Hall [22] and Petch [23] in the early 1950’s revealed that many metals and alloys show a strong dependence between their grain size and yield strength, as shown in Figure 2-8.

The change in strength being approximately in accordance with the Hall-Petch equation:

\[ \sigma_y = \sigma_0 + \frac{k_y}{\sqrt{d}} \]  

(2.18)

where, \( k_y \) is a constant and \( d \) is the mean grain diameter. The Hall-Petch relationship could just as easily be used to include other regular microstructural arrays, such as the spacing of twin and phase boundaries.

### 2.2.2 Measurement of Strain Rate Sensitivity in a Material

The strain rate sensitivity of a material measured experimentally may be defined as a function of flow stress by the following expression [24]:

\[ \sigma = K(\dot{\varepsilon})^m \]  

(2.19)
where $K$ is a constant and $m$ is the strain rate sensitivity. Applying equation (2.19) to two strain rates and eliminating $K$, $m$ can be easily obtained from experimental measurements using equation (2.20):

$$m = \frac{\ln(\sigma_2/\sigma_1)}{\ln(\dot{\varepsilon}_2/\dot{\varepsilon}_1)}$$  \hspace{2cm} (2.20)

where, $\sigma_1$ and $\sigma_2$ are the stresses at the lower strain rate, $\dot{\varepsilon}_1$ and higher strain rate, $\dot{\varepsilon}_2$ respectively. In general $m$ varies between 0.02 and 0.2 for homologous temperatures between 0 and 0.9 (90% of the melting point in Kelvin). At these rate sensitivities, a doubling of the strain rate would result in a materials yield stress increasing by about 15% at most. Not all metals however exhibit high strain rate sensitivity; aluminium and some of its alloys have either zero or negative $m$ [24].

### 2.2.3 Influence of Stress or Strain State on Material Response.

There is some evidence that the stress state can have an influence on a material’s behaviour when a material is tested under uni-axial strain conditions as revealed by results from materials tested using different testing techniques. Stress-strain data on aluminium 6061-T651 and 7075 alloys, reported by Nicholas [25] using a tension version of the split Hopkinson bar, showed an apparent increase of 12% in the flow stress over the strain rate range of $4 \text{ s}^{-1}$ to $600 \text{ s}^{-1}$ for Al 6061-T651 and a 20% increase over a quasi-static rate to $800 \text{ s}^{-1}$ for Al 7075-T6. These results appeared to conflict with general observations from compressive tests where little or no strain rate sensitivity is observed in aluminium alloys up to strain rates of $\sim 10^3 \text{ s}^{-1}$. However, work conducted in tension on aluminium alloys, by previous experimentalists, such as Smith [26], Steidel & Makerov [27], Lindholm et al. [28] and Hoge [29] on Al 6061 T6 have also reported some degree of strain rate sensitivity in the flow stress at strain rates above $\sim 10 \text{ s}^{-1}$ supporting this claim.

There is also evidence from literature that strain state can have some influence on the development of a metallic material’s microstructure. While it may not be so obvious in examining microstructures in strained copper, nickel, or aluminium, there are some metallic systems where dramatic effects have been observed. For example, work carried out by Johnson et al. [30], has shown that the volume fraction of $\alpha'$-martensite
created in 304 stainless steel is considerably higher when the alloy subjected to triaxial strain conditions such as that experienced in shock loading than when subjected to biaxial straining as experienced by a deep drawing operation, which in turn showed a considerably greater volume fraction of martensite than when subjected to uni-axial strain conditions typical of a standard tensile/compression test.

2.3 Constitutive Modelling

2.3.1 Development of Constitutive Models

A simple tensile (or compression) test will measure a material’s response to an applied stress. By recording the load and displacement during a test the stress-strain response can be calculated, and thus various parameters such as yield stress and flow stress can be derived. It is well known that most materials show a significant change in mechanical response when tested under varying strain rates and temperatures, and tests can be carried out over a range of temperatures and strain rates to determine the effect on the stress parameters. However, the development of a mathematical model that would allow the mechanical behaviour of a material to be predicted over a wide range of parameters for use in engineering design outside the range of test conditions practically possible in the standard laboratory would of be of significant advantage. The basic objective of researchers in the mechanical behaviour of materials has been to develop an equation to show how the stress parameter, \( \sigma \), evolves as a function of strain, \( \varepsilon \), temperature, \( T \), and strain rate, \( \frac{d\varepsilon}{dt} \) (or \( \dot{\varepsilon} \)), in the form [31]:

\[
\sigma = f\left(\varepsilon, \frac{d\varepsilon}{dt}, T\right)
\]  

(2.21)

The problem with this equation of state is that that plastic deformation is irreversible and path dependent (dependent on the accumulation of damage) and the mode of failure which a material undergoes. Therefore, deformation history has to be included in any equation and equation (2.21) can be modified to [31]:

\[
\sigma = f(\varepsilon, \frac{d\varepsilon}{dt}, T, \text{deformation\_history})
\]  

(2.22)

Many constitutive equations have been proposed by various investigators, these may be of the empirically or physically based form. Empirical constitutive equations are
the simplest models and are easy to calibrate with a minimum of experimental data in the form of a few stress-strain curves at several strain rates and temperatures. The most commonly used of these equations is the Johnson-Cook (J-C) model [32], however, their development and application consists of a general curve fitting procedure and they do not address the more fundamental questions on how the plastic deformation takes place and how the micromechanical processes are connected to the overall plastic flow.

A number of models have been developed over the years which are based on the physical mechanisms taking place in a material during deformation. The most important physical mechanism in the deformation in metals is that of the generation and motion of dislocations, and two of the most notable models based on this mechanism are the Zerilli-Armstrong (Z-A) [33] and the Mechanical Threshold Stress (MTS) models [34].

2.3.2 Johnson-Cook Equation

One of the earliest attempts, and subsequently widely used equations, to express the mechanical behaviour of a material is the Johnson-Cook model [32]. This empirically based constitutive equation expresses in a very simple form the flow stress as a function of plastic strain, strain rate and temperature and describes very well the response of a number of metals tested at strain rates between 0.001 s\(^{-1}\) to 1000 s\(^{-1}\). The flow stress is given as:

\[
\sigma = \left[\sigma_0 + K\varepsilon^p\right] \left[1 + C \ln \dot{\varepsilon}^*\right] \left[1 - T^* M\right]
\]  

(2.23)

where, the three groups of terms in parentheses represent work-hardening, strain rate, and thermal effects respectively. The three terms are multiplied by each other; the rate of work hardening therefore increases for decreasing temperature and increasing rate. The constants, \(K\), \(n\), represent the effects of strain hardening, \(\sigma_0\) is the stress at zero plastic strain generally taken as the yield stress, \(C\) is the strain constant, \(M\) is the thermal softening fraction, \(\dot{\varepsilon}^*\) is the dimensionless plastic strain term calculated from:

\[
\dot{\varepsilon}^* = \frac{\dot{\varepsilon}}{\dot{\varepsilon}_0}
\]  

(2.24)
where $\dot{\varepsilon}$ is the test strain rate and $\dot{\varepsilon}_0$ is the reference strain rate (the lowest strain rate used). $T^*$ is referred to in the original J-C model paper [32] as the homologous temperature and calculated by the expression:

$$T^* = \frac{(T - T_r)}{(T_m - T_r)}$$  \hspace{1cm} (2-25)

where $T$ is the test temperature, $T_r$ is the reference temperature, and $T_m$ is the melting point of the metal/alloy being tested. One observation of the $T_r$ term in the equation, which in most modern publications is referred to as a reference temperature and therefore is taken as the lowest test temperature, is that in their original paper Johnson and Cook referred to this term as room temperature [32]. The disadvantage of using the original definition for this term is that the model cannot be applied to test data acquired at below room temperature. Consequently the equation constants will be affected, depending on which definition for $T_r$ is used, and therefore cannot be compared with constants quoted in published literature unless the method for determining them has been declared.

A more generic problem with the J-C model is that it assumes a linear increase in flow stress with the logarithm of strain rate. Experimental evidence has shown that substantial deviations can occur at strain rates of the order $10^3 \text{ s}^{-1}$ and above [35], especially for Body-Centred-Cubic (BCC) metals [36].

### 2.3.3 Zerilli-Armstrong Equations

The Zerilli-Armstrong constitutive model (Z-A) is a physically based constitutive equation based on thermally activated dislocation motion [33]. The different effects of work hardening and strain rate effects on hardening are incorporated into the equation and the effects of grain size can also be included. Zerilli and Armstrong made the point that the crystal structure of a metal or alloy will have its own constitutive behaviour, dependent on the dislocation characteristics of that particular structure (see section 2.2.1). While there are several generations of the model, their initial model addressed metallic materials with either FCC or BCC crystal structures, while they treated Hexagonal-Close-Packed (HCP) crystal structures as behaving the same as BCC structures. The flow stress is given by:
\[
\sigma = C_0 + \left[ C_1 + C_2 \varepsilon^{\frac{1}{n}} \right] \exp\left( -C_3 + C_4 \ln \varepsilon \right) T + C_5 \varepsilon^n
\]  
(2.26)

where, \( \varepsilon \) is strain, \( \dot{\varepsilon} \) is strain rate, \( T \) is the test temperature, and \( C_0, C_1, C_2, C_3, C_4, C_5, \) and \( n \) are constants. When applied to FCC metals, the constants, \( C_1 = C_5 = 0 \), while when applied to BCC metals, the constant, \( C_2 = 0 \). In this model, the different effects of work hardening, strain rate effects on hardening and thermal softening are additive. In BCC metals the plastic strain hardening contribution to the flow stress is characterised from assuming an isothermal power law dependence on strain \( (C_5 \varepsilon^n) \), where the stress-strain curves are translated upward and downward by strain rate and temperature increases, and are better represented by this equation than the J-C model.

The constant \( C_0 \) is the athermal portion of the equation and characterises the thermal and rate independent interactions of dislocations with long range barriers such as grain boundaries. This may be represented by the Hall-Petch relationship given by equation (2.18), and could just as easily be used to include other long range regular microstructural barriers, such as the spacing of twins and phase boundaries. It should be noted however, that the constants \( C_1, C_2, C_3, C_4, C_5 \) and \( n \) still largely need to be obtained experimentally, and hence the basis of the constitutive equation is still rooted in experiments.

### 2.3.4 Mechanical Threshold Stress Model

Another model that is commonly used, which is based on dislocation dynamics in metallic materials much like the Z-A model, is the MTS model developed by a group at Los Alamos National Laboratory (LANL), New Mexico, USA [34]. As with the Z-A model, the MTS model contains a thermal and an athermal component of stress. Where it is fundamentally different is that instead of making the assumption that the thermal component of flow stress depends explicitly on strain, it takes into account how the internal structure of a material evolves, based on dislocation mechanics, and in this respect represents the most advanced description of a material’s constitutive behaviour to date.

It is a fundamental assumption of the mechanical tensile test that the flow stress of a material depends on the current structure and that the structure evolves with increasing strain. The evolution of this structure is related to the balance of dislocation
generation and recovery processes, which is dependent not only on the temperature and the applied traction, but also on the strain rate. The evolution of structure is not the same at low and at high strain rates, as at higher strain rates the ratio of dislocation generation to dislocation annihilation will increase. This will have the effect of changing the values of immobile and mobile dislocation densities leading to different total densities and as a natural consequence different flow stresses [21, 34]. Therefore, instead of using strain as a state parameter, comparison is instead made at constant structure, where the mechanical threshold stress (i.e. flow stress at 0 K) is used as the parameter. The MTS model gives the flow stress in the following form:

\[
\sigma = \sigma_a + \left( S\dot{\sigma}_i + S\dot{\sigma}_e \right) \frac{\mu}{\mu_0}
\]  

(2.27)

where, \(\sigma\) is the flow stress, \(\mu\) is the shear modulus, and \(\mu_0\) is the shear modulus at 0 K. \(\sigma_a\) is the athermal component of the flow stress associated with the interaction of dislocations with long range barriers such as grain boundaries, or second phases (discussed in section 2.2.1) and may be represented by the Hall-Petch relationship equation (2.18). The parameter \(\dot{\sigma}_i\) is the rate-dependent portion of the yield stress mainly due to intrinsic barriers such as the strong Peierls stress in BCC metals at low temperature or at high strain rates (section 2.2.1). It is further assumed that this term does not evolve after yielding. The term \(\dot{\sigma}_e\) is the strain hardening component of flow stress, and evolves with strain due to dislocation accumulation (work hardening) and annihilation (recovery). \(\dot{\sigma}_e\) may be determined from the extended Voce law [34]:

\[
\frac{d\dot{\sigma}_e}{d\varepsilon} = \theta_0 \left( 1 - \frac{\dot{\sigma}_e}{\dot{\sigma}_0} \right)^\kappa
\]  

(2-29)

where, \(\kappa \geq 1\) is a non-dimensional parameter that characterises the shape of the hardening law. \(\theta_0\) is the initial slope of the stress-strain curve at the moment of yielding (where hardening is due to dislocation accumulation only), and \(\dot{\sigma}_0\) is the saturation stress (where the strain hardening rate is zero), as shown in Figure 2-9.
Figure 2-9. Description of the saturation stress $\dot{\sigma}_s$ and the integral slope at yielding $\theta_0$, parameters from the strain curve.

The evolution law of stress with strain can be numerically evaluated by moving the $de$ expression in equation (2.29) to the right-hand side of the equation and calculating $\dot{\sigma}$ at each strain increment. The value of $\theta_0$ changes with strain rate and temperature.

$S_i$, and $S_c$, are the Arrhenius form for a temperature scaling factor, which specifies the ratio between applied stress and the mechanical threshold stress component for the rate dependent term for the yield stress, $\dot{\sigma}_i$, and the strain hardening component term of flow stress $\dot{\sigma}$ respectively.

$$S_i(\dot{\varepsilon}, T) = \left\{ 1 - \left[ \frac{kT}{\mu b^3 g_{0i}} \ln \left( \frac{\dot{\varepsilon}_{0i}}{\dot{\varepsilon}} \right) \right]^{q_i} \right\}^{\gamma_i} \quad (2.30)$$

$$S_c(\dot{\varepsilon}, T) = \left\{ 1 - \left[ \frac{kT}{\mu b^3 g_{0c}} \ln \left( \frac{\dot{\varepsilon}_{0c}}{\dot{\varepsilon}} \right) \right]^{q_{0c}} \right\}^{\gamma_{0c}} \quad (2.31)$$

where, $k$ is the Boltzmann constant, $b$ is the magnitude of the Burgers vector, and $T$ is temperature in Kelvin. $\dot{\varepsilon}_{0i}$ and $\dot{\varepsilon}_{0c}$ are the reference strain rates, $q_i$ and $p_i$ are constants describing the dislocation glide resistance profile, while $g_{0i}$ and $g_{0c}$ are the normalised activation energies. The shear modulus of a material ($\mu$), which is a function of
temperature can be determined by using impact excitation or other acoustic techniques over a range of temperatures from which \( \mu_0 \) can be determined using the description:

\[
\mu = \mu_0 - \left( \frac{D_0}{\exp\left(\frac{T_0}{T}\right) - 1} \right) \tag{2.32}
\]

where, \( D_0 \) and \( T_0 \) are material constants.

**Adiabatic Heating**

Equations (2.27) to (2.32) provide a good description of the plastic response of a material when tested isothermally. However, for strain rates above \( 1 \, \text{s}^{-1} \) this is not the case as performing mechanical work on the sample will raise the sample temperature and lead to significant thermal softening (section 2.1.3). The temperature rise due to adiabatic heating, \( \Delta t \), is given as:

\[
\Delta t = \frac{\psi}{\rho C_p} \int \sigma(\dot{\varepsilon}) \dot{\varepsilon} \, d\varepsilon \quad \text{when } \dot{\varepsilon} \geq 1 \, \text{s}^{-1} \tag{2.33}
\]

where, \( \psi \) is the conversion factor from mechanical work to heating, \( \rho \) is the density and, \( C_p \) is the specific heat capacity of the material. The majority of the deformation energy used during plastic deformation is dissipated as heat, with the remainder retained as strain energy associated with lattice deformation. The conversion factor from the mechanical work to heating (\( \psi \)) is commonly set at 95% [37]. However, there is no agreement on the value of \( \psi \) that works for all materials. The parameter, \( \psi \), has been found to be history dependent on both strain and strain rate [38-41]. Two-phase materials and crystalline materials possessing a crystal structure of low symmetry (i.e. hexagonal) will have a larger \( \psi \) value than cubic-lattice materials, especially at lower strains. The temperature dependence of the specific heat capacity, \( C_p \), is determined from:

\[
C_p = C_{p0} + C_{p1} T + \frac{C_{p2}}{T^2} \tag{2.34}
\]

where, \( C_{p0}, C_{p1}, \) and \( C_{p2} \) are constants. If the temperature dependence of specific heat capacity is not available, constant \( C_{p1} \) and \( C_{p2} \) can be set to zero, which implies that the specific heat capacity for a material is constant with temperature.
Determination of $\dot{\sigma}_i$ and $g_{0i}$.

Substituting equation (2.30) into equation (2.27), $\dot{\sigma}_i$ can be found if $\sigma$ and $\sigma_a$ are known and $\dot{\varepsilon}_s$ is zero. Assuming $\dot{\varepsilon}_s$ (at yield) is zero when plastic strain is zero, we get the following relationship:

$$\sigma = \sigma_a + \left[1 - \left( \frac{kT}{g_0 b^3 \mu} \ln \frac{\dot{\varepsilon}_0}{\dot{\varepsilon}_s} \right) \right] \frac{\dot{\sigma}_i \mu}{\mu_0}$$

(2.35)

Rearranging the equation in the form for a straight line gives:

$$\left( \frac{\sigma - \sigma_a}{\mu} \right)^{p_i} = \left( \frac{\dot{\sigma}_i}{\mu} \right)^{p_i} \left( \frac{1}{g_0} \right)^{q_i} \left( \frac{kT}{b^3 \mu} \ln \frac{\dot{\varepsilon}}{\dot{\varepsilon}_0} \right)^{q_i} + \left( \frac{\dot{\sigma}_i}{\mu_0} \right)^{p_i}$$

(2.36)

$$Y = M \times X + C$$

The modified Arrhenius (Fisher) plots based on equation (2.36) can be used to determine the normalised activation energy, $g_{0i}$, and the intrinsic thermally activated portion of the flow stress $\dot{\sigma}_i$ (shown in Figure 2-10). The parameters $p_i$ and $q_i$ for the metal systems tested have been suggested to be $\frac{1}{2}$ and $\frac{1}{2}$ respectively [34 & 38]. Alternative values can be obtained depending on the assumed shape of the activation energy profile or the obstacle force-distance profile. However, if assuming a rectangular force-distance profile, then the parameters $p_i$ and $q_i$ are given as $\frac{1}{2}$ and 1 [34 & 38] respectively which give reasonable values of $g_{0i}$. The reference strain rate $\dot{\varepsilon}_0$ in the equation is assumed to be $10^{-3}$ s$^{-1}$, Boltzmann’s constant $k$ is $1.3806503 \times 10^{-23}$ J/K, and the magnitude of the Burger’s vector $b$ is $2.48 \times 10^{-10}$ m [39].
Determination of $\sigma_{o_w}$ and $g_{o_w}$

Once estimates have been obtained for $\hat{\sigma}_i$ and $g_{0i}$, the value of $S_i \hat{\sigma}_i$ can be calculated for a particular strain rate and temperature. From equation (2.27) we get:

$$\hat{\sigma}_\varepsilon = \frac{1}{S_\varepsilon} \left[ \frac{\mu_0}{\mu} (\sigma - \sigma_a) - S_i \hat{\sigma}_i \right]$$  \hspace{1cm} (2.37)

The strain hardening component $\hat{\sigma}_\varepsilon$ is given by a modified Voce law [40, 41]:

$$\frac{d\sigma_{\varepsilon}}{d\varepsilon_p} = \theta = \theta_0 \left[ 1 - \frac{\hat{\sigma}_\varepsilon}{\hat{\sigma}_{\varepsilon_0}} \right]^k$$  \hspace{1cm} (2.38)

where $\hat{\sigma}_{\varepsilon_0}$ is the saturation stress and $\theta_0$ is the initial hardening modulus. Integrating equation (2.38) gives:

$$\hat{\sigma}_\varepsilon = \hat{\sigma}_{\varepsilon_0} \left[ 1 - \left( \frac{1 - \varepsilon (1 - k)}{\hat{\sigma}_{\varepsilon_0}} \right) \theta_0 \right]^{\frac{k}{k + 1}}$$  \hspace{1cm} (2.39)
The saturation stress $\hat{\sigma}_{vs}$ is related to strain rate and temperature through:

$$\ln \hat{\sigma}_{vs} = -\frac{1}{g_{0,v}} \frac{kT}{b^v \mu} \ln \left( \frac{\dot{\varepsilon}}{\dot{\varepsilon}_{0,v}} \right) + \ln \hat{\sigma}_{0,v} \quad (2.40)$$

$$Y = M \ X + C$$

The value of $\hat{\sigma}_{vs}$ can be determined from a plot of $\ln \hat{\sigma}_{vs}$ versus $\frac{kT}{\mu b^v} \ln \left( \frac{\dot{\varepsilon}}{\dot{\varepsilon}_{0,v}} \right)$ as shown in Figure 2-11.

![Fisher plot based on equation (2.40) used to determine $\sigma_{0,v}$ and $g_{0,v}$](#)

**Figure 2-11.** Fisher plot based on equation (2.40) used to determine $\sigma_{0,v}$ and $g_{0,v}$.

### 2.4 Conclusions

This chapter has briefly introduced the mechanical testing techniques available for generating material data over a wide range of strain rates, with emphasis on the challenging set of problems encountered in conducting tensile tests in the dynamic strain rate regime. The importance the microstructure of the material has on the dynamic deformation behaviour of metallic materials has also been shown with particular emphasis on the influence the grain size and crystal structure has on the flow curves. The most common constitutive material models, both of the empirical and physical type have been introduced which capture the behaviour of metallic materials within the quasi-static to dynamic range.
The following chapters evaluate these topics in turn. Chapter 3 investigates the practical aspects of tensile testing metallic materials within the dynamic strain rate range; while Chapter 4 and 5 develops the methodology of processing of the tensile data acquired over the quasi-static to dynamic range in order to acquire meaningful material properties and compares it with data acquired from compression experiments using the SHPB. Chapter 6 assesses the degree to which the constitutive models discussed in this chapter represent the material behaviour over this strain rate range.
CHAPTER 3

Evaluation of Dynamic Tensile Testing

3.1 Introduction

This chapter details an experimental investigation into dynamic tensile tests carried out using a servo-hydraulic machine; the objectives being: (1) to investigate the relationships between the stress wave propagation and amplitude during loading, and the conditions necessary for stress equilibrium in the specimen to ensure the specimen remains under homogenous deformation throughout the test; (2) to assess the methods available for measurement of load and displacement at strain rates up to $10^3$ s$^{-1}$; (3) the interpretation and processing of the stress-strain signal produced under dynamic loading conditions so that valid stress-strain data in a material test can be obtained; and (4) the evaluation of DIC as a strain measurement/analysis technique.

3.2 Experimental Procedure

3.2.1 Tensile Test Specimen

The geometry of the tensile test samples used in this work was initially based on the American ASTM standard E8/E 8M-08 [42]. However, it was found that a number of modifications were necessary to be made to these round tensile specimens such as adding M6 threads to the ends of the sample in order to fit the grips available. Additionally, owing to the need to image the sample during testing, the original 2.29 mm gauge length diameter was increased to 3.5 mm. It was anticipated that difficulties would be experienced when trying to measure loads at strain rates above $10^2$ s$^{-1}$ and hence a dynamometer gauge section was included on the specimen. In this section, the material remains linear-elastic and therefore enables the load to be measured using strain gauges bonded to the surface. Figure 3-1 shows the final tensile specimen geometry.
Tensile specimens were prepared from oxygen free electrolytic (OFE) copper rod, aluminium 6061-T651 alloy plate, and tantalum-2.5 wt. % tungsten wrought plate. All materials were initially cut into $65 \times 8 \times 8 \text{ mm}^3$ blanks and then turned to a 6 mm diameter bars on a CNC lathe. The Al 6061-T651 and Ta-2½% W samples were cut in the plane of the plate, parallel to each of the two cross rolled directions. The OFE copper samples were cut along the axis of the rod. The thread was then added to the first 13 mm of both ends of all the tensile samples using a die. Finally, the gauge length profile was cut using a CNC lathe.

3.2.2 Test Equipment

Uniaxial tensile tests were carried out in the high strain rate range (i.e. $>10^2 \text{ s}^{-1}$) using an Instron VHS 20/25 servo-hydraulic test machine. The Instron VHS 20/25 servo-hydraulic machine could be operated in either closed or open loop control mode. The closed-loop mode ensures that the actuator moves at a constant speed. In open-loop mode the control system does not have a feedback signal, and therefore the actuator speed is not as well controlled. The maximum speed of the actuator for use in closed-loop mode is typically up to 1.0 m/s, whereas in open loop mode, a higher maximum speed can be achieved (up to 30 m/s). The load was applied by the servo-hydraulic actuator from the top of the sample and the load measured at the bottom stationary end of the sample.

In order to overcome the inertia effect it was necessary to incorporate a lost motion (slack adapter) device into the load train to provide a free travel distance thus enabling the actuator to accelerate and achieve a consistent velocity before loading the specimen. The general arrangement of the load train is shown in Figure 3-2.

![Figure 3-1. Tensile sample geometry](image)
3.2.3 Load Measurement

The load was measured using a PCB 222B piezoelectric load cell and by direct measurement on the sample surface using a Vishay EA-06-062AQ-350 strain gauge bonded to the dynamometer section of each sample. Vishay M-Bond AE10 adhesive was used to bond the strain gauge to the sample and Vishay 330-DFV wire leads were then soldered to the strain gauge terminals. The strain gauges were calibrated by loading the samples quasi-statically in an Instron test machine to a maximum of 55% of the anticipated yield load for the specimen. It was assumed that the strain gauge calibration would not be affected by loading rate and therefore a quasi-static test was used to calibrate the strain gauge for measuring at high rates.

Figure 3-2. Lost motion device for intermediate and high strain rate testing [2].
3.2.4 Strain Measurement

At all strain rates, strain data can be obtained from the crosshead position, although it is considered that this is an unreliable means of measuring specimen strain. Conversely, contact extensometers at the high strain rates do not have the necessary response characteristics and are physically too fragile. Therefore Digital Image Correlation (DIC) was the preferred method of strain measurement.

In the measurement of strain the primary interest is in recording the specimen displacement in the axial direction and where the out-of-plane displacement was expected to be minimal at least until the sample began to neck. Therefore, the majority of tests at 760 s\(^{-1}\) were conducted using 2-D DIC and only a limited number of tests were carried out using 3-D DIC. In order to assess the accuracy of the DIC technique in recording strain data, a number of low and intermediate strain rate tests were carried out on an Instron 5584 tensile test machine. Strain measurements during these tests used an Instron 2620-601 dynamic extensometer with a 12.5 mm gauge length in conjunction with DIC.

Measurement by DIC requires a speckled pattern to be applied onto the surface of the test sample’s gauge length. In the present study 30 to 60 minutes before testing, the samples were etched and sprayed white with an acrylic based matt paint, to provide a plain featureless background. Once the white background paint had dried, a speckled pattern covering \(\sim 50\%\) of the surface using black matt paint was then applied. A short time period of approximately 30 minutes was required after applying the paint and before tensile testing in order to allow sufficient time for the paint to bind to the surface of the specimen, but not long enough for the paint to become too dry and brittle, and subsequently becoming detached from the sample at low strains early in the test.

A Photron SA1.1 high speed camera with a frame rate of 75,000 Hz and a resolution of \(64 \times 480\) pixels was used to capture the sample deformation optically. Two 1.25 kW halogen lights were use to illuminate the sample and were automatically switched on a few seconds before loading to minimise heating of the sample. Aluminium foil was also used to reflect light to ensure the light was evenly distribute from multiple direction to improve the quality of the image. During the high strain rate tests, the
output signals from the strain gauge amplifier, the load cell, and the displacement from the Instron crosshead, were all connected to an Imatek C2008 data acquisition system with the PC running ImpAcqt software. This system is designed for analysing data from high speed tests. The Imatek system was set to trigger from the displacement of the actuator at a point just before the end of travel of the lost motion device when the sample begins to be loaded. The trigger from the Imatek system was connected to the high speed camera to start recording. A synchronisation pulse was sent from the camera back into the Imatek system so that the exact time at which images captured were known. Figures 4-3 and 4-4 show the experimental set-up.

The images captured were analysed using Aramis-v5.4 software produced by GOM mbH. Before analysis the images were divided up into 13 × 13 pixel facet with a two pixel overlap. Uniaxial strain data was extracted from the results by selecting two facets, one at either end of the gauge length. The strain was then calculated from measuring the displacement distance between the two points, using the sample in the unloaded condition as the reference distance. This is effectively using the DIC method as a non-contact optical extensometer rather than a full-field strain measurement technique.

Figure 3-3. General overview of the experimental configuration for high strain rate testing using the DIC method.
3.2.5 Determination of Dynamic Stress Equilibrium

The SHPB quantitative criterion for the condition of dynamic stress equilibrium was used to determine if the material tests in the high rate testing range > 100 s⁻¹ were valid (see Section 2.1.3). The number of round trips the stress wave made during the test, \( N \), was estimated from equation (2.17). The load rising time, \( t_r \), was measured from the stress strain curves and defined; as the time between the introduction of the load to the peak stress. The length of the specimen was taken as the distance between the grips, which included the gauge length, the shoulder areas, and the dynamometer section, a total distance of 38.41 mm. The elastic stress wave was calculated using equation (2.15), where Young’s Modulus (\( E \)) and the density (\( \rho \)) of the test materials were obtained from the open literature [43-45].

3.2.6 Assessment of Load Measurement and Data Processing Methods

Stress-strain curves at a strain rate of \(~760\) s⁻¹ (actuator speed of 20 mm/s) were produced using load measurements obtained from:
(1) The piezoelectric load cell.
(2) The strain gauge, calibrated up to a maximum of 55% of the expected yield load for each specimen.
(3) The strain gauge, where load readings were adjusted to reflect the readout from the load cell, thereby effectively using the load cell to calibrate the strain gauge.
(4) A 3rd order polynomial fit curve produced from the readings obtained from the piezoelectric load cell.

From each of these curves, the stress at 0.2% strain and the UTS were determined. The data acquired by each method were tabulated and a comparison of the results was made assessing each method for its consistency by standard deviation.

3.2.7 Evaluation of DIC as a Measurement and Analytical Technique

A number of tests were carried out at low and intermediate strain rates, measuring strain by both high speed imaging and by extensometer in order to compare the strain data acquired by both techniques. The imaging of the stress distribution by DIC obtained at low to dynamic strain rates was also evaluated the value of the information it provided. In addition the possibility of utilising the DIC images in extending the true stress-true strain curve beyond the instability point was also investigated. This was carried out by measuring the neck curvature and its radius during the development of the neck to eventual failure using software produced by Dunnett & Balint [46, & 47], the measurements being fed into a correction factor developed by Bridgeman [3].

3.3 Results and Discussion

3.3.1 Load Signal and Dynamic Equilibrium

The typical load signals produced at a strain rate of 760 s⁻¹ obtained from the piezoelectric load cell for the tantalum, aluminium and copper test materials are shown in Figure 3-5. The traces show the load rising time followed by considerable load oscillation. In all three specimens the load rises gradually to about 0.2 kN in the initial 0.06 ms and then rises rapidly to a peak, of about 5 kN, 4.2 kN and 3.3 kN for the tantalum, copper and aluminium test samples respectively. The initial gradual
increase in load can be attributed to the compression of the damping layer. The load rising time to the peak load thereafter was about 0.115, 0.121 and 0.134 ms and the total rise time therefore was taken as 0.175, 0.181, and 0.194 ms for the tantalum aluminium, and copper test materials respectively.

![Graph](image)

*Figure 3-5. Load history trace of the metallic materials tested at 760 s⁻¹.*

The data obtained from these traces are tabulated in Table 3-1, from which the elastic wave velocity and number of round trips the stress wave makes in the sample were calculated from equation (2.17).

<table>
<thead>
<tr>
<th>Material</th>
<th>Sample/fixture length (m)</th>
<th>Young’s Modulus (GPa) [43-45]</th>
<th>Density (kg/m³) [43-45]</th>
<th>Elastic stress wave velocity (m/s)</th>
<th>Average load rising rate (ms)</th>
<th>Number of round-trips of the stress wave</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al-6061</td>
<td>38.4 x10⁻³</td>
<td>70</td>
<td>2,700</td>
<td>5,092</td>
<td>0.175</td>
<td>11.60</td>
</tr>
<tr>
<td>OFE Cu</td>
<td>38.4 x10⁻³</td>
<td>118</td>
<td>8,940</td>
<td>3,633</td>
<td>0.181</td>
<td>8.56</td>
</tr>
<tr>
<td>Ta-W</td>
<td>38.4 x10⁻³</td>
<td>186</td>
<td>16,600</td>
<td>3,347</td>
<td>0.194</td>
<td>8.46</td>
</tr>
</tbody>
</table>

As can be seen from Table 3-1, the calculated number of round-trips the stress wave made in each of the test samples at this strain rate (760 s⁻¹) was 11.60, 8.56, and 8.46 for the tantalum, aluminium, and copper test samples respectively. Accepting the criterion for dynamic stress equilibrium within a test sample as specified for SHPB
validation of $N \geq 3$, clearly shows the criteria for a valid test has been met. However, while the qualitative criterion of an acceptable or unacceptable test by visual inspection of the stress-strain curve is a more objective measure, it shows that a considerable amount of oscillation is present when the load reading is taken from the piezoelectric load cell. This visual inspection strongly indicates that the SHPB quantitative criterion was not met. On the other hand, load readings as recorded by the strain gauge show significantly reduced oscillation (Figure 3-9) and are considerably more visually acceptable. It is likely that the load cell is recording movement of the stress wave within either the entire length or partial length of the load train, while the strain gauge only records the stress wave within the specimen itself. It is not possible to calculate the number of round-trips the stress wave makes within the load train to any degree of accuracy owing to the composite nature of the materials involved. However, a rough calculation based merely on the length of the load train indicates it would be approximately 1, which would explain the large oscillation observed in the load cell signal.

Close examination of the high speed video images of the Ta-W wrought material at a strain rate of $760 \text{ s}^{-1}$ showed a neck forming at the top end of the specimen gauge length, before a new neck forms further down the gauge section. This early necking halts on the formation of this second neck and it is this second neck that proceeds to failure. As the load is applied from the top, the stress wave will naturally travel from the top and down the specimen, before being reflected and returning back up the sample. As the top of the specimen gauge length sees the stress wave first, the formation of this neck early during the load application indicates that the specimen may not be in a state of dynamic stress equilibrium during the test.

The high speed video images show the time inception of the first neck was 0.13 ms and was interrupted from further growth by the inception of the second neck at 0.25 ms. The distance between the two necks was measured as $\sim 12.95 \text{ mm}$, and the region connecting them did not yield but gradually unloaded. DIC images of this double necking of the specimen are shown in Figure 3-6.
Figure 3-6. DIC images of double necking of the Ta-2½%W sample. The image shows the strain distribution in the sample, colour coded from blue-green-yellow-red representing the low to high strain present within the tensile specimen respectively.

The phenomenon of multiple necking at high strain rates has been observed in other materials, such as steel, aluminium, copper, and lead solder [48-52]. The tensile curves where this double necking phenomenon occurs can show peaks corresponding to the formation of the two separate necks in the gauge section of the tensile specimen [46]. Examination of the load-time curves for the tests conducted in the Ta-W wrought material at this strain rate (760 s⁻¹) does show two peaks (Figure 3-7).
The first and major peak occurs at 0.186 ms, while a much shallower peak (difficult to see precisely on the load-time trace) occurs at 0.282 ms before progressing to failure (a time interval of 0.096 ms between the two peaks). These peaks correspond closely with the times in the necking sequence observed in the high speed video images. No double peaks in the load-time traces were observed at quasi-static or intermediate rates.

This double necking sequence was not observed in high speed video images of tests carried out at this high strain rate (760 s⁻¹) in the aluminium alloy samples. The fact that the double necking phenomenon is not observed in the aluminium alloy suggests that the strain rate threshold for double necking to occur is dependent on the test material. This may simply be due to the fact that the elastic stress wave velocity in the aluminium as calculated by equation (2.15) is ~5 m/s, about 1.5 times faster than that in the tantalum alloy of ~3.3 m/s (see Table 3-1) resulting in the dynamic stress equilibrium being achieved in a shorter time in aluminium than that in the tantalum sample.

The dynamics of plastic instability are quite complicated, and have attracted a considerable amount of interest among researchers over the years. As early as 1885, Considere [53], proposed the concept of necking failure, he suggested that localised...
necking begins when the strain hardening rate is equal to the thickness reduction rate. Work reported by Yazzie et al. [48] on the tensile testing of pure tin solder noted that double necking only took place above a critical strain rate, which in this work was found to be about 10 s\(^{-1}\). They concluded that the length of the gauge section plays an important role in the deformation of the specimen, as the double necking only occurred for specimens with relatively long gauge sections, and occurred at both ends of the gauge section. Using Finite Element Analysis (FEA) they concluded that the local plastic strain rate, closely related to the displacement rate, played an important role in the potential sites for necking. Therefore, in a specimen with a relatively long strain gauge section the maximal principal plastic strain reaches the necking criterion globally throughout the gauge section, while the maximal principal strain rate criterion is met locally close to the both ends of the gauge section. While in the case of specimens of shorter gauge length, the maximal principal strain rate is not localised at either end of the gauge length section but eventually single necking takes place approximately mid-gauge section. They also indicate that the critical length of the gauge section for double necking to occur may also depend on other parameters, such as the constitutive behaviour of the material. This would indicate that the mechanical/microscopic properties of a material (such as elastic modulus and formability) may also influence the strain rate threshold at which double necking occurs, and therefore whether a test achieves an acceptable criterion for dynamic stress equilibrium.

As has been mentioned in Section 2.1.3, in dynamic testing it is impossible to reach the state of stress equilibrium achieved in quasi-static testing and therefore only a dynamic stress equilibrium condition is sought, which has been defined in SHPB as the stress wave making at least three round-trips in the specimen before yielding takes place. Whether this definition is suitable for all materials or for larger test samples, as is the case for the standard sized tensile specimen, needs to be determined. However, these observations would suggest that the critical strain rate at which double necking occurs in a tensile sample is both material and geometry dependent.

In order to try and assess the strain dynamics occurring in the tantalum alloy tensile specimen at this strain rate, Finite Element Analysis (FEA) was carried out using ABAQUS/Explicit [54]. Material and geometric parameters for the specimen were
used in the simulation with one end of this specimen fixed (clamped) using “encastre” boundary conditions and a load applied via a striker at the top end to prevent non-axial movement of the simulated specimen (See Appendix). Figure 3-8 shows 2-D axis symmetric results of the analysis.

![Figure 3-8. FEA images of double necking in tantalum specimens at 760 s⁻¹. The image shows the strain distribution in the sample, colour coded from blue-green-yellow-red, representing the low to high strain values present within the tensile model image respectively.](image)

The series of images in the FE Analysis (Figure 3-8) shows axial strain contours in the strain gauge region of a specimen over the duration of a tensile test. Strain within the specimens gauge section is homogenous up to about 0.28 ms from application of the load at the top end of the gauge section, after which two heterogeneous strain distributions start to develop. The strain in these two areas develops further with time while the remainder of the gauge section undergoes strain relaxation. At about 0.38 ms the lower of the high strain areas (necking) starts to relax while the other continues to develop to failure. While the FEA sequence differs somewhat from that observed in the high speed video recordings of the “real life” test, the FE analysis does show the occurrence of the double necking in the tantalum at this strain rate. Further development using the FE analysis model could be used to give an indication of the
strain rate threshold dependency of the tensile specimen material and geometry for double necking to occur.

3.3.2 Processing and Evaluating the Stress-Strain Curve

Figure 3-9 shows a typical stress-strain curve produced from the load cell and strain gauge readings for an OFE copper test sample. The amplitude of oscillation in the stress-strain curve from readings taken from the strain gauge is considerably lower than that recorded by the load cell. The peak to peak average oscillation from the load cell signal is approximately 70 MPa, while that recorded by the strain gauge is about 22 MPa. The stress-strain curve produced from the strain gauge readings calibrated under quasi-static conditions within the elastic region can often be higher or lower than the readings produced by the load cell. The example shown in Figure 3-9 shows only a small difference between the two curves. Adjustment of the curve obtained by the strain gauge was made by minimising (least squares) the difference between it and those obtained from the load cell, effectively using the load cell as a calibration reference for the strain gauge at these loading rates (see Figure 3-9).

Figure 3-9. A typical example of stress as function of strain traces provided by a load cell and strain gauge for OFE copper.
The method for extracting the stress values at 0.2% off-set strain from these three curves are shown in Figure (3-10). Engineering stress-engineering plastic strain curves were then produced, as the example in Figure (3-11) shows. Additionally, a polynomial curve fit (3rd order) using the load cell data is also shown in Figure (3-11). Table (3-2) shows a comparison of the UTS and stress at 0.2% strain data extracted from these four curves for all the tests carried out at a 760 s⁻¹ strain rate.

Figure 3-10. Example of the determination of stress at 0.2% off-set strain for copper.

Figure 3-11. Example of engineering stress as a function of engineering plastic strain for data acquired and processed by four different methods for OFE Copper.
Table 3-2. Comparison of methods for determining stress data.

<table>
<thead>
<tr>
<th>Material</th>
<th>Stress Measurement Method</th>
<th>Engineering Stress (MPa)</th>
<th>0.2% Strain</th>
<th>UTS</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al-6061</td>
<td>Load cell</td>
<td>327.9 ± 27.4</td>
<td>342.3 ± 22.7</td>
<td></td>
</tr>
<tr>
<td></td>
<td>Strain gauge-calibration</td>
<td>309.9 ± 15.7</td>
<td>334.0 ± 16.0</td>
<td></td>
</tr>
<tr>
<td></td>
<td>Strain gauge-adjusted</td>
<td>300.4 ± 11.1</td>
<td>325.7 ± 9.1</td>
<td></td>
</tr>
<tr>
<td></td>
<td>Polynomial fit</td>
<td>299.7 ± 13.7</td>
<td>324.3 ± 9.2</td>
<td></td>
</tr>
<tr>
<td>OFE Cu</td>
<td>Load cell</td>
<td>395.0 ± 23.2</td>
<td>426.1 ± 5.6</td>
<td></td>
</tr>
<tr>
<td></td>
<td>Strain gauge-calibration</td>
<td>401.5 ± 19.8</td>
<td>428.6 ± 28.8</td>
<td></td>
</tr>
<tr>
<td></td>
<td>Strain gauge-adjusted</td>
<td>370.8 ± 5.8</td>
<td>394.8 ± 3.2</td>
<td></td>
</tr>
<tr>
<td></td>
<td>Polynomial fit</td>
<td>365.3 ± 3.9</td>
<td>393.1 ± 2.2</td>
<td></td>
</tr>
<tr>
<td>Ta-W wrought</td>
<td>Load cell</td>
<td>489.1 ± 13.9</td>
<td>497.3 ± 12.8</td>
<td></td>
</tr>
<tr>
<td></td>
<td>Strain gauge-calibration</td>
<td>430.2 ± 60.1</td>
<td>463.4 ± 56.4</td>
<td></td>
</tr>
<tr>
<td></td>
<td>Strain gauge-adjusted</td>
<td>439.8 ± 22.3</td>
<td>472.3 ± 13.8</td>
<td></td>
</tr>
<tr>
<td></td>
<td>Polynomial fit</td>
<td>428.7 ± 18.0</td>
<td>461.6 ± 11.3</td>
<td></td>
</tr>
</tbody>
</table>

A comparison of the data obtained by the various test and data processing methods shows that the mean UTS and stress at 0.2% offset strain values as obtained from the load cell are often considerably higher than those values obtained by the strain gauge-adjusted and polynomial fit methods. This is clearly due to the large oscillation present in the load signal caused by “system ring” (section 2.1.3), which also accounted for the large variation present in the values as indicated by the standard deviation. The mean readings obtained from the quasi-statically calibrated strain gauges can also be high and have a large standard deviation. In this case the stress-strain traces produced by this method varied considerably in comparison with those produced by the load cell. It is evident from these results that the calibration of the strain gauges under quasi-static conditions did not prove sufficiently accurate for testing under dynamic conditions. Data obtained by the adjustment of the strain gauge curves using the load cell as a reference and those determined by fitting a polynomial curve from the load cell readings gave the most consistent values with the least variation. Both these methods gave consistently similar mean values for stress at 0.2% offset strain and UTS in all three materials.

3.3.3 Assessment of Strain Measurement by High Speed Imaging

Figure (3-12) shows the engineering stress-engineering strain curves for the Ta-2.5%W alloy where the strain has been measured by extensometer and by two high speed DIC cameras at a strain rate of $6.75 \times 10^{-1} \text{ s}^{-1}$. The strain readings obtained by
DIC appear to closely follow those obtained by extensometer up to the UTS value of the material. At this point necking (localised extension) of the tensile sample starts and the measured strain will depend on which two points on the specimen are taken as the gauge length. The gauge length of the extensometer was 12.5 mm, while the gauge length measured by the DIC image covered approximately the entire gauge length of the specimen (~17.5mm). The total strain in the specimen as calculated by equation (2-2) will contain proportionately much less of the necked region in the DIC measurement than that from the extensometer measurement and therefore give a proportionately smaller figure of total strain. There is also a difference in strain readings within the elastic region of the stress-strain curve. This is the region of the curve in which the stress rises quickly and therefore any difference in synchronising the strain readings with the load readings will be at its greatest.

Figure 3-12 also shows a difference in the strain reading between images obtained by the two high speed cameras. Again the only difference in strain reading was in the elastic region and after the UTS. The reasons for the difference being the same as those mentioned above between the DIC readings and those obtained from the extensometer, although the difference was much smaller. A comparison of strain readings at UTS obtained by DIC measurement and those by extensometer are shown in Table 3-3.

![Figure 3-12. Example of engineering stress as a function of engineering plastic strain for data acquired using an extensometer and by the imaging of the specimen by two synchronised high speed cameras (Strain rate 0.675 s⁻¹).](image-url)
Table 3-3. Comparison of strain readings obtained from the extensometer and DIC

<table>
<thead>
<tr>
<th>Material</th>
<th>Strain Rate (s(^{-1}))</th>
<th>% Strain at UTS</th>
<th>% Difference in DIC Reading</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>Extensometer</td>
<td>DIC</td>
</tr>
<tr>
<td>Ta-W</td>
<td>(1 \times 10^{-3})</td>
<td>21.8 ± 0.5</td>
<td>18.6 ± 1.6</td>
</tr>
<tr>
<td>Ta-W</td>
<td>0.675</td>
<td>14.4 ± 1.7</td>
<td>13.9 ± 2.3</td>
</tr>
<tr>
<td>Al-6061</td>
<td>0.675</td>
<td>3.3 ± 0.3</td>
<td>3.4 ± 0.3</td>
</tr>
</tbody>
</table>

What is most noticeable from Table 3-3 is the larger standard deviation in the readings obtained with DIC. This is undoubtedly due to the variation in gauge length used when measuring the DIC images, as opposed to the constant 12.5 mm gauge length of the extensometer. It is apparent that greater care needs to be taken when measuring strain via this method in order to maintain consistency. The lower percentage strain values recorded via DIC are as discussed above due to the larger gauge length used for measuring by this method.

Table 3-4 shows the standard deviation of the strain readings (%) obtained by DIC from those obtained by extensometer covering strain ranges at different parts of the engineering stress–engineering strain curve up to the UTS value. From this Table it can be seen that the larger deviation is in the earlier part of the curve, principally in the elastic region, for the reasons mentioned above.

Table 3-4. Standard deviation of the DIC strain readings with those obtained by an Extensometer (Ta-2.5% W alloy at \(1 \times 10^{-3}\) s\(^{-1}\))

<table>
<thead>
<tr>
<th>% Strain Range as Measured by an Extensometer</th>
<th>0–0.5</th>
<th>0.5–1</th>
<th>1–2</th>
<th>2–4</th>
<th>4–8</th>
<th>8–16</th>
<th>16–UTS</th>
</tr>
</thead>
<tbody>
<tr>
<td>Standard Deviation of Strain measured by DIC</td>
<td>0.58</td>
<td>0.23</td>
<td>0.18</td>
<td>0.11</td>
<td>0.06</td>
<td>0.11</td>
<td>0.05</td>
</tr>
</tbody>
</table>

3.3.4 Analysis of the DIC Technique

**Strain Distribution**

Figure 3-13 shows a series of pictorial DIC images obtained on the aluminium-6061 alloy during a tensile test at a strain rate of 0.675 s\(^{-1}\). The figure shows the strain distribution within the specimen, which is represented by the colour scale shown on the right hand side.
Figure 3-13. A series of DIC images of an Aluminium 6061 alloy tensile specimen over the duration of a tensile test

Image 2 of the 3-D DIC image sequence clearly shows elastic stress distribution along the gauge length of the sample (green colour) and the initiation of the neck (yellow/red colour). Image 3 shows an increase in the strain at the region of the specimen where the necking is occurring and a reduction/relaxation in elastic strain in the rest of the specimen gauge length. The series of images can be calibrated to give an approximate quantitative value of the stress distribution throughout the duration of the tensile test and the stress distribution can be used to give an indication whether the tensile specimen is in dynamic equilibrium indicating any double/multiple necking in the specimen if it occurs. However, strain distribution images are typically lost in the neck region in the later part of the deformation process owing to the paint becoming detached from the specimen.

Imaging at the higher strain rates (>10^2 s^-1) is challenging and puts great demands on the DIC system. The image capture rate of the system and spatial resolution demands at these rates can mean the system has difficulties tracking the speckle pattern during the entire test resulting in strain distribution images being periodically lost on areas of the tensile sample. Optimising size and distribution of the speckles applied to the surface of the gauge section of the tensile sample can improve the quality and consistency of the images, but achieving this is difficult. Care must also be taken that
the DIC imaging represents the strain taking place in the test sample and not simply tracking movement of the paint. This can be largely overcome by ensuring the paint is well bonded to the surface of the specimen, and this may be achieved by lightly etching the surface prior to applying the paint and applying as thin a layer as practically possible.

While imaging the strain distribution in a tensile specimen during a test can be achieved using 2-D DIC, more detailed imaging can be better produced using 3-D DIC. While 2-D DIC will track in-plane displacement in the specimen, 3-D DIC is required to track out of plane displacement, such as that which occurs during the necking of the sample. There are a number of difficulties when measuring diametrical contraction in a specimen using 2-D image correlation, first there are difficulties in correlating data close to the edges of the image/specimen, which makes diametrical measurement inaccurate in those areas. Secondly, the circular cross sectional nature of the tensile specimen means that some areas of the specimen will be out of the depth of focus of the lens which makes results unreliable. These problems can be overcome by using a 3-D DIC system which enables full field displacement data to be collected. The 3-D measurement of the out-of-plane displacements allows radial contraction of a specimen in the direction perpendicular to the field of view. This is of particular importance where the specimen undergoes significant anisotropic deformation, an effect that is common in wrought products.

**Measurement of Stress during Necking**

At the point when necking takes place in a tensile sample the true stress-true strain trace as determined by equations (2.6) and (2.7) becomes invalid as the uni-axial stress distribution is disrupted by the geometrical irregularity in the specimen (see section 2.1.1). However, DIC images allow the continuous measurement of the radius of the neck at its thinnest part and measurement of the neck’s radius of curvature, which can be fed into the Bridgeman correction factor [3] allowing the true stress-true strain curve to be extended for almost the entire test. Figure 3-14 shows an example of a true stress-strain curve extended beyond the instability point using software [47] developed for utilisation of the correction factor.
DIC images suitable for making these measurements were successfully captured at dynamic strain rates up to 760 s\(^{-1}\). The measurements can be successfully made using 2-D DIC, providing the sample undergoes isotropic deformation during necking. As mentioned above, 3-D DIC is more suitable for making measurements in materials that deform anisotropically during necking. However, in materials where severe anisotropic deformation takes place in the necked region, there is a need to measure diametrical contractions in two orthogonal directions simultaneously during the tensile test and in these situations Zanganeh et al. [55] have proposed a method to measure deformation anisotropy using a three camera system.

### 3.4 Conclusions

The conclusions of the evaluation of dynamic tensile testing are as follows:

1) Qualitative assessment of the stress equilibrium within a sample at a strain rate of 760/s, made by visual examination of the stress-strain curve produced from load cell measurements indicated that the specimens were not in dynamic equilibrium, while the stress equilibrium by strain gauges attached directly to the samples suggested that dynamic equilibrium was achieved.
2) The quantitative criterion defined for stress equilibrium in SHPB experiments indicated that dynamic stress equilibrium within the test sample was present when the stress wave is assumed to be travelling back and forth within the sample only. The present results suggest that the load cell records the stress wave in the entire or partial load train, while the strain gauge only records the stress waves within the test specimen, and hence the different indications from the two assessment methods.

3) High speed video images indicated a difference in the high rate deformation behaviour of the materials tested. Double necking was observed in the Ta-W wrought samples at a strain rate of 760 s\(^{-1}\). A double peak in the force-time curve and FEA analysis confirmed that double necking did occur in the sample at this strain rate. However, double necking was not observed in the aluminium alloy at this strain rate. Previous work [40] concluded that the length of the gauge section plays an important role in the deformation of a specimen, strongly suggesting that dynamic stress equilibrium is both material and geometry dependent, and that each material and specimen geometry combination will have its own strain rate threshold at which double necking occurs.

4) The acquisition of useful data from dynamic tensile test curves is not as simple a task as in quasi-static testing, and appropriate data processing methods need to be developed. Extracting material data (i.e. stress at 0.2% yield stress and UTS) from the stress-strain curves recorded using the piezoelectric load cell at high strain rates gave results that were affected by excessive oscillation in the load train and consequently showed a large standard of deviation. Removal of the elastic region of the curve and the large initial overshoot allowed a third order polynomial curve to be fitted to the load cell signal which gives consistent data to be extracted such as the UTS, and by projecting the curve back (if necessary) to the 0.2% off-set strain for a 0.2% proof stress value.

5) The load recorded from strain gauges attached to the dynamometer section of the tensile specimen showed considerably less oscillation in the load signal than that recorded by the piezoelectric load cell at high strain rates. However, calibrating the strain gauge quasi-statically within the elastic region of the material prior to
testing gave a large variation in results, indicating that this method of calibrating
the strain gauge was not suitable at higher strain rates and strains.

6) Calibration of the strain gauge using the load cell reading as a reference gave
results with a low standard deviation similar to that achieved from fitting a
polynomial curve to the load cell trace.

7) Stress-strain curves at intermediate strain rates (~ 0.675 s⁻¹) produced by DIC as a
measurement of strain corresponded well with those produced using a dynamic
extensometer within the plastic region of the curve up to the instability point
(UTS). Plastic strain readings beyond the instability point will depend on where
the strain is measured on the specimen gauge length (a measurement made over a
larger gauge length measurement will indicate a smaller strain than one taken over
a shorter length). A large difference in readings was also present at low strain
readings within the elastic region of the curve. In this region the stress rises
extremely rapidly (due to tensioning of the load frame) with strain increasing the
sensitivity in synchronising the data acquisition during the test.

8) The DIC technique showed that it can image the strain distribution within the test
sample gauge length by a colour scale, giving a semi-quantitative value of strain.
Areas of high strain are clearly visible in the neck region with a relaxation of
stress in the rest of the sample. It has been shown that this technique can be used
to determine if the test sample is in stress equilibrium, the formation of a double
neck in the specimen can indicate stress equilibrium has not been achieved.
However, the work showed that strain distribution in the neck region is lost as
necking progresses because tracking of the speckled pattern becomes impossible
once the paint fragments and becomes detached from this region of the specimen.
This work also showed that care needs be taken so that the DIC image is
representative of the strain taking place in the specimen surface and not simply
tracking movement of the paint. This was best achieved by ensuring the strain
gauge portion of the test sample is lightly etched before applying paint and that
the paint coating is as thin as practically possible. Additionally, the paint needed
be applied no more than 60 minutes before testing to avoid excessive drying of the
paint causing it to flake off under moderate amounts of strain.
9) Dimensional measurements taken in the neck region of the test specimen by the DIC technique at dynamic strain rates showed it can be used as a means to extend the true stress-true strain curve beyond the UTS point.
CHAPTER 4

Effect of Strain Rate and Temperature

4.1 Introduction

This chapter details an experimental investigation into the mechanical behaviour of metallic materials under uni-axial tensile tests conditions, conducted using standard and high strain rate tensile machines using standard sized tensile samples covering the quasi-static to dynamic regime at test temperatures of 203 K to 373 K. The materials tensile data acquired from the dynamic testing regime are determined and processed using the practices laid out in Chapter 3. The objectives are: (1) to determine the mechanical response of these metallic materials within this strain rate and temperature test range, and (2) to assess the merits and practical application of the methods laid out in Chapter 3 in determining the mechanical behaviour of a range of metallic materials.

4.2 Experimental Procedure

4.2.1 Test Materials

Three metals/alloys were chosen for evaluation; Oxygen Free Electrolytic (OFE) copper produced as a wrought rod, Aluminium 6061-T6 (Al 6061-T6) alloy supplied as cross rolled wrought plate, and a tantalum-2.5 weight % tungsten alloy (Ta-2.5%W) produced as a cross rolled wrought plate and in a Hot Isostatic Pressed (HIP) form.

Copper was chosen as a standard model material for this work as its properties are well documented in literature. Copper is a Face Centred Cubic (FCC) metal with a low stacking fault energy. OFE copper generally refers to a group of wrought high conductivity copper alloys that have been electrolytically refined to reduce the level of oxygen to 0.001% or below [44]. OFE Copper has an excellent homogeneity of microstructure and under quasi-static strain rates has typical mechanical properties as shown in Table 4-1.
Aluminium 6061-T6 is an alloy which has been shown to have low strain rate sensitivity when tested in compression in the quasi-static to dynamic strain rate regime so was expected to show little change in mechanical properties over this test range [24-29]. Aluminium is a FCC metal with a high stacking fault energy, and Al 6061 is a good general purpose precipitation hardened aluminium alloy with magnesium and silicon as its major alloying elements. Its mechanical properties depend greatly on the temper, or heat treatment, and in the T6 temper condition under quasi-static strain rate conditions are as summarised in Table 4-1 [45].

The Ta-2.5%W alloy was selected for this investigation owing to its favourable mechanical properties of high strength, and extensive ductility. Tantalum is a body centred cubic metal (BCC) and like all BCC metals, exhibits deformation behaviour that is markedly influenced by impurities, alloying additions, crystallographic texture, temperature, and strain rate [43]. The addition of tungsten to tantalum provides a higher strength and better corrosion resistance than pure tantalum while maintaining good fabrication characteristics and may be fabricated in a variety of forms.

The Ta-2.5%W alloy was supplied in two fabricated forms, namely, wrought plate and HIP bars. The plate material procured from Cabot Supermetals was fabricated as electron beam melted and cross rolled wrought plate (1168 mm × 650 mm × 10.16 mm) in an annealed condition to ASTM specification R05252 B708. The mechanical properties of the plate under quasi-static strain rates conditions are summarised in Table 4-1. The HIP bars were manufactured from powder produced by the Plasma Rotating Electrode Process (PREP). This process was favoured as it produces spherical powder allowing a high cold packing density (termed tap density) to be obtained. Powder consolidation was then completed at Birmingham University at a temperature of 1300 °C under a pressure of 200 MPa for 2 hours. No mechanical data is available for the material manufactured in this form.
Table 4-1. Typical mechanical properties of as supplied materials [43-45]

<table>
<thead>
<tr>
<th>Mechanical Properties</th>
<th>Al-6061-T6 Plate</th>
<th>OFE Cu Bar</th>
<th>Ta-2.5%W Plate</th>
</tr>
</thead>
<tbody>
<tr>
<td>Elastic Modulus (GPa)</td>
<td>68.9</td>
<td>118</td>
<td>179</td>
</tr>
<tr>
<td>0.2% Yield Stress (MPa)</td>
<td>275</td>
<td>69</td>
<td>241</td>
</tr>
<tr>
<td>UTS (MPa)</td>
<td>310</td>
<td>220</td>
<td>379</td>
</tr>
<tr>
<td>Reduction in Area at Failure (%)</td>
<td>-</td>
<td>-</td>
<td>80</td>
</tr>
<tr>
<td>Elongation at Failure (%)</td>
<td>12</td>
<td>45</td>
<td>30</td>
</tr>
<tr>
<td>Shear Modulus (GPa)</td>
<td>26</td>
<td>42</td>
<td>65</td>
</tr>
<tr>
<td>Hardness (HV)</td>
<td>100</td>
<td>90-105</td>
<td>90</td>
</tr>
<tr>
<td>Poisson’s Ratio</td>
<td>0.33</td>
<td>0.31</td>
<td>0.35</td>
</tr>
<tr>
<td>Density (g/cm³)</td>
<td>2.70</td>
<td>8.91</td>
<td>16.6</td>
</tr>
<tr>
<td>Melting Point (K)</td>
<td>933</td>
<td>1356</td>
<td>3239</td>
</tr>
</tbody>
</table>

4.2.2 Tensile Test Sample Preparation

The geometry and manufacture of the OFE copper, Al 6061-T6 and Ta-2.5%W wrought tensile test samples used in this work is described in Section 3.2.1. Specimens were prepared taking care not to alter the properties of the material in the as supplied condition. Additionally, samples were also produced from Ta-2.5%W Hot Isostatic Pressings (HIP). Ta-2.5%W HIP material blanks were cut using an electric discharge machining (EDM) to minimise material wastage. The thread was then cut into the first 12 mm of both ends of all the tensile samples using a die. Finally, the gauge length profile was cut using a CNC lathe.

The surface finish of a number of the test specimens was measured using a Taylor-Hobson Form Talysurf PGI series 2 machine in order to ensure variation in the surface finish of the specimens did not significantly affect fracture properties of the samples. The machine uses a diamond tip to follow the profile of the sample and calculated a variety of roughness measurements. The surface roughness is characterised by the arithmetic mean of the absolute differences between the profile and the mean profile height and is defined as $R_a$. Table 4-2 shows $R_a$ values measured in the gauge length for a number of samples. There was little variation in the standard deviation between samples of the sample material while average surface roughness of the samples prepared from the Ta-2.5%W plate; aluminium plate and copper rod were $< 0.81 \mu m$, $1.15 \mu m$ and $0.51 \mu m$ respectively.
Table 4-2. Surface roughness measurements from tensile specimens prepared from Ta-2.5%W, Al 6061-T6 wrought plates and OFE Cu rod.

<table>
<thead>
<tr>
<th>Metal/Alloy</th>
<th>Surface Roughness $R_a$ (μm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>OFE Copper</td>
<td>0.51 ± 0.10</td>
</tr>
<tr>
<td>Al 6061-T6</td>
<td>1.15 ± 0.03</td>
</tr>
<tr>
<td>Ta-2.5%W</td>
<td>0.81 ± 0.09</td>
</tr>
</tbody>
</table>

4.2.3 Tension Testing Machines

Uniaxial tensile tests were carried out in the quasi-static strain rate range ($10^{-3}$ to $10^{-1}$ s$^{-1}$), intermediate strain rate range ($10^{-1}$ s$^{-1}$ to $10^{2}$ s$^{-1}$) and in the dynamic strain rate range (above $10^{2}$ s$^{-1}$). The quasi-static rate tests were conducted on an Instron 5584 or 4466 tensile test machine, the intermediate rate tests on the Instron 5584 (capable of maximum extension speed of 720 mm/min.) and the high rate tests on an Instron VHS 20/25 servo-hydraulic test machine. At both the intermediate and high strain rates it was necessary to incorporate a lost motion device into the test set-up to provide a free travel distance to enable the actuator to accelerate to the desired velocity before loading the specimen. The general experimental test set-up is shown in Figure 3-2.

4.2.4 Load Measurement

A 10 kN (Instron number 2525-804) or 150 kN strain gauge load cell (Instron number 25250171) was used on the Instron 5584 or 4466 tensile test machines at the quasi-static and intermediate strain rates. At high strain rates the load was measured using a PCB 222B piezoelectric load cell and by measuring the load directly in the sample using a Vishay EA-06-062AQ-350 strain gauge bonded to the dynamometer section of each sample as described in Section 3.2.3.

4.2.5 Strain Measurement

An extensometer, generally considered the most accurate method to measure strain, was used for the low and intermediate strain rates. The extensometer was an Instron 2620-601 series dynamic strain gauge extensometer with a 12.5 mm gauge length and giving up to 40 % strain, accurate for direct measurement operating within the temperature range of 193 K to 473 K and operated at a 50 Hz frequency. All data collection and control of the load frame was carried out using Bluehill software. However, the measurement of strain by extensometry is only suitable for strain rates
up to ~10 s\(^{-1}\), beyond which they don’t have the necessary dynamic response characteristics. At the higher strain rates, Digital Image Correlation (DIC) was used as described in Sections 2.1.3 and 3.2.4.

4.2.6 Environmental Chamber

An environmental chamber was used for the low temperature (203 K and 223 K) and high temperature tests (373 K), with liquid nitrogen cooling being used for the low temperature tests. A thermocouple was attached to each of the test samples to ensure the material was at the desired temperature before testing.

4.2.7 Experimental Test Conditions

The tensile behaviour of the OFE copper, aluminium 6061-T6 and the tantalum-tungsten alloy in the wrought and HIP condition was carried out principally over four strain rates covering the quasi-static to high strain rate ranges (0.001 s\(^{-1}\), 0.675 s\(^{-1}\), 280 s\(^{-1}\) and 760 s\(^{-1}\)) and at temperatures of 203 K, 223 K, 294 K and 373 K.

4.2.8 Metallography/Fractography

Cross sections were cut from the Ta-2.5%W and Al 6061-T6 wrought alloy plate materials to reveal the microstructure in both cross rolled directions (X and Y). Cross sections were taken of the copper rod material parallel to the drawing direction while the Ta-2.5%W HIP alloy was cross-sectioned to reveal the microstructure along the iso-static pressing axis direction. These cross sections were then mounted and ground and polished using colloidal silica to a 25 nm grade finish, for subsequent analysis, using a JEOL-JS7000F Field Emission Gun Scanning Electron Microscope (FEGSEM) to reveal the grain structure, size and morphology. Grain orientation was obtained using Electron Backscatter Diffraction (EBSD). The grain boundaries were defined as the mismatch angle between two adjacent grains of 5 degrees. The average grain size from the EBSD mapped areas were determined by two methods, namely, the ASTM [56] value, and the intercept method. The fracture surfaces of the tensile samples were also imaged using the FEGSEM and where appropriate an EDX map of the fracture surface made.
4.3 Results

4.3.1 Aluminium 6061-T6 Alloy

Figure 4-1 shows the true stress-true plastic strain curves for the Al 6061-T6 alloy at a strain rate of 0.001/s and at the test temperatures of 294 K and 203 K for tensile samples cut parallel to the two cross rolled directions X and Y. Instantly noticeable from these figures is that, despite the material being specified as cross-rolled, there is some anisotropy in the plate material. Although the difference is relatively slight, tensile samples cut parallel to the X-direction produce a curve at a higher flow stress value than those cut parallel to the Y-direction. On average the tests at 294 K give a yield stress that is ~0.5% higher in the X-direction than the Y-direction, the curve from samples cut in X-direction rise quickly and then settles to a value of ~4% higher for the rest of the trace (i.e. after the initial increase after yielding, the slope of the curve is the same for both cross rolled directions). The tests at 203 K give a yield stress value ~5% higher in the X-direction than the Y-direction, the slope of the curves thereafter again being the same in both directions, ~7% higher in the X-direction. It appears from these curves that the material is in a slightly more worked condition in the X-direction than the Y-direction. Samples were cross-sectioned parallel to both cross-rolled directions in the un-strain condition for metallographic examination, however, etching of this alloy to reveal the grain structure proved difficult and no images of the grain morphology could be obtained.

Figures 4-2 shows the true stress-true plastic strain curves produced for the Al 6061-T6 alloy at a strain rate of 0.675 s\(^{-1}\), again at the test temperatures of 294 K and 203 K. As with the lower strain rate, the test samples cut parallel to the X-direction give a higher flow stress reading than those cut parallel to the Y-direction. However, in both cases determining the 0.2% off-set stress by conventional means still leaves a portion of the elastic region and the slope of the curve in the plastic region also seems to be flatter than would be expected. The Instron 5584 test machine was working close to its maximum specified velocity and either there was some compliance issue with the physical apparatus or there was a problem with the data recording capabilities of the equipment.
Figure 4-1. True stress-true plastic strain curves for Al 6061-T651 alloy at a strain rate of 0.001 s$^{-1}$ at 203 and 294K.

Figure 4-2. True stress-true plastic strain curves for Al 6061-T651 alloy at a strain rate of 0.675 s$^{-1}$ at 203 and 294K.

Figure 4-3 shows the raw true stress-true plastic strain curves for Al 6061-T6 alloy subjected to a strain rate of 760 s$^{-1}$ at a test temperature of 294 K as recorded by the adjusted strain gauge readings (described in Section 3.3.2). While the curves show considerable oscillation, it is apparent that the tensile samples cut parallel to the X-
direction give a higher flow stress reading than those cut parallel to the Y-direction, the stress at 0.2% off-set strain being ~5 MPa higher. The curves for samples cut in the X-direction, initially rise more rapidly than those cut parallel to the Y-direction before finding a similar gradient but at a true flow stress value of ~15 MPa higher.

![True stress-true plastic strain curves for Al 6061-T6 alloy at a strain rate of 760 s⁻¹ at 294 K.](image)

Figures 4-4 and 4-5 show the average true stress-true plastic strain curves for the aluminium alloy at the three different strain rates for the test temperatures of 294 K and 203 K respectively. These graphs clearly show that after a small difference 0.2% yield stress between the samples cut from the plate at perpendicular orientations. The slopes of the curves are approximately the same after the initial 0.01 true plastic strain for tests carried out at strain rates of 0.001/s at both temperatures (294 K and 203 K), and 760/s at 294 K.
The traces at 0.675 s$^{-1}$ for the tests carried out at both 294 K and 203 K seem to show that the yield point may not have been reached at 0.02% strain, and the plastic part of the curve thereafter is considerably flatter than those carried out at other strain rates.
Examination of the few flow curves at this strain rate produced by measuring the strain directly from images of the specimen by DIC also produced flatter than expected curves. It was concluded therefore that the load cell was slow in recording at this strain rate and a delay in the load reading with respect to strain was present. The stress-strain curves obtained at a strain rate of 0.675 s⁻¹ were therefore considered to be compromised and that the stress-strain curves were not representative of the material behaviour at these strain rates. In view of this fact the true stress-true plastic strain curves are excluded from further evaluation of the materials behaviour at this strain rate. However regarding the true UTS readings, although slightly compromised by the strain readings, the results seem to be reasonably representative of those expected at this strain rate and used for further analyse together with the strain at fracture, measured directly from the specimens.

Table 4-3 shows the mechanical data obtained from these tests (4 to 6 tests at each test condition), giving the true UTS, true plastic strain at UTS, true stress at 0.2% off-set strain, and strain at fracture (measure manually from fracture sample).

**Table 4-3 Mechanical Properties of Aluminium 6061-T6**

<table>
<thead>
<tr>
<th>Tensile Orientation</th>
<th>Test Temperature (K)</th>
<th>Strain Rate (s⁻¹)</th>
<th>True UTS (MPa)</th>
<th>True Strain at UTS (%)</th>
<th>True Stress at 0.2% Off-Set Strain (MPa)</th>
<th>Strain at Fracture (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>X</td>
<td>373</td>
<td>280</td>
<td>327.8 ± n/a</td>
<td>9.5 ± n/a</td>
<td>269.6 ± n/a</td>
<td>22.5 ± n/a</td>
</tr>
<tr>
<td></td>
<td>294</td>
<td>0.001</td>
<td>335.9 ± 3.7</td>
<td>7.4 ± 0.4</td>
<td>284.9 ± 1.0</td>
<td>14.5 ± 0.7</td>
</tr>
<tr>
<td></td>
<td></td>
<td>0.675</td>
<td>346.9 ± 1.6</td>
<td>n/a</td>
<td>n/a</td>
<td>17.2 ± 0.6</td>
</tr>
<tr>
<td></td>
<td></td>
<td>280</td>
<td>364.2 ± 8.8</td>
<td>8.8 ± 0.2</td>
<td>290.1 ± 5.8</td>
<td>20.4 ± 3.0</td>
</tr>
<tr>
<td></td>
<td></td>
<td>760</td>
<td>358.5 ± 12.5</td>
<td>8.3 ± 1.5</td>
<td>306.6 ± 11.8</td>
<td>22.2 ± 3.1</td>
</tr>
<tr>
<td></td>
<td>223</td>
<td>280</td>
<td>393.0 ± 4.7</td>
<td>10.4 ± 0.3</td>
<td>327.9 ± 5.7</td>
<td>20.2 ± 0.8</td>
</tr>
<tr>
<td></td>
<td></td>
<td>0.001</td>
<td>376.3 ± 1.8</td>
<td>8.3 ± 0.3</td>
<td>314.8 ± 2.5</td>
<td>15.9 ± 1.6</td>
</tr>
<tr>
<td></td>
<td></td>
<td>0.675</td>
<td>380.6 ± 2.8</td>
<td>n/a</td>
<td>n/a</td>
<td>19.2 ± 0.8</td>
</tr>
<tr>
<td>Y</td>
<td>373</td>
<td>280</td>
<td>328.0 ± 1.6</td>
<td>8.9 ± 0.3</td>
<td>279.1 ± 5.4</td>
<td>22.5 ± n/a</td>
</tr>
<tr>
<td></td>
<td>294</td>
<td>0.001</td>
<td>325.7 ± 4.2</td>
<td>7.9 ± 0.7</td>
<td>285.5 ± 2.8</td>
<td>17.9 ± 1.5</td>
</tr>
<tr>
<td></td>
<td></td>
<td>0.056</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>16.9 ± n/a</td>
</tr>
<tr>
<td></td>
<td></td>
<td>0.280</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>16.9 ± 2.0</td>
</tr>
<tr>
<td></td>
<td></td>
<td>0.562</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>18.3 ± n/a</td>
</tr>
<tr>
<td></td>
<td></td>
<td>0.675</td>
<td>331.7 ± 2.6</td>
<td>n/a</td>
<td>n/a</td>
<td>17.5 ± 2.0</td>
</tr>
<tr>
<td></td>
<td></td>
<td>280</td>
<td>346.9 ± 8.7</td>
<td>8.9 ± 1.0</td>
<td>303.2 ± 11.2</td>
<td>22.1 ± 2.3</td>
</tr>
<tr>
<td></td>
<td>203</td>
<td>0.001</td>
<td>354.5 ± 4.3</td>
<td>8.7 ± 1.1</td>
<td>305.0 ± 4.8</td>
<td>20.0 ± 0.7</td>
</tr>
<tr>
<td></td>
<td></td>
<td>0.675</td>
<td>364.7 ± 2.4</td>
<td>n/a</td>
<td>n/a</td>
<td>22.8 ± 1.5</td>
</tr>
</tbody>
</table>
The values in Table 4-3 have been plotted in Figures 4-6, to 4-9. Figure 4-6 shows the change in UTS and 0.2% off-set yield stress with respect to temperature. The true UTS value decreases by 30 MPa to 40 MPa between test temperatures of 203 to 294 K at a strain rate of 0.001 s\(^{-1}\). At a strain rate of 280 s\(^{-1}\) it decreased by about the same amount ~30 MPa between 223 K to 294 K, and by ~65 MPa between 223 K and 373 K.

Figure 4-6. True stress as a function of temperature for Al 6061-T6 at strain rates of 0.001/s and 280/s.

Figure 4-7. True strain as function of temperature for Al 6061-T6 at strain rates of 0.001/s and 280/s.
Figure 4.7 shows the true strain at UTS and at failure as a function of test temperature for the Al 6061-T6 alloy at strain rates of 0.001 s\(^{-1}\) and 280 s\(^{-1}\). The graph strongly suggests that there is little change in the true plastic strain in this alloy over the temperature range tested.

Figure 4-8 shows the change in true stress with strain rate. The true UTS value increases from 330 MPa to 352 MPa from a strain rate of 0.001 s\(^{-1}\) to a strain rate of 760/s, approximately a 10% increase. The stress at 0.2% off-stress increases from approximately 285 MPa to 305 MPa an increase of approximately 7%.

Figure 4-8. True stress as a function of strain rate for Al 6061-T6 at a temperature of 294 K.

Table 4-4 tabulates the strain rate sensitivity of this alloy over this strain rate range using equation (2-19). The strain rate sensitivity of this alloy was found to be 0.004 to 0.005 for the 0.2% off-set stress and 0.010 to 0.012 for UTS. This clearly demonstrates the low strain rate sensitivity of the alloy over this test range.

Figure 4-9 shows the true plastic strain as a function of strain rate for the Al 6061-T6 alloy. The trend line of strain at UTS is relatively flat showing only a small increase with strain over this strain rate range (~0.9% increase in elongation). The strain at
failure trend line is considerably steeper showing an increase in elongation over this
strain rate range of ~7.7%.

Table 4.4. Strain rate sensitivity of true yield stress and true UTS
for Al 6061-T6 at 294 K

<table>
<thead>
<tr>
<th></th>
<th>$\sigma_1$</th>
<th>$\sigma_2$</th>
<th>$\dot{\varepsilon}_1$</th>
<th>$\dot{\varepsilon}_2$</th>
<th>$m$</th>
</tr>
</thead>
<tbody>
<tr>
<td>True Yield Stress (X)</td>
<td>284.9</td>
<td>306.6</td>
<td>0.001</td>
<td>760</td>
<td>0.005</td>
</tr>
<tr>
<td>True Yield Stress (Y)</td>
<td>285.5</td>
<td>303.2</td>
<td>0.001</td>
<td>760</td>
<td>0.004</td>
</tr>
<tr>
<td>True Plastic UTS (X)</td>
<td>306.6</td>
<td>358.5</td>
<td>0.001</td>
<td>760</td>
<td>0.012</td>
</tr>
<tr>
<td>True Plastic UTS (Y)</td>
<td>303.2</td>
<td>346.9</td>
<td>0.001</td>
<td>760</td>
<td>0.010</td>
</tr>
</tbody>
</table>

Figure 4-9. True plastic strain as a function of strain rate for Al 6061-T651.

Figure 4-10 shows typical images of the fracture surface of the tensile samples in this
alloy. The circular nature of the fracture area shows the isotropic nature of the local
deformation (i.e. necking) in the sample that occurred prior to fracture. Figure 4-11
shows a more detailed image of the fracture surface, displaying the cup and cone
arrangement typical of a ductile fracture. The EDX images also show second phase
particles products of the manufacturing process located in the centre of the recesses in
the fracture surface. There is a clear correlation between the magnesium and silicon,
suggesting atomic bonding between the two elements, while a similar correlation is
suggested between the iron and chromium.
Figure 4-10. Fractography of a typical Al 6061-T6 tensile specimen showing the deformation that takes place in the neck and fracture area (Strain rate of 760 s$^{-1}$ at a temperature of 294 K).
4.3.2 OFE Copper

Figures 4-12 and 4-13 show the true stress-true plastic strain curves produced at 0.001 s\(^{-1}\), 0.675 s\(^{-1}\) and 760 s\(^{-1}\) strain rates for OFE copper at test temperatures of 294 K and at strain rates of 0.001 s\(^{-1}\) and 0.675 s\(^{-1}\) at 203 K respectively.
Figure 4-12. True stress-true plastic strain curves for OFE copper at various test strain rates at a test temperature of 294 K.

Figure 4-13. True stress-true plastic strain curves for OFE copper at various test strain rates at a test temperature of 203K.

Figure 4-14 shows the averaged true stress-true plastic strain curve for the copper at the different rates and temperatures. As was the case for the aluminium alloy the flow stress curves for the intermediate strain rate (0.675 s$^{-1}$) are compromised for exactly
the same reason, namely a slow load cell response. Table 4-5 shows the mechanical data obtained from these tests, giving the true UTS, true plastic strain at UTS, true stress at 0.2% off-set strain, and the strain at fracture.

![Figure 4-14. Average true stress-true plastic strain curves for OFE copper at different test strain rates and temperatures.](image)

<table>
<thead>
<tr>
<th>Test Temperature (K)</th>
<th>Strain Rate (s⁻¹)</th>
<th>True UTS (MPa)</th>
<th>True Plastic Strain at UTS (%)</th>
<th>True 0.2% Off-Set Stress (MPa)</th>
<th>Strain at Failure (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>373</td>
<td>280</td>
<td>389.0 ± 3.1</td>
<td>8.0 ± 1.6</td>
<td>352.9 ± 5.7</td>
<td>28.1 ± 1.4</td>
</tr>
<tr>
<td>294</td>
<td>0.001</td>
<td>335.4 ± 1.0</td>
<td>1.1 ± 1.4</td>
<td>329.9 ± 3.3</td>
<td>16.7 ± 4.3</td>
</tr>
<tr>
<td></td>
<td>0.675</td>
<td>385.1 ± 1.9</td>
<td>n/a</td>
<td>n/a</td>
<td>25.0 ± 1.4</td>
</tr>
<tr>
<td></td>
<td>280</td>
<td>432.3 ± 0.9</td>
<td>12.6 ± 2.1</td>
<td>368.1 ± 5.6</td>
<td>29.1 ± 0.8</td>
</tr>
<tr>
<td></td>
<td>760</td>
<td>453.5 ± 8.4</td>
<td>15.6 ± 1.0</td>
<td>371.6 ± 8.3</td>
<td>33.7 ± 4.9</td>
</tr>
<tr>
<td>223</td>
<td>280</td>
<td>515.0 ± 4.9</td>
<td>20.0 ± 0.9</td>
<td>432.3 ± 24.3</td>
<td>31.9 ± 1.6</td>
</tr>
<tr>
<td>203</td>
<td>0.001</td>
<td>407.6 ± 3.7</td>
<td>10.9 ± 1.3</td>
<td>358.4 ± 2.5</td>
<td>24.9 ± 1.3</td>
</tr>
<tr>
<td></td>
<td>0.675</td>
<td>445.0 ± 3.9</td>
<td>n/a</td>
<td>n/a</td>
<td>29.2 ± 0.8</td>
</tr>
</tbody>
</table>

The values in Table 4-5 are plotted in Figures 4-15 to 4-18. Figure 4-15 and 4-16 show the relationship between true stress (UTS and 0.2% off-set stress) and true plastic strain with temperature respectively. It may be noted that the yield stress for the copper under quasi-static conditions as determined in this work differs considerably from that given in the material data sheet (Table 4.1). Figure 4-15
shows that the yield stress is considerably less sensitive than the UTS to an increase in temperature. The yield stress declines by ~30 MPa over the 294 K to 203 K temperature range, while the UTS declines by 60 MPa to 70 MPa over the same variation in temperature. The true plastic strain varies at UTS is shown in Figure 4-16 to be highly sensitive to variation in temperature over the test temperature range, while the true strain at failure is shown to be less sensitive.

![Figure 4-15. True stress as a function of temperature for OFE copper.](image1)

![Figure 4-16. True plastic strain as a function of temperature for OFE copper.](image2)
The strain rate sensitivity of the true stress shown in Figure 4-17 shows the greater sensitivity of true UTS (increasing by ~118 MPa) as compared to the yield stress (increasing by ~38 MPa) over the quasi-static to dynamic range tested.

Figure 4-17. True stress as a function of strain rate for OFE copper.

Table 4-6 tabulates the calculated strain rate sensitivity from equation (2.20) for this metal over this strain rate range (0.001 s\(^{-1}\) to 760 s\(^{-1}\)). The strain rate sensitivity, \(m\), is shown to be 0.009 for the true yield stress (at 294 K) and 0.022 for the UTS at a test temperature of 294 K.

Table 4-6. Strain rate sensitivity of true yield stress and true plastic UTS for OFE copper at 294 K

<table>
<thead>
<tr>
<th></th>
<th>(\sigma_1)</th>
<th>(\sigma_2)</th>
<th>(\dot{\varepsilon}_1)</th>
<th>(\dot{\varepsilon}_2)</th>
<th>(m)</th>
</tr>
</thead>
<tbody>
<tr>
<td>True Yield Stress</td>
<td>329.9</td>
<td>371.6</td>
<td>0.001</td>
<td>760</td>
<td>0.009</td>
</tr>
<tr>
<td>True UTS</td>
<td>335.4</td>
<td>453.5</td>
<td>0.001</td>
<td>760</td>
<td>0.022</td>
</tr>
</tbody>
</table>

Figure 4-18 shows how the true plastic strain varies with strain rate. Figure 4-18 strongly suggests that the true plastic strain at UTS and at fracture increases by a greater amount at the higher temperature (294 K) than at the lower temperature (203 K) over the strain rate ranged tested.
Figure 4.18. *True plastic strain as a function of strain rate for OFE copper.*

Figure 4.19 shows images of the typical fracture surface of the OFE copper samples over the test regime. The images show the sample failed in a ductile manner and that plastic deformation prior to failure tended to be isotropic in nature.
Figure 4-19. Fractography of a typical OFE copper tensile specimen showing the deformation that takes place in the neck and fracture area (Strain rate of 760 s\(^{-1}\) at 294 K).

4.3.3 Ta-2.5 wt % W Wrought Alloy

Figures 4-20 to 4-22 show the flow curves for the Ta-2.5%W wrought alloy at strain rate of 0.001 s\(^{-1}\), 0.675 s\(^{-1}\) and 760 s\(^{-1}\) respectively. The figures display flow curves for tensile samples cut in the plan of the plate parallel to the two cross rolled directions X and Y. The flow curves for the 0.675 s\(^{-1}\) strain rate were as previously mentioned compromised due to a slow load cell response. The flow curves at a strain rate of 760 s\(^{-1}\) showed varying degrees of oscillation. Average flow curves were extracted from these curves and are shown in Figure 4-23.
Figure 4-20. True stress-true plastic strain curves for Ta-2.5% W wrought alloy at a strain rate of 0.001 s$^{-1}$ at 203 K and 294 K.

Figure 4-21. True stress-true plastic strain curves for Ta-2.5% W wrought alloy at a strain rate of 0.675 s$^{-1}$ at 203 K and 294 K.
As can be seen from Figures 4.20 to 4.22 the cross rolled plate in this material is much more isotropic than that shown for the Al 6061-T6 plate material. Test samples cut parallel to both cross rolling directions gave a similar mechanical response.
Metallographic examination of the grain structure (Figure 4-24), however did show a degree of anisotropy.

![Figure 4-24](image)

Figure 4-24. EBSD images of the Ta-2.5%W wrought alloy plate material showing variation in the grain morphology with respect to spatial direction. Cross sections in the (a) X-direction (rolling direction from top to bottom of photograph), and (b) Y-direction, cross rolling directions (rolling direction from top to bottom of photograph).

Figure 4-24 reveals that the grain structure parallel to the X-direction is slightly more elongated than in the Y-direction, and therefore showing signs of greater mechanical working in that direction. Table 4-6 gives a summary of the grain structure for the alloy and shows the degree of elongation of the grains parallel to the X-cross rolled direction (45.5 μm by 22.5 μm), and the equi-axed nature of the grains in the Y-cross rolled direction (32.9 μm by 33.0 μm).
Table 4-7. Grain structure of the Ta-2.5% wrought alloy by ASTM and intercept method (H* horizontal intercept, V* vertical intercept on image).

<table>
<thead>
<tr>
<th>Section</th>
<th>ASTM</th>
<th>Area μm²</th>
<th>Threshold Angle deg (boundary definition)</th>
<th>Number of grains</th>
<th>Average line intercept H* μm</th>
<th>Average line intercept V* μm</th>
<th>Average line intercept μm</th>
</tr>
</thead>
<tbody>
<tr>
<td>X</td>
<td>7.1</td>
<td>960</td>
<td>5.0</td>
<td>372</td>
<td>32.9</td>
<td>33.0</td>
<td>33.0</td>
</tr>
<tr>
<td>Y</td>
<td>7.1</td>
<td>946</td>
<td>5.0</td>
<td>436</td>
<td>22.5</td>
<td>45.4</td>
<td>33.9</td>
</tr>
</tbody>
</table>

The mechanical properties obtained from the flow curves are shown in Table 4-8. The table shows the true UTS, true strain at UTS, true stress at 0.2% off-set strain, and strain at failure.

Table 4-8. Mechanical Properties of Ta-2.5% W wrought alloy.

<table>
<thead>
<tr>
<th>Tensile Orientation</th>
<th>Test Temperature (K)</th>
<th>Strain Rate (s⁻¹)</th>
<th>True UTS (MPa)</th>
<th>True Strain at UTS (%)</th>
<th>True Stress at 0.2% Off-Set Strain (MPa)</th>
<th>Strain at Fracture (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>X</td>
<td>373</td>
<td>280</td>
<td>469.8 ± 26.1</td>
<td>22.8 ± 4.8</td>
<td>353.5 ± 20.3</td>
<td>40.8 ± 9.9</td>
</tr>
<tr>
<td></td>
<td>294</td>
<td>0.001</td>
<td>430.6 ± 11.1</td>
<td>31.7 ± 1.3</td>
<td>210.2 ± 13.5</td>
<td>51.3 ± 3.0</td>
</tr>
<tr>
<td></td>
<td></td>
<td>0.675</td>
<td>442.1 ± 8.9</td>
<td>n/a</td>
<td>n/a</td>
<td>43.6 ± 2.0</td>
</tr>
<tr>
<td></td>
<td></td>
<td>280</td>
<td>545.3 ± 14.6</td>
<td>18.9 ± 0.7</td>
<td>410.2 ± 17.8</td>
<td>42.5 ± 1.6</td>
</tr>
<tr>
<td></td>
<td></td>
<td>760</td>
<td>535.4 ± 22.5</td>
<td>16.7 ± 3.3</td>
<td>416.8 ± 7.9</td>
<td>38.7 ± 2.7</td>
</tr>
<tr>
<td></td>
<td>223</td>
<td>280</td>
<td>663.2 ± n/a</td>
<td>16.9 ± n/a</td>
<td>386.1 ± n/a</td>
<td>31.0 ± n/a</td>
</tr>
<tr>
<td></td>
<td>760</td>
<td>535.4 ± 22.5</td>
<td>16.7 ± 3.3</td>
<td>416.8 ± 7.9</td>
<td>38.7 ± 2.7</td>
<td></td>
</tr>
<tr>
<td></td>
<td>203</td>
<td>0.001</td>
<td>465.7 ± 14.4</td>
<td>29.2 ± 4.3</td>
<td>304.9 ± 24.7</td>
<td>49.9 ± 1.0</td>
</tr>
<tr>
<td></td>
<td></td>
<td>0.675</td>
<td>492.8 ± 24.1</td>
<td>n/a</td>
<td>n/a</td>
<td>39.4 ± 4.0</td>
</tr>
<tr>
<td>Y</td>
<td>373</td>
<td>280</td>
<td>470.0 ± n/a</td>
<td>24.0 ± n/a</td>
<td>345.0 ± n/a</td>
<td>45.0 ± n/a</td>
</tr>
<tr>
<td></td>
<td>294</td>
<td>0.001</td>
<td>400.4 ± 11.8</td>
<td>26.0 ± 3.8</td>
<td>195.6 ± 14.4</td>
<td>50.6 ± 4.0</td>
</tr>
<tr>
<td></td>
<td></td>
<td>0.675</td>
<td>413.5 ± 17.9</td>
<td>n/a</td>
<td>n/a</td>
<td>40.3 ± 7.7</td>
</tr>
<tr>
<td></td>
<td></td>
<td>760</td>
<td>530.8 ± 8.9</td>
<td>14.5 ± 1.5</td>
<td>428.3 ± 21.4</td>
<td>38.0 ± 5.0</td>
</tr>
<tr>
<td></td>
<td>203</td>
<td>0.001</td>
<td>462.5 ± 28.7</td>
<td>21.5 ± 7.5</td>
<td>302.8 ± 7.6</td>
<td>46.8 ± 7.7</td>
</tr>
<tr>
<td></td>
<td></td>
<td>0.675</td>
<td>484.6 ± 14.2</td>
<td>n/a</td>
<td>n/a</td>
<td>40.8 ± 2.8</td>
</tr>
</tbody>
</table>

Figure 4-25 shows the true stress as a function of temperature at a strain rate of 0.001/s, while Figure 4-26 shows the true plastic strain as a function of temperature. The true stress at UTS shown in Figure 4-25 reveals a greater temperature sensitivity at the high strain rate, but there is little change with temperature in the yield stress value at this rate. At the lower rate, the yield stress shows a large sensitivity to the change in temperature. The true plastic strain in Figure 4-26 generally shows only a small increase at UTS and at failure with temperature, although the true strain at UTS in the higher strain rate (280 s⁻¹) does suggest a substantial increase with increase in temperature of nearly 6 % over the 223 K to 373 K temperature range.
Figure 4-27 shows how the true stress increases with increase in strain rate at a test temperature of 294 K. What is noticeable in this figure is that the true yields stress is much more strain rate sensitive than the true UTS.

Figure 4-25. True stress as a function of temperature for Ta-2.5% W wrought alloy.

Table 4-9 gives a quantitative value of the strain rate sensitivity, \( m \), calculated from equation (2-20). The strain rate sensitivity for the yield stress is between 0.051 and 0.058, while that for the UTS is calculated as 0.016 to 0.021.

Table 4-9. Strain rate sensitivity of true yield stress and true UTS for Ta-2.5%W wrought alloy at 294 K

<table>
<thead>
<tr>
<th></th>
<th>( \sigma_1 )</th>
<th>( \sigma_2 )</th>
<th>( \dot{\epsilon}_1 )</th>
<th>( \dot{\epsilon}_2 )</th>
<th>( m )</th>
</tr>
</thead>
<tbody>
<tr>
<td>True Yield Stress (X)</td>
<td>210.2</td>
<td>416.8</td>
<td>0.001</td>
<td>760</td>
<td>0.051</td>
</tr>
<tr>
<td>True Yield Stress (Y)</td>
<td>195.6</td>
<td>428.3</td>
<td>0.001</td>
<td>760</td>
<td>0.058</td>
</tr>
<tr>
<td>True Plastic UTS (X)</td>
<td>430.6</td>
<td>535.4</td>
<td>0.001</td>
<td>760</td>
<td>0.016</td>
</tr>
<tr>
<td>True Plastic UTS (Y)</td>
<td>400.4</td>
<td>530.8</td>
<td>0.001</td>
<td>760</td>
<td>0.021</td>
</tr>
</tbody>
</table>
Figure 4-26. True strain as a function of temperature for Ta-2.5% W wrought alloy.

Figure 4-27. True stress as a function of strain rate for Ta-2.5% W wrought alloy, tested at 294 K.
Figure 4-28 shows the true plastic strain at UTS and at failure as a function of strain rate for the Ta-2.5%W wrought alloy. The reduction in the true strain value recorded at UTS and at failure with increase in the strain rate are clearly apparent from the Figure 4-28.

Figure 4-29 shows a typical profile of the Ta-2.5%W wrought fracture surface. The image shows how the ductility of the alloy, within the neck region deformation down to a small area before failure of the specimen occurs. This image of the localised deformation also reveals that the sample deformed in an anisotropic manner, before failure, as shown by the elongated nature of the fracture surface.

Figure 4-28. True plastic strain as a function of strain rate for Ta-2.5% W wrought alloy at 203 K and 294 K.
Metallography of the fractured tensile specimens reveals the deformation of the grain structure as displayed in Figures 4-30 and 4-31. Figure 4-30 shows the elongated grains in the majority of the neck region, while Figure 4-31 show the grain structure closer to the fractured surface which has undergone the greatest degree deformation. Figure 4-31 clearly shows (from the large reduction in the grain size) that there has been recrystallisation of the grain structure in this region, strongly suggesting adiabatic heating (section 2.1.3) has occurred during deformation at this strain rate (760 s\(^{-1}\)).
Figure 4-30. EBSD image of the Ta-2.5 wt % W wrought alloy tensile specimen neck region showing deformation elongation of the grains (strain rate of 760 s⁻¹ at a temperature of 294 K).

Figure 4-31. EBSD images at different magnification of the Ta-2.5 wt % W wrought alloy tensile specimen neck region near to the fracture surface showing extreme grain deformation and recrystallisation (strain rate 760 s⁻¹ at a temperature of 294 K).
4.3.4 Ta-2.5 wt % W HIP Alloy

Figure 4-32 shows the true stress-true plastic strain flow curves obtained for the Ta-2.5%W HIP alloy, while Figure 4-33 shows the average flow curve for each test condition. Again it is noticeable that the 0.675 s\(^{-1}\) strain rate is compromised by the slow response of the load cell. Table 4-10 shows the mechanical data obtained from these flow curves, giving the true UTS, true stress at 0.2% off-set true strain, true plastic strain at UTS and the true plastic strain at failure.

![Graph showing true stress-true strain curves for Ta-2.5%W HIP alloy at different strain rates and temperatures.](image)

Figure 4-32. True stress-true strain curves for Ta-2.5%W HIP alloy at different strain rates and temperatures.

The data in Table 4-10 is plotted in Figures 4-34 to 4-37. Figures 4-34 and 4-35 shows the relationship between true stress and true plastic strain as a function of temperature respectively. Figure 4-34 indicates that the yield stress is much more sensitive than the UTS to changes in temperature.

Table 4-10. Mechanical Properties of Ta-2.5%W HIP alloy.

<table>
<thead>
<tr>
<th>Test Temperature (°C)</th>
<th>Strain Rate (s(^{-1}))</th>
<th>True UTS (MPa)</th>
<th>True 0.2% Off-Set Stress (MPa)</th>
<th>True Plastic Strain at UTS (%)</th>
<th>True Plastic Strain at Failure (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>294</td>
<td>0.001</td>
<td>473.9 ± 11.2</td>
<td>294.9 ± 10.8</td>
<td>22.1 ± 3.4</td>
<td>35.6 ± 4.3</td>
</tr>
<tr>
<td></td>
<td>0.675</td>
<td>473.5 ± 10.3</td>
<td>n/a</td>
<td>n/a</td>
<td>25.3 ± 3.2</td>
</tr>
<tr>
<td></td>
<td>280</td>
<td>616.9 ± 4.0</td>
<td>516.0 ± 9.9</td>
<td>14.4 ± 0.1</td>
<td>30.2 ± 5.0</td>
</tr>
<tr>
<td>203</td>
<td>0.001</td>
<td>492.6 ± 7.0</td>
<td>398.2 ± 7.5</td>
<td>6.9 ± 1.6</td>
<td>15.8 ± 2.1</td>
</tr>
</tbody>
</table>
Figures 4-36 and 4-37 show the changes in true stress and true plastic strain as a function of strain rate respectively. Figure 4-36 shows the yields stress is much more strain rate sensitive than the UTS. Table 4-11 attempts to put a value to this strain rate sensitivity using equation (2-20).

Table 4-11. Strain rate sensitivity of true yield stress and true UTS for Ta-2.5%W HIP alloy

<table>
<thead>
<tr>
<th></th>
<th>$\sigma_1$</th>
<th>$\sigma_2$</th>
<th>$\dot{\epsilon}_1$</th>
<th>$\dot{\epsilon}_2$</th>
<th>$m$</th>
</tr>
</thead>
<tbody>
<tr>
<td>True Yield Stress</td>
<td>294.9</td>
<td>516.0</td>
<td>0.001</td>
<td>280</td>
<td>0.045</td>
</tr>
<tr>
<td>True Plastic UTS</td>
<td>473.9</td>
<td>616.9</td>
<td>0.001</td>
<td>280</td>
<td>0.021</td>
</tr>
</tbody>
</table>
Figure 4-34. True Stress as a function of temperature for a strain rate of 0.001 s$^{-1}$.

Figure 4-35. True plastic strain as a function of temperature for a strain rate of 0.001 s$^{-1}$. 
Figure 4-36. True stress as a function of strain rate for Ta-2.5% W HIP alloy at a test temperature of 294 K.

Figure 4-37. True plastic strain as a function of strain rate for Ta-2.5% W HIP alloy at a test temperature of 294 K.

Cross-sections of the Ta-2.5%W HIP material are shown in Figure 4-38. The relative grain orientations are colour coded and indicate that the each grain is roughly 110 μm in size which roughly corresponds to the original powder size and are largely single crystal in nature. Figure 4-38 clearly shows a certain level of porosity between the
grains where the powder particles have not bonded. This amount of porosity determined on the plane section was found to be about 0.8%. Table 4-12 summaries the grain size measurements of this material.

![Figure 4](image1.png)

**Figure 4-38.** EBSD images of the Ta-2.5% W HIP alloy material showing the equiaxed grain structure, (a) ×40 magnification, (b) ×140 magnification.

**Table 4-12. Grain structure of Ta-2.5%W HIP alloy**

<table>
<thead>
<tr>
<th>Section &amp; Mag.</th>
<th>ASTM</th>
<th>Area μm²</th>
<th>Threshold Angle deg (boundary definition)</th>
<th>Number of grains</th>
<th>Average line intercept H* μm</th>
<th>Average line intercept V* μm</th>
<th>Average line intercept μm</th>
</tr>
</thead>
<tbody>
<tr>
<td>Axial</td>
<td>3.6</td>
<td>10981</td>
<td>6.8</td>
<td>136</td>
<td>104.3</td>
<td>96.7</td>
<td>100.5</td>
</tr>
<tr>
<td>Cross</td>
<td>3.4</td>
<td>12018</td>
<td>6.8</td>
<td>104</td>
<td>143.2</td>
<td>103.3</td>
<td>123.3</td>
</tr>
</tbody>
</table>

Figure 4-39 shows a typical image of the fracture surface of the tensile specimen. The sample has clearly fractured inter-granularly in a brittle manner.
Figure 4-39. Fractography of a Ta-2.5 wt % W HIP alloy tensile specimen showing the isotropic deformation of the neck and fracture area tested at strain rate of 280 s\(^{-1}\) at a temperature of 294 K.

4.4 Discussion

4.4.1 Effect of Crystal Structure

Figures 4-40 and 4-41 shows the difference in flow curves for the three wrought metals/alloy investigated here, over the quasi-static (0.001 s\(^{-1}\)) to dynamic (760 s\(^{-1}\)) strain rate range, and the temperature range respectively. The traces show typical characteristics of flow curves obtained from BCC and FCC metals/alloys.
Figure 4-40. True stress-true plastic strain curves for copper, aluminium-6061, and tantalum-2.5% tungsten wrought alloys with respect to strain rate.

Figure 4-41. True stress-true plastic strain curves for copper, aluminium-6061, and tantalum-2.5% tungsten wrought alloys with respect to test temperature.

The yield stress of the Ta-W alloy is shown to be highly sensitive to changes in strain rate and temperature. The large Peierls stress (Section 2.2.1) in BCC metallic
materials is the rate controlling mechanism and results in a large increase in the yield stress with increase in strain rate and decrease in temperature. The yield stress at a test temperature of 294 K rises from 210 MPa to 416.8 MPa, an increase of ~215 MPa over the quasi-static to dynamic strain rate (760 s\(^{-1}\)) range investigated. This is reflected in the strain rate sensitivity value, \(m\), shown in Table 4-13 which was determined to be approximately 0.055. This class of material also tends to have a strain-hardening rate which is insensitive to changes in temperature and strain rate (see Section 2.2.1), as reflected in the lower strain rate sensitivity of the UTS value of approximately 0.019 (Table 4-13).

Table 4-13. Comparison of strain rate sensitivity values for the wrought metals tested over the 0.001 s\(^{-1}\) to 760 s\(^{-1}\) strain rate range.

<table>
<thead>
<tr>
<th>Metal/Alloy</th>
<th>Strain Rate Sensitivity Value, (m)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Yield Stress</td>
</tr>
<tr>
<td>OFE Cu (FCC)</td>
<td>0.009</td>
</tr>
<tr>
<td>Al 6061-T6 (FCC)</td>
<td>0.004 – 0.005</td>
</tr>
<tr>
<td>Ta-2.5%W (BCC)</td>
<td>0.051 – 0.058</td>
</tr>
</tbody>
</table>

In contrast the flow curves of typical FCC metal/alloys such as the copper and aluminium investigated in this study (Figures 4-40 and 4-41), show the yield stress at 294 K to increase by only about 20 MPa for the aluminium alloy and by 40 MPa for the copper over the quasi-static to dynamic range tested. The yield stress in FCC materials is only weakly dependent on the temperature and strain rate (Section 2.2.1). In addition, the stress-strain curves at different temperatures and strain rates diverge upon further deformation indicating their strain hardening behaviour is rate dependent. The flow curves for copper show this quite clearly in Figure 4-40, rising rapidly after yielding at a strain rate 760 s\(^{-1}\) as compared to the shallow rise at a strain rate of 0.001 s\(^{-1}\). This is reflected in the strain rate sensitivity factors shown in Table 4-13 which show a higher sensitivity at UTS than at the yield stress. The rapid hardening of the metal also facilitates the increases in elongation of the sample before failure.

Aluminium and many of its alloys also possess a FCC crystal structure and Figure 4-40 and Table 4-13 show as with copper the low strain rate sensitivity of the yield stress. However, unlike copper, where the flow stress rises quickly as deformation
proceeds the strain rate sensitivity in the aluminium alloy increases only slightly with increase in strain rate. Aluminium has a high stacking fault energy (SFE) of approximately 250 mJ/m² as compared with copper of approximately 90 mJ/m² [58-59] which means dislocations are not confined to specific planes, and they can cross slip to overcome obstacles, thereby allowing the dislocations that produce the deformation greater mobility. Cross slip is more difficult for materials with low stacking fault energy like copper and consequently the material will work harden more rapidly.

A difference between the FCC and BCC materials was also revealed on examination of the fractures surfaces. The crystal structure of the FCC materials showed that the localised plastic deformation prior to fracture was isotropic in nature revealing a circular or close to circular fracture area (Figures 4-10 and 4-19). The crystal structure of the BCC material produced a fracture surface very elongated in shape (Figure 4-29). All wrought materials within the test regime investigated showed that fracture was ductile in nature. The aluminium alloy showed classic failure by micro-void nucleation at second phase particles, growth of these micro-voids by slip in the crystal lattice, and coalescence of the voids by plastic strain localisation in the ligaments between the voids in form of internal necking or shear bands. The result is the cup and cone appearance revealed in Figure 4-11.

4.4.2 Effect of Tantalum-Tungsten Manufacture Route

An alternative power metallurgy processing route was adopted to compare the mechanical response of the tantalum-tungsten alloy produced by this alternative method with the material produced in the conventionally processed wrought plate condition. Figure 4-42 compares the flow curves for the HIP and wrought materials at a strain rate 0.001 s⁻¹ and at test temperatures of 203 K, and 294 K. Also included is the wrought material at a strain rate range of 760 s⁻¹ at 294 K. The general shape of the stress-strain curve observed previously for the wrought Ta-W alloy is preserved in the powder metallurgy product at the 0.001 s⁻¹ strain rate at a test temperature of 294 K as shown in Figure 4-42. The powder metallurgy product exhibits a higher flow stress than the wrought plate at this test condition, but closely follows the gradient of the flow curves produced in the wrought material. HIP products tend to contain a large amount of interstitial content especially at the grain boundaries that originates
from the oxide layer which surrounds the pre-consolidated powder particles. This oxide layer increases the flow stress but reduces the ductility. At a lower test temperature, the flow curve for the HIP material shows a different profile, with a steeper gradient and lower elongation than the wrought material, suggesting the grain boundary region may become embrittled at this temperature (203 K).

![Graph](image)

*Figure 4-42. Difference in true stress-true plastic strain curves for wrought and HIP Ta-W alloys.*

Table 4-14 shows the similarity between the strain rate sensitivity of the wrought and HIP produced material. The strain rate sensitivity of the yield stress being ~0.054 and ~0.045 for the wrought and HIP product respectively while the UTS value being ~0.021 and ~0.019 for the wrought and HIP respectively over the quasi-static to dynamic strain rate range. The values show the large strain rate sensitivity of the yield stress compared to the UTS.

*Table 4-14. Comparison of strain rate sensitivity values for the wrought and HIP Ta-W processed materials at a test temperature of 294 K*

<table>
<thead>
<tr>
<th>Metal/Alloy</th>
<th>Strain Rate Range (s⁻¹)</th>
<th>Strain Rate Sensitivity Value, m</th>
<th>Yield Stress</th>
<th>UTS</th>
</tr>
</thead>
<tbody>
<tr>
<td>Wrought Plate</td>
<td>0.001 – 760</td>
<td>0.045</td>
<td>0.021</td>
<td></td>
</tr>
<tr>
<td>HIP Bar</td>
<td>0.001 – 280</td>
<td>0.051 – 0.058</td>
<td>0.016 – 0.021</td>
<td></td>
</tr>
</tbody>
</table>
4.5 Conclusions

1. The cross rolled Al 6061-T6 alloy showed a degree of anisotropy in its mechanical response as revealed by the true stress–true plastic strain curves traces from sample cut in the plane of the plate at perpendicular directions. This anisotropy was more pronounced at the lower test temperature (203 K) than at room temperature (294 K). The response of the alloy to change in test temperature over the 203 K to 373 K range showed the true UTS and 0.2 % off-set stress decreased slightly with increase in temperature and there was likewise little change in the true plastic strain at UTS or at failure over this temperature range. In general the alloy at these test conditions responded in a typical manner to that expected of aluminium and many of its alloys. Over the strain rate range tested, there was only a small increase in the 0.2% off-set yield stress or UTS with increase in strain rate, with strain rate sensitivities of ~0.005 and ~0.012 respectively. Little change in the true plastic strain at UTS or at failure was also observed over test range. Examination of the fracture surfaces showed that all test samples failed in a ductile manner displaying classic failure by micro-void nucleation at second phase particles, growth and coalescence of these micro-voids to failure, revealing a cup and cone fracture surface.

2. The mechanical response of the OFE copper material over the strain rate and temperature test range conditions in this work respond generally as expected of this metal. The yield stress was shown to be less sensitive to an increase in test temperature than the UTS, over the 203 K to 294 K temperature range. The true plastic strain at UTS was also more sensitive than the strain at failure over this temperature range. The strain rate sensitivity of the UTS was also shown to be greater than that of the yield stress, calculated to be 0.022 and 0.009 respectively. The true plastic strain at UTS and failure with strain rate showed a greater sensitivity at the higher test temperature of 294 K as opposed to 203 K. All samples failed in a ductile manner over the test conditions performed in this study and local plastic deformation prior to failure tended to be isotropic in nature.
3. Despite the grain structure of the wrought Ta-W alloy showing the grains parallel to X-direction being more elongated in shape than those in the Y-direction, the flow curves revealed only a small amount of anisotropy between the two directions. The response of the alloy to test temperature over the 203 to 373 temperature range showed the UTS to be more sensitive than the yield stress at dynamic rates (280 s\(^{-1}\)), although the yield stress was more sensitive at the quasi-static rate (0.001 s\(^{-1}\)). The true plastic strain increases only slightly with increase in test temperature, with the strain at UTS showing the greatest increase at the quasi-static strain rate although at the higher rate (280 s\(^{-1}\)) a large increase is seen in the strain at failure. The response of the alloy to an increase in strain rate is much more marked, with the stress at 0.2% off-set strain showing a much greater sensitivity than the UTS, with the strain rate sensitivity being > 0.05 and < 0.02 respectively. A large decrease in plastic strain at UTS and at failure was also seen in the alloy over the strain rate range tested, with elongation decreasing from 39% down to 17% at UTS and 51% down to 31% at failure. All samples failed in a ductile manner, with images of the fractured specimens showing the anisotropic nature of the local plastic strain prior to fracture. Metallographic cross sections of the most extensively deformed region of the dynamic strain rate samples (760 s\(^{-1}\)) showed recrystallisation of the grains had occurred, strongly suggesting that adiabatic heating had taken place during deformation at this rate.

4. The flow curves of Ta-W HIP alloy material mirror those of the wrought material (same strain hardening slope) for tests carried out at room temperature, but with a flow stress approximately 80 MPa higher. As with the wrought material, the yield stress was shown to be more sensitive than the UTS to change in temperature and strain rate. The strain rate sensitivity of the material at room temperature was found to be 0.045 and 0.021 for the yield stress and UTS respectively, which was similar to that, obtain for the wrought material. Fracture of the material tends to be brittle and transgranular in nature. At low temperature (203 K) the flow curve suggests embrittlement had occurred, probably at the grain boundaries.
CHAPTER 5

Comparison of Tensile and Compression Tests

5.1 Introduction

This chapter details an experimental investigation into the mechanical behaviour of Al 6061-T6 and Ta-W2.5% W materials under compression over the quasi-dynamic to dynamic strain rate range. The objectives are to compare the flow curves and strain rate sensitivities of these materials in compression with those obtained in tension as detailed in Chapter 4 under a similar strain rate range. The results are compared with those findings reported in open literature.

5.2 Experimental Procedure

5.2.1 Compression Test Sample Preparation

The specimen geometry for compression testing was 5 mm diameter cylindrical bars 5 mm in length. The sample length to diameter ratio was chosen to minimise errors due to inertial effects during high rate testing [8, 60]. To produce parallel and flat bar ends, the plate of each material were first machined flat to a thickness of 5.2 mm on a milling machine. The plates were then ground on both sides using a surface grinder until a thickness of 5 mm was reached. The surface grinding operation produced plates which were parallel to within 20 μm over a 100 mm length, giving an estimated ± 1 μm deviation over a 5 mm diameter sample. A 100 grit silicon carbide grinding wheel (grade 39C from Norton abrasives) was used to achieve good surface finish on the final pass. The cylindrical bars were then cut from the plate using wire electric discharge machining (EDM). Using wire EDM ensured that the cylindrical bar axis was perpendicular to the ground faces and that there were no burrs left on the sample.

5.2.2 Quasi-Static Compression Testing Method

An Instron 5584 test machine was used to perform the low rate uniaxial compression tests. The experimental arrangement is shown in figure 3.5. The samples were placed between two small tungsten carbide platens in which were then compressed between larger steel platens. Tungsten carbide platens were used to reduce wear and friction
between the sample and platens during testing. Dry molybdenum disulphide paste was used to lubricate contact between the sample and platens. The upper platen was connected to an Instron 2525-171 load cell located on the moving crosshead, with force on the sample measured to an accuracy of ± 0.25 %.

Figure 5-1. Experimental Set-up for Quasi-Static Compression Tests [2].

A constant crosshead speed of 5 m/s was used during compression of the sample, giving an initial strain rate of $10^3$ s$^{-1}$. Sample length during the test was measured with the Instron’s position encoder. However, compliance of the test machine was significant at high loading, resulting in errors in the recorded length. Compliance of the machine and test fixture was measured and taken into account by recording the position output whilst increasing the load (up to 50 kN) with no test sample (platens in contact), the position output showing how compliant the machine and test fixture was under increasing load. The recorded extension for each test specimen was then corrected for by subtracting the extension due to compliance at the applied test force.

The test temperature was controlled in an environmental chamber with electric heating and liquid nitrogen cooling. Each sample was soaked for at least 10 minutes
before testing. Tests were conducted at three temperatures; 203 K, 294 K and 394 K. Thermocouples were used to check that the platen temperature was within ± 2 K of the desired test temperature.

5.2.3 High Rate Compression Testing Method

High rate compression test testing was performed using a Split Hopkinson Pressure Bar (SHPB). The SHPB apparatus is driven by a vacuum gun and the equipment set-up is shown in Figures 2-3 and 5-2. A projectile bar was fired at the incident bar by opening the evacuated vacuum gun to atmospheric pressure. The velocity of the projectile was controlled using sized orifice plates. In this experimental test it was assumed that one-dimensional wave theory was used to derive the stress and strain rate in the sample.

![SHPB Equipment Test Set-up.](image)

The pulse were recorded using strain gauges (Vishay CAE-06-062UW-120) on the incident and transmitted bars connected to high rate strain amplifiers (FYLDE FE-H359) and a digital storage oscilloscope sampling at 100 MHz. Recorded values were found to be within 1 % of those expected when a four point bend was performed on the bars. Two gauges attached on opposite sides of each bar ensured bending of the bars during the test would be cancelled out. Vishay M-Bond AE10 adhesive was used to bond the gauges to the bars. The glue line was kept as thin as possible to minimise any damping effects introduced by the epoxy glue. This was achieved by covering the gauge with a silicone backing pad and clamping the gauges down with a jubilee clip whilst the glue cures. This provided an even clamping pressure over the gauge and ensures a consistent glue line.
Low mass thin copper foil strand lead wires were soldered onto the strain gauges keeping the mass of solder to a minimum. The low mass wires and solder were important to avoid inertia effects pulling the wires off the bars. The copper foil lead wires were then connected to conventional lead wires which fed to the strain amplifiers.

**Temperature Control**

An electronic firing system for the SHPB test was developed to enable precise control of the test temperature with minimal heating of the bars. Test pieces were soaked for at least 10 minutes in a temperature chamber prior to testing. A test piece was transferred to the SHPB bar and aligned in the test position in a PTFE holder, which helped to insulate and reducing the rate of heat loss during this set-up period. The bars were then brought together and on contact with the sample an electronic solenoid value fired the gun. Using this arrangement, the time the test sample remained in contact with the bars before loading could be minimised and accurately controlled. The time the sample was in contact with the bars was found to be 250 ± 10 ms using high speed video.

Determination of the test piece temperature at load was established by embedding a thermocouple in a number of samples and monitoring the change in temperature from leaving the temperature, coming in contact with the bars and at loading. Figure 5.3 shows an example of the cooling curve for the Ta-2.5%W plate material.
From these temperature experiments the desired test temperature at loading could be established with confidence, and a repeatability of ± 2 °C was achieved.

**Processing Test Data**

The raw data obtained from a typical SHPB test contains a large amount of oscillation in the load/stress signal. The key stages in processing that data include removing the elastic region (0.2% off-set strain) together with any excessive load signal overshoot. The data at the end of the test, where the load signal tails off significantly is also removed. A polynomial trend line was then fitted to each test curve, from which trend line equations were established. The trend line for each test was, if necessary, projected back to the 0.2% off-set strain position, and the strain at this point was zeroed to give a plot of the true stress against true plastic strain. From these trend lines the 0.2% off-set strain stress can be determined and the true flow stress as a function of true plastic strain can be displayed. An average trend line determined from the trend lines from each test can also be superimposed on the graph.

For the quasi-static compression data the 0.2% off-set strain was identified. The compression and tensile test are not a good method for measuring the elastic modulus.
of a material, often being different for each test, so the slope of the elastic region on each test curve was used to identify the 0.2% off-set strain. The 0.2% off-set strain was zeroed to give a plot of the true stress as a function of true plastic strain. A polynomial fit was used for each curve to determine the average stress-strain flow curve for each test condition.

The average true stress-plastic strain curves for all test conditions, both these determined from SHPB tests and quasi-static tests were used for subsequent fitting of constitutive material models.

5.3 Results

5.3.1 Aluminium 6061-T6 alloy

Figure 5-4 shows the true stress-true strain curves for the quasi-static compression tests obtained at a strain rate of 0.001 s\(^{-1}\) at three test temperatures of 203 K, 294 K and 373 K. This figure shows the curves in their completeness including the elastic region. Figure 5-5 shows these curves in the true stress-true plastic strain form.

Figure 5-4. Quasi-static (strain rate 0.001 s\(^{-1}\)) compression true stress-true strain curves for Al 6061-T6.
Table 5-1 shows the mechanical data obtained from the true stress-true plastic strain curves in Figure 5-5. The data includes the 0.2% off-set strain yield stress (0% true plastic strain) and the true plastic stress at 0.05, 0.1, 0.2 and 0.3.

*Table 5-1. Average true stress as a function of true strain for Al6061-T651 alloy at a 0.001 s⁻¹ strain rate*

<table>
<thead>
<tr>
<th>Test Temperature (K)</th>
<th>Yield Stress</th>
<th>True Stress (MPa) at True Plastic Strain</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>0.05</td>
</tr>
<tr>
<td>203</td>
<td>322.0 ± 5.0</td>
<td>378.8 ± 2.9</td>
</tr>
<tr>
<td>294</td>
<td>307.1 ± 2.8</td>
<td>356.0 ± 1.0</td>
</tr>
<tr>
<td>373</td>
<td>294.5 ± 3.0</td>
<td>333.0 ± 2.0</td>
</tr>
</tbody>
</table>

The true stress values in Table 5-1 have been plotted in Figure 5-6 as a function of test temperature, and this figure shows how the true stress decreases with an increase in temperature.
Figure 5-6. True stress as a function of temperature and true strain for Al 6061-T6 at a 0.001 s$^{-1}$ strain rate.

Figure 5-7 shows the true stress-true strain curve in its entirety obtained for the Al 6061-T6 alloy at a strain rate of 2450 s$^{-1}$ at the indicated various temperatures. As is expected with SHPB test the stress output contains a large amount of oscillation because of system ringing and was therefore difficult to extract the material properties of the alloy from the experimental data in this form. The curves were processed by determining the stress at 0.2% off-set strain and removing this elastic region of the curve. A polynomial fit to the experimentally obtained curve was then obtained as shown in Figure 5-8. Figure 5-9 shows the polynomial curves obtained for each test and the average curve (least squares) for each test condition.
Figure 5-7. True stress as a function of true strain for Al 6061-T6 at a 2450 s\(^{-1}\) strain rate on SHPB.

\[ y = -2503x^2 + 825.14x + 379.96 \]

Figure 5-8. Example of the determination of 0.2\% off-set stress and polynomial curve fit for Al 6061-T6 at a 2450 s\(^{-1}\) strain rate on SHPB.
Table 5-2 shows the mechanical data obtained from the processed SHPB curves for a strain rate of 2540 s\(^{-1}\). The true stress is recorded at 0.2\% yield stress (zero plastic strain), 0.05, 0.1, 0.2, and 0.3 plastic strain.

**Table 5-2. Average true stress as function of true strain for Al 6061-T6 alloy at a strain rate of 2540 s\(^{-1}\)**

<table>
<thead>
<tr>
<th>Test Temperature (K)</th>
<th>Yield Stress</th>
<th>True Stress (MPa) at True Plastic Strain</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>0.05</td>
</tr>
<tr>
<td>203</td>
<td>394.3 ± 6.9</td>
<td>426.6 ± 8.3</td>
</tr>
<tr>
<td>294</td>
<td>340.8 ± 15.0</td>
<td>383.8 ± 8.0</td>
</tr>
<tr>
<td>373</td>
<td>334.7 ± 2.9</td>
<td>355.1 ± 0.1</td>
</tr>
</tbody>
</table>

The data in Table 5-2 are plotted in Figure 5-10 as true stress as a function of true plastic strain at test temperatures of 203 K, 293 K and 373 K.
The yield stress data obtained from the quasi-static and SHPB tests are plotted in Figure 5-11 as function of strain rate. This yield stress data at the two different strain rates is also used in Table 5-3 to calculate the strain rate sensitivity, $m$, of the alloy using equation (2.20).
Table 5-3. Strain rate sensitivity of the true yield stress in compression for Al 6061-T6 alloy.

<table>
<thead>
<tr>
<th>Temperature (K)</th>
<th>$\sigma_1$</th>
<th>$\sigma_2$</th>
<th>$\dot{\varepsilon}_1$</th>
<th>$\dot{\varepsilon}_2$</th>
<th>$m$</th>
</tr>
</thead>
<tbody>
<tr>
<td>203</td>
<td>322.0</td>
<td>394.3</td>
<td>0.001</td>
<td>2450</td>
<td>0.014</td>
</tr>
<tr>
<td>294</td>
<td>307.1</td>
<td>340.8</td>
<td>0.001</td>
<td>2450</td>
<td>0.007</td>
</tr>
<tr>
<td>373</td>
<td>294.5</td>
<td>334.7</td>
<td>0.001</td>
<td>2450</td>
<td>0.009</td>
</tr>
</tbody>
</table>

The strain rate sensitivity was found to be 0.007 at a test temperature 294 K and deviates little from this at a test temperature of 373 K. at a test temperature of 203 K the strain rate sensitivity appears to increase twice that obtained at 294 K.

5.3.2 Tantalum-2.5 wt % Tungsten Wrought Alloy

The true stress-true strain flow curves obtain form quasi-static compression experiments are shown in Figure 5-12 in their entirety. Figure 5-13 shows these curves in the true stress-true plastic strain form, where the elastic part of the curve has been removed at 0.2% off-set strain.

Figure 5-12. Quasi-static compression true stress-true strain curves for Ta-2.5 % W wrought alloy.
The data extracted from Figure 5-13 is tabulated in Table 5-4 showing true stress values at 0.2% yield stress (zero plastic strain), 0.05, 0.1, 0.2 and 0.3 true plastic strain at the various test temperatures.

Table 5-4. Average true stress as function of true compressive strain for Ta-2.5 % W wrought alloy at a strain rate of 0.001 s⁻¹.

<table>
<thead>
<tr>
<th>Test Temperature (K)</th>
<th>Yield Stress</th>
<th>0.05</th>
<th>0.1</th>
<th>0.2</th>
<th>0.3</th>
</tr>
</thead>
<tbody>
<tr>
<td>203</td>
<td>297.2 ± 5.0</td>
<td>404.8 ± 1.5</td>
<td>486.7 ± 2.4</td>
<td>600.0 ± 5.9</td>
<td>657.8 ± 9.5</td>
</tr>
<tr>
<td>294</td>
<td>204.6 ± 4.2</td>
<td>319.5 ± 3.5</td>
<td>385.5 ± 2.3</td>
<td>466.7 ± 2.0</td>
<td>512.2 ± 2.8</td>
</tr>
<tr>
<td>373</td>
<td>168.0 ± 3.9</td>
<td>279.3 ± 2.9</td>
<td>333.7 ± 1.3</td>
<td>394.0 ± 2.2</td>
<td>432.2 ± 3.3</td>
</tr>
</tbody>
</table>

The data in Table 5-4 is plotted in Figure 5-14 graphically showing true stress as a function of test temperature. The curves show how the true stress value decreases with increase in test temperature.
Figure 5-14. True stress as a function of temperature and true compressive strain for Ta-2.5% W wrought alloy at a 0.001 s\(^{-1}\) strain rate.

Figure 5-15 shows the true stress flow curves for the Ta-2.5%W wrought alloy at a strain rate of 2200 s\(^{-1}\) at the three test temperatures. The raw SHPB data was processed as described for the Al 6061-T6 alloy in Section 5.3.1. The resulting processed curves are shown in Figure 5-16, together with an average curve for each test condition.
Figure 5-15. True stress as a function of true strain for Ta-2.5% W wrought alloy at a 2200 s⁻¹ strain rate on SHPB.

Table 5-5 shows the mechanical data obtained from the curves from Figure 5-16. The true stress value at 0.2% yield stress (zero plastic strain), 0.05, 0.1, and 0.2, at the three test temperatures are shown. The true stress-true plastic strain curves did not continue to 0.3 true plastic strain, so no data is available.
Table 5-5. Average true stress as function of true compressive strain for Ta-2.5 % W wrought alloy at a strain rate of 2200 s\(^{-1}\)

<table>
<thead>
<tr>
<th>Test Temperature (K)</th>
<th>True Stress (MPa) at True Plastic Strain</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Yield Stress</td>
</tr>
<tr>
<td>203</td>
<td>662.7 ± 3.9</td>
</tr>
<tr>
<td>294</td>
<td>553.3 ± 3.3</td>
</tr>
<tr>
<td>373</td>
<td>481.8 ± 4.6</td>
</tr>
</tbody>
</table>

Figure 5-17 shows a plot of the true stress as a function of temperature for the alloy at a strain rate of 2200 s\(^{-1}\), and demonstrates the fall in stress with rise in temperature.

Figure 5-18 shows how the yield true stress increases with strain rate over the quasi-static to 2200 s\(^{-1}\) test range. Table 5-6 shows the strain sensitivity, \(m\), for this alloy calculated from equation (2-20) over the strain rate range at each of the test temperatures investigated.
Table 5-6 appears to suggest that the strain rate sensitivity increases with increase in test temperature, the value $m$ increasing from 0.055 to 0.072 over the 203 K to 373 K temperature range.

Table 5-6. Strain rate sensitivity of the true yield stress in compression for Ta-2.5 % W wrought alloy.

<table>
<thead>
<tr>
<th>Temperature (K)</th>
<th>$\sigma_1$</th>
<th>$\sigma_2$</th>
<th>$\dot{\varepsilon}_1$</th>
<th>$\dot{\varepsilon}_2$</th>
<th>$m$</th>
</tr>
</thead>
<tbody>
<tr>
<td>203</td>
<td>297.2</td>
<td>662.7</td>
<td>0.001</td>
<td>2200</td>
<td>0.055</td>
</tr>
<tr>
<td>294</td>
<td>204.5</td>
<td>553.3</td>
<td>0.001</td>
<td>2200</td>
<td>0.068</td>
</tr>
<tr>
<td>373</td>
<td>168.0</td>
<td>481.8</td>
<td>0.001</td>
<td>2200</td>
<td>0.072</td>
</tr>
</tbody>
</table>

5.3.3 Tantalum-2.5 wt % Tungsten HIP Alloy

Figure 5-19 shows the true stress-true strain curves for the Ta-2.5%W HIP alloy under quasi-static strain rate (0.001 s$^{-1}$) at three test temperatures and the true stress-true plastic strain is presented in Figure 5-20.
Figure 5-19. Quasi-static (strain rate of 0.001 s$^{-1}$) true stress-true compressive strain curves for Ta-2.5%W HIP alloy.

Figure 5-20. Quasi-static (strain rate of 0.001 s$^{-1}$) true stress-true compressive plastic strain curves for Ta-2.5%W HIP alloy.

Mechanical data for the alloy obtained from the flow curves in Figure 5-20 are tabulated in Table 5-7 for the quasi-static strain rate condition and the three test temperatures, 203 K, 294 K and 373 K.
Table 5-7. Average true stress as function of true plastic strain for Ta-2.5%W HIP alloy at a strain rate of 0.001 s\(^{-1}\).

<table>
<thead>
<tr>
<th>Test Temperature (K)</th>
<th>True Stress (MPa) at True Plastic Strain</th>
<th>Yield Stress</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>0.05 0.1 0.2 0.3</td>
<td></td>
</tr>
<tr>
<td>203</td>
<td>379.6 ± 8.8 497.0 ± 2.6 577.7 ± 5.5</td>
<td>701.3 ± 11.9</td>
</tr>
<tr>
<td>294</td>
<td>293.2 ± 1.7 399.8 ± 2.9 464.7 ± 3.5</td>
<td>553.7 ± 3.8</td>
</tr>
<tr>
<td>373</td>
<td>250.6 ± 1.6 349.7 ± 2.3 398.7 ± 2.3</td>
<td>461.3 ± 3.1</td>
</tr>
</tbody>
</table>

Figure 5-21 displays the data in Table 5-7 in graphical form and shows the decrease in flow stress with increase in the test temperature for the alloy at a strain rate of 0.001/s.

![Graph showing the decrease in flow stress with increase in test temperature](image)

Figure 5-21. True stress as a function of temperature and true strain for Ta-2.5%W HIP alloy at a 0.001 s\(^{-1}\) strain rate.

Figure 5-22 show the true stress true strain flow curves for the alloy at a strain rate of 2090 s\(^{-1}\) at the three test temperatures. The oscillation present in SHPB raw data was process in a similar way as that with the Al 6061-T6 and Ta-W wrought alloys shown previously. The true stress-true plastic strain form is shown in Figure 5-22, together with an average curve for each test condition.
Figure 5-22. True stress as a function of true strain for Ta-2.5%W HIP alloy at a 2090 s⁻¹ strain rate on SHPB.

Figure 5-23. True stress as a function of true plastic strain for Ta-2.5%W HIP alloy at a 2090 s⁻¹ strain rate on SHPB.
Data extracted from the flow curves in Figure 5-23 are shown in Table 5-8. The true stress as a function temperature is plotted in graphical form in Figure 5-24. Few of the flow curves continued to 0.2 plastic strain therefore only the data and curves up to 0.1 are included.

Table 5-8. Average true stress as function of true strain for Ta-2.5%W HIP alloy at a strain rate of 2090 s⁻¹.

<table>
<thead>
<tr>
<th>Test Temperature (K)</th>
<th>Yield Stress</th>
<th>0.05</th>
<th>0.1</th>
<th>0.2</th>
<th>0.3</th>
</tr>
</thead>
<tbody>
<tr>
<td>203</td>
<td>754.5 ± 1.4</td>
<td>785.7 ± 2.9</td>
<td>805.0 ± 5.3</td>
<td>n/a</td>
<td>n/a</td>
</tr>
<tr>
<td>294</td>
<td>643.0 ± 8.3</td>
<td>684.2 ± 8.3</td>
<td>709.2 ± 12.6</td>
<td>n/a</td>
<td>n/a</td>
</tr>
<tr>
<td>373</td>
<td>547.0 ± 0.4</td>
<td>588.5 ± 5.8</td>
<td>616.0 ± 8.6</td>
<td>n/a</td>
<td>n/a</td>
</tr>
</tbody>
</table>

Figure 5-24. True stress as a function of temperature and true plastic strain for Ta-2.5%W HIP alloy at a 2090 s⁻¹ strain rate.

Figure 5-25 shows how the yield stress varies with strain rate at the three test temperatures. Table 5-9 shows the strain rate sensitivity, \( m \), of the material over the strain rate and test temperature range investigated, which remains fairly consistent covering the small range of 0.47 to 0.054.
Table 5-9. Strain rate sensitivity of the true yield stress for Ta-2.5%W HIP alloy in compression.

<table>
<thead>
<tr>
<th>Temperature (K)</th>
<th>$\sigma_1$</th>
<th>$\sigma_2$</th>
<th>$\dot{\varepsilon}_1$</th>
<th>$\dot{\varepsilon}_2$</th>
<th>$m$</th>
</tr>
</thead>
<tbody>
<tr>
<td>203</td>
<td>379.6</td>
<td>754.5</td>
<td>0.001</td>
<td>2090</td>
<td>0.047</td>
</tr>
<tr>
<td>294</td>
<td>293.2</td>
<td>643.0</td>
<td>0.001</td>
<td>2090</td>
<td>0.054</td>
</tr>
<tr>
<td>373</td>
<td>250.6</td>
<td>547.0</td>
<td>0.001</td>
<td>2090</td>
<td>0.054</td>
</tr>
</tbody>
</table>

5.4 Discussion

Figures 5-26 to 5-28 compare the true stress-true plastic strain traces in tension and compression for Al 6061-T6, Ta-2.5%W wrought and HIP alloys respectively. Table 5-10 summaries the strain rate sensitivity calculated for each material at each temperature and strain rate range. The slope of flow curves for the Al 6061-T6 alloy at quasi-static conditions are identical in tension and compression (Figure 5-26), the flow stress being about 6% higher in the compression test. At the higher rates 760 s$^{-1}$ in tension and 2540 s$^{-1}$ in compression, the slope of the compression curve is slightly steeper than that obtained in tension. Table 5-10, however, shows that the strain rate
sensitivity of the yield stress and at higher strains over the strain rate ranges tested is similar both in tension and compression.

Figure 5-26. Comparison of true stress-true plastic strain curves at quasi-static and dynamic strain rates in tension and compression for the Al 6061-T6 alloy.

Table 5-10. Summary of Al 6061 and Ta-2.5%W alloys strain rate sensitivity in tension and compression.

<table>
<thead>
<tr>
<th>Material</th>
<th>Strain State</th>
<th>Temp (K)</th>
<th>Strain Range (s⁻¹)</th>
<th>Strain Rate Sensitivity at True Plastic Strain</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td></td>
<td>0.01 - 2540</td>
<td>0.014 0.008 0.008 n/a</td>
</tr>
<tr>
<td>Al 6061-T6 (FCC)</td>
<td>Compression</td>
<td>203</td>
<td>0.001 - 2540</td>
<td>0.007 0.005 0.006 n/a</td>
</tr>
<tr>
<td></td>
<td></td>
<td>294</td>
<td></td>
<td>0.009 0.004 0.005 n/a</td>
</tr>
<tr>
<td></td>
<td>Tension</td>
<td>294</td>
<td>0.001 - 760</td>
<td>0.005 - - 0.005</td>
</tr>
<tr>
<td>Ta-2.5%W Wrought</td>
<td>Compression</td>
<td>203</td>
<td>0.001 - 2200</td>
<td>0.055 0.038 0.028 n/a</td>
</tr>
<tr>
<td>(BCC)</td>
<td></td>
<td>294</td>
<td></td>
<td>0.068 0.043 0.033 n/a</td>
</tr>
<tr>
<td></td>
<td>Tension</td>
<td>294</td>
<td>0.001 - 760</td>
<td>0.051 - - 0.02</td>
</tr>
<tr>
<td>Ta-2.5%W HIP (BCC)</td>
<td>Compression</td>
<td>203</td>
<td>0.001 - 2090</td>
<td>0.047 0.031 0.023 n/a</td>
</tr>
<tr>
<td></td>
<td></td>
<td>294</td>
<td></td>
<td>0.054 0.037 0.029 n/a</td>
</tr>
<tr>
<td></td>
<td>Tension</td>
<td>294</td>
<td>0.001 - 280</td>
<td>0.038 - - 0.014</td>
</tr>
</tbody>
</table>
Typical of aluminium and many of its alloys, the strain rate sensitivity of the Al 6061-T6 alloy is very low, giving a value of 0.007 in compression and ~0.005 in tension. The slightly higher value obtained in compression is can be attributed to the wider strain rate range of 0.001 s$^{-1}$ to 2540 s$^{-1}$, as opposed to the 0.001 s$^{-1}$ to 760 s$^{-1}$ range in tension. This would indicate that this alloy is very strain rate independent over the quasi-static and dynamic range in both modes of strain.

These results agree well with work carried out in compression by Lindholm & Bassey [61], and Jiang & Chen [62] who found that while there was a small degree of strain rate dependence in pure aluminium, higher strength aluminium alloys such as Al 6061-T6 tended to be strain rate independent in the $10^{-4}$ s$^{-1}$ to $10^{3}$ s$^{-1}$ strain rate range. Similarly, work by Holt et al. [63] and Maiden & Green [64] found little strain rate sensitivity in this alloy in the quasi-static to 910 s$^{-1}$ and 560 s$^{-1}$ strain rate range respectively.

In contrast to compression however, a number of authors have reported that a certain amount of strain rate sensitivity is present in this alloy in tension. Nicholas [25] suggested an increase of 12% in the flow stress between the quasi-static strain rate and 600 s$^{-1}$ range in tension although they found no strain rate sensitivity between the quasi-static to 4 s$^{-1}$ strain rate range. A number of other researchers have reported they have found that the flow stress increases within the quasi-static to dynamic range. Smith [26] reported an increase flow stress of 15% occurs at a strain of 0.05 within the $1.7 \times 10^{-4}$ s$^{-1}$ and 192 s$^{-1}$ strain rate range, Steidel & Makerov [27] found an increase of ~5% at 69 s$^{-1}$ compared with the values at quasi-static strain rate conditions, while Lindholm et al. [28], reported a 7% increase in the yield stress at $10^{3}$ s$^{-1}$ compared to values at quasi-static rates. Hoge [29] reported a larger increase in the yield stress of 28% going from a strain rate of $5 \times 10^{-1}$ s$^{-1}$ to 65 s$^{-1}$, and yield stress and UTS data compiled by Jiang & Chen [62, 65-70] showed the strain rate sensitivity increased above $1 \times 10^{1}$ s$^{-1}$. Harding et al [67] on similar high strength alloys found an increase of 25% over the quasi-static to 800 s$^{-1}$ and 180 s$^{-1}$ range. All these findings appear to conflict with those found in compression where little or no strain rate sensitivity is observed in this alloy or indeed other similar high strength aluminium alloys up to strain rates of $1 \times 10^{3}$ s$^{-1}$. However, the work here has clearly shown that while this alloy possesses a small amount of strain rate sensitivity in tension over the
quasi-static to 760 s\(^{-1}\) strain rate range, this was not substantially greater than that observed in compression covering the quasi-static to 2540 s\(^{-1}\) range.

Figure 5-27 shows the flow curves for the Ta-2.5\%W wrought alloy produced at quasi-static strain rates in tension and compression, and at 760 s\(^{-1}\) in tension and 2200 s\(^{-1}\) in compression. The flow curve in tension at the higher rate, while 23% lower than that produced in compression, mirrors it well, showing a similar strain hardening rate. At the quasi-static rate for some reason the flow curve in compression shows a higher gradient. The strain rate sensitivity of the alloy shown in Table 5-10 shows that at yield stress the sensitivity is comparable at 0.051 in tension and 0.068 in compression. At higher strains the rate sensitivity in tension is calculated to be less than that in compression.

Figure 5-27. Comparison of true stress-true plastic strain curves at quasi-static and dynamic strain rates in tension and compression for the Ta-2.5\%W wrought alloy.

Figure 5-28 shows the flow curves obtained for the Ta-2.5\%W HIP alloy at quasi-static strains in tension and compression. Again the flow curve in tension shows a shallower raise in the flow stress. The strain rate sensitivity value shown in Table 5-10 for this material shows that it is less strain rate sensitive than the wrought material.
Figure 5-28. *Comparison of true stress-true plastic strain curves at quasi-static strain rate in tension and compression for the Ta-2.5%W HIP alloy.*

5.5 Conclusions

1. A small amount of rate sensitivity was observed in the Al 6061-T6 alloy in tension over the quasi-static to 760 s\(^{-1}\) strain rate range, but does not appear to differ substantially from values obtained in compression covering the quasi-static to 2450 s\(^{-1}\) range. While the flow curve in compression at quasi-static conditions was some 6% higher than that produced in tension the shape of the flow curves in tension and compression were extremely similar. The flow curve produced in compression at a strain rate of 2450 s\(^{-1}\) showed a slightly steeper gradient that those show in tension and under quasi-static conditions under compression.

2. A comparison of the strain rate sensitivity for the Ta-2.5% W wrought alloy showed the values obtained to be comparable in tension and compression. The flow curves were similar in shape although the quasi static trace in compression did show a steeper rise.

3. The Ta-2.5% W HIP material was shown to be less strain rate sensitive than the wrought material in both tension and compression.
CHAPTER 6
Constitutive Modelling

6.1 Introduction

In this chapter, tensile flow curves obtained from the experimental tests described in Chapter 4 are used to assess the degree to which three of the most commonly used constitutive material models represent the material behaviour of these metals/alloys in the quasi-static to dynamic tensile testing range. The models evaluated include; the empirically based Johnson-Cook (J-C) model, and the physically based Zerilli-Armstrong (Z-A) and Mechanical Threshold Stress (MTS) models. The objectives are: (1) to derive model constants for these materials in the quasi-static to dynamic strain rate range, (2) to assess how well the models represent these materials at these test conditions, (3) to compare the model data with those determined by other test methods, such as those obtained in compression in Chapter 5 and those reported in literature, and (4) to generally validate the proposed modelling procedure, using a readily available data analysis software package.

6.2 Experimental Procedure

6.2.1 Determination of Material Model Constants

Average true stress-true plastic strain flow curves experimentally determined as described in Chapter 4 were selected as representative of the material’s tensile behaviour over the quasi-static to dynamic regime. Average true stress-true plastic strain flow curves in OFE copper, Ta-2.5%W and Al 6061-T6 wrought alloys at strain rates of 0.001 s$^{-1}$ at test temperatures of 203 K and 294 K and 760 s$^{-1}$ at a test temperature of 294 K, were chosen as the basis for curve fitting by the constitutive material models. Many of the flow curves produced at other rates and temperatures were often compromised by the low number of tests completed or by equipment compliance issues, such as those at 0.675 s$^{-1}$ (see Section 4-3). Some flow curve data at 0.001 s$^{-1}$ at 203 K and 294 K was also available from the Ta-2.5%W HIP material and used for modelling as was some compression data from experimental tests reported in Chapter 5.
The constitutive material models utilised in this work were the Johnson-Cook (as described in section 2.3.2), Zerilli-Armstrong (as described section 2.3.3), and Mechanical Threshold Stress (as described in section 2.3.4) models. As mentioned in section 2.3.2, the Johnson-Cook model contains a term $T_r$ which is described in the original publication as room temperature [30], while a number of later publications describe this term as the reference temperature. Both versions have been represented in this work. No considerations of the adiabatic effects taking place in the materials were taken into account in either J-C or the Z-A models.

In the case of the J-C and Z-A models, the average true stress-true plastic strain data determined by experiment for each material at each strain rate and temperature were displayed in two columns in Microsoft Excel and the experimental strain rate and test temperature in the appropriate next two columns. The calculated corresponding true stress data at each test condition was represented by the modelling equation, with each of the equation’s constants represented in single cells.

As a starting point, constants available from literature were used as values for the model. The difference between the experimentally determined stress and the calculated model stress at each strain value was determined and the total difference over the testing range totalled in one cell. The Excel Solver optimised the fitting constants to the stress-strain data for the temperatures and strain rates represented by minimising the total difference. All curves were given the same weighting in order to prevent those which failed at a higher value of strain having a greater influence over the calculated model curve than those which failed at lower strain values. The constants for each material over the test regime were thus established. The data are presented in a series of Figures comparing the curves produced by the model with the curves determined experimentally. The equation constants established for each material are shown in a Table for each model.

For the MTS model, each stage in the determination of the various constants for the model was written in Microsoft Excel and use was made of available parameters in the published literature in determining these constants. For this reason modelling the curves for OFE copper only was attempted.
6.2.2 Degree of Model Fit

The degree of model fit for each material at each condition can be qualitatively assessed by visually compared the flow curves established experimentally with those curves produced by the model calculation. However, a quantitative value was also determined by the parameter, $\delta$, indicating the degree of fit defined as:

$$\delta = \frac{\sum_{i=1}^{j} |\sigma_{\text{experimental}}(e_i) - \sigma_{\text{calculated}}(e_i)|}{\sigma_{\text{experimental}}(e_i)}$$

(6.1)

where, the $\sigma_{\text{experimental}}$ is the experimental stress, $\sigma_{\text{calculated}}$ is the calculated stress from the constitutive material models described in equations (2.23), (2.26) and (2.27). The percentage degree of fit for each material strain rate and test temperature was established and shown in a table for each model.

6.3 Results

6.3.1 Johnson-Cook Calculated Curves

Figure 6-1 shows the Johnson-Cook fit for OFE copper experimental tensile data using the equation in its original form where $T^*$ is represented by $(T - T_{\text{Room}})/(T_{\text{Melt}} - T_{\text{Room}})$ room temperature being taken in this case as 294K, while Figure 6-2 shows the model fit where a reference temperature ($T_{\text{Reference}}$), in this case the lowest test temperature is used in place of the room temperature ($T_{\text{Room}}$). As can be seen from Figures 6-1 and 6-2, the J-C model slightly under estimates the initial yield stress, represented by term, $\sigma_0$, and slightly overestimates the rate of hardening, (i.e. the curve slope), for the quasi-static strain rate at both the 294 K and 203 K test temperatures. However, the model fits well for the high strain rate condition of 760 s$^{-1}$. 
Figure 6-1. Comparison of true stress-true plastic strain curves as calculated using the original Johnson-Cook equation (2.23) with the experimentally determined average true stress-true plastic tensile strain curves for OFE copper.

Figure 6-2. Comparison of true stress-true plastic strain curves as calculated using the reference temperature in the Johnson-Cook equation (2.23) with the experimentally determined average true stress-true plastic tensile strain curves for OFE copper.
Figures 6-3 and 6-4 show the Johnson-Cook model fit for the Al-6061 alloy for samples under tension in both the X- and Y-cross rolling directions. Again both versions of the J-C equation are used, Figure 6-3 representing $T^*$ in its original form and Figure 6-4 using a reference temperature (lowest test temperature). The model gives a good fit at all test conditions represented in both the X- and Y-cross rolling directions. Again there is a slight underestimate by the model of the initial yield stress, $\sigma_0$, but the rate of hardening is in good agreement with the experimental data.

Figure 6-3. Comparison of true stress-true plastic strain curves as calculated using the original Johnson-Cook equation (2.23) with the experimentally determined average true stress-true plastic tensile strain curves for Al 6061-T6 Alloy.
Figure 6-4. Comparison of true stress-true plastic strain curves as calculated using the reference temperature version of the Johnson-Cook equation (2.24) with the experimentally determined average true stress-true plastic tensile strain curves for Al-6061 alloy.

Figures 6-5 and 6-6 show the Johnson-Cook model fit curves for the tantalum-tungsten wrought alloy in tension for both representations of $T^*$, while Figure 6-7 shows the model fit for the tantalum-tungsten HIP alloy. As can be seen from the Figures for the wrought alloy, the model curves give a general trend with the experimentally determined curves but do not follow them closely. The initial yield stress, $\sigma_0$ for the room temperature quasi-static and dynamic strain rate conditions are again underestimated, although to a greater degree than that for the copper and aluminium materials. The rate of hardening is underestimated for the quasi-static test conditions and slightly over estimated at the dynamic a strain rate of 760 s$^{-1}$. On the other hand, the model curves for the tantalum-tungsten HIP alloy in tension do give a close representation of the experimentally determined curves.
Figure 6.5. Comparison of true stress-true plastic strain curves as calculated using the original Johnson-Cook equation (2.23) with the experimentally determined average true stress-true plastic tensile strain curves for Ta-W Wrought alloy.

Figure 6-6. Comparison of true stress-true plastic strain curves as calculated using the Johnson-Cook equation (2.23) and a reference temperature with the experimentally determined average true stress-true plastic tensile strain curves for Ta-W Wrought Alloy.
Table 6-1 shows the Johnson-Cook model constants determined for each material for both the original temperature definition and using a reference temperature. Table 6-2 shows how well the model curves represent the experimentally determined curves by a percentage degree of fit figure for each material using equation (6.1). The percentage degree of fit figure is listed for each test condition and an overall percentage figure for each material at all test conditions is also given.

The J-C model equation was applied to the compression data determined in Chapter 5 for the Ta-2.5%W and Al 6061-T6 wrought alloys but no sensible model fit curves were achieved over the quasi-static to 2000 s\(^{-1}\) plus, test condition range reported.
Table 6-1. Johnson-Cook parameters determined from experimental tension data.

<table>
<thead>
<tr>
<th>Material</th>
<th>Version</th>
<th>Parameters</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>$\sigma_0$ (MPa)</td>
</tr>
<tr>
<td>OFE Copper</td>
<td>Original</td>
<td>328.0</td>
</tr>
<tr>
<td></td>
<td>Later</td>
<td>359.1</td>
</tr>
<tr>
<td>Al-6061 T6 Alloy (X-direction)</td>
<td>Original</td>
<td>297.6</td>
</tr>
<tr>
<td></td>
<td>Later</td>
<td>318.5</td>
</tr>
<tr>
<td>Al-6061 T6 Alloy (Y-direction)</td>
<td>Original</td>
<td>286.7</td>
</tr>
<tr>
<td></td>
<td>Later</td>
<td>307.5</td>
</tr>
<tr>
<td>Ta-W Wrought Alloy (X-direction)</td>
<td>Original</td>
<td>263.0</td>
</tr>
<tr>
<td></td>
<td>Later</td>
<td>314.0</td>
</tr>
<tr>
<td>Ta-W Wrought Alloy (Y-direction)</td>
<td>Original</td>
<td>252.9</td>
</tr>
<tr>
<td></td>
<td>Later</td>
<td>319.1</td>
</tr>
<tr>
<td>Ta-W HIP Alloy</td>
<td>Original</td>
<td>-</td>
</tr>
<tr>
<td></td>
<td>Later</td>
<td>384.6</td>
</tr>
</tbody>
</table>

Table 6-2. Degree of fit for true stress–true strain curves calculated using the Johnson-Cook equation (2.24) from experimental tension data.

<table>
<thead>
<tr>
<th>Material</th>
<th>Formula</th>
<th>Percentage Degree of Fit</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>294K, 0.001 s$^{-1}$</td>
</tr>
<tr>
<td>OFE Copper</td>
<td>Original</td>
<td>0.7</td>
</tr>
<tr>
<td></td>
<td>Later</td>
<td>0.5</td>
</tr>
<tr>
<td>Al-6061 Alloy (X)</td>
<td>Original</td>
<td>0.2</td>
</tr>
<tr>
<td></td>
<td>Later</td>
<td>0.0</td>
</tr>
<tr>
<td>Al-6061 Alloy (Y)</td>
<td>Original</td>
<td>0.1</td>
</tr>
<tr>
<td></td>
<td>Later</td>
<td>0.1</td>
</tr>
<tr>
<td>Ta-W Wrought (X)</td>
<td>Original</td>
<td>2.7</td>
</tr>
<tr>
<td></td>
<td>Later</td>
<td>2.6</td>
</tr>
<tr>
<td>Ta-W Wrought (Y)</td>
<td>Original</td>
<td>4.3</td>
</tr>
<tr>
<td></td>
<td>Later</td>
<td>4.3</td>
</tr>
<tr>
<td>Ta-W HIP Alloy</td>
<td>Later</td>
<td>0.9</td>
</tr>
</tbody>
</table>

The quantitative value for the degree of fit of a model curve with that of the experimentally determined curve shows generally that a good fit for all materials is achieved with only the quasi-static strain rate at room temperature test conditions in the tantalum-tungsten wrought alloy showing any significant percentage deviation (~4.3%).
6.3.2 Zerilli-Armstrong Calculated Curves

Figures 6-8 to 6-12 show the Zerilli-Armstrong model curve fits to experimentally determined tension data for the OFE copper, aluminium-6061 alloy and tantalum-tungsten wrought and HIP alloys respectively. As can be seen from Figure 6-8 the fit to the model curves for the copper is generally poor at the lower strain values. For the aluminium-6061 alloy shown in Figure 6.9, the model curves do not represent the experimentally determined curves at the quasi-static strain rates carried out at room temperature at all and are also generally disappointing at the other test conditions. The model trace for the room temperature quasi-static condition is merely represented by a straight line passing through the middle of the experimentally determined curve it represents. Figure 6-10, re-plots the model trace so that the constant $C_0$ which represents the yields stress in the Z-A equation is pinned at the experimental value obtained at $0.001 \text{ s}^{-1}, 294\text{K}$. Again the model curves are a poor fit to the experimental data.

![Figure 6-8. Comparison of true stress-true plastic strain curves as calculated using the original Zerilli-Armstrong equation (2.26) with the experimentally determined average true stress-true plastic tensile strain curves for OFE copper.](image-url)
Figure 6-9. Comparison of true stress-true plastic strain curves as calculated using the original Zerilli-Armstrong equation (2.26) with the experimentally determined average true stress-true plastic tensile strain curves for Al 6061-T6 Alloy.

Figure 6-10. Comparison of true stress-true plastic strain curves as calculated using the original Zerilli-Armstrong equation (2.26) with the experimentally determined average true stress-true plastic tensile strain curves for Al 6061-T6 Alloy ($C_0$ pinned at the experimental value obtained for $0.001 \, s^{-1}, 294K$).
In contrast to the copper and aluminium, the model curves for the tantalum-tungsten wrought and HIP alloys show visibly a much closer fit to the experimentally determined tensile flow curves, as shown in Figures 6-11 and 6-12 respectively.

Figure 6-11. Comparison of true stress-true plastic strain curves as calculated using the Zerilli-Armstrong equation (2.26) with the experimentally determined average true stress-true plastic strain tensile curves for Ta-W Wrought Alloy.
**Figure 6-12.** Comparison of true stress-true plastic strain curves as calculated using the Zerilli-Armstrong equation (2.26) the experimentally determined average true stress-true plastic tensile strain curves for Ta-W HIP alloy.

Figures 6-13 and 6-14 show the Z-A model curves fitted to the Ta-2.5%W and Al 6061-T6 compression data reported in Chapter 5 respectively. At low strain rates, the model overestimates the experimental curves at low strains and underestimates at high strains. At dynamic strain rates the model curve underestimates at low strain and over estimates at high strain.
Figure 6-13. Comparison of true stress-true plastic strain compression curves as calculated using the Zerilli-Armstrong equation (2.26) the experimentally determined average true stress-true plastic compression strain curves for Ta-W wrought alloy.

Table 6-3 displays the Zerilli-Armstrong model constants determined for each material within the quasi-static to dynamic tensile and compression range these tests
were conducted. Table 6-4 and 6-5 displays a quantitative percentage degree of fit value for each model curve. The averaged percent which the model deviates from the experimentally determined data for each material and test condition is shown as well as the overall percentage degree of fit for each material covering the whole test regime.

**Table 6-3. Zerilli-Armstrong parameters determined from experimental data.**

<table>
<thead>
<tr>
<th>Material</th>
<th>Parameters</th>
<th>C₀</th>
<th>C₁</th>
<th>C₂</th>
<th>C₃</th>
<th>C₄</th>
<th>C₅</th>
<th>n</th>
</tr>
</thead>
<tbody>
<tr>
<td>OFE Copper</td>
<td></td>
<td>337.7</td>
<td>n/a</td>
<td>371641</td>
<td>0.0301</td>
<td>0.000889</td>
<td>n/a</td>
<td>n/a</td>
</tr>
<tr>
<td>Al 6061-T6 Alloy (X)</td>
<td></td>
<td>313.7</td>
<td>n/a</td>
<td>371641</td>
<td>0.0313</td>
<td>0.000734</td>
<td>n/a</td>
<td>n/a</td>
</tr>
<tr>
<td>Al 6061-T6 Alloy (Y)</td>
<td></td>
<td>304.2</td>
<td>n/a</td>
<td>371641</td>
<td>0.0325</td>
<td>0.000813</td>
<td>n/a</td>
<td>n/a</td>
</tr>
<tr>
<td>Al 6061-T6 Alloy (X) *</td>
<td></td>
<td>290.6*</td>
<td>n/a</td>
<td>371641</td>
<td>0.0292</td>
<td>0.000732</td>
<td>n/a</td>
<td>n/a</td>
</tr>
<tr>
<td>Al 6061-T6 Alloy (Y) *</td>
<td></td>
<td>287.8*</td>
<td>n/a</td>
<td>371641</td>
<td>0.0305</td>
<td>0.000790</td>
<td>n/a</td>
<td>n/a</td>
</tr>
<tr>
<td>Al 6061-T6 Alloy (Compression)</td>
<td></td>
<td>324.9</td>
<td>n/a</td>
<td>1000</td>
<td>0.0059</td>
<td>0.006656</td>
<td>n/a</td>
<td>n/a</td>
</tr>
<tr>
<td>Ta-W Wrought (X)</td>
<td></td>
<td>162.0</td>
<td>653.1</td>
<td>n/a</td>
<td>0.0054</td>
<td>0.00031</td>
<td>429.4</td>
<td>0.629</td>
</tr>
<tr>
<td>Ta-W Wrought (Y)</td>
<td></td>
<td>117.3</td>
<td>816.9</td>
<td>n/a</td>
<td>0.0054</td>
<td>0.00028</td>
<td>480.1</td>
<td>0.644</td>
</tr>
<tr>
<td>Ta-W Wrought (Compression)</td>
<td></td>
<td>123.3</td>
<td>835.0</td>
<td>n/a</td>
<td>0.0045</td>
<td>0.00024</td>
<td>534.8</td>
<td>0.650</td>
</tr>
<tr>
<td>Ta-W HIP Alloy</td>
<td></td>
<td>100.0</td>
<td>800.0</td>
<td>n/a</td>
<td>0.0049</td>
<td>0</td>
<td>700.0</td>
<td>0.700</td>
</tr>
</tbody>
</table>

* C₀ pinned at the experimental value obtained for 0.001 s⁻¹, 294K.

**Table 6-4. Degree of fit for true stress–true strain curves calculated using the Zerilli-Armstrong equation (2.26) from experimental tension data.**

<table>
<thead>
<tr>
<th>Material</th>
<th>Percentage Degree of Fit</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>294K, 0.001 s⁻¹</td>
</tr>
<tr>
<td>OFE Copper</td>
<td>1.1</td>
</tr>
<tr>
<td>Al 6061-T6 Alloy (X)</td>
<td>3.4</td>
</tr>
<tr>
<td>Al 6061-T6 Alloy (Y)</td>
<td>3.1</td>
</tr>
<tr>
<td>Al 6061-T6 Alloy (X) *</td>
<td>7.4</td>
</tr>
<tr>
<td>Al 6061-T6 Alloy (Y) *</td>
<td>5.6</td>
</tr>
<tr>
<td>Ta-W Wrought (Y)</td>
<td>1.8</td>
</tr>
<tr>
<td>Ta-W HIP Alloy</td>
<td>3.0</td>
</tr>
</tbody>
</table>

* C₀ pinned at the experimental value obtained for 0.001 s⁻¹, 294K.

**Table 6-5. Degree of fit for true stress–true strain curves calculated using the Zerilli-Armstrong equation (2.26) from experimental compression data.**

<table>
<thead>
<tr>
<th>Material</th>
<th>Percentage Degree of Fit</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>203K, 0.001 s⁻¹</td>
</tr>
<tr>
<td>Al 6061-T6</td>
<td>2.8</td>
</tr>
<tr>
<td>Ta-W Wrought</td>
<td>8.2</td>
</tr>
</tbody>
</table>
6.3.3 Mechanical Threshold Stress Calculated Curves

Calculation of the structural evolution parameters and thermal activation parameters for MTS model using the experimental data at a strain rate range of 0.001 s\(^{-1}\) to 760 s\(^{-1}\) and a temperature range of 203 K to 294 K produced the Fisher plots shown in Figures 6-15 and 6-16 respectively. As can be seen, over the small number of data sets available and over the narrow strain rate and temperature range tested, there was little change with temperature and not much mechanical evolution with applied strain. The use of parameter published in open literature did not fit the data established in this work. Possible reasons being the MTS model used in this study was written using Microsoft Excel, which made use of the extended Voce law, constants in open literature often had made use of the hyperbolic tangent law, also, the experimental was established using a number of different test methods, such as SHPB and the Taylor test. Therefore a satisfactory model curve fit to the experimental tension data established here was not possible.

![Fisher plot based on equation (2.36), used to calculate the structure evolution parameters for OFE copper.](image)
6.4 Discussion

6.4.1 Evaluation of the Johnson-Cook Model for FCC Metals/Alloys

The Johnson-Cook model equation represents a good fit to the experimentally determined flow curves for both the OFE copper and aluminium-6061 alloy over the quasi-static to dynamic and test temperature range evaluated (Figures 6-1 to 6-4). The deviation of the model for these materials was better than 1%. Examination of the Tables 6-1 and Figures 6-1 to 6-4 shows that using a reference temperature for \( T^* \) in equation (2-25) instead of room temperature has little effect on the model’s flow curve, although the model’s constants can be quite different. The only advantage of using the room temperature in the original definition is that the value of \( \sigma_0 \) is closer to the yield stress of the material at quasi-static conditions as quoted in literature.

The model constants determined for the aluminium 6061-T6 alloy showed a slight variation in parameters between samples cut parallel to X- and Y-cross rolled directions. The strain hardening exponent, \( n \), was slightly higher in the X- than in the Y-direction, and like wise the parameter \( \sigma_0 \) (signifying stress at zero plastic strain), indicating that the plate material in the Y-cross rolled direction was less well worked,
and therefore containing a lower density of dislocations. Table 6-5 compares these model constants with values available in open literature for the Al 6061-T6 alloy.

Table 6-6. Comparison of Johnson-Cook model constant determined from experimental tension data for Al 6061-T6 with those obtained from literature.

<table>
<thead>
<tr>
<th>Reference</th>
<th>Parameters</th>
<th></th>
<th></th>
<th></th>
</tr>
</thead>
<tbody>
<tr>
<td>Current work (X-direction)</td>
<td>σ₀</td>
<td>297.6</td>
<td>297.3</td>
<td>0.634</td>
</tr>
<tr>
<td>(Room temp.)</td>
<td>B</td>
<td>318.5</td>
<td>312.3</td>
<td>0.631</td>
</tr>
<tr>
<td>Current work (X-direction)</td>
<td>n</td>
<td>286.7</td>
<td>390.2</td>
<td>0.853</td>
</tr>
<tr>
<td>(Reference temp.)</td>
<td>C</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Current work (Y-direction)</td>
<td>M</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>(Room temp.)</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Current work (Y-direction)</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>(Reference temp.)</td>
<td>σ₀</td>
<td>307.5</td>
<td>434.6</td>
<td>0.850</td>
</tr>
<tr>
<td>Zhu et al. [68] 25mm gauge length</td>
<td>B</td>
<td>203 ± 43</td>
<td>244 ± 67</td>
<td>0.427 ± 0.16</td>
</tr>
<tr>
<td>Zhu et al. [68] 50mm gauge length</td>
<td>n</td>
<td>236 ± 46</td>
<td>430 ± 54</td>
<td>0.376 ± 0.02</td>
</tr>
<tr>
<td>D. Lesuer et al [69]</td>
<td>C</td>
<td>324</td>
<td>114</td>
<td>0.42</td>
</tr>
</tbody>
</table>

The model constants for Al 6061-T6 obtained from literature shown in Table 6-5 have been taken from work by Zhu et al. [68] and Lesuer et al [69]. Those obtained by Zhu et al., were established by fitting the model to experimental data determined from a series of dynamic tensile tests using a servo hydraulic high strain rate testing machine. This work was carried out on flat specimens of 25 mm and 50 mm gauge length at strain rates of 22 s⁻¹ to 200 s⁻¹ and 5 × 10⁻⁴ s⁻¹ to 120 s⁻¹ respectively. As in this study no account of adiabatic effects were taken into account but the constitutive model fitted the experimental data very well. The model constants determined by Lesuer et al [66], were established over a greater strain rate and temperature range than that by Zhu et al. Examination of Table 6.5 shows there is considerable variation in the values for the constants in each of the studies, notably the parameter σ₀ which in J-C model represents the stress at zero plastic strain and n representing the strain hardening are higher in this study than those determined by Zhu et al. [68] or Lesuer et al. [69].

While the value of strain hardening coefficient n will depend on the strain range over which the coefficient is optimised, as can be seen from the values published by Zhu et al [68], there is considerable variation in the parameter n simply due to the size of the tensile specimen. Their work showed that value of n was lower in the longer 50 mm
gauge length specimen than in the shorter 25 mm gauge length specimen, indicating the strain hardening of the material is dependent on the specimen’s gauge length. In this present study the specimen’s gauge length is considerably shorter than those used in the work by Zhu et al., being 17.78 mm in length, and therefore suggesting the reason for the higher strain hardening coefficient $n$ value.

Table 6-6 compares the constants established in this study with those provided by Johnson & Cook [32] for OFE copper. The test data was primarily obtained from torsion and SHPB tensile tests over a wide range of strain rates (quasi-static to about 400 s$^{-1}$) and temperatures. These parameters are commonly used for representing the constitutive behaviour of OFE copper. As can be seen, they differ considerably from those obtained in this work. The constants $B$ and $n$ characterising the work hardening behaviour of the material are considerably higher, and there is considerable difference in the constant $\sigma_0$ which represents the yield stress.

![Table 6-7. Comparison of Johnson-Cook model constants determined from experimental tension data for OFE copper with those obtained from literature.

<table>
<thead>
<tr>
<th>Reference</th>
<th>$\sigma_0$</th>
<th>$B$</th>
<th>$n$</th>
<th>$C$</th>
<th>$M$</th>
</tr>
</thead>
<tbody>
<tr>
<td>Current work (room temp.)</td>
<td>328.0</td>
<td>354.5</td>
<td>0.790</td>
<td>0.0091</td>
<td>n/a</td>
</tr>
<tr>
<td>Current work (reference temp.)</td>
<td>359.1</td>
<td>413.1</td>
<td>0.861</td>
<td>0.0097</td>
<td>0.983</td>
</tr>
<tr>
<td>Johnson &amp; Cook [32]</td>
<td>90</td>
<td>292</td>
<td>0.31</td>
<td>0.025</td>
<td>1.09</td>
</tr>
</tbody>
</table>

Work by Marias et al. [70] on OFE copper using data obtained from SHPB compression tests also found the parameters determined by Johnson & Cook [32] deviated substantially from their results at low strains but at higher strains there was closer convergence. They concluded that the annealed condition of copper of their test specimen and those used to establish Johnson-Cook parameter differed somewhat.

As would be expected, this work and those in literature show the degree to which the Johnson-Cook constants depend not only on the material, but also its mechanical history, strain condition and size of the test sample. The specific constants therefore need determining for each specific material, its mechanical and thermal history and its application.
6.4.2 Evaluation of the Johnson-Cook model to BCC metals/alloys

The model fits to the experimentally determined tantalum-tungsten wrought alloy flow curves were not as visually satisfactory, deviating somewhat at low values of plastic strain (Figures 6-5 and 6-6). The model data also deviated somewhat at the later stages from the experimentally determined curves. In this case the deviation parameter over this test regime could be out by as much as 4.3% for the quasi-static condition and with an overall deviation of no better than 1.6% over the test regime. Table 6-7 shows the Johnson-Cook constants determined in this work for the Ta-2.5%W wrought alloy together with those available in literature.

*Table 6-8. Comparison of Johnson-Cook model constant determined from experimental tension data for wrought tantalum-2.5% tungsten with those obtained from literature.*

<table>
<thead>
<tr>
<th>Reference</th>
<th>Parameters</th>
<th></th>
<th></th>
<th></th>
<th></th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>$\sigma_0$</td>
<td>$B$</td>
<td>$n$</td>
<td>$C$</td>
<td>$M$</td>
</tr>
<tr>
<td>Current work (X) (Room temp.)</td>
<td>263.0</td>
<td>475.2</td>
<td>0.840</td>
<td>0.0406</td>
<td>n/a</td>
</tr>
<tr>
<td>Current work (X) (Reference temp.)</td>
<td>314.0</td>
<td>505.9</td>
<td>0.766</td>
<td>0.0396</td>
<td>0.516</td>
</tr>
<tr>
<td>Current work (Y) (Room temp.)</td>
<td>252.9</td>
<td>552.4</td>
<td>0.931</td>
<td>0.0473</td>
<td>n/a</td>
</tr>
<tr>
<td>Current work (Y) (Reference temp)</td>
<td>319.1</td>
<td>622.3</td>
<td>0.847</td>
<td>0.0467</td>
<td>0.437</td>
</tr>
<tr>
<td>Chen &amp; Gray [38]</td>
<td>390</td>
<td>700</td>
<td>0.575</td>
<td>0.0400</td>
<td>0.500</td>
</tr>
<tr>
<td>Chen &amp; Gray [38] (larger strains)</td>
<td>270</td>
<td>650</td>
<td>0.325</td>
<td>0.0375</td>
<td>0.600</td>
</tr>
</tbody>
</table>

The constants published by Chen & Gray [38] were established from SHPB compression tests using solid cylindrical samples 6.35 mm in diameter by 6.35 mm in length. These samples were tested at strain rates from the quasi-static at 77 K and 298 K to strain rates of 1000 s$^{-1}$ to 8000 s$^{-1}$ at test temperatures of 77 K to 1273 K, the effects of adiabatic heating at high strain rates being neglected at low strains. The constants are in general a good match with those established by Chen & Gray, although the work hardening exponential, $n$, value is higher in this work.

The formulation of the Johnson-Cook model, equation (2.23), which was originally derived for FCC metals, presumes that the stress-strain curves diverge upon increasing deformation after yielding as is the case for copper and nickel [32]. In an unalloyed BCC metals such as tantalum the strain hardening rate is insensitive to the strain rate and temperature change within the quasi-static to dynamic range investigated. Although the temperature and strain rate sensitivity of tantalum is significantly modified by the addition of tungsten owing to an increase in the
importance of dislocation-solute interactions, work by Chen & Gray [38] showed that substantial deviation of the model predictions can be expected at higher strains due to this divergence of the model.

The model fit for the tantalum-tungsten HIP alloy was much more visually expectable with the deviation parameter being better than 1%. The manufacturing technique may have had an influence on the mechanical behaviour of the material (see Section 4.4.2).

6.4.3 Evaluation of the Zerilli-Armstrong model for FCC metals/alloys

The Zerilli-Armstrong model equation as described for FCC metals and alloys did not give a good fit for either the OFE copper or aluminium 6061-T6 alloy. The equation for the FCC crystal structure means that $C_1$ and $C_5$ are set to zero, which results in the yield stress (the point at zero plastic strain) as specified by $C_0$ with little scope for variation with change in temperature and strain rate. This results in the model simply taking an averaged value for $C_0$ for the tests giving the lowest yield stress, resulting in a trace at this test condition being represented by a straight line which simply passes through the middle of the experimental curve (Figure 6.9). Pinning the value of $C_0$ at the lowest yield stress value did not give any substantial improvement to the model fit.

The Z-A model does not allow enough degree of change in the yield stress experienced over this strain rate test range in tension for these FCC metallic materials. While the change in the yield stress for the OFE copper is only ~40 MPa and the aluminium 6061-T6 alloy only ~20 MPa, as compared to the much greater change in BCC tantalum-tungsten wrought alloy of ~225 MPa, it seems unreasonable to assume little or no change will take place over this quasi-static to dynamic strain rate range. As was stated in Section 2.2.1, slip in a FCC crystal lattice will have a lower temperature and strain rate dependence than that in a BCC crystal lattice as the activation energy for flow on a close packed plane is extremely small. The basis of the Z-A model for an FCC metal assumes that the extremely low activation energy requirement to initiate slip means there is no provision for variation in the critical resolved yield stress for the material with change in strain rate and temperature. The development of many of these constitutive equations have often been developed from high strain rate data acquired from SHPB compression tests, where the yield stress
may not be as sensitive to a change strain rate/temperature. Over the quasi-static to dynamic/temperature tensile range investigated here using standard sized tensile samples, the 0.2% off-set stress varies enough for the Z-A equation to be unsuitable for modelling FCC metallic materials. Tables 6-5 and 6-6 show the constants determined for the OFE copper and those acquired from literature. No Zerilli-Armstrong constants for Al 6061-T6 were available in open literature.

Table 6-9. Zerilli-Armstrong constants established for OFE copper with those in open literature.

<table>
<thead>
<tr>
<th>Reference</th>
<th>Parameters</th>
<th>C₀</th>
<th>C₂</th>
<th>C₃</th>
<th>C₄</th>
<th>n</th>
</tr>
</thead>
<tbody>
<tr>
<td>Current work</td>
<td></td>
<td>337.7</td>
<td>371641</td>
<td>0.0301</td>
<td>0.000889</td>
<td>0.5</td>
</tr>
<tr>
<td>Gray et al [71]</td>
<td></td>
<td>11</td>
<td>1350</td>
<td>0.0011</td>
<td>0.000025</td>
<td>0.7025</td>
</tr>
<tr>
<td>Zerilli &amp; Armstrong [33]</td>
<td></td>
<td>46.5</td>
<td>890</td>
<td>0.0028</td>
<td>0.000115</td>
<td>0.5</td>
</tr>
</tbody>
</table>

The parameters established for copper by Gray et al. [71] were obtained from experimental results conducted from compression experiments using SHPB at a strain rate of 200 s⁻¹ covering a temperature range of 298 K to 1073 K. The model constants determined by Zerilli & Armstrong [33] were established from experimental tension tests results of Johnson & Cook covering using a tensile tester and SHPB over a range of quasi-static to 400 s⁻¹ at temperatures of 300 K, 500 K and 735 K. The constants listed for copper by these three studies varied considerably. The constant C₂ established for both the copper and aluminium 6061-T6 alloy for this work (Table 6-5) were unreasonably large, as C₂ should be describing the evolution of flow stress with strain and its variation with temperature and strain rate.

Another problem with the Z-A model in its original form for FCC metals is that it demands that the strain exponent be set to 0.5. Chen & Gray [72] have shown that the equation in this form produces a curve that misses the experimental data for many FCC metals. In their paper, Armstrong & Zerilli [33] make the simple argument that for FCC metals, strain hardening is proportional to the square root of the plastic strain. However, Chen & Gray found that the best fit for FCC metals was when this exponent was derived as a variable for each material. The square root value of the exponent only works well if all dislocations are mobile and evenly distributed in the crystal [71-75]. It neglects the concept of a saturation stress and that materials do not work harden indefinitely. Materials with substantial dislocation generation, recovery and
storage rates cannot be described by this simple argument [73-75], and consequently no unique relationship exist between these three effects, each being rate and temperature dependent which will vary with material.

6.4.4 Evaluation of the Zerilli-Armstrong model for BCC metals/alloys

The Zerilli-Armstrong model equation describes the tantalum-tungsten wrought alloy extremely well. For BCC metallic crystal structures the model allows for the expected wide range of yield strength at different strain rates and test temperatures.

Table 6-10. Zerilli-Armstrong constants obtained for Ta-2.5%W and those open literature.

<table>
<thead>
<tr>
<th>Material</th>
<th>Parameters</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>C₀</td>
</tr>
<tr>
<td>Ta-W Wrought (X) Tensile</td>
<td>162</td>
</tr>
<tr>
<td>Ta-W Wrought (Y) Tensile</td>
<td>117</td>
</tr>
<tr>
<td>Ta-W Wrought Compression</td>
<td>123</td>
</tr>
<tr>
<td>Chen &amp; Gray [38]</td>
<td>140</td>
</tr>
<tr>
<td>Chen &amp; Gray [38] large strains</td>
<td>55</td>
</tr>
</tbody>
</table>

The constants published by Chen & Gray [38] were established from SHPB compression tests using solid cylindrical samples 6.35 mm in diameter by 6.35 mm in length. These samples were tested at strain rates from the quasi-static at 77 and 298 K to strain rates of 1000 to 8000 s⁻¹ at test temperatures of 77 to 1273 K, the effects of adiabatic heating at high strain rates was neglected at low strains. The constants published by Chen & Gray show how the equation constants are dependent on the strain range over which they are determined. The constants in this work were established over the whole strain range, zero plastic strain to failure and are likewise much closer representation of those established by Chen & Gray over greater strain range, rather than those established over the large strain region only.

6.4.5 Evaluation of the MTS model

The MTS model contains many material constants and the dependence of the model on these constants is highly nonlinear. As a result, fitting the model to uniaxial stress experimental data must involve nonlinear regression in order to obtain those material constants and as a general rule, requires an educated guess of the initial values of
those fitting parameters. The parameter determination process can be described as turning knobs to achieve a good fit. Use can be made of readily available published data for OFE Cu [34, 35] and for Ta-2.5%W alloy by Chen & Gray [38]. However, researchers have customarily used two different strain hardening laws in determining the models constants. These are the extended Voce law as described by equation (2-29), which is in the form of a simple power law, while the other is represented by hyperbolic tangent functions shown in equation (6.2).

\[
\frac{d\hat{\sigma}}{d\varepsilon} = \theta_0 \left\{1 - \frac{1}{\tanh(\alpha)} \tanh\left(\frac{\alpha \hat{\sigma}_\varepsilon}{\hat{\sigma}_0}\right)\right\}
\]  

(6.2)

where \(\alpha\) is a constant \(\geq 0\), while \(\theta_0\) and \(\hat{\sigma}_0\) have the same meanings as those in equation (2.29). As can be seen these two strain hardening laws are apparently different but both describe the same material hardening behaviour. Using parameters from published in literature therefore raises the question whether the set of parameters obtained based on one hardening law can be applied to the other. The model available here made use of the extended Voce law for the structural evolution. The hyperbolic tangent law which describes the strain hardening law more accurately may be more suitable over such a narrow range.

6.5 Conclusions

1. The Johnson-Cook equation represents the OFE Cu and Al 6061-T6 alloy experimental data well, slightly under estimating the initial yield stress in both materials and slightly over estimating the rate of hardening in the copper at quasi-static strain rates. For the Ta-W wrought alloy, the J-C model gives a general trend to the experimental data only, under estimating the yield stress and over estimating rate of hardening at high strain rates and over estimating the yield stress and under estimating the hardening rate at quasi-static rates. The model constants determined in this work generally agreed well with those in literature although the constant \(n\) was very dependent on the strain range it is optimised over and on the specimen gauge length. The J-C model equation was also applied to the compression data in quasi-static to 2000 s\(^{-1}\) plus strain rate range. No sensible model fit of the curves to the experimental data was
achieved. The simplicity of the J-C model is its merit, but may limit its capability to handle more complex behaviour, for instance, the Peierls stress contribution found in BCC materials. Also at higher strain rates the adiabatic effect needs to be taken into account by incorporating an additional part to the equation.

2. The Z-A model for FCC metals in the form shown in equation (2-26) does not represent FCC materials in tension and over the quasi-static to dynamic strain rate range well. The equation giving little scope for variation in the yield stress over this range of test condition in tension. The strain exponent as a fixed constant for all FCC materials set at 0.5 needs to be treated as an adjustable parameter for specific materials. For BCC materials the Z-A model gives a good fit for the Ta-W wrought alloy under tension within the test range conditions described in this work. The fit was not so good for the compression data covering a large strain rate range. Again, adiabatic effects need to be taken into account at the high strain rates and an addition to the equation needs to be incorporated.

3. The MTS model does not give a good fit for the limited strain rate and temperature test range conditions investigated in this study. The structural evolution over this range changed little with temperature and strain rate and determination of the model constants was not possible.
CHAPTER 7

Conclusions and Further Work

7.1 Introduction

The primary aim of this thesis was to develop an understanding of the problems of tensile testing metallic materials in the quasi-static to dynamic strain rate regime. The work identified the steps necessary to extract meaningful material data from the raw data provided by the test equipment. This data was compared with data acquired from other commonly used mechanical testing techniques covering a similar strain rate range. The data was also applied to a number of the most commonly used constitutive material models and the constants derived from this data compared with those published in open literature. A summary of the main conclusions from this work are highlighted below, together with a proposal for further work.

7.2 Evaluation of Dynamic Tensile Testing

Care must be taken to ensure that a dynamic equilibrium is maintained in the tensile sample during testing. This may be done by visual examination of the stress-strain curve and/or by calculation using criteria established by criteria established in SHPB tests. High speed video imaging of the tensile sample during the test can also give an indication if dynamic stress equilibrium is maintained throughout the test. The work strongly suggests that each material and specimen geometry will have its own strain rate threshold at which stress equilibrium is maintained.

Appropriate methods of processing of raw dynamic data need to be developed in order to extract meaningful data from high strain rate tensile experiments. Methods described in Chapter 2, namely, measuring the stress directly from strain gauges attached to the dynamometer section of the tensile specimen, calibrated using the piezoelectric load cell reading as a reference, gave results with a low standard of deviation. Data obtained from this method seemed to provide legitimate material properties over the quasi-static to dynamic strain rate range tested.

The DIC technique showed it could be used at the dynamic strain rate range as a non-contact extensometer. As an image technique to assess the strain distribution within
the test sample gauge length, it was shown to be able to determine if stress equilibrium is maintained in the test sample over the duration of the test. DIC was also shown to be capable of providing images capable of being measured in order to extend the true stress-true strain curve beyond the instability point in the true stress-strain curve.

7.3 Effect of Strain Rate and Temperature

A number of metallic materials were successfully evaluated in the quasi-static to dynamic strain rate range, producing data expected of those materials. Al-6061-T6 and OFE copper response was typical of FCC materials, their yields stress being only weekly dependent on an increase in strain rate and temperature. The strain hardening of OFE copper was shown to be strongly dependent on strain rate owing to its low-stacking fault energy, while Al 6061-T6 due to it high stacking fault energy was shown to be largely strain rate independent within this strain rate regime. The Ta-2.5%W alloy in contrast showed behaviour typical of BCC materials. Its yield stress was highly strain rate dependent but its strain hardening rate was largely strain rate independent. This was the case for both the wrought and HIP material, although the strain rate sensitivity of the HIP material’s yield stress was slightly than that of the wrought material.

7.4 Comparison of Tensile and Compression Tests

Flow curves obtained for both the Ta-2.5%W wrought, HIP’d and Al 6061 alloys in compression tended to be at a higher stress value than those obtained in tension. However, the strain rate sensitivity in compression in these alloys did not appear to differ substantially from the values obtained in tension over the quasi-static to dynamic range tested.

7.5 Constitutive Modelling

The J-C model represents the experimental tension data obtained for the FCC materials of OFE copper and Al 6061-T6 behaviour well within the quasi-static to dynamic range tested. The model only gives a very general trend in the case of the BCC Ta-2.5%W alloy. The Z-A model equation for FCC materials does not allow for the variation in yield stress obtained in tension over the quasi-static to dynamic range,
and therefore does not represent the behaviour of this alloy well over this strain rate range. For the BCC Ta-2.5%W alloy the Z-A model gives a good fit to the tension data, but did not give a good representation of the compression data which covered a wider strain rate range and where adiabatic effects need to be considered. The MTS model required a greater number of data sets covering a greater strain rate and temperature range in order to determine the model constants.

7.6 Further Work

The work has highlighted a number of areas where further work is required to improve the high strain rate tensile testing methodology and advance the understanding of the material mechanics taking place during testing within the dynamic strain rate range. Of primary importance to the high strain rate tensile test is the measurement of the strain by the DIC method. While the images obtained in this work were suitable for using the DIC method as a non-contact optical extensometer, the variable quality of the images were often not suitable for strain distribution maps across the specimen gauge length. Improvement of the quality and reproducibility of these images requires a method to achieve the optimum speckle size and distribution and also an improvement in the adhesion of the underlying paint to the specimen’s surface. An experimental programme using different paints, spraying techniques and test sample surface preparation methods could improve this process.

DIC images for extending the true stress-true strain curve beyond the instability point also need to be improved. Measurement of the radius of curvature of the neck and the radius of the thinnest part of the neck requires a sharp silhouette image of the specimen gauge length necking region. At these high strain rates the quality of the images are dependent on the lighting, background, and the distance of the camera from the specimen. Some work is required to optimise these conditions. When a material deforms in an anisotropic manner during necking, where out of plane motion of the necked region takes place, as shown by the wrought Ta-2.5%W alloy (Figure 4-29), 3-D images will be required. 3-D Images obtained from a two camera set-up at the higher dynamic strain rates require experimental optimisation of variables such as, the camera shutter speed, lighting and distance of the camera from the specimen.
Further work is also necessary to ensure that a tensile test at these high strain rates gives legitimate results. An investigation into the occurrence of double necking taking place at the higher strain rates and how the dimensions of the specimens and test material properties influence the strain rate threshold at which this happens could provide valuable information regarding future tensile testing within this strain rate regime. A combination of strain distribution maps provided by the DIC technique and FE analysis would be invaluable in this type of investigation on a variety of different metals and alloys.

Developing a more practical form of the MTS model programme is also necessary to advance this work. The large number of constants that need to be determined in the MTS model, require a greater number of data sets (i.e. combinations of test strain rates and temperatures) than that obtained in this work. While it would be difficult to extend the strain rate range covered in this work, a larger number of tests could be carried out within the 0.001 s⁻¹ to 760 s⁻¹ range, together with a greater number of tests covering the 173 K to 473 K temperature range. Also, variations on the MTS model such as the use of the hyperbolic tangent law (equation 6.2) are worth investigating owing to the greater amount of material data that is available in published literature.


44. Copper CDA 11- material property data sheet. VFM. (2009)
47. T. Dunnett & D.S. Balint, Correction factor software, current PhD work, Mechanics of Materials, School of Mechanical Engineering, Imperial College, (2013).
54. ABAQUS Inc., ABAQUS Analysis User’s Manual V. 6.7 (207).


APPENDIX

Conditions used for the 2-D Axis Symmetry FEA Model

Finite Element Analysis (FEA) was carried out using ABAQUS/Explicit [54]. The element type used for both the tensile test piece and the striker was CAX4R (4 node bilinear axisymmetric quadrilateral, reduced integration, hour glass control). A total time displacement (“time step”) of 0.48 ms was used for the model. Typical true stress-true plastic strain material data established from experimental tests at these test conditions, such a yield stress (i.e. zero plastic stain) and strain hardening rate were used in the model to represent the material behaviour. The bottom shoulder of the specimen was clamped using “encastre” boundary conditions. Boundary conditions were applied to the striker in the X- and Z- directions with a velocity of 13292 mm/s in the Y-direction, giving strain rate of 747 s⁻¹.

The other material properties used in the FEA model were as follows:

**Tantalum-2.5% tungsten tensile specimen:**
- Young’s Modulus: 179 GPa [43]
- Poisons ratio: 0.35 [43]
- Density: 16.6 g/cm³ [43]

**Material properties for striker:**
- Young’s Modulus: 1000 GPa
- Poisons ratio: 0.1
- Density: 7.8 g/cm³

The FEA images were displayed showing maximum plane strain, colour coded from blue-green-yellow-red, representing the lower to higher strain values present within the tensile model image respectively.