EFFECT OF TEMPER CONDITIONS ON TORSIONAL BEHAVIOR OF SELECTED 2000 SERIES ALUMINUM ALLOYS AT HIGH STRAIN RATES

A Thesis Submitted to the College of Graduate and Postdoctoral Studies In Partial Fulfillment of the Requirements For the Degree of Master of Science In the Department of Mechanical Engineering University of Saskatchewan Saskatoon

By

Nima Gharraie

© Copyright Nima Gharraie, January 2021. All rights reserved. Unless otherwise noted, Copyright of the material in this thesis belongs to the author
PERMISSION TO USE

In presenting this thesis in partial fulfillment of the requirements for a Postgraduate degree from the University of Saskatchewan, I agree that the Libraries of this University may make it freely available for inspection. I further agree that permission for copying of this thesis in any manner, in whole or in part, for scholarly purposes may be granted by the Prof. Akindele Odeshi who supervised my thesis work or in his absence, by the Head of the Department or the Dean of the College in which my thesis work was done. It is understood that any copying or publication or use of this thesis or parts thereof for financial gain shall not be allowed without my written permission. It is also understood that due recognition shall be given to me and to the University of Saskatchewan in any scholarly use which may be made of any material in my thesis.

Requests for permission to copy or to make other uses of materials in this thesis/dissertation in whole or part should be addressed to:

Head of the Department of Mechanical Engineering
57 Campus Drive
University of Saskatchewan
Saskatoon, Saskatchewan S7N 5A9 Canada.

OR

Dean
College of Graduate and Postdoctoral Studies
University of Saskatchewan
116 Thorvaldson Building, 110 Science Place
Saskatoon, Saskatchewan S7N 5C9 Canada
Abstract

In recent times, aluminum alloys have been widely used in different industrial sectors, including transportation, packaging, electrical, etc. Due to their lightweight and high specific strength and modulus, they are highly favored in the aerospace and automobile industries. Over the years, aluminum alloys have been produced in different grades (1000, 2000, …, 8000) due to the high demand for newer alloys with better properties. This thesis focuses on deformation and damage mechanisms in AA2017, AA2024, and AA2624 aluminum alloys under high strain rate torsional loading.

The mechanical response of the as-received aluminum alloys under high strain-rate torsional loading using torsional split Hopkinson bar was investigated. Cylindrical test specimens were subjected to high strain-rate torsional loading using angles of twist (ϕ) of 4°, 8° and 12°. Heat treatment processes were conducted on two sets of naturally aged specimens to obtain T651 and O temper condition. The effect of temper conditions on the microstructural evolution in the investigated alloys was investigated. Optical and scanning electron microscopy techniques were used to observe the morphology or composition of the second phase particles, grain structure, and crystallographic texture of the investigated alloys before and after rapid torsional deformation.

Results from this study showed that the strength and ductility of the investigated alloys are dependent on the temper condition. Increasing the angle of twist resulted in the generation of higher strains and strain rates in the alloys for all temper conditions. Al-Cu-Mg alloys (AA2024, AA2624) have higher peak flow stress in the artificially aged condition than Al-Cu alloy (AA2017). The annealed specimens showed lower peak flow stress and higher ductility than their age-hardened counterparts. Artificial aging of AA2024 alloy led to lower plasticity and failure at lower stress levels than the naturally aged AA2024 alloy. In the naturally aged condition, AA2024 and AA2624 alloys exhibit the highest, and the lowest strength, respectively. Fractography of the AA2024-T651 specimen revealed a fracture mode that exhibits ductile features. Micrographs of the investigated alloys after torsional loading did not indicate heterogeneous deformation leading to strain localization and the development of adiabatic shear bands.
Acknowledgments

I would like to express my gratitude to my supervisor, Dr. Akindele Odeshi, for his guidance through each process stage. I appreciate his support throughout my master's degree program, without which this thesis would not have been possible. The door to Dr. Odeshi’s office was always open to me whenever I ran into a problem or had questions about my research, and he steered me in the right direction whenever he thought I needed it. The contributions of Prof. Ike Oguocha and Prof. Duncan Cree, my advisory committee members, are greatly appreciated; their valuable and insightful comments contributed to the success of this thesis. My appreciation also goes to Prof. Szpunar for permitting me to use his EBSD facility.

My gratitude also goes to Mr. Tonye Jack and Mr. Reza Khatib for their time, advice, and patience in training me on most of the equipment I needed to use in my MSc program.

I would like to express my appreciation to my parents for their support and encouragement to see me succeed. To my siblings, Raha and Iman, whose help was what sustained me this far.

I would like to thank ALCOA for providing one of the investigated alloys (AA2624) for this study at no cost.
Dedication

To my family for all their support
Table of Contents

PERMISSION TO USE .............................................................................................................. ii

Abstract ..................................................................................................................................... iii

Acknowledgments .................................................................................................................... iv

Dedication ................................................................................................................................. v

Table of Contents ..................................................................................................................... vi

List of Tables ............................................................................................................................. ix

List of Figures .......................................................................................................................... x

List of Abbreviations ............................................................................................................... xv

List of Symbols ......................................................................................................................... xvi

CHAPTER 1 ................................................................................................................................. 1

INTRODUCTION ....................................................................................................................... 1

1.1 Overview ............................................................................................................................. 1

1.2 Motivation ........................................................................................................................... 2

1.3 Research objectives ............................................................................................................. 3

1.4 Thesis arrangement ............................................................................................................. 4

CHAPTER 2 ................................................................................................................................. 5

LITERATURE REVIEW .............................................................................................................. 5

2.1 Aluminum .......................................................................................................................... 5

2.2 Aluminum alloys classification .......................................................................................... 6

2.2.1 Wrought composition families .................................................................................... 7

2.3 Precipitation hardening of 2000 series aluminum alloys .................................................. 9

2.3.1 Al-Cu alloys ................................................................................................................ 10
<table>
<thead>
<tr>
<th>Section</th>
<th>Title</th>
<th>Page</th>
</tr>
</thead>
<tbody>
<tr>
<td>2.3.2</td>
<td>Al-Cu-Mg alloys</td>
<td>13</td>
</tr>
<tr>
<td>2.4</td>
<td>Dislocation-particles interactions</td>
<td>15</td>
</tr>
<tr>
<td>2.5</td>
<td>Deformation behavior of metallic materials under mechanical loading</td>
<td>17</td>
</tr>
<tr>
<td>2.6</td>
<td>Shear and Torsional test</td>
<td>21</td>
</tr>
<tr>
<td>2.7</td>
<td>Adiabatic Shear Band (ASB)</td>
<td>23</td>
</tr>
<tr>
<td>2.8</td>
<td>Hardening curve of metallic alloys under the dynamic shock loading</td>
<td>24</td>
</tr>
<tr>
<td>2.9</td>
<td>Fracture</td>
<td>26</td>
</tr>
<tr>
<td>2.9.1</td>
<td>Ductile fracture</td>
<td>27</td>
</tr>
<tr>
<td>2.9.2</td>
<td>Brittle fracture</td>
<td>28</td>
</tr>
<tr>
<td>2.10</td>
<td>Elemental composition of the investigated 2000 series aluminum alloys</td>
<td>30</td>
</tr>
</tbody>
</table>

CHAPTER 3 .................................................................................................................. 33

MATERIALS AND METHODOLOGY ................................................................................. 33

3.1     Materials | 33   |
3.2     Heat treatment procedure | 33   |
3.3     Torsion test | 34   |
3.4     Microstructural investigation | 37   |
| 3.4.1  | Optical microscopy and scanning electron microscopy | 38   |
| 3.4.2  | Electron backscatter diffraction (EBSD) and energy dispersive spectroscopy (EDS) | 40   |

CHAPTER 4 .................................................................................................................. 42

RESULTS AND DISCUSSION ......................................................................................... 42

4.1     Microstructure of the investigated alloys before mechanical loading | 42   |
<p>| 4.1.1  | Morphology of second phase particles in AA2017 | 43   |
| 4.1.2  | Morphology of second phase particles in AA2024 | 45   |
| 4.1.3  | Morphology of second phase particles in AA2624 | 47   |</p>
<table>
<thead>
<tr>
<th>Section</th>
<th>Page</th>
</tr>
</thead>
<tbody>
<tr>
<td>4.2 Energy dispersive spectroscopy (EDS)</td>
<td>49</td>
</tr>
<tr>
<td>4.3 EBSD measurements of the as-received aluminum alloys</td>
<td>55</td>
</tr>
<tr>
<td>4.4 Torsion test</td>
<td>59</td>
</tr>
<tr>
<td>4.4.1 Effects of strain rate mechanical response to torsional loading</td>
<td>59</td>
</tr>
<tr>
<td>4.4.2 Effects of the temper condition on dynamic mechanical response</td>
<td>68</td>
</tr>
<tr>
<td>4.5 Microstructural evolution of the test specimens after mechanical loading</td>
<td>82</td>
</tr>
<tr>
<td>4.5.1 Deformed AA2017 alloy</td>
<td>83</td>
</tr>
<tr>
<td>4.5.2 Deformed AA2024 alloy</td>
<td>87</td>
</tr>
<tr>
<td>4.5.3 Deformed AA2624 alloy</td>
<td>92</td>
</tr>
<tr>
<td>4.5.4 Microstructure on transverse section</td>
<td>94</td>
</tr>
<tr>
<td>CHAPTER 5</td>
<td>97</td>
</tr>
<tr>
<td>CONCLUSIONS AND RECOMMENDATIONS</td>
<td>97</td>
</tr>
<tr>
<td>5.1 Summary</td>
<td>97</td>
</tr>
<tr>
<td>5.2 Conclusions</td>
<td>98</td>
</tr>
<tr>
<td>5.3 Recommendations for future work</td>
<td>98</td>
</tr>
<tr>
<td>References</td>
<td>100</td>
</tr>
<tr>
<td>APPENDIX A</td>
<td>110</td>
</tr>
<tr>
<td>COPYRIGHT PERMISSIONS</td>
<td>110</td>
</tr>
<tr>
<td>Chapter 2, Figure 2.1.</td>
<td>110</td>
</tr>
<tr>
<td>Chapter 2, Figure 2.2.</td>
<td>111</td>
</tr>
<tr>
<td>Chapter 2, Figure 2.6.</td>
<td>112</td>
</tr>
<tr>
<td>Chapter 2, Figure 2.8.</td>
<td>114</td>
</tr>
<tr>
<td>Chapter 2, Figure 2.9.</td>
<td>115</td>
</tr>
<tr>
<td>Chapter 2, Figure 2.11.</td>
<td>116</td>
</tr>
</tbody>
</table>
List of Tables

Table 2.1. Chemical composition of the investigated as-received 2000 series aluminum alloys (wt. %) [1]. .......................................................... 31
Table 2.2. Result of ICP-MS for investigated alloys (wt. %). ............................................... 31
Table 2.3. Temper designation of the as-received alloys [1]. ..................................................... 31
Table 3.1. Etching time used for each investigated alloy with different temper conditions [1]. . 38
List of Figures

Figure 2.1. Aluminum-rich corner of Al-Cu-Mg phase diagram which indicates the phase-field as a function of composition 190 °C and 500 °C [12]................................................................. 10

Figure 2.2. Aluminum-rich corner of Al-Cu phase diagram which indicates the metastable solvus boundaries of the GP zones $\theta''$, $\theta'$, and $\theta$ [13]. Note: the [20] and [21] references come with the figure................................................................. 12

Figure 2.3. Schematic of a dislocation cutting a particle. (a) a dislocation collide a particle (b) the dislocation passing through the particle (c) the dislocation cut through the particle into half [33]................................................................. 16

Figure 2.4. Schematic of looping mechanism and subsequent dislocation passages (a) a dislocation passing through two particles, (b) the dislocation loops around the particles, and (c) the distance between the particles decreased after the dislocation passage [23]............... 17

Figure 2.5. Engineering stress-strain curve. The dashed line indicates the offset yield strength. 19

Figure 2.6. typical response of a work-hardenable material to plastic deformation under different loading conditions [38]. ................................................................. 20

Figure 2.7. (a) Schematic of pure shear loading. (b) Schematic of torsional deformation [42]. .. 22

Figure 2.8. Characteristics of adiabatic shear bands in AA2219 aluminum alloy captured by optical microscope: (a) Deformed bands and (b) Transformed bands [51]. ....................... 24

Figure 2.9. Strain hardening- strain curve in materials with (a) Twinning-controlled deformation and (b) Slip-controlled deformation [57]. ................................................................. 26

Figure 2.10. (a) Highly ductile material in which the material neck down to a point (b) most common type of tensile fracture in which necking moderately occurs (c) brittle fracture in which no necking occurs. Note: Adapted from Callister [42]................................................................. 27

Figure 2.11. Diagram showing the effect of three stress states on microvoid morphology: (a) elongated dimples in the shearing direction generated by pure shear stresses in quench-hardened AISI 4340 [58]; (b) tensile stresses produce equiaxed dimples in AZ31 magnesium alloy [59]. 28

Figure 2.12. (a) Schematic of crack propagation along grain boundaries in intergranular fracture (b) Schematic of crack propagation through the interior of grains in transgranular fracture [42]. ................................................................. 29
Figure 3.1. Schematic of cylindrical specimen (a) dimensions and (b) cutting directions (ND: normal direction, RD: rolling direction, and TD: transverse direction). ........................................ 33
Figure 3.2. Schematic of torsional split Hopkinson bar (TSHB) system........................................ 35
Figure 3.3. (a) Torsional split Hopkinson bar (TSHB), (b) the loading arm and the clamp parts in the TSHB, and (c) the place in which specimen is placed between the transmitter and incident bar. .......................................................................................................................... 36
Figure 3.4. Voltage calibration linear relationship for high strain rate torsional test. ............... 36
Figure 3.5. (a) Optical microscope and (b) Scanning electron microscope (SEM) used in this study........................................................................................................................................... 39
Figure 3.6. Electron backscatter diffraction (EBSD) system....................................................... 41
Figure 4.1. Optical micrographs of the as-received alloys (a) AA2017-T451, (b) AA2024-T351, and (c) AA2624-T351 showing the morphology of second phase particles........................................ 43
Figure 4.2. Optical micrographs showing the distribution of second phase particles in the (a) AA2017-T451, (b) AA2017-T651, and (c) AA2017-O alloy before mechanical loading. ......... 44
Figure 4.3. SEM micrographs showing the distribution of second phase particles in (a) AA2017-T451, (b) AA2017-T651, and (c) AA2017-O before mechanical loading. ......................... 45
Figure 4.4. Optical micrographs showing the distribution of second phase particles in the (a) AA2024-T351, (b) AA2024-T651, and (c) AA2024-O alloy before mechanical loading. .............. 46
Figure 4.5. SEM micrographs showing the distribution of second phase particles in the (a) AA2024-T351, (b) AA2024-T651, and (c) AA2024-O alloy before mechanical loading. .............. 47
Figure 4.6. Optical micrographs showing the distribution of second phase particles in the (a) AA2624-T351, (b) AA2624-T651, and (c) AA2624-O before mechanical loading. .......... 48
Figure 4.7. SEM micrographs showing the distribution of the precipitates in the (a) AA2624-T351, (b) AA2624-T651, and (c) AA2624-O before mechanical loading. ......................... 49
Figure 4.8. EDS maps of major elements in second phase particles of all three as-received samples.......................................................................................................................... 51
Figure 4.9. Elemental mapping of second phase particles in as-received AA2017. (a) point scanning and (b) map scanning................................................................. 53
Figure 4.10. Elemental mapping of second phase particles in as-received AA2024. (a) point scanning and (b) map scanning................................................................. 54
Figure 4.11. Elemental mapping of second phase particles in as-received AA2624. (a) point scanning and (b) map scanning.

Figure 4.12. EBSD orientation maps (a, b, c), local average misorientation map (d, e, f), deformed and recrystallized fraction maps (g, h, i) obtained for AA2017-T451 (a, d, g), AA2024-T351 (b, e, h), and AA2624-T351 (c, f, i).

Figure 4.13. Grain size distribution for AA2017-T451, AA2024-T351, AA2624-T351.

Figure 4.14. Local average misorientation of all three as-received aluminum alloys.

Figure 4.15. Recrystallized, recovered, and deformed fraction of all as-received alloys.

Figure 4.16. Summary of peak flow stresses of AA2017, AA2024, AA2624 aluminum alloys in T451/T351, T651, and O temper conditions under torsional loading using $\phi=12^\circ$.

Figure 4.17. Typical shear stress-shear strain curves obtained for AA2017-T451 specimens subjected to torsional loading at $\phi=4^\circ$, $\phi=8^\circ$, and $\phi=12^\circ$.

Figure 4.18. Typical shear stress-shear strain curves obtained for AA2017-T651 specimens subjected to torsional loading at $\phi=4^\circ$, $\phi=8^\circ$, and $\phi=12^\circ$.

Figure 4.19. Typical shear stress-shear strain curves obtained for AA2017-O specimens subjected to torsional loading at $\phi=4^\circ$, $\phi=8^\circ$, and $\phi=12^\circ$.

Figure 4.20. Typical shear stress-shear strain curves obtained for AA2024-T351 specimens subjected to torsional loading at $\phi=4^\circ$, $\phi=8^\circ$, and $\phi=12^\circ$.

Figure 4.21. Typical shear stress-shear strain curves obtained for AA2024-T651 specimens subjected to torsional loading at $\phi=4^\circ$, $\phi=8^\circ$, and $\phi=12^\circ$.

Figure 4.22. Typical shear stress-shear strain curves obtained for AA2024-O specimens subjected to torsional loading at $\phi=4^\circ$, $\phi=8^\circ$, and $\phi=12^\circ$.

Figure 4.23. Typical shear stress-shear strain curves obtained for AA2624-T351 specimens subjected to torsional loading at $\phi=4^\circ$, $\phi=8^\circ$, and $\phi=12^\circ$.

Figure 4.24. Typical shear stress-shear strain curves of AA2624-T651 specimens subjected to torsional loading at $\phi=4^\circ$, $\phi=8^\circ$, and $\phi=12^\circ$.

Figure 4.25. Typical shear stress-shear strain curves obtained for AA2624-O specimens subjected to torsional loading at $\phi=4^\circ$, $\phi=8^\circ$, and $\phi=12^\circ$.

Figure 4.26. (a) shear stress-shear strain curves and (b) strain hardening rate curves of AA2017 aluminum alloy at temper conditions of T451, O, and T651 under torsional loading with $\phi=4^\circ$. 
Figure 4.27. (a) shear stress-shear strain curves and (b) strain hardening rate curves of AA2017 aluminum alloy at temper conditions of T451, O, and T651 under torsional loading with \( \phi = 8^\circ \). 70
Figure 4.28. (a) shear stress-shear strain curves and (b) strain hardening rate curves of AA2017 aluminum alloy at temper conditions of T451, O, and T651 under torsional loading with \( \phi = 12^\circ \). 71
Figure 4.29. (a) shear stress-shear strain curves and (b) strain hardening rate curves of AA2024 aluminum alloy at temper conditions of T351, O, and T651 under torsional loading with \( \phi = 4^\circ \). 73
Figure 4.30. (a) shear stress-shear strain curve and (b) strain hardening rate curve of AA2024 aluminum alloy at temper conditions of T351, O, and T651 under torsional loading with \( \phi = 8^\circ \). 74
Figure 4.31. (a) shear stress-strain curves and (b) strain hardening rate curves of AA2024 aluminum alloy at temper conditions of T351, O, and T651 under torsional loading with \( \phi = 12^\circ \). 75
Figure 4.32. (a) shear stress-shear strain curves and (b) strain hardening rate curves of AA2624 aluminum alloy at temper conditions of T351, O, and T651 under torsional loading with \( \phi = 4^\circ \). 77
Figure 4.33. (a) shear stress-shear strain curves and (b) strain hardening rate curves of AA2624 aluminum alloy at temper conditions of T351, O, and T651 under torsional loading with \( \phi = 8^\circ \). 78
Figure 4.34. shear stress-shear strain curves and (b) strain hardening rate curves of AA2624 aluminum alloy at temper conditions of T351, O, and T651 under torsional loading with \( \phi = 12^\circ \). 79
Figure 4.35. shear stress-shear strain curves of all investigated alloys in T351/T451 temper conditions using \( \phi = 12^\circ \). 80
Figure 4.36. shear stress-shear strain curves of all investigated alloys in T651 temper conditions using \( \phi = 12^\circ \). 81
Figure 4.37. shear stress-shear strain curves of all investigated alloys in Annealed (O) temper conditions using \( \phi = 12^\circ \). 82
Figure 4.38. Photograph and schematic showing sectioning of test specimens used for microstructural investigation. 82
Figure 4.39. Optical micrographs of AA2017-T451 after torsional loading using (a) \( \phi = 8^\circ \) (b) \( \phi = 12^\circ \). 83
Figure 4.40. SEM micrographs of AA2017-T451 after torsional loading using (a) \( \phi = 8^\circ \) (b) \( \phi = 12^\circ \). 84
Figure 4.41. Optical micrographs of AA2017-T651 after torsional loading using (a) $\phi=8^\circ$ (b) $\phi=12^\circ$. .......................................................... 85
Figure 4.42. SEM micrographs of AA2017-T651 after torsional loading using (a) $\phi=8^\circ$ (b) $\phi=12^\circ$. .................................................................................. 86
Figure 4.43. Optical micrographs of AA2017-O after torsional loading using (a) $\phi=8^\circ$ (b) $\phi=12^\circ$. .................................................................................. 86
Figure 4.44. SEM micrographs of AA2017-O after torsional loading using $\phi=12^\circ$ .......... 87
Figure 4.45. SEM micrographs of AA2024-T351 after torsional loading using (a) $\phi=8^\circ$ (b) $\phi=12^\circ$. .................................................................................. 88
Figure 4.46. Optical micrographs of AA2024-T651 after torsional loading using (a) $\phi=8^\circ$ (b) $\phi=12^\circ$. .................................................................................. 88
Figure 4.47. SEM micrographs of AA2024-T651 after torsional loading using (a) $\phi=8^\circ$ (b) $\phi=12^\circ$. .................................................................................. 89
Figure 4.48. SEM micrographs showing the (a) upper section (b) lower section of the broken AA2024-T651. .................................................................................. 90
Figure 4.49. SEM micrographs of fracture surface of the AA2024-T651, showing ductile shear fracture. .................................................................................. 91
Figure 4.50. SEM micrographs of AA2024-O after torsional loading using $\phi=12^\circ$ .......... 92
Figure 4.51. SEM micrographs obtained for a specimen of AA2624-T351 after torsional loading using (a) $\phi=8^\circ$ (b) $\phi=12^\circ$. .................................................................................. 93
Figure 4.52. SEM micrographs obtained for specimens of (a) AA2624-T651, and (b) AA2624-O alloys after torsional loading using $\phi=12^\circ$. .................................................................................. 94
Figure 4.53. Photograph and schematic showing sectioning of test specimens in the direction perpendicular to the axis of the cylindrical test specimen .................................................................. 95
Figure 4.54. SEM micrographs of (a)AA2017-T451, (b) AA2024-O alloys after torsional loading using $\phi=12^\circ$. .................................................................................. 96
Figure 4.55. SEM micrographs of AA2624-T651 alloy after torsional loading using $\phi=12^\circ$ .... 96
## List of Abbreviations

<table>
<thead>
<tr>
<th>Abbreviation</th>
<th>Description</th>
</tr>
</thead>
<tbody>
<tr>
<td>AA</td>
<td>Aluminum Association</td>
</tr>
<tr>
<td>ASB</td>
<td>Adiabatic shear band</td>
</tr>
<tr>
<td>DSB</td>
<td>Deformed shear band</td>
</tr>
<tr>
<td>TSB</td>
<td>Transformed shear band</td>
</tr>
<tr>
<td>GB</td>
<td>Grain boundary</td>
</tr>
<tr>
<td>OM</td>
<td>Optical microscope/microscopy</td>
</tr>
<tr>
<td>SHPB</td>
<td>Split Hopkinson pressure bar</td>
</tr>
<tr>
<td>TSHB</td>
<td>Torsional split Hopkinson bar</td>
</tr>
<tr>
<td>SEM</td>
<td>Scanning electron microscope/microscopy</td>
</tr>
<tr>
<td>TEM</td>
<td>Transmission electron microscope/microscopy</td>
</tr>
<tr>
<td>HREM</td>
<td>High resolution electron microscope</td>
</tr>
<tr>
<td>EDS</td>
<td>Energy dispersive spectroscopy</td>
</tr>
<tr>
<td>EBSD</td>
<td>Electron backscatter diffraction</td>
</tr>
<tr>
<td>SFE</td>
<td>Stacking fault energy (mJ/m²)</td>
</tr>
<tr>
<td>ALCOA</td>
<td>Aluminum company of America</td>
</tr>
<tr>
<td>GP</td>
<td>Guiner-Preston</td>
</tr>
<tr>
<td>GPB</td>
<td>Guiner-Preston-Bagaryatsky</td>
</tr>
<tr>
<td>SP</td>
<td>Second phase</td>
</tr>
<tr>
<td>DSC</td>
<td>Differential scanning calorimetry</td>
</tr>
<tr>
<td>IPF</td>
<td>Inverse pole figure</td>
</tr>
<tr>
<td>UTS</td>
<td>Ultimate tensile strength (Pa)</td>
</tr>
<tr>
<td>SSS</td>
<td>Supersaturated solid solution</td>
</tr>
<tr>
<td>OR</td>
<td>Orientation relationship</td>
</tr>
<tr>
<td>ICP-MS</td>
<td>Inductivity coupled plasma-mass spectrometry</td>
</tr>
<tr>
<td>Wt%</td>
<td>Weight percentage</td>
</tr>
<tr>
<td>SHR</td>
<td>Strain hardening rate (Pa)</td>
</tr>
</tbody>
</table>
List of Symbols

\( \tau \) \quad \text{Shear stress (Pa)}
\( G \) \quad \text{Shear modulus (Pa)}
\( r \) \quad \text{Particle radian (rad)}
\( f \) \quad \text{Volume fraction of precipitates}
\( \varepsilon \) \quad \text{Strain}
\( \sigma \) \quad \text{Applied stress (Pa)}
\( S_u \) \quad \text{Ultimate tensile strength (Pa)}
\( S_r \) \quad \text{Relative thermal softening}
\( \phi \) \quad \text{Angle of twist (°)}
\( \gamma \) \quad \text{Shear strain}
\( \rho \) \quad \text{Radius of the specimen (mm)}
\( \dot{\gamma} \) \quad \text{Shear strain rate (s}^{-1})$
\( \gamma_T \) \quad \text{Transmitted shear strain}
\( \gamma_R \) \quad \text{Reflected shear strain}
CHAPTER 1

INTRODUCTION

1.1 Overview

For several decades, the use of lightweight metals has been popular in the aerospace and automobile industries, where high-strength and high-damage tolerance and lightweight are essential. Aluminum-based alloys are just one of the examples of these popular metals due to their special characteristics. Aluminum alloys are produced in different grades (1000, 2000, 3000, 4000, 5000, 6000, 7000, and 8000) by adding a specific amount of various major and minor alloying elements [1]. The emergence of different new alloys in each series has resulted from the demand for improvement in the mechanical, physical, and other properties of existing aluminum alloys [2].

The designations of aluminum alloys include the temper condition which is indicated by a letter and a number (e.g., AA2024-T3). The letter could be F for as fabricated alloy, H for strain hardened alloys and T for heat treated alloys. Therefore, tempered condition provide information on the mechanical and thermal processing conditions that contribute to different properties in the aluminum alloys [3]. One of the most commonly used processes for strengthening aluminum alloys is precipitation hardening. This heat treatment process leads to the generation of fine precipitates in aluminum alloys, which pin down the dislocations and hinder their motion, making the alloy stronger and harder [4]. In this thesis, the effect of precipitation hardening in three different temper conditions of natural aging (T351/T451), artificial aging (T651), and O (annealing) are investigated. For an alloy to be age hardenable, some conditions must be met. There should be a significant solubility of solute atoms in the solvent, and the limit of solubility should reduce if the temperature decreases. Finally, the precipitate-matrix interface should be coherent. Only some of the aluminum alloy grades are heat-treatable (e.g., 2000, 6000, 7000, and some of the 8000 series), and the remaining grades (e.g., 1000, 3000, 4000, and 5000 grades) are not heat-treatable and are usually strengthened by cold working.
1.2 Motivation

Aluminum alloys are utilized in different parts of the moving structures like wings, fittings, gears, shafts, etc. Therefore, these alloys can be exposed to different types of loading while they are in service. High strain-rate loading is an example of such loadings that can cause premature failure. For instance, in an airplane, a collision of flying debris or a bird with an aircraft is an example of high strain-rate loading. Rotating machine axels or driving shafts can also experience high strain-rate torsional loading. These loadings can lead to catastrophic failure. Extensive research on most commercial aluminum alloys under static and dynamic impact loading conditions has been carried out. However, little research has been done on the torsional behavior of most aluminum alloys at high strain rates. To fill this knowledge gap, it is crucial to understand the deformation of these alloys and failure mechanism under high strain-rate loading in pure shear, i.e., torsion. Furthermore, with the acquired microstructural and mechanical data, it will be possible to simulate the response of the alloy to high strain-rate torsional loading. This data can also be used for microstructure design to further improve aluminum alloys against failure under high strain-rate torsional loading.

Some studies have been previously carried out on AA2017, AA2024, and AA2624 aluminum alloys at the University of Saskatchewan. This has been only conducted under compressive loading at high strain rates. In one of these studies, the effects of temper conditions on the adiabatic shear failure of AA2024 aluminum alloy under dynamic impact loading was investigated [5]. In this research, the AA2024 samples in naturally and artificially aged conditions were subjected to quasi-static compressive loading and high strain-rate dynamic impact loading. Failure was observed to start with shear strain localization leading to the formation of adiabatic shear bands (ASBs). Cracks and fractures occurred along these ASBs. The tendency for the formation of the ASBs in the naturally aged conditions was reported to be greater than for alloy samples in artificially aged conditions.

In another research at the University of Saskatchewan, AA2017 aluminum alloys in T451, T651, and O temper conditions were subjected to quasi-static compressive loading and dynamic impact loading, and the effect of temper conditions and strain rates were investigated [6]. The result indicated homogenous deformation under quasi-static loading with crack initiation at the edge of the cylindrical specimen, and their radial propagation into the specimen. At high strain rates,
thermal softening occurred, resulting in mechanical instability that caused the alloy's strain localization and adiabatic shear failure. The amount of strain-softening that occurred in the specimen with O temper condition was reported to be greater than those of specimens in T451 and T651 temper conditions.

The crash-worthiness of AA2624-T351 aluminum alloy was estimated by subjecting it to dynamic impact loading along the normal and rolling direction at the University of Saskatchewan [7]. It was observed that the samples subjected to load in the normal direction exhibited higher dynamic impact strength than those samples subjected to loading along the rolling direction. Also, it was concluded that that the dynamic mechanical strength and crash-worthiness are higher for AA2624-T351 aluminum alloys than for AA2024-T351 aluminum alloys.

These are only some examples of those numerous studies conducted on the behavior of the 2000 aluminum alloys under high strain rate and quasi-static loading. These studies demonstrate the importance of the temper conditions on the response aluminum alloys under various loading conditions. In the current study, the deformation and damage behavior of three aluminum alloys of AA2017, AA2024, and AA2624 in naturally aged, artificially aged, and annealing temper conditions under high strain rate torsional loading were investigated and discussed.

1.3 Research objectives

The major goal of this research was to investigate the deformation and behavior mechanisms in the selected 2000 series (AA2017, AA2024, and AA2624) aluminum alloys under high strain rate torsional loading. To achieve this main research goal, the following specific objectives will be realized:

1. Determine the effect of temper conditions on the torsional behavior of AA2017, AA2024, and AA2624 aluminum alloys during the high strain-rate loading in torsion.
2. Investigate the microstructural evolution in the selected alloys under high strain-rate loading in torsion.
3. Identify the failure mechanisms in the alloys at high strain rates in torsion.
1.4 Thesis arrangement

Chapter one of this thesis consists of the introduction, motivation, and objectives to provide an overview of the research topic and the expected outcomes. Chapter two contains a literature review and the results of previous studies similar to the research topic, as reported by other researchers. Chapter three includes detailed information about the material used in this study and the methodology used to realize research objectives. The experimental and test results are provided and discussed in detail in Chapter four. These include the results of mechanical tests and microstructural investigations of the alloys before and after high strain-rate torsional loading. Chapter five contains a summary of this work and the key conclusions that are drawn from research findings obtained in this study. Recommendations for future work are also provided in Chapter 5.
CHAPTER 2

LITERATURE REVIEW

2.1 Aluminum

Aluminum can be considered one of the most important nonferrous metals due to its unique properties, including good workability, lightweight, high stiffness, thermal and electrical conductivity, and corrosion resistance. These properties have broadened the usage of aluminum in different industrial sectors as transportation, packaging, electrical application, mechanical equipment, and construction. Aluminum can be fabricated using different techniques. It can be cast, forged, or rolled to any desired thickness down to very thin foil or sheet. These sheets can be stamped, spun, drawn. Aluminum has a density of 2.70 g/cm\(^3\) which is about three times lower than that of steel (7.83 g/cm\(^3\)). It means that for an equivalent volume, steel is three times heavier than aluminum. The combination of lightweight and good strength of some aluminum alloys makes aluminum very popular for aerospace and automotive application in which a strong light structure is an asset [3][8].

However, aluminum alloys have noticeable disadvantages compared to steel in different aspects. Aluminum alloys are at a significant cost disadvantage as the cost per kilo of aluminum is four to five times higher than that of carbon steel. From an engineering viewpoint, the modulus of elasticity of aluminum is about three times lower than that of steel. However, it should be noted that there has been significant improvement in the modulus of elasticity of aluminum alloys compared to pure aluminum. Also, fatigue, wear, and creep properties are relatively poor compared to steel or other structural metals [8].

Pure aluminum is soft and ductile, and it does not have an excellent strength to be utilized for its mechanical properties; rather, it is mostly used for its physical properties. Combining with alloying elements, heat treatment, and cold working provide aluminum alloys with good mechanical properties for structural applications [8].
In aluminum, oxidation does not occur the same way as in steel. When aluminum reacts with oxygen, it forms a very thin layer of aluminum oxide that resists further oxidation and serves as a corrosion-resistance barrier. This thin layer adheres strongly to the metal and has a color similar to pure aluminum and cannot be seen with naked eye. In steel, however, the oxidized layer of steel flakes off, and oxidation of steel exposed to an oxygen-containing atmosphere will proceed unabated [8].

The addition of different alloying elements influences the mechanical properties of a metal, such as hardness, ductility, corrosion resistance, ability to be heat-treated, etc. Depending on the situation and the demand for a specific mechanical property, a different proportion of these alloying elements can be added to the aluminum base metal. The vast range of mechanical properties makes the higher cost of aluminum alloys, compared to pure aluminum, more reasonable [3][8].

2.2 Aluminum alloys classification

Aluminum alloys are mainly divided into two major groups based on fabrication methods. Wrought alloy is a term given to those alloys which are shaped by mechanical forming such as rolling, extrusion, and drawing, which can be followed by machining for complex geometries [3][9]. These alloys are rolled, forged, or extruded from an ingot or billet [9]. On the other hand, cast aluminum alloys are a term given to those alloys that are shaped by casting rather than mechanical forming. They usually have low melting points and high fluidity. Cast aluminum alloys are not used for products that need further processing, such as machining or forming. They acquire their shapes by solidifying in a mold cavity having the desired shape and geometry [3][9].

As the properties of wrought and cast aluminum alloys are noticeably different, the processes needed to achieve those properties are also different. Furthermore, these two groups have their classification system, and the alloys are divided into families for simplification. This literature review will focus on only wrought aluminum alloys since the wrought alloys are investigated in this graduate research study.
2.2.1 Wrought composition families

Depending on the alloying addition, aluminum alloys are classified into 1xxx, 2xxx, 3xxx, 4xxx, 5xxx, 6xxx, 7xxx, and 8xxx family series. The 1xxx series consists of pure aluminum with no alloy addition. They contain only minor amounts of impurities which are sometimes removed for some special applications. This family of aluminum cannot be heat treated but can be strengthened slightly by strain hardening [3]. Unique characteristics for these alloys are high electrical and thermal conductivity, good workability, and excellent corrosion resistance in many environments [8].

The main alloying element in the 2xxx series aluminum alloys is copper, whose content can range between 1 and 10 wt%. Magnesium is sometimes added to these alloys as the secondary alloy addition. Copper provides high strength to this family of alloys if it is heat-treated and aged hardened. Mechanical properties of these alloys after solution heat treatment are often similar to low-carbon steel [8]. These alloys, after heat treatment, have higher yield strength, lower elongation to fracture, and lower atmospheric corrosion resistance than the pure aluminum (1xxx) and non-heat-treatable alloys. The 2xxx series alloy can be liable to intergranular corrosion. Hence, the alloys are usually coated with high purity aluminum or aluminum-magnesium-silicon alloy from the 6xxx series to protect them from corrosion [8].

The base alloying element in the 3xxx series aluminum alloys is manganese. These alloys are not heat treatable, and they do not have a high strength even when they are work hardened. However, they have excellent formability, ductility, and corrosion resistance. However, the strengths of 3xxx aluminum alloys are about 20% higher than that of pure aluminum. The optimum percentage of manganese as a major element in these series is up to 1.2%, and it is not widely used in alloys as a major alloying element for aluminum. However, it is commonly used in other series as a secondary alloying element to increase strength [8].

Silicon is the main alloying element in these 4xxx series aluminum alloy. Silicon can be efficiently added to the alloys up to 12%. Depending on the percentage of silicon and other secondary alloying constituents, these alloys can be both heat-treatable and non-heat-treatable. If there is a high percentage of silicon and a lower percentage of other alloying elements, the alloy tends to have
better flow and surface finish characteristics. Those alloys with a substantial portion of magnesium and copper besides silicon can gain a remarkable strength by age hardening [8][9]. A high percentage of silicon in alloys lowers the melting point, making the alloy suitable for the wire welding process and as brazing alloy for joining aluminum where the brazing alloy is required to have a lower melting point than the base alloy. This family of alloys doesn't have a good resistance against atmospheric corrosion, and they tend to change color with time on exposure to a humid environment [8].

The 5xxx series aluminum alloys are based on magnesium. They are not heat-treatable and can only be strengthened by work hardening. Magnesium addition gives aluminum alloys in this series excellent toughness, especially when manganese is added as a secondary alloying element. The alloys have good welding characteristics and excellent resistance against seawater corrosion, making them ideal materials for use in marine environments. The amount of magnesium in these alloys should not exceed 3% if they are to be used in temperatures higher than 65 °C where they become susceptible to stress corrosion cracking [8].

The 6xxx series of aluminum alloys are based on magnesium and silicon as the principal alloying elements. Magnesium silicide (Mg₂Si) forms in these alloys, which makes them well heat treatable. Alloys in this series possess good corrosion resistance, formability, and machinability. As a result, they are widely used for producing complex shapes via the extrusion process. However, the forming process should be done soon after quenching as this series of alloys will age naturally right after solution heat treatment. After solution heat treatment, this family of aluminum alloys is formed to T4 temper condition, and it can be strengthened by further precipitation heat treatment to develop T6 temper [8].

The main alloying element in 7xxx aluminum alloys is zinc, whose content can range between 1 and 8 percent. These alloys are usually combined with a small amount of copper and magnesium that make the alloys heat-treatable and achieve good strength. The higher strength alloys in this series have the highest strength of all commercial aluminum alloys [10]. However, these alloys with higher strength have lower stress corrosion cracking and atmosphere corrosion resistance than those with less strength. These alloys are often slightly overaged to achieve a good combination of strength, corrosion resistance, and fracture toughness [8].
This 8xxx family of aluminum alloys is produced in various chemical compositions of less frequently used major alloying elements, and their characteristics depend on the major alloying element [3]. For instance, in alloys that contain lithium (e.g., AA8090), lower density and higher stiffness can be obtained. Dispersion-strengthened AL-Fe-Ce alloys (e.g., AA8019), have excellent elevated-temperature strength [8].

2.3 Precipitation hardening of 2000 series aluminum alloys

Precipitation hardening is the most common method for strengthening 2000 series aluminum alloys. This process is done by homogenizing the alloy at a specific temperature between solidus and solvus and quenching to room temperature to obtain a supersaturated solid solution. This process, called solution heat treatment, is followed by the precipitation heat-treatment process. The supersaturated solid solution is heated to an elevated temperature within the $\alpha + \beta$ phase region. At this temperature, precipitates of the second phase will begin to form as finely dispersed particles. These particles impede the dislocation motion and result in strengthening the alloy. In the process of aging, the strength of the alloy will increase gradually with time until it reaches a peak called peak hardness, and then it will decrease subsequently with a further increase in aging time. This process is also called age hardening as the strength increases with time.

Commercial 2000 series aluminum alloys contain Cu and Mg as the major elements with the addition of minor alloying elements like Si, Zn, Mn, etc. As the concentration of these major and minor elements changes, the characteristics of the alloy and properties will differ [11][1]. Depending on the percentage of major elements, the compositions of the 2000 series aluminum alloys can be in the forms of Al-Cu or Al-Cu-Mg. The concentrations of major and minor elements in these two forms are different, and as a result, these alloys may lie within different phase fields. Figure 2.1 shows the possible phase fields that would exist during the aging process of the 2000 series aluminum alloys.
The phase fields \((\alpha + \theta)\) or \((\alpha + \theta + S)\) could be related to the aged Al-Cu alloys, while \((\alpha + S)\), \((\alpha + S + T)\) or \((\alpha + T)\) phase fields could be related to Al-Cu-Mg alloys [1]. The compositions of these phases are discussed in subsequent sections. AA2024 aluminum alloy, for example, is in \(\alpha + S\) phase-field which is widely used in airplane wing and fuselage outer skin material [11]. Precipitation in these alloys starts from the formation of metastable precipitates, which are mentioned as GP zones, and subsequently, replacement of the GP zones with \(\theta''\) and \(\theta'\) phases, which is the structural cause of hardening, to more stable precipitates [13][14].

### 2.3.1 Al-Cu alloys

Al-Cu alloys are the most studied alloy systems among the 2000 series. This alloy is widely used in structural components of engine pistons in the automotive and aerospace industry. Although the alloy comprises of only two elements, the microstructural evolution is complicated. The sequence of the precipitation depends on the degree of supersaturation and aging temperature [13]. Different authors represent the sequence of the precipitation in Al-Cu aluminum alloys as:

\[
\text{Supersaturated solid solution (SSS) \rightarrow GP I zones \rightarrow } \theta'' \rightarrow \theta' \rightarrow \theta
\]  

(2.1)

where SSS stands for the supersaturated solid solution which is obtained from the solution heat treatment, GP is an abbreviation for Guiner and Preston [15], who first observed these precipitates.
by X-ray scattering in 1938 [14], $\theta''$ (also known as GP II) is $Al_3Cu$, $\theta'$ is $Al_2Cu$, and $\theta$ is $Al_2Cu$ among which $\theta''$ and $\theta'$ are both metastable phases and $\theta$ is stable phase [13]. These GP zones are said to appear in different morphologies (needle, platelet, sphere) depending on the atomic size of the solute and solvent elements. The first original model implemented by Guiner and Preston showed GP zones as Cu-rich platelets coherent with \{100\} aluminum matrix atomic planes. Due to the smaller size of copper atoms compared to the aluminum atoms ($r_{Al} = 0.143\text{nm}$, $r_{Cu} = 0.128\text{nm}$), the GP zones' matrix planes collapse towards the Cu layers [14]. Although GP zones have been known for more than 70 years, there is still no definite conclusion on their formation, structure, evolution, and chemical composition due to the size of these precipitates, which are typically tens of nanometers small. Furthermore, different models for GP zones have been proposed. A study conducted by Gerold [16] describes GP I zones as single-layered platelets of Cu surrounded by an elastic distortion field, and this theory was accepted for a long time. However, this was followed by so many researches that focused on the morphology, surrounding strain field, and percentage of Cu of GP I zones that contained contradictory results [14]. A study conducted by Matsubara and Cohen [17], Gerold and Bubeck [18], and Auvray et al. [19] show 100% Cu atoms in the GP I zones. The study proposed by Fontaine et al. [20] suggests 50% of Cu and 50% of Al. Also, smaller content of Cu of Cu-rich planes (between 25% and 45%) was observed by Höno et al. [21]. Fujita and Lu [18] also declared new results that suggest that before the formation of GP I zones, three-dimensional clusters named "GP pre-zones" have been observed. These GP pre-zones are composed of several Cu-rich planes separated by the matrix planes, and they can evolve to GP I stage or directly to $\theta''$ stage. Figure 2.2 shows the aluminum-rich side of the phase diagram of the Al-Cu alloys, which shows the solvus boundaries of GP zones $\theta''$, $\theta'$, and $\theta$. 
Figure 2.2. Aluminum-rich corner of Al-Cu phase diagram which indicates the metastable solvus boundaries of the GP zones $\theta''$, $\theta'$, and $\theta$ [13]. Note: the [20] and [21] references come with the figure.

Complete precipitation sequence occurs when the alloy is aged at a temperature below the GP zones solvus [13][22]. In these alloys, as the aging temperature increases and exceeds the GP zones solvus line, for a specific mol fraction of Cu, the phase $\theta''$ starts to precipitate instead of GP zones.

In the aging process of Al-Cu alloys, the size of the precipitates continuously grows until the equilibrium volume fraction of second-phase particles is reached. Furthermore, larger precipitates coarsen with the growing particles (Ostwald ripening effect), resulting in losing coherency [23]. The coherency decreases as the precipitates changes to the equilibrium phase, which means that the $\theta''$ phase is more coherent to the aluminum-rich matrix than $\theta'$ and subsequently, $\theta'$ phase is more coherent than $\theta$ [6][24]. This loss of coherency between the precipitates and the Al-rich matrix will be leading to less resistance to slip as the dislocation will loop around the incoherent precipitates instead of cutting through them [23].
2.3.2 Al-Cu-Mg alloys

Al-Cu-Mg alloys are another alloy system in heat-treatable 2000 series aluminum alloys. In these alloys, the formation of the precipitates is a function of the ratio of the major alloying elements (Cu/Mg) [25]. Wang and Starink [26] reported that Bagaryatsky was the first person to propose a four-stage precipitation sequence for the age-hardening process of Al-Cu-Mg alloys. This sequence is proposed as below:

\[
\text{Supersaturated solid solution (SSS) \rightarrow GPB zones \rightarrow S''/GPB2 \rightarrow S' \rightarrow S} \quad (2.2)
\]

where clusters are predominantly Cu-Mg, GPB (also termed co-clusters or GPB I) is Guiner-Preston-Bagaryatsky, S'', and S' are both metastable phases, and S is the stable phase [24][26]. S'' is an orthorhombic phase which is coherent with the Al matrix and has a probable composition of \(Al_{10}Cu_{3}Mg_{3}\). The S phase is incoherent with the Al matrix and it is the equilibrium \(Al_{2}CuMg\) phase. S' which is considered semi-coherent with the Al matrix phase, reported to be a distorted version of the S phase so it is not considered as a separate phase [24].

Research has been shown that Cu-Mg clusters form within a short time during natural and artificial aging. Cu-Mg clusters and the S phase are known as the first and last stage of the hardening during the aging process [24]. The GPB zones and other precipitate structures before reaching the stable S phase formation are said to be the dominant precipitates for strengthening in the aging sequence. In contrast, the formation of the S-phase is accompanied by softening or over-aging conditions [27].

Ringer et al. [28][29] reported that Cu and Mg clusters form at the early stage of heat treatment (at around 190 °C), and they are believed to be precursors of the GPB zones. Bagaryatsky reported that the GPB zones are short-range ordering of Cu and Mg solute atoms. The most accepted structure for the S phase, according to some authors [24][26], is the Perlitz and Westgren (PW) model. In this model, on the basis of X-ray diffraction (XRD), S phase with a composition of \(Al_{2}CuMg\) has an orthorhombic structure with lattice parameters \(a_{S} = 0.400 \text{ nm}, b_{S} = 0.923 \text{ nm}, c_{S} = 0.714 \text{ nm}\). The structure of the S'' phase, however, has not yet been clearly described. Several structures including orthorhombic, cubic, tetragonal, and monoclinic have been observed for the
S" phase. S' phase has the same structure as the S phase but with different lattice parameters of 
\[ a_{S'} = 0.404 \text{ nm}, \quad b_{S'} = 0.925 \text{ nm}, \quad c_{S'} = 0.718 \text{ nm} \] [26]. GPB zones are believed to have an ordered structure with cylindrical shapes and diameters of 1-2 nm, and 4 nm in length which are along \( (110)_{Al} \) directions [25].

In recent years, other researchers have suggested different sequences of precipitation for Al-Cu-Mg other than Bagaryatsky's model. Gouma et al. [25] proposed a model with Cu/Mg weight ratio of 0.11 to 0.29 described as below:

Supersaturated solid solution (SSS) \( \rightarrow \) Cu/Mg clusters or GPB zones \( \rightarrow \) S" \( \rightarrow \) S' \( \rightarrow \) S \quad (2.3)

Ringer et al. [28][29] proposed the precipitation sequence model in Al-1.1Cu-1.7Mg alloy that was aged between 120 to 200 °C. They suggested that the age hardening happens firstly due to the formation of Cu and Mg co-clusters and secondly the probable nucleation of GPB zones around co-cluster sites. They could not confirm the presence of the S" phase. Therefore, they proposed their suggesting precipitation sequence as below:

Supersaturated solid solution (SSS) \( \rightarrow \) Cu/Mg co-clusters \( \rightarrow \) GPB zones \( \rightarrow \) S \quad (2.4)

In this sequence, the authors suggested the nucleation of the co-clusters represent the initial stage of age hardening, and the GPB zone is the dominant precipitate at the peak of the strengthening. At the same time, S is the phase wherein softening occurs.

In another study on the AA2024 (0.49 wt% Si) and AA2324 (0.08 wt% Si) Al-Cu-Mg alloys, Wang and Starink [26] investigated two possible variants using transmission electron microscopy (TEM) and differential scanning calorimetry (DSC). They suggested two distinct S phase precipitates, type I and type II, in the precipitation sequence of these alloys. Furthermore, by increasing the content of silicon, the evolution of the type I S phase is enhanced, but the formation of the type II S phase will be suppressed [30]. They proposed that the type II S phase is more stable than type I. These two types have the same lattice parameters with slightly different orientation relationships (OR), whereas S' and S in previous sequences were reported to have somewhat different lattice parameters with the same OR [26]. This sequence was represented as below:
Supersaturated solid solution (SSS) → clusters → S”/GPB2 → S (Type I) → S (Type II) \hspace{1cm} (2.5)

By observing precipitation sequence in AA2024 alloy during aging at 170 °C, Sha et al. [11] observed solute clusters in all stages of precipitation. They coexisted with the particles of GPB zones and the S phase. Therefore, they provided a new precipitation sequence as below:

Supersaturated solid solution (SSS) → Solute clusters → Solute clusters + GPB zone → Solute clusters + GPB zones + S → S \hspace{1cm} (2.6)

Furthermore, the possibility of the coexistence of phases in other stages of the precipitation sequence was reviewed again by Charai et al. [31]. Their work on Al-0.9%Cu-1.4%Mg alloy at the aging temperature around 200 °C with a high-resolution electron microscope (HREM) coupled with localized Fourier analysis confirmed the simultaneous coexistence of pre-precipitates (clusters and GPB zones) and larger phases like S”, S’, and S [30].

### 2.4 Dislocation-particles interactions

Based on the particles' structure and the particle-matrix interface's physics, the interface between two phases can be one of two types: coherent interface and incoherent interface [23]. In a coherent system, the matrix and second phase particle have the same crystal structure [23]. It is expected that the dislocation passing through one phase can easily move to the matrix lattice of the other phase and the atomic arrangement and the slip system are continuous. There is a small lattice misfit due to the differences between the lattice parameters, resulting in the development of elastic strain fields around the boundary of the phases [23][32]. These strain fields between the host lattice and the precipitates will slow down the dislocation motion. The strengthening contribution due to the misfit strains is found to be:

\[ \tau \propto G \varepsilon^{3/2} (rf)^{1/2} \hspace{1cm} (2.7) \]

in which $\varepsilon$ is misfit strain, $G$ is shear modulus, $r$ is particle radius, and $f$ is the volume fraction of the precipitates. At the coherent interface, dislocations cut through the particles, which result in
the shearing of the particles. However, the stress level is much higher than passing through the matrix. Figure 2.3 shows the schematic of a particle being sheared by a dislocation.

![Figure 2.3](image)

**Figure 2.3.** Schematic of a dislocation cutting a particle. (a) a dislocation collide a particle (b) the dislocation passing through the particle (c) the dislocation cut through the particle into half [33].

In an incoherent system, there is a large misfit between the interfaces as the structures of the phases are different. Furthermore, the slip systems are not continuous as it was in the coherent system [32]. Unlike a coherent system that the dislocations cut through the particles, in an incoherent system, dislocations will loop around the particles. With each passage of dislocations, the effective distance between the adjacent particles decreases (as shown in Figure 2.4), and it leads to higher stress necessary for the dislocations to loop around precipitates and subsequently lead to hardening. The relation between the stress and the distance between the particles is given as:

\[ \tau = \frac{Gb}{l} \]  

(2.8)

Where \( G \) is the shear modulus, \( b \) is the Burgers vector, and \( l \) is the distance between the particles. In some alloying systems, however, both cutting through and looping around the particles can occur simultaneously.
Figure 2.4. Schematic of looping mechanism and subsequent dislocation passages (a) a dislocation passing through two particles, (b) the dislocation loops around the particles, and (c) the distance between the particles decreased after the dislocation passage [23].

2.5 Deformation behavior of metallic materials under mechanical loading

The behavior of the metallic materials and how they respond to mechanical loading is microstructure-dependant regardless of the type of loading and test conditions. The response of metallic materials to an applied load could fall into three categories: Elastic deformation, plastic deformation, and fracture. Elastic deformation is a non-permanent deformation, and the material will return to its primary form after releasing the applied load. The applied load stretches (straining) the inter-atomic bonds that hold the atoms in position in elastic deformation. However, these bonds keep the atoms' original position and return to the original position (deformation recovery). In plastic deformation, as the applied load exceeds the elastic limit, the changes will be permanent. In this case, the inter-atomic bonds will break when the load is applied, and new bonds will be formed. Furthermore, when the load is removed, the material will not return to its original form. At the first stage of plastic deformation, strain hardening occurs due to the interaction of the dislocations, and this continues to a point where stress is maximum which is called the ultimate tensile strength (UTS) [34]. For many metallic materials, the elastic deformation can last only to the strain of about 0.005.
The term fracture refers to the separation of the material subjected to mechanical loading into two or more parts. The damage is permanent that can follow plastic deformation under mechanical loading, and it is not recoverable, unlike plastic deformation. It involves two steps of crack initiation and propagation. Stress-strain curves obtained mechanical testing of metallic specimens, indicate the mechanical properties of the specimen. The shape and magnitude of this curve depend on the composition of the specimen, applied heat treatment, strain rate, the testing temperature, and the magnitude of the applied stress. However, the general form of engineering tensile test is derived from the load-elongation measurements on the specimen. The engineering stress plotted on the vertical axis of the stress-strain curve is the average longitudinal stress in the tensile specimen. It is obtained from the division of the load subjected to the specimen ($P$) and the original cross-sectional area of the specimen ($A_0$) [35].

$$
\sigma = \frac{P}{A_0}
$$

(2.9)

The engineering strain, $\epsilon$, plotted on the horizontal axis of the stress-strain curve is the average linear strain which is obtained from the change in length divided by original length ($L_0$):

$$
\epsilon = \frac{L - L_0}{L_0}
$$

(2.10)

where $L$ is final length and $L_0$ is the original length. Stress-strain curves are usually described by different parameters: tensile strength and yield strength, which are the strength parameters, and percentage elongation and percentage reduction in area, which indicate ductility. In Figure 2.5, the region in which the stress is directly proportional (linear) is the elastic region. When the stress exceeds the value corresponding to the yield strength, the deformation will become permanent or plastic [35].

After the yield point, the stress required for further plastic deformation increases as the plastic strain increases. At this point, strain hardening begins. As the specimen continues to elongate, its cross-sectional area decreases uniformly along the gage length. During this deformation, the volume of the specimen remains constant. This change continues to a point where the stress in the
metal is maximum. This point on the engineering stress-strain curve is the maximum load divided by the original cross-sectional area, and it is called tensile strength or ultimate tensile strength (UTS):

\[ S_u = \frac{P_{\text{max}}}{A_0} \] (2.11)

After this point, the deformation is no longer uniform, and the decrease in cross-sectional area is greater than the increase in deformation load from the strain hardening. This usually happens at a point of the specimen that is slightly weaker than the other parts due to defect(s). Subsequently, further plastic deformation will concentrate at this point and result in local thinning of the specimen. This term is called necking. As the cross-sectional area decreases more rapidly than the load is increased by the strain hardening, the actual load needed to deform the specimen drops down. The engineering stress decreases subsequently until the fracture occurs.

Figure 2.5. Engineering stress-strain curve. The dashed line indicates the offset yield strength.
One of the factors influencing the shape and magnitude of the stress-strain curve during the tensile and compression test is strain rate. Strain rate is the change in the strain of the specimen with respect to time. The dynamic and fracture behavior of the specimen subjected to a high strain rate is different from that of the specimen subjected to quasi-static loading [34]. Furthermore, a specimen subjected to sudden load at a high strain rate can experience a more catastrophic failure than that of deformed at low strain rates [1]. The typical strain rate for a quasi-static loading condition in which the plastic deformation mechanism is slip and/or twinning ranged between $10^{-5}$ and $10^{-1}$ s$^{-1}$. Failure is proceed by necking for tensile test and buckling for compression test [34]. At much higher strain rates, as in dynamic shock loading (or dynamic impact loading) with strain rate around $10^3$ s$^{-1}$ or higher, plastic deformation is controlled by the combination of strain hardening effect of dislocation multiplication and thermal softening [1]. For metallic alloys subjected to a high-velocity impact, only around 10% of the projectile's kinetic energy is used for deforming the specimen. The rest of the kinetic energy will be converted to the thermal energy that increases the sample's temperature [36]. The behaviors of metallic materials to deformation under different loading conditions are described in Figure 2.6 [37].

![Image of stress-strain curve]

Figure 2.6. typical response of a work-hardenable material to plastic deformation under different loading conditions [38].
Figure 2.6 indicates that a metallic material under quasi-static loading conditions where the temperature change is negligible exhibits continuous strain hardening until failure occurs. The adiabatic curve is typical for a material subjected to dynamic impact loading that experiences temperature increase during loading. In this case, the material softens after a maximum peak stress $\gamma_{\text{max}}$ is reached. This peak is followed by strain softening. In the localization curve, if localized thermal softening increases, the capacity of carrying load decreases, and the stress collapse along the heat paths and leading to strain localization.

Regarding thermal softening, there is no specific and clear reason that leads to this phenomenon and it is seemingly not the same in different alloys [37]. In the literature, it is suggested that the flow softening variation occurs mainly due to dynamic recovery, dynamic recrystallization, or dynamic precipitate coarsening [39][40]. For example, for most Al-Zn-Mg system alloys, dynamic recovery is the balancing factor for work hardening. The dynamic precipitation and precipitate coarsening at higher temperatures have been the main reason for flow softening and drop in stress during hot deformation [37]. An investigator [41] also suggested that temperature rise in the metal during deformation, dynamic precipitation during the initial stages of straining, subgrain coarsening during straining, and texture softening are the main reasons for thermal softening. An equation for the quantitative analysis of relative softening in deformed metals as suggested by Verlinden et al. [41] is given as:

$$S_r(\%) = \left(\frac{\sigma_p - \sigma}{\sigma_p}\right) \times 100$$

(2.12)

where the $\sigma_p$ is the peak stress (MPa), $\sigma$ is the stress (MPa) at different stress levels beyond the peak stress on the true stress-strain curves.

### 2.6 Shear and Torsional test

For understanding the concept of the torsional behavior of the investigated alloys in this study, it is important to review information on the behavior of structural materials under shear loading. For performing a test using pure shear force as shown in Figure 2.7(a), the shear stress ($\tau$) calculated according to this formula:
\[ \tau = \frac{F}{A_0} \]  
(2.13)

where \( F \) is the load applied parallel to the upper and lower faces, which have an equal area of \( A_0 \). Furthermore, shear strain (\( \gamma \)) is the tangent of the strain angle \( \theta \), as shown in Figure 2.7(a).

Torsional behavior is a variation of pure shear, wherein a specimen is twisted as shown in Figure 2.7(b) [42]. Torsional forces create a rotational motion along the longitudinal axis of one end of the specimen relative to the other end. Examples of machine elements that are subjected to torsional loading include axels, driveshafts, and twist drills, among others [23][42].

Along the axis, shear stress (\( \tau \)) varies from zero to a maximum value on the surface, as given in equation 2.14.

\[ \tau = \frac{T r}{I_p} \]  
(2.14)

where \( r \) is the radius of the shaft, \( I_p \) is the polar moment of inertia of the shaft. The maximum shear strain (\( \gamma \)) on the surface of the shaft is:
\[ \gamma = \frac{r\phi}{L} \quad (2.15) \]

where \( \phi \) is the angle of twist in radian and \( L \) is the twist length of the shaft. The shear modulus \( (G) \) for the material is given by:

\[ G = \frac{\tau L}{I_p \phi} \quad (2.16) \]

### 2.7 Adiabatic Shear Band (ASB)

Failure in metallic alloys subjected to high strain-rate (around \( 10^3 \text{ s}^{-1} \) and higher) deformation usually occurs by intense strain localization along narrow paths (~ 5-100 \( \mu \text{m} \)) called Adiabatic Shear Bands (ASBs) [43]. During impact loading, near-adiabatic heating can occur that causes the heat generation along certain narrow paths that would not conduct away to the other part of the metal [44][45]. The initiation of these bands was first reported by Zener and Holloman in 1944 [46].

ASBs have been observed in various types of structural materials, including metals and alloys, composites, ceramics, when subjected to ballistic impact, forging, and high-speed machining/cutting [47][48]. Shear bands are usually undesirable and lead to intense thermal softening, loss of load-carrying capacity, and failure. The formation of ASBs results in further deformation taking place within these narrow bands [43]. The existence of these bands is believed to be due to the initial strain hardening, which is compensated by the occurrence of thermal softening. It is because of the adiabatic conditions in which the heat will not transfer during the impact process [47]. Generally, adiabatic shear bands are generated fast, and they undergo localized fast heating and rapid cooling by the surrounding matrix [49]. ASB instability occurs when thermal heating during high strain rate deformation occurs more quickly than thermal dissipation [7]. The formation of ASBs in a metallic alloy under the high strain-rate loading is believed to initiate fracture as they are potential sites for crack initiation and propagation [1].

ASB is categorized into two different types, namely deformed shear band (DSB) and transformed shear band (TSB) [7][50]. The difference between these two types is that the black-etching color
of DSBs consists of elongated and highly distorted grains. TSB consists of ultrafine subgrains which are different from the bulk materials and can be seen under the optical microscope (OM). Figure 2.8 shows the difference between these two bands.

Figure 2.8. Characteristics of adiabatic shear bands in AA2219 aluminum alloy captured by optical microscope: (a) Deformed bands and (b) Transformed bands [51].

The formation of TSB results from further development of deformed bands, and its occurrence has been attributed to phase transformation, dynamic recovery, or dynamic recrystallization [52]. White color shear bands have frequently been observed in impacted steel, classified as transformed bands, exhibiting white-etching characteristics under the optical microscope [49][50]. White etching characteristics of the transformed bands are believed to be due to fine grains formation or the occurrence of phase transformation in the bands [53]. The presence of these bands has been associated with early failure in different mechanical parts in different applications like wind turbine gearbox bearings which is one of the most frequently cited. Recent studies have shown that white etching areas are composed of refined nano-ferrite, which are supersaturated with carbon and have a hardness 30 to 50 percent higher than that of the surrounding matrix [54].

2.8 Hardening curve of metallic alloys under the dynamic shock loading

At room temperature, slip and twinning are two major deformation mechanisms that result in plasticity in metallic materials [55]. Furthermore, the stress hardening of these materials could
happen by one or both of the mechanisms depending on which mechanism is more likely to occur for that specific metallic material [1][56]. In a metallic alloy such as magnesium, in which the deformation process is controlled by twinning mechanism, twinning will be influenced by parameters like strain rate, temperature, grain size, and initial texture. It has been observed that by increasing the strain rate or decreasing the temperature, the possibility of twinning will increase [56]. The magnesium alloys' typical strain hardening rate-strain curve includes four stages, as shown in Figure 2.9(a). Stage I is the elasto-plastic regime which coincides with the onset of primary \{10\bar{1}2\} extension twinning and followed by microscopic yielding. In stage II, primary twin formation takes place. In this stage, the flow stress starts to increase with a low work hardening rate. In stage III dense secondary \{10\bar{1}2\} extension twins form, and the strain hardening rate increases as the twins act as barriers on the way of dislocation paths. Stage IV is where the strain hardening rate drops to the low of negative values and saturation of the \{10\bar{1}2\} extension twins occur. Furthermore, the hardening curve possesses a hump as the process moves from stage III to stage IV [1][57].

In FCC structure alloys like aluminum alloys, where the dominant plastic deformation mechanism is mostly slip, the strain hardening rate-strain curve does not have any hump, as shown in Figure 2.9(b). It is suggested that the hump-free curve is due to the absence of twinning activity in the deformed alloy. Consequently, it implies that in alloys that the deformation process has a slip-controlled mechanism, there is a lack of dense secondary \{10\bar{1}2\} extension twins formation in stage III. Stage I, where the hardening rate is decreasing, is the elasto-plastic regime, the region of dynamic recovery in high SFE metals where the deformation process accommodates by slip [56]. Stage IV is where the thermal softening becomes dominant over strain-hardening, resulting in a drop in the strain hardening rate. In this region, the hardening curve can drop to below zero [1].
2.9 Fracture

Fracture occurs when a specimen is separated into two or more pieces due to the application of mechanical load, including quasi-static loading, high strain rate loading, cyclic loading (fatigue), and creep (loading at high temperature). For metals, a fracture is classified into two types: ductile and brittle fracture. Depending on the characteristic of the material (brittle or ductile), the type of stress (tensile, compressive, torsion, creep), the rate of loading (strain rate), and temperature, the type of fracture can vary. Generally, this classification is based on the ability of the material to experience plastic deformation before fracture. Ductile metals usually show a noticeable amount of plastic deformation with high energy absorption before failure. Brittle materials, however, exhibit little or no plastic deformation with low energy absorption before fracture.

The fracture process involves two steps in response to stress: crack initiation and crack propagation. Ductile fracture shows extensive plastic deformation in the vicinity of the advancing crack. In ductile fracture, the crack length extends slowly such that it takes some time before leading to fracture. These cracks are usually stable until the stress imposed on the material extends the crack to the extent that the effective cross-sectional area can no longer carry the load. On the other hand, brittle fracture shows almost no plastic deformation and the crack propagates rapidly. Cracks in brittle material are extremely unstable, and once the crack is initiated, it propagates spontaneously without any increase in the applied load.

Figure 2.9. Strain hardening-strain curve in materials with (a) Twinning-controlled deformation and (b) Slip-controlled deformation [57].
2.9.1 Ductile fracture

In ductile fracture mode, there are voids inside the material, or they will nucleate as the material starts to deform. The nucleation occurs by particle cracking or failure of the interface between the matrix and a precipitate. The voids will grow until they link together or coalesce to form an elliptical crack path [42]. In ductile materials, necking usually occurs before fracture. In the most common form of tensile fracture for ductile metals, the fracture is preceded by a moderate necking. Highly ductile metals, neck down to a point before a fracture occurs. On the other hand, there would be no necking during the brittle fracture process. Figure 2.10 shows the schematic of the plastic deformation of three types of materials.

![Diagram of plastic deformation of materials](image)

Figure 2.10. (a) Highly ductile material in which the material neck down to a point (b) most common type of tensile fracture in which necking moderately occurs (c) brittle fracture in which no necking occurs. Note: Adapted from Callister [42].

In tension, ductile fracture initiates at the center of the material resulting in a cup and cone fracture. In this type of fracture, the central interior region of the surface is covered by segments of voids called dimples, a fibrous and irregular appearance that expresses plastic deformation in ductile fracture [42].
The geometry of the dimples depends on the type of stress applied to the specimen. If the specimen is subjected to shear stress, the dimples would be elongated in the direction of the shear stresses and have a parabolic shape. In pure shear stress, C-shaped elongated dimples would be in the opposite direction on two matching surfaces (Figure 2.11(a)). For uniaxial tensile loading, the dimples will have spherical shapes, and each dimple would be one-half of the microvoid that formed and separated during the fracture (Figure 2.11(b)). It has been observed that the fracture energy increases with increasing the depth and width of the dimples [23]. On the other hand, if the specimen is subjected to shear stress, the dimples would be elongated in the direction of the shear stresses and have a parabolic shape. In pure shear stress, C-shaped elongated dimples would be in the opposite direction on two matching surfaces (Figure 2.11(b)).

Figure 2.11. Diagram showing the effect of three stress states on microvoid morphology: (a) elongated dimples in the shearing direction generated by pure shear stresses in quench-hardened AISI 4340 [58]; (b) tensile stresses produce equiaxed dimples in AZ31 magnesium alloy [59].

2.9.2 Brittle fracture

A brittle fracture occurs with very little deformation. Crack propagation proceeds in brittle materials rapidly and can occur even with no further increase in stress. In this type of fracture, the direction of the crack is almost perpendicular to the stress axis, and the fractured surface is relatively flat, as shown in Figure 2.10(c). The brittle fracture may leave distinctive patterns on the fracture surface. These patterns could be V-shaped marking, called chevron, near the center of the fracture that points back to the crack origin or fan-like lines and ridges that radiate from the initiation of the crack. The lines are usually coarse enough to be seen with the naked eye. However,
in very hard and fine-grained metals, the lines are not discernible [42]. Crack propagation in low-energy brittle fracture may occur in two forms. There are intergranular fracture and transgranular that is also called cleavage fracture (Figure 2.12)

For intergranular fracture, the crack propagates along the grain boundaries and the fracture plane exposes grain boundary surface and a faceted morphology [23]. There are different processes that may result in intergranular fracture. For example, initiation and coalescence of microvoids at inclusions or second phase particles located along grain boundaries [23][42].

In ferritic steels, intergranular fracture occurs when the segregation of impurities like carbon reinforces grain boundaries. Furthermore, by segregating other impurities like phosphorus, the fracture mode can be changed from cleavage to intergranular. This phenomenon is called temper embrittlement [60].

In cleavage or transgranular fracture, the crack propagates along specific crystallographic planes. Transgranular fracture commonly occurs in specific BCC and HCP metals. It can also occur in FCC metals when subjected to severe environmental conditions, high strain rates, or very low temperatures [23]. The fracture surface in cleavage is usually flat, with river pattern flows in the
direction of the crack. The process of cleavage fracture consists of three steps: 1. Plastic deformation, 2. Crack initiation, and 3. Crack propagation [42]. Cleavage fracture could form different fracture surfaces when they cross sub-boundaries, twin boundaries, and grain boundaries [60]. River patterns may be generated if a crack propagates across a grain boundary and the advancing crack is reoriented in search of weak cleavage planes in new grain. The direction of the river patterns aligns with the direction of the cleavage cracks propagation. Creating the cleavage steps is another possibility if the crack overpasses a low angle twist boundary and the cleavage crack intersect with screw dislocations [23][60].

2.10 Elemental composition of the investigated 2000 series aluminum alloys

The materials selected for investigation in this study were AA2017, AA2024, and AA2624 aluminum alloys received in naturally aged conditions. The AA2017 alloy is received in T451 temper condition while AA2024 and AA2624 are supplied in T351 temper. Table 2.1 provides the range of elemental composition of the AA2017 and AA2024 samples obtained from literature sources [1]. The composition of the AA2624 alloy was obtained from the manufacturer, ALCOA Canada. Table 2.2 provides the chemical compositions of the alloys, as obtained from chemical analysis using inductivity coupled plasma-mass spectrometry (ICP-MS) at the Department of Geology, University of Saskatchewan by a former students, who worked on the alloy in our research group. It shows that the major alloying additions to aluminum base metal are Cu, Mg, Mn, and Fe for the three alloys. In addition, AA2017 contains a significant amount of Zn. The heat treatment processes that were used to achieve the T351 and T451 temper conditions are provided in Table 2.3 [1].
Table 2.1. Chemical composition of the investigated as-received 2000 series aluminum alloys (wt. %) [1].

<table>
<thead>
<tr>
<th></th>
<th>Al</th>
<th>Cu</th>
<th>Mg</th>
<th>Mn</th>
<th>Fe</th>
<th>Si</th>
<th>Zn</th>
<th>Ti</th>
<th>Cr</th>
<th>Others</th>
</tr>
</thead>
<tbody>
<tr>
<td>AA2017</td>
<td>Bal</td>
<td>3.5-</td>
<td>0.4-</td>
<td>0.4-</td>
<td>≤0.7</td>
<td>0.2-</td>
<td>≤0.25</td>
<td>≤0.15</td>
<td>≤0.10</td>
<td>≤0.15</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>4.5</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>0.8</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>1.0</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>0.8</td>
</tr>
<tr>
<td>AA2024</td>
<td>Bal</td>
<td>3.8-</td>
<td>1.2-</td>
<td>0.3-</td>
<td>≤0.5</td>
<td>≤0.5</td>
<td>≤0.25</td>
<td>≤0.15</td>
<td>≤0.10</td>
<td>≤0.15</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>4.9</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>1.8</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>0.9</td>
</tr>
<tr>
<td>AA2624</td>
<td>Bal</td>
<td>4.1</td>
<td>1.4</td>
<td>0.6</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
</tbody>
</table>

Table 2.2. Result of ICP-MS for investigated alloys (wt. %).

<table>
<thead>
<tr>
<th></th>
<th>Al</th>
<th>Cu</th>
<th>Mg</th>
<th>Mn</th>
<th>Fe</th>
<th>Zn</th>
<th>Ti</th>
<th>Cr</th>
<th>Si+others</th>
</tr>
</thead>
<tbody>
<tr>
<td>AA2017</td>
<td>90.57</td>
<td>3.39</td>
<td>0.57</td>
<td>0.58</td>
<td>0.19</td>
<td>0.14</td>
<td>0.01</td>
<td>0.03</td>
<td>4.50</td>
</tr>
<tr>
<td>AA2024</td>
<td>91.26</td>
<td>3.90</td>
<td>1.34</td>
<td>0.60</td>
<td>0.22</td>
<td>0.01</td>
<td>0.02</td>
<td>0.02</td>
<td>2.62</td>
</tr>
<tr>
<td>AA2624</td>
<td>93.79</td>
<td>3.44</td>
<td>1.24</td>
<td>0.53</td>
<td>0.06</td>
<td>0.00</td>
<td>0.02</td>
<td>0.00</td>
<td>0.92</td>
</tr>
</tbody>
</table>

Table 2.3. Temper designation of the as-received alloys [1].

<table>
<thead>
<tr>
<th>Temper designation</th>
<th>Heat treatment procedure</th>
<th>Product forms</th>
</tr>
</thead>
<tbody>
<tr>
<td>T351</td>
<td>Solution heat treatment followed by a stress relief by stretching to about 1-3% for rod/bar or 1.5-3 % for plate finished bar before naturally aged</td>
<td>Plate; rolled or cold-finished bar</td>
</tr>
</tbody>
</table>
As can be concluded from Tables 2.1 and 2.2, all the three as-received alloys contain Cu, Mg, and other elements. However, in AA2017, the ratio of Cu to Mg is nearly 6, whereas the ratio in AA2024 and AA2624 is around 2.8. Furthermore, due to the low content of Mg, AA2017 is considered an Al-Cu alloy, while AA2024 and AA2624 are considered as Al-Cu-Mg alloys because of the higher amount of magnesium [24].
CHAPTER 3

MATERIALS AND METHODOLOGY

3.1 Materials

In this thesis, three aluminum alloys from the 2000 series were selected for investigation. These test specimens are AA2017, AA2024, and AA20624 aluminum alloys. The alloy compositions are provided in Section 10 of Chapter 2. Regarding the geometry and dimensions, all the specimens were machined along the rolling direction into a hollow cylindrical sample, as depicted in Figure 3.1. The specimens have an outer diameter of 13.8 mm and an inner diameter of 13 mm with a wall thickness of 0.4 mm and a gauge length of 3.8 mm.

![Diagram of cylindrical specimen](image)

Figure 3.1. Schematic of cylindrical specimen (a) dimensions and (b) cutting directions (ND: normal direction, RD: rolling direction, and TD: transverse direction).

3.2 Heat treatment procedure

Some of the machined as-received specimens were heat-treated to T651 temper condition and some into O temper condition to investigate the effect of temper conditions on mechanical response to high strain rate under torsional loading. The as-received alloys AA2017-T451 and AA2624-T351 were artificially aged at 100 °C for 10 hours, followed by air-cooling to obtain a temper condition of T651. Also, alloy AA2024-T351 was artificially aged at 191 °C for 10 hours followed by air-cooling to obtain the same temper condition. For O temper condition, which is
annealing, all specimens were heated to 413 °C for three hours, followed by a two-step cooling process. Firstly, the specimens were slow cooled at approximately 30 °C per 30 minutes to 260 °C and then air-cooled to ambient temperature.

3.3 Torsion test

High strain-rate torsion tests were done using a torsional split Hopkinson bar (TSHB) equipment, consisting of an incident bar, a transmitter bar, and a clamp with a torque discharging mechanism, as shown in Figures 3.2 and 3.3. The procedure of the TSHB started with placing the specimens between the two bars. The loading end of the incident bar was twisted by applying torque which was stored between the loading bar end and the clamp [43]. When the clamp was opened, the torque was released, generating elastic waves along the incident bar, rapidly deforming the specimen. The loading arm was subjected to torsional loading at different angles of twist to create different strain rates in the specimens. The waves were captured by the strain gauges attached to the bars, which were in turn connected to a signal conditional and amplification system. The connected digital oscilloscope stores the signals that were imported to a computer for further analysis.

The elastic waves were stored in the oscilloscope as voltage and time data. These data needed to be converted to elastic strain data for further analysis. This was done by obtaining a conversion factor by calibrating the equipment prior to the major mechanical tests. Calibrating was done by subjecting a solid specimen to high strain rate torsional loading at angles of twist of 0, 2, 4,...,20 degrees (intervals of 2 degrees) while the end of the transmitter bar was pinned down. During this process, the generated voltages corresponding to each angle of twist were captured. By using equation 3.1, the shear strain for each point was calculated.

\[ \gamma = \frac{\rho \phi}{L} \]  

(3.1)

where \( \rho \) is the radius of the specimen, \( \phi \) is the angle of twist (in radian), and \( L \) is the total length from the loading arm to the end of the transmitter bar. Two linear plots of strain versus voltage were made for incident and transmitter bars by calculating the shear strain for each point. Linear relations were further used to calculate the shear strain corresponding to voltage recorded by the
oscilloscope. Figure 3.4 shows the linear relation between the shear strain and voltage generated in the incident bar.

The calculated shear strains are then used in obtaining shear stress, strain, and strain rate data using the equations below [61]:

\[
\tau = \frac{GD^3}{8D_s^2t_s} \gamma_T
\]  

(3.2)

\[
\gamma = \frac{2CD_s}{L_sD} \int_0^t \gamma_R \, dt
\]  

(3.3)

\[
\dot{\gamma} = \frac{2CD_s}{L_sD} \gamma_R
\]  

(3.4)

where \(D\), \(D_s\), and \(t_s\) are the diameter of the input/output bar, mean diameter of the thin wall of the specimen, and wall thickness of the specimen, respectively; \(G\), \(C\), and \(L_s\) are shear modulus of bar material, speed of wave sound in the bar, and the gage length of the specimen, respectively; and \(\gamma_T\) and \(\gamma_R\) are respectively transmitted and reflected strain pulses.

Figure 3.2. Schematic of torsional split Hopkinson bar (TSHB) system.
Figure 3.3. (a) Torsional split Hopkinson bar (TSHB), (b) the loading arm and the clamp parts in the TSHB, and (c) the place in which specimen is placed between the transmitter and incident bar.

Figure 3.4. Voltage calibration linear relationship for high strain rate torsional test.

After the system preparation and calibration, specimens of the alloys in the three different temper conditions (O, T351, or T451 and T651) were subjected to high strain rate tests. This mechanical testing was done using three angles of twists of 4, 8, and 12 degrees. By increasing the angle of
twist, the strain rate increases, and it is possible to observe the damage evolution in the samples for each temper condition as the strain rate changes. Each test was repeated thrice, and the reported shear stress-shear strain curves are averages of the three tests. Finally, the stress, strain, and strain rate data were calculated and used in generating shear stress-shear strain curves and strain hardening curves.

3.4 Microstructural investigation

Microstructural evaluation of the samples started with metallographic sample preparation. The specimens intended to be observed before the torsion loading process were cut from the aluminum alloy bars and cold mounted. The machined specimens, which were subjected to torsional loading, were cut into two halves. The cutting process was done using a low-speed precision cutting machine so that no further additional stresses were applied to the specimens during cutting. The cutting process was along longitudinal and transverse directions to observe the deformation on both transverse and longitudinal sections. Finally, all cut specimens were cold mounted.

The cold mounting process was conducted using acrylic resin, a mixture of VersoCit-2 powder, and liquid. The cold mounting of the specimens was followed by pre-grinding and fine grinding before polishing (diamond and oxide) to get a mirror-like surface finish for microscopy. Pre-grinding was done with 320 and 500 grades SiC emery papers. For fine grinding, a 9 μm MD-Largo cloth with a 9 μm MD-Largo suspension was used. Lastly, the diamond polishing was done with 3 μm MD-Mol with 3 μm MD-Mol suspension and 1 μm MD-Nap clothes with 1 μm MD-Nap suspension. Regarding oxide polishing, 0.04 μm OP-Chem cloth with 0.04 μm oxide polishing suspension (OPS) was used. OPS is a colloidal silica suspension. Etching before optical and scanning microscopy helps the specimen to reveal the microstructure of interest. The etching was done using a chemical consisting of 25 ml methanol, 25 ml HNO₃, 25 ml HCL, and one drop of HF. The etching time for each specimen depends on the temper condition of each alloy. Table 3.1 shows the etching time used for each alloy [1].
Table 3.1. Etching time used for each investigated alloy with different temper conditions [1].

<table>
<thead>
<tr>
<th>Alloy</th>
<th>AA2017</th>
<th>AA2024</th>
<th>AA2624</th>
</tr>
</thead>
<tbody>
<tr>
<td>Temper</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>condition</td>
<td>T451</td>
<td>T651</td>
<td>O</td>
</tr>
<tr>
<td>T351</td>
<td>T651</td>
<td>T5</td>
<td>O</td>
</tr>
<tr>
<td>T351</td>
<td>T651</td>
<td>O</td>
<td></td>
</tr>
<tr>
<td>Etching</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>time (s)</td>
<td>50</td>
<td>50</td>
<td>10-15</td>
</tr>
<tr>
<td></td>
<td>50</td>
<td>10-15</td>
<td>5</td>
</tr>
<tr>
<td></td>
<td>50</td>
<td>50</td>
<td>5</td>
</tr>
</tbody>
</table>

3.4.1 Optical microscopy and scanning electron microscopy

After the etching, the microstructure of the samples was investigated using optical microscopy (OM) and scanning electron microscopy (SEM). For optical microscopy, a Nikon MA100 inverted microscope with a PAX-it! image analysis system (Figure 3.5(a)) was used, and for scanning electron microscopy, a JEOL-JSM6010LV scanning electron microscope (Figure 3.5(b)) with the accelerated voltage of 20 kV was used.
Figure 3.5. (a) Optical microscope and (b) Scanning electron microscope (SEM) used in this study.

With OM, it is possible to observe the structure of second phase particles in the investigated alloys. However, the pictures taken by an optical microscope do not offer the high magnification needed for a detailed understanding of the morphology of the second phase particles. For better observation of second phase particles, SEM was used after OM, in which the shape and color difference of the second phase particles could be more distinctly observed. Under the optical microscope, micrographs were taken using three magnifications (100x, 500x, and 1000x). The observation was conducted in various parts of the deformed samples in search of the changes in precipitate regimes. The SEM was used to investigate the samples at higher magnifications (up to 10000x).
3.4.2 Electron backscatter diffraction (EBSD) and energy dispersive spectroscopy (EDS)

EBSD measurements were conducted to obtain detailed information about crystallographic orientation, local misorientation, deformed and recrystallized fraction of the samples. The tests were conducted on a plane perpendicular to the axis of the as-received AA2017 and AA2024 aluminum alloy rods, and a plane perpendicular to the rolling direction of the AA2624 aluminum alloy plate samples in the naturally aged conditions before mechanical loading. A Hitachi SU6600 field emission scanning electron microscope equipped with an Oxford Instruments HKLNordlys EBSD detector (Figure 3.6) and an energy dispersive spectroscopy (EDS) detector was used for observation of the texture of the investigated aluminum alloys. This machine also coupled with the AZTEC 2.0 data acquisition software to acquire the electron diffraction patterns. Electron microscopy was done at an accelerating voltage of 15 kV, and scanning was carried out with a binning of 8 x 8 pixels. Furthermore, all micrographs were processed with the Oxford Instruments Channel 5 post-processing software. By post-processing of the acquired micrographs, crystallographic orientation maps, local average misorientation data, and maps, recrystallized and deformed fraction data and maps, grain size distribution data were captured. EBSD Micrographs were taken at a low magnification of 200x to include a large area of the specimens. Each micrograph were captured in a process lasting for more than 12 hours.

Energy dispersive spectroscopy was done with the same machine on the same as-received alloys to obtain the data regarding the composition of the second phase particles observed with scanning electron microscopy. Analysis was conducted on a plane perpendicular to the axis of the AA2017 and AA2024 aluminum alloys rods and perpendicular to the rolling direction of the AA2624 plate. This analysis was done at low magnification of 200x to include a large area of the alloys. The data was captured in the form of Kα maps and a table of weight percentages for each element.
Figure 3.6. Electron backscatter diffraction (EBSD) system.
CHAPTER 4

RESULTS AND DISCUSSION

In this chapter, the results obtained from the experimental investigation are provided and discussed. Firstly, the microstructures of the AA2017, AA2024, and AA2624 aluminum alloys at different temper conditions of naturally aged, artificially aged, and annealed before mechanical loading are presented and discussed. The second subsection includes shear stress-shear strain and strain hardening rate curves obtained from TSHB tests. These results include the shear flow stress, shear strain, and strain rate generated during the mechanical test. Lastly, the microstructural evolution in the investigated alloy during the high strain-rate torsional loading is discussed.

4.1 Microstructure of the investigated alloys before mechanical loading

Figure 4.1 shows typical optical micrographs obtained for the polished and etched as-received specimens before mechanical loading. These micrographs were taken on the plane perpendicular to the rolling direction of the investigated alloys. Although the grain structures of all samples can be observed, the grain boundaries could not be easily identified. The HF-containing etchant preferentially attacked the second phase particles and not the grain boundaries for the optimum etching time. As observed in Figure 4.1a-c, spherical and irregularly shaped second phase particles are finely dispersed within the microstructure of AA2624-T351 aluminum alloy, whereas only coarse second phase particles with irregular shapes are observed in the AA2024 alloy. In AA2017 alloy, a combination of the fine and coarse second phase particles can be observed. The difference in the grain structure of the alloys could be because of the difference in pre-machining processes as the AA2017-T451 and AA2024-T351 were rolled into a rod, whereas AA2624-T351 was priorly rolled into a plate. These particles appeared in different colors, shapes, and sizes. SEM was used to observe the second phase particles in greater detail. The ultrafine precipitates that are formed during solution heat treatment or precipitation heat treatment and contribute to strengthening cannot be observed under an OM or SEM. These precipitates are so small that they can only be investigated using high-resolution electron microscope (HREM) or transmission electron microscope.
4.1.1 Morphology of second phase particles in AA2017

The microstructure of the AA2017 aluminum alloys in T451, T651, and O temper conditions before mechanical loading are provided in Figures 4.2 and 4.3. Optical microscopy and SEM have been used to produce these micrographs, respectively. Fine second phase particles are dispersed in the aluminum matrix in all the three investigated temper conditions. The size of the second phase particles seems to be bigger in AA2017-T451 and AA2017-T651 compared to AA2017-O (Figure 4.2). AA2017 is an Al-Cu aluminum alloy with a very low amount of Mg that is strengthened by the formation of $\text{Al}_2\text{Cu}$, $\text{Al}_3\text{Cu}$ precipitates following $\text{GP} \rightarrow \theta'' \rightarrow \theta' \rightarrow \theta$ sequence during precipitation heat treatment process [62][24]. The addition of Mg as a minor alloying element can result in the formation of the $\text{Al}_2\text{CuMg}$ ($\text{S}$) phase which can co-exist with $\text{Al}_2\text{Cu}$ ($\theta$) [1].

The result of the scanning electron microscopic study on the AA2017 samples (Figure 4.3) revealed semi-spherical shapes and irregularly shaped particles within the matrix of the alloy in T451 temper. More particle clustering is observed in the AA2017-T451 compared to the AA2017-T651 and AA2017-O. Tiny white particles were observed in the microstructure of AA2017-O alloy under the scanning electron microscope. These tiny white particles could not be observed under the optical microscope. They have a needle-like shape and are dispersed within the aluminum matrix in AA2017-O alloy (Figure 4.3(c)). In the AA2017-T451 and AA2017-T651 alloys, however, these particles were not seen. That would be because they probably dissolved within the aluminum matrix during aging treatment (Figure 4.3a, b). Energy dispersive spectroscopy (EDS) technique is needed for detecting the
composition of these particles. Rezaei et al. [63], in a study on 6061 aluminum alloy, investigated these white particles as Si-rich precipitates by using energy dispersive spectroscopy (EDS).

Figure 4.2. Optical micrographs showing the distribution of second phase particles in the (a) AA2017-T451, (b) AA2017-T651, and (c) AA2017-O alloy before mechanical loading.
Figure 4.3. SEM micrographs showing the distribution of second phase particles in (a) AA2017-T451, (b) AA2017-T651, and (c) AA2017-O before mechanical loading.

### 4.1.2 Morphology of second phase particles in AA2024

The optical and SEM micrographs showing the distribution of the second phase particles in AA2024 aluminum alloys before mechanical loading are provided in Figures 4.4 and 4.5. These micrographs indicate their distribution for all three annealed, naturally, and artificially aged AA2024 aluminum alloys. AA2024 is considered an AL-Cu-Mg aluminum alloy due to its higher magnesium content compared to AA2017 aluminum alloy. In these types of alloys, aluminum combines with the alloying elements, which are copper and magnesium, and form phases like
Al_2CuMg and Al_{10}Cu_3Mg_3 which are referred to as S and S” phases, respectively [24]. These are the phases that are formed during the age-hardening of this alloy and contribute to hardening. Optical micrographs of AA2024 aluminum alloy show that in AA2024-T351, very coarse second phase particles with some irregularly shaped grey particles are widely distributed within the continuous aluminum-rich phase matrix (Figure 4.4(a)). In the artificially aged AA2024, the second phase particles appear to be finer (Figure 4.4(b)). Although second phase particles clustered together in AA2024-O (Figure 4.4(c)), AA2024-T351 exhibits more clustering of the second phase particles.

A more detailed investigation using scanning electron microscopy (Figure 4.5) shows coarse, second phase particles in AA2024-T351 clustered together and formed irregularly shaped particles. In contrast, in AA2024-T651, second phase particles were observed to have needle-like shapes with less clustering of particles. In AA2024-O aluminum alloy, needle-like second phase particles clustered together and formed irregularly shaped particles. Grain boundaries are observed on SEM micrographs of the annealed alloy as the second phase particles are arranged along the grain boundaries. In a previous research by Tiamiyu [1] AA2024-O aluminum alloy, the formation of the second phase particles along the grain boundaries was also detected. It was pointed out that the precipitation along grain boundaries substantially increased in the annealed AA2017, AA2024, and AA2624 aluminum alloys after high strain-rate impact loading.

Figure 4.4. Optical micrographs showing the distribution of second phase particles in the (a) AA2024-T351, (b) AA2024-T651, and (c) AA2024-O alloy before mechanical loading.
Figure 4.5. SEM micrographs showing the distribution of second phase particles in the (a) AA2024-T351, (b) AA2024-T651, and (c) AA2024-O alloy before mechanical loading.

4.1.3 Morphology of second phase particles in AA2624

The micrographs of AA2624 aluminum alloys in different temper conditions of T351, T651, and O before mechanical loading are provided in Figures 4.6 and 4.7. Both optical and scanning electron microscopes were used to obtain these micrographs. AA20624, just like AA2024, is an Al-Cu-Mg alloy that can be strengthened by forming the S” and S’ phases, which are coherent and semi-coherent phases that form during age hardening processes. Figures 4.6a and 4.6b show that finely dispersed second phase particles are widely dispersed in the continuous aluminum-rich
phase in AA2624-T351 and AA2624-T651. In the annealed AA2624 alloy, second phase particles clustered together and formed irregularly shaped big particles. A high resolution imaging with the scanning electron microscope (Figure 4.7(a)) reveals that finely dispersed spherical particles exist in AA2624-T351. These particles partly clustered together and formed irregularly shaped bigger particles. In AA2624-T651 (Figure 4.7(b)), second phase particles are coarser, and more irregularly shaped clusters are observed. Furthermore, irregularly shaped particles clustering along the grain boundaries were also observed in the annealed AA2624-O sample (Figure 4.7(c)). In the age-hardened AA2624 alloys, clustering of second-phase particles around the grain boundaries was not observed. The presence of grain boundary precipitates in all annealed samples gives the impression that it could be possible that these precipitates contribute to weakening in annealed specimens and subsequently strength decrease [64].

![Figure 4.6. Optical micrographs showing the distribution of second phase particles in the (a) AA2624-T351, (b) AA2624-T651, and (c) AA2624-O alloy before mechanical loading.](image-url)
Figure 4.7. SEM micrographs showing the distribution of the precipitates in the (a) AA2624-T351, (b) AA2624-T651, and (c) AA2624-O before mechanical loading.

4.2 Energy dispersive spectroscopy (EDS)

The compositional analysis results of the second phase particles in the as-received AA2017-T451, AA2024-T351, and AA2624-T351 using energy dispersive spectroscopy (EDS) are presented in Figure 4.8. Considering the Al Kα1 map of all three samples, the aluminum-rich phase is in red color, and the dark shape features are either second phase particles or cavities created in the sample by chemical attack of second phase particles by the etchant. As it is previously observed in the micrographs captured by scanning electron microscope in Figures 4.3(a), 4.5(a), and 4.7(a), second
phase particles are more coarse in naturally aged AA2024 alloy compared to the other two alloy with the same temper condition. Similarly, dark features, which represent second phase particles in Al K\(\alpha\)1 maps, were observed more in AA2024-T351 compared to the other two alloys.

The Cu L\(\alpha\)1-2 analysis indicates the amount of copper element, which is the major alloying element for all three investigated alloys. The grey irregular-shaped particles rich in Cu appeared as shiny particles in Cu L\(\alpha\)1-2 maps. From these maps, second phase particles in AA2024-T351 have a higher concentration of shiny particles, which have more Cu content compared to the AA2017-T451 and AA20624-T351 alloys. Furthermore, in Table 2.1, a higher percentage of the Cu has been reported for AA2024 alloy compared to the other investigated alloys with 3.8 to 4.9 weight percentage.

The Mg K\(\alpha\)1-2 maps show the intensity of the magnesium within the investigated alloys. Second phase particles with less concentration of Mg element have been indicated in a darker color in these maps. In the AA2017-T451 Mg K\(\alpha\)1-2 map, which is an Al-Cu alloy with much less content of Mg element, second phase particles appeared in darker colors compared to the AA2024-T351 and AA2624-T351, which are Al-Cu-Mg alloys with more content of Mg. Table 2.1 also shows lower content of Mg for AA2017-T451 compared to the other investigated samples.

In Mn K\(\alpha\)1 and Fe K\(\alpha\)1 maps, both Mn-containing and Fe-containing particles were observed in brighter colors within the matrix of the AA2017-T451 and AA2024-T35 alloys. However, in AA2624-T351 alloy, those particles became dimmer. It could be due to the lower content of these elements in AA2625-T351 alloy. Tables 2.1 and 2.2 also show less Mn and almost no Fe in the AA2624-T351 alloy. It has been observed that the presence of iron-rich precipitates in artificially aged AA2017-T651 may result in the formation of \(Al_3Fe\) or \(Al_{12}(Fe,Mn)_3Si\) which co-exist with \(Al_2Cu(\theta)\) and \(Al_2CuMg\) (S) equilibrium phases [65].

Si was also observed in the Si K\(\alpha\)1 EDS maps as shiny particles, and these were mostly seen in the AA2017-T451 and AA2024-T351, in which higher content of Si element was recorded in Tables 2.1 and 2.2. It has been reported that the addition of the Si element increases the formation of clusters during natural aging [66]. Furthermore, the alloy with higher content of Si is expected
to have a large number of clusters and subsequently higher yield strength, tensile strength, and instantaneous strain hardening rate (SHR). Therefore, it is expected the Si-content in AA2017-T451 and AA2024-T351 improves yield strength and flow stress.

Figure 4.8. EDS maps of major elements in second phase particles of all three as-received samples.
To obtain more accurate results regarding the composition of the second phase particles in AA2017-T451, AA2024-T351, and AA2624-T351 alloys, point scanning, and map scanning were conducted on these particles in the as-received alloys using EDS, the results of which are summarized in Figures 4.9 to 4.11. In each micrograph, spectrum 1 is for the point map, while spectrum 2 is for the selected area map of the second phase particles. Spectrum 1 is specified with a white dot, whereas spectrum 2 is shown with a white rectangular around a part of a second phase particle.

The spectrum captured from a point within a coarse second phase particle in AA2017 alloy shows that this particle is rich in copper and aluminum with the weight percentage of 47.9 and 41.4, respectively (Figure 4.9(a)). Furthermore, the presence of $Al_2Cu$ in the aluminum matrix of AA2017 is expected. Other minor elements like iron, zinc, silicon, manganese, and magnesium were also observed in very small weight percentages (< 4 wt. %). On the other hand, the spectrum captured from a point within a coarse white second phase particles in AA2024 alloy, which were observed everywhere within the aluminum matrix of this alloy, shows a higher percentage of copper than aluminum than the AA2017 alloy with weight percentages of 65.7 and 27.6, respectively (Figure 4.10(a)). Scanning a point in a fine second phase particle of the AA2624 alloy showed a lower weight percentage for copper compared to the other two alloys which are about 35.2 wt.% (Figure 4.11(a)). That could be because of the fact that second phase particles are more dispersed within the aluminum matrix than clustering together as observed in AA2624 aluminum alloy.

Results from spectrum 2 of AA2017, AA2024, and AA2624 aluminum alloys are presented in Figures 4.9(b), 4.10(b), and 4.11(b), respectively. A higher weight percentage of Cu and Al in AA2017-T451 alloy shows the high possibility of the existence of $Al_2Cu$. A significant percentage of Fe, Mn, and Si elements in AA2017-T451 alloy were also observed (Figure 4.9(b)). According to Mirjalili [65], by containing a high percentage of Fe, Mn, and Si elements, the formation of $Al_3Fe$ or/and $Al_{12}(Fe,Mn)_3Si$ can be expected. A higher weight percentage of Mg was observed for AA2024-T351 and AA2624-T351 alloys, which are Al-Cu-Mg alloys, compared to AA2017-T451, which is an Al-Cu alloy. Results show 0.6 Wt.% and 0.8 Wt.% for the AA2024-T351 and AA2624-T351 alloys, respectively, whereas only 0.2 Wt.% for the AA2017-T451 alloy (Figures
4.10(b) and 4.11(b)). Having a lower ratio of Cu/Mg in AA2024-T351 and AA2624-T351 alloys can increase the possibility of the formation of $Al_2CuMg$ composition along with $Al_2Cu$ in AA2024-T351 and AA2624-T351 alloys. In AA2024-T351, the presence of $Al_3Fe$, $Al_{12}(Fe,Mn)_3Si$ can also be expected due to the significant weight percentage of Fe, Mn, and Si elements (Figure 4.10(b)). On the contrary, AA2624 exhibited a small percentage of Mn with 0.2 Wt.%, and 0 Wt.% for Fe and Si elements (Figure 4.11(b)). Furthermore, formation of $Al_2CuMg$ and $Al_2Cu$ phases are more expected in AA2624-T351 [24].

![Elemental mapping of second phase particles in as-received AA2017. (a) point scanning and (b) map scanning.](image)

Figure 4.9. Elemental mapping of second phase particles in as-received AA2017. (a) point scanning and (b) map scanning.
Figure 4.10. Elemental mapping of second phase particles in as-received AA2024. (a) point scanning and (b) map scanning.

Figure 4.11. Elemental mapping of second phase particles in as-received AA2624. (a) point scanning and (b) map scanning.
4.3 EBSD measurements of the as-received aluminum alloys

Electron backscatter diffraction (EBSD) was used to obtain detailed information regarding the crystallographic orientation, local misorientation, deformed and recrystallized fraction of the samples. EBSD scans were conducted on the as-received AA2017, AA2024, and AA2624 aluminum alloys in the naturally aged condition (T451/T351) before mechanical testing. The results obtained from the post-processing of the raw data obtained from EBSD measurements are presented in Figures 4.12 to 4.15. These measurements were conducted on a plane perpendicular to the rolling direction of the AA2624-T351 plate, as well as the AA2017-T451 and AA2024-T351 rods.

EBSD inverse pole figure (IPF) maps provided in Figure 4.12(a, b, c) show the grains' morphology in each as-received alloy. These maps clearly show that the AA2017-T451 and AA2024-T351 alloys have smaller and more equiaxed grains, while AA2624-T351 showed large and elongated grains. Figure 4.13 shows information about the grain size distribution of all three as-received samples obtained from the processed EBSD data. It is evident that AA2624-T351 alloy has more grains with diameters greater than 25 μm, whereas AA2017-T451 and AA2024-T351 had fewer grains larger than 25 μm in diameter, and the grains were evenly distributed within the microstructure.
Figure 4.12. EBSD orientation maps (a, b, c), local average misorientation map (d, e, f), deformed and recrystallized fraction maps (g, h, i) obtained for AA2017-T451 (a, d, g), AA2024-T351 (b, e, h), and AA2624-T351 (c, f, i).

The average grain size of AA2017-T451, AA2024-T351, and AA2624-T351 alloys was obtained from the direct grain area measurement as 19 μm, 22 μm, and 40 μm, respectively. From the well-known Hall-Petch relationship, there will be a higher volume of grain boundaries per unit area in alloys with finer grain sizes to hinder dislocation motion and will require more stress to cause slip. From this principle, it can be predicted that the AA2624-T651 should show less flow stress than the other two alloys if other microstructural variables are similar.
Figures 4.12(d, e, f) provide the local average misorientation maps of all three as-received alloys. The legend below the maps simply shows the degrees of local misorientation within the grains. This means that the closer the grains are to the red color (maximum), the higher is the degree of misorientation. Furthermore, the increased misorientation indicates higher elastic stored energy and dislocation density within the material [67]. Since this energy is usually stored within dislocations, it can also serve as a qualitative or approximate measure of the dislocation density within the material.

From the local average misorientation maps, it can be seen that a shade of green and yellow distributed within the grains of AA2624-T351 (Figure 4.12(f)), whereas in AA2017-T451 and AA2024-T351 alloy, the color of the grains is closer to dark green and blue (Figures 4.12(d) and (e)). It means that more dislocation density and higher energy are stored in the AA2624-T351 alloy. The graph in Figure 4.14 confirms this idea as higher misorientation angles have more relative frequencies for AA2624-T351. Furthermore, in AA2024-T351 alloy, more grains have a higher degree of misorientation than AA2017-T451, which means more energy is stored within the AA2024-T351 compared to AA2017-T451, and subsequently, the former is expected to show more strength than the latter.

Figures 4.12(g, h, i) show the recrystallized, recovered, and deformed grains in the microstructure of the as-received alloys. It can be observed that there is a good relationship between the local
average misorientation maps and recrystallized fraction maps. AA2624-T351 shows more red- and yellow-colored grains than blue compared to the other two specimens, meaning more deformed and recovered grains exist in these samples than recrystallized grains. Also, in AA2017-T351, there are more recrystallized grains, whereas in AA2024-T351, fewer recrystallized grains and more recovered grains exist.

Figure 4.14. Local average misorientation of all three as-received aluminum alloys.

Figure 4.15 also shows the relative frequencies for recrystallized, recovered, and deformed fractions. A larger deformed fraction in AA2624-T351 alloy can be expected to have more stored energy than the other two samples with less deformed fraction. Also, AA2017 exhibited the most recrystallized fraction, resulting in less stored energy among all three alloys. Having more deformed grains means having more dislocation density and more stored energy. Therefore, it is expected that AA2624-T351 will exhibit more strength compared to the other two samples if other microstructural variables are the same. Both Figures 4.12(d, e, f), and 4.14, indicate more angle of misorientation within the grains for the AA2624-T351, which has the most deformed fraction.
In conclusion with respect to the EBSD maps, AA2624-T351 alloy has the largest grain size among all alloys and exhibited the highest stored energy and the most deformed grains. On the other hand, AA2017-T451 alloy had the highest fraction for recrystallized grains compared to AA2024-T351 and AA2624-T351 alloys, which mean less stored energy and less degree of misorientation, which may result in more ductility in the alloy.

The EBSD results contain relatively detailed information regarding the microstructural characteristics of the investigated alloys and the probable reaction of these alloys under mechanical loading. This data would enhance the understanding of the results and behavior of the alloys during mechanical testing.

4.4 Torsion test

4.4.1 Effects of strain rate mechanical response to torsional loading

The shear stress-shear strain curves of AA2017, AA2024, and AA2624 aluminum alloys in different temper conditions (T351/T451, T651, and O) under the high strain-rate torsional loading are provided in Figures 4.16 to 4.25. Various strain rates obtained from the different angles of twist (ϕ) used in the test have an influence on these curves. For torsional loadings with ϕ=12°, the strain rate is more than twice that of the torsional loading with ϕ=4° for all investigated samples. It shows that the higher the angle of twist, the higher the strain rate will be. On the other hand, the magnitude
of strain rate generated in the aluminum alloys is different for different temper conditions. In annealed samples, the generated strain rate is higher than the other investigated temper conditions. It is due to the absence of precipitates that contribute to the strengthening and presence of more recrystallized grains than deformed ones that contribute to strengthening. In the annealing process, the hardness and strength of the material are reduced, and the material becomes more ductile as the dislocation annihilation occurs at higher temperatures. A decrease in shear flow stress in annealed samples of all alloys was investigated in the shear stress-shear strain curves compared to the age-hardened specimens. In these shear stress-shear strain curves, the competition between work hardening and thermal softening during the deformation will determine the maximum peak shear flow stress [68]. In shear stress-shear strain curves for all test specimens, an initial elastic deformation occurs, which is rapidly followed by plastic deformation.

The final result of the shear stress-shear strain curves of the AA2017, AA2024, and AA2624 aluminum alloys under high strain rate torsional loading at different temper conditions are summarized in Figure 4.16. As in the high strain-rate torsional loading test, the peak flow stress is an important factor that resistance to plastic deformation.

It can be observed that all alloys in O temper condition exhibited the lowest peak flow stress compared to the same alloys in aged conditions. AA2624 alloys in O temper condition showed the highest peak flow stress of 202 MPa compared to the AA2017 and AA2024 alloys with 186 and 177 MPa in the same temper condition, respectively. AA2017 and AA2024 alloys in the naturally aged condition of T451/T351 have higher peak flow stress than the same alloys in the artificially aged condition of T651, whereas AA2624-T651 alloys have higher peak flow stress than AA2624-T351. Lower peak flow stress in artificially aged AA2017-T651 alloys could be due to over-aging of these specimens, which results in a decrease in peak flow stress of these materials. Also, the decrease of peak flow stress in AA2024-T651 alloys could be due to the higher susceptibility of these alloys to fracture under high strain rate torsional loading.

In conclusion, AA2024 alloys exhibited highest peak flow stresses in both naturally and artificially aged temper conditions. Furthermore, all alloys in O temper condition exhibited lower peak flow stresses than the age hardened samples.
Figure 4.16. Summary of peak flow stresses of AA2017, AA2024, AA2624 aluminum alloys in T451/T351, T651, and O temper conditions under torsional loading using $\phi=12^\circ$.

The result of torsional loading under different angles of twist combined with the discussion over each condition is presented in the shear stress-shear strain curves in this subsection (Figure 4.17 To 4.25). Figure 4.17 shows the deformation behavior of the AA2017-T451 aluminum alloy. As the angle of twist changed from $\phi=4^\circ$ to $\phi=8^\circ$, peak shear flow stress and strain rate increased. At $\phi=4^\circ$, a peak flow stress of 183 MPa was recorded. At $\phi=8^\circ$, peak flow stress increased to 281 MPa and the strain rate increased from 530 /s to 990 /s. As the angle of twist was raised to $\phi=12^\circ$, the peak flow stress increases to 322 MPa. A sudden drop in peak flow stress follows due to the dominance of thermal softening generated as a result of temperature increase in the sample. In AA2017-T651 (Figure 4.18), as the angle of twist changed from $\phi=4^\circ$ to $\phi=8^\circ$, shear flow stress and the strain-hardening increased significantly. For $\phi=4^\circ$ and $\phi=8^\circ$, the flow stresses are 127 MPa and 262 MPa, respectively. When the angle of twist was increased to 12°, less significant increase in strain hardening and the flow compared to the other loading conditions in $\phi=4^\circ$ and $\phi=8^\circ$ was observed. Comparing AA2017-T451 and AA2017-T651, the change in the magnitude of the strain has been insignificant. In AA2017-O (Figure 4.19), the peak flow stresses for $\phi=4^\circ$, $\phi=8^\circ$, and $\phi=12^\circ$ were determined to be 109, 167, and 186 MPa, respectively. It can be observed that the shear flow stresses in the annealed specimen are much lower than in the age-hardened samples.
These annealed samples also exhibited more ductility than excellent strength compared to the aged specimens.

Figure 4.17. Typical shear stress-shear strain curves obtained for AA2017-T451 specimens subjected to torsional loading at $\phi=4^\circ$, $\phi=8^\circ$, and $\phi=12^\circ$.

Figure 4.18. Typical shear stress-shear strain curves obtained for AA2017-T651 specimens subjected to torsional loading at $\phi=4^\circ$, $\phi=8^\circ$, and $\phi=12^\circ$. 
Figure 4.19. Typical shear stress-shear strain curves obtained for AA2017-O specimens subjected to torsional loading at $\phi=4^\circ$, $\phi=8^\circ$, and $\phi=12^\circ$.

In the shear stress-shear strain curves of AA2024 alloys, the overall deformation pattern was almost the same for the three temper conditions; which means by increasing the angle of twist, the strain rate and flow stress increased. Figure 4.20 shows the deformation behavior of the AA2024-T351 alloy under torsional loadings at different angles of twist. As the angle of twist changes from $\phi=4^\circ$ to $\phi=8^\circ$ the flow stress increased by 102%, and as the angle of twist changed from $\phi=8^\circ$ to $\phi=12^\circ$ the flow stress increased only by 10% percent. AA2024-T651 alloy experienced a very small peak flow stress in $\phi=4^\circ$ with the magnitude of 70 MPa comparing to the peak flow stresses in higher angles of twist of $\phi=8^\circ$ and $\phi=12^\circ$ which reported as 334 and 339 MPa, respectively (Figure 4.21).
Figure 4.20. Typical shear stress-shear strain curves obtained for AA2024-T351 specimens subjected to torsional loading at $\phi=4^\circ$, $\phi=8^\circ$, and $\phi=12^\circ$.

Figure 4.21. Typical shear stress-shear strain curves obtained for AA2024-T651 specimens subjected to torsional loading at $\phi=4^\circ$, $\phi=8^\circ$, and $\phi=12^\circ$. 
Figure 4.22. Typical shear stress-shear strain curves obtained for AA2024-O specimens subjected to torsional loading at $\phi=4^\circ$, $\phi=8^\circ$, and $\phi=12^\circ$.

For AA2024-T651, as the angle of twist changes from $\phi=8^\circ$ to $\phi=12^\circ$ the flow stress drops at lower strain. Early drop in flow stress could be due to more intense adiabatic heat generation in the sample with a bigger angle of twist that results in having a more dominant thermal softening. AA2024-O samples showed a bigger strain rate and strain magnitude due to the absence of coherent precipitates that contribute to hardening (Figure 4.22).

Comparing AA2024-T351 and AA2024-T651, the magnitude of generated strain in the naturally aged AA2024-T351 is more than that of artificially aged AA2024-T651. It means that the artificially aged AA2024 aluminum alloy has lower plasticity than the naturally aged one. The brittleness of the artificially aged samples could be as a result of the existence of S phases which contribute to a reduction in ductility in Al-Cu-Mg alloys [69]. In AA2624 specimens, the strain rate and the flow stress increase by increasing angle of twist without any significant drop in the magnitude of peak flow stress.
Figure 4.23. Typical shear stress-shear strain curves obtained for AA2624-T351 specimens subjected to torsional loading at $\phi=4^\circ$, $\phi=8^\circ$, and $\phi=12^\circ$.

Figure 4.24. Typical shear stress-shear strain curves of AA2624-T651 specimens subjected to torsional loading at $\phi=4^\circ$, $\phi=8^\circ$, and $\phi=12^\circ$. 

66
Figure 4.25. Typical shear stress-shear strain curves obtained for AA2624-O specimens subjected to torsional loading at $\phi=4^\circ$, $\phi=8^\circ$, and $\phi=12^\circ$.

The artificially aged AA2624-T651 showed slightly more strength and a higher peak flow stress than the naturally aged AA2624-T451 alloy (Figure 4.23). By comparing the AA2624-T651 and AA2624-T351 alloys, a higher strain rate was observed in the naturally aged alloys. It can be concluded that the artificially aged AA2624-T651 has a better deformation resistance than the naturally aged AA2624-T351 samples. Although the peak flow stress is remarkably less than aged samples, AA2624-O alloy has experienced a peak during strain hardening in the loading condition with $\phi=12^\circ$ (Figure 4.25). It reached the value of 202 MPa, which is slightly higher than the peak flow stress for other alloys with the same temper condition. AA2624-O showed higher strain and strain rate values due to their lower mechanical strength and the absence of precipitates that contribute to hardening. The stress increases after the strain of 0.6 could result from dislocation multiplication that leads to strain hardening of the sample.
4.4.2 Effects of the temper condition on dynamic mechanical response

4.4.2.1 AA2017 Aluminum alloy

In age hardenable series of aluminum alloys, the nature of the precipitates directly influences the work hardening behavior [1]. In the precipitation sequence, whether the precipitates are coherent, semi-coherent, or incoherent with the aluminum matrix, an atomic plane can affect the response of these aluminum alloys to deformation [1][70]. The shear stress-shear strain and the SHR curves in Figures 4.26 to 4.34 show the response of the AA2017, AA2024, and AA2624 aluminum alloys under the torsional loading with different angles of twist of $\phi=4^\circ$, $\phi=8^\circ$, $\phi=12^\circ$.

The results of high strain rate torsional loading with $\phi=4^\circ$ in AA2017 alloys (Figure 4.26(a)) illustrate that AA2017-T451 has a higher peak flow stress than the alloy in the other temper conditions. The peak flow stress for AA2017-T451 is 183 MPa magnitude, while the peak flow stress for AA2017-T651 and 2017-O are 127 MPa and 109 MPa, respectively. Also, the SHR curve for AA2017-T451 (Figure 4.26(b)) exhibit a peak after the yield point, followed by a drop, which implies the strain hardening became dominant, and then the thermal softening overcame the strain hardening effect result in a drop in the magnitude of the SHR. In the annealed samples, the SHR drops quickly after yielding without the formation of any further peak. It explains the lower peak flow stress in the annealed samples.
Figure 4.26. (a) shear stress-shear strain curves and (b) strain hardening rate curves of AA2017 aluminum alloy at temper conditions of T451, O, and T651 under torsional loading with $\phi$=4°.

Comparing the AA2017-T451 and AA2017-T651 under loading condition using $\phi$=8° (Figure 4.27(a)), it can be observed that that AA2017-T451 alloy has a higher peak flow stress. The strain rate for AA2017-T451 is 990/s, which is lower than that for artificially aged AA2017-T651 alloy, which is 1140. This means that AA2017-T451 has better resistance to deformation when deformed using this angle of twist. However, AA2017-T651 has a slightly steeper hump (Figure 4.27(b)), which means a better SHR after the yield point.
Figure 4.27. (a) shear stress-shear strain curves and (b) strain hardening rate curves of AA2017 aluminum alloy at temper conditions of T451, O, and T651 under torsional loading with $\phi=8^\circ$.

AA2017-O alloy, deformed at torsional loading using $\phi=8^\circ$ (Figure 4.27(a)), has a lower peak shear flow stress compared to the aged samples, but with an maximum strain of near 0.36, which is higher than the aged specimens. It means that annealed specimen exhibited better ductility.
Shear stress-shear strain curve for AA2017 alloys under torsional loading using an angle of twist of $\phi=12^\circ$ (Figure 4.28(a)) shows that AA2017-T451 has the highest peak flow stress among the three temper conditions of this alloy. The lower strength of AA2017-T651 compared to AA2017-T451 could be due to overaging during the heat treatment process. AA2017-O alloy exhibited much lower peak flow stress compared to the aged samples due to the absence of precipitates like $\theta''$ and $\theta'$ which result in strengthening. Furthermore, the resistance to plastic deformation would be less in the annealed sample, resulting in higher strain rates.
Figure 4.28(b) indicate peaks in the SHR curves for both AA2017-T451 and AA2017-T651 alloys. The AA2017-T451 has a steeper peak which means the SHR in this sample is higher. The plateau pattern in the SHR curves of all samples means there is a balance between strain hardening and thermal softening that keeps the graphs in a stable horizontal trend until the thermal softening becomes dominant and the SHR drops. It can be observed that thermal softening becomes dominant in AA2017-T451 at lower strains compared to the other temper conditions.

In conclusion, for the AA2017 specimens, AA2017 at naturally aged condition offered a better strength under torsional loadings at all angles of twist, and the annealed AA2017 has the lowest strength due to its lower mechanical properties.

**4.4.2.2 AA2024 Aluminum alloy**

The Shear stress-shear strain curves for AA2024 alloys under high strain rate torsional loading using $\phi=4^\circ$ (Figure 4.29(a)) show that this alloy in naturally aged condition (T351) exhibits much higher flow stress than the other two temper conditions. Also, a strain rate of 480/s was generated in AA2024-T351. This is lower than AA2024-T651 and AA2024-O, which are 500/s and 510/s, respectively. This explains the higher flow stress in the naturally aged alloy as it exhibits less resistance to deformation. SHR curve for AA2024-T651 alloy (Figure 4.29(b)) shows a sudden drop in SHR followed by an increase due to strain hardening, which created a peak and again another drop to the point where SHR becomes negative. An increase in SHR in this alloy after yielding could be due to the multiplication of dislocations. The thermal softening is because of the conversion of about 90% of torsional loading energy into heat, which increases the temperature of the test samples during high strain rate torsional loading.

The Shear stress-shear strain curve for AA2024 alloys deformed with the $\phi=8^\circ$ (Figure 4.30(a)) shows that peak flow stresses and maximum strains increased significantly compared to the same alloy deformed using the $\phi=4^\circ$. Both AA2024-T351 and AA2024-T651 have almost the same peak flow stress, but the maximum strain for the AA2024-T651 is smaller. Also, the area under the curve of AA2024-T351 is bigger than that of the AA2024-T651, which means it exhibited better toughness.
Figure 4.29. (a) shear stress-shear strain curves and (b) strain hardening rate curves of AA2024 aluminum alloy at temper conditions of T351, O, and T651 under torsional loading with $\phi=4^\circ$.

SHR curve for AA2024-T351 and AA2024-T651 specimens (Figure 4.30(b)) shows almost the same pattern peak at low strain, whereas there is no peak in the SHR curve for the AA2024-O alloy. Furthermore, all alloys showed a plateau due to the balance between strain hardening and thermal softening followed by a drop as the thermal softening has become dominant.
Figure 4.30. (a) shear stress-shear strain curve and (b) strain hardening rate curve of AA2024 aluminum alloy at temper conditions of T351, O, and T651 under torsional loading with $\phi=8^\circ$.

It can be observed from the shear stress-shear strain curves for $\phi=12^\circ$ (Figure 4.31(a)), that artificially aged AA2024-T651 has the highest strength compared to the naturally aged and annealed AA2024 alloy. However, the generated maximum strain in AA2024-T651 is much lower than in other temper conditions, and some of the specimens failed.
Figure 4.31. (a) shear stress-strain curves and (b) strain hardening rate curves of AA2024 aluminum alloy at temper conditions of T351, O, and T651 under torsional loading with $\phi=12^\circ$.

AA2024-T651 sample fractured when deformed using $\phi=12^\circ$. This could be due to the presence of excessive S phases, which contribute to fracture of Al-Cu-Mg aluminum alloy like AA2024 [69]. Observation of the SHR curve of AA2024 specimens (Figure 4.31(b)) shows the generation of the peak at the AA2024-T351 and AA2024-T651. The hump in AA2024-T651 alloy has been sharper with a higher magnitude, which means a higher SHR has been generated in AA2024-T651 than AA2024-T351. However, the SHR curve drops to the negative region at a much lower strain in AA2024-T651 compared to the AA2024-T351 alloy. It means that thermal softening has become dominant at a lower strain in the artificially aged AA2024.
4.4.2.3 AA2624 Aluminum alloy

Shear stress-shear strain curves of AA2624 alloy under high strain torsional loading using $\phi=4^\circ$ are presented in Figure 4.32(a). AA2624-T351 and AA2624-T651 exhibit slightly the same deformation patterns, although AA2624-T651 exhibited slightly higher peak flow stress. AA2624-O has a much lower peak flow stress due to its lower mechanical properties. However, the AA2624-O sample exhibits better ductility since the total strain is higher. The SHR curves in Figure 4.32(b) show no peak in the aged specimen nor in the annealing one. It could be because strain hardening could not overcome the thermal softening effect as not much dislocation multiplication occurred due to a small magnitude of strain rates.

The Shear stress-shear strain curves of AA2624 under torsional loading with $\phi=8^\circ$ (Figure 4.33(a)) shows a much higher strain compared to the loading condition using $\phi=4^\circ$. The aged-hardened samples of AA2624-T351 and AA2624-T651 exhibited almost the same peak flow stress. The AA2624-O alloy also exhibits lower peak flow stress, with the total strain that is significantly higher than those of the aged samples.
The SHR curves in Figure 4.33(b) show the generation of a smooth hump for AA2624-T651, which subsequently drops to the stable state (plateau), where work hardening and thermal softening are balanced. Finally, a drop to the negative magnitude occurred as thermal softening became dominant. It can be concluded that the AA2624-T651 has an excellent combination of strength and ductility compared to the naturally aged AA2624-T351 alloy, although the former has more SHR at the beginning of the deformation.

Figure 4.34(a) shows the shear stress-shear strain curves of AA2624 alloys under torsional loading with $\phi=12^\circ$. The aged specimens of AA2624-T351 and AA2624-T651 have shown similar
deformation patterns. The final thermal softening occurred earlier in the AA2624-T351 alloy compared to the AA2624-T651 alloy. AA2624-O exhibited significantly higher total strain compared to the aged samples. Furthermore, strain hardening becomes dominant at around strain of 0.6, which is followed by a drop, indicating the point at which thermal softening became dominant again. AA2624-O is the only annealed sample that showed further strain hardening of all the annealed AA2017 and AA2024 test specimens.

Figure 4.33. (a) shear stress-shear strain curves and (b) strain hardening rate curves of AA2624 aluminum alloy at temper conditions of T351, O, and T651 under torsional loading with $\phi=8^\circ$.

The SHR curve for AA2624 alloys under torsional loading with $\phi=12^\circ$ is presented in Figure 4.34(b). The aged samples of AA2624-T351 and AA2624-T651 have shown a peak, followed by
a drop due to thermal softening. The peak generated in the artificially aged sample is slightly higher than for the naturally aged AA2624 alloy, which means more dominant strain hardening has occurred. By comparing the age-hardened AA2624 alloys, it can be observed that thermal softening became permanently dominant at a lower strain in the naturally aged alloy (AA2624-T351).

Figure 4.34. shear stress-shear strain curves and (b) strain hardening rate curves of AA2624 aluminum alloy at temper conditions of T351, O, and T651 under torsional loading with \( \phi = 12^\circ \).
4.4.2.4 Comparison of the torsional behavior of the three alloys

shear stress-shear strain curves of the same temper conditions were organized in the same graph to compare the effect of temper conditions on the torsional behavior of the investigated aluminum alloys. These graphs are compared at a higher angle of twist (ϕ=12°) to observe the results at higher strain rates. In the as-received naturally aged condition (Figure 4.35), AA2024-T351 alloy has shown the highest flow stress, whereas AA2624-T351 exhibited lower flow stress than the other two alloys. Data obtained from the EBSD maps showed more deformed grains, more stored energy, and bigger grain size for AA2624-T351 alloy. Although more stored energy and deformed fraction contribute to the hardening of the sample, the effect of bigger grains which result in less strength, seems to be more dominant in the AA2624-T351, and it exhibited the lowest strength in the shear stress-shear strain curve in Figure 4.35. A higher strain rate in AA2624-T351 also is due to its lower deformation resistance than the other two alloys.

Figure 4.35. shear stress-shear strain curves of all investigated alloys in T351/T451 temper conditions using ϕ=12°.

In Figure 4.36, the torsional response of the investigated aluminum alloys in T651 temper conditions are compared. From the curves, it can be observed the AA2024-T651 alloy has a lower strain than the other alloys. After reaching the peak flow stress of 339 MPa, this alloy experiences a dominant thermal softening followed by the specimen’s failure. This could be due to the
excessive formation of the S phases, which are said to contribute to the sample's brittleness [69]. On the other hand, AA2624-T651 alloy is exhibiting slightly better strength and less strain rate due to more deformation resistance in this sample. It should be mentioned that no failure occurred in the AA2017-T651 and AA2624-T651.

![Shear Stress-Shear Strain Curves](image)

Figure 4.36. shear stress-shear strain curves of all investigated alloys in T651 temper conditions using $\phi=12^\circ$.

In Figure 4.37, the mechanical response of the alloys in O temper conditions are compared. It shows that AA2024-O and AA2017-O have almost the same flow stress, whereas AA2624-O exhibits much less strength and more ductile behavior. The total strain is more than twice compared to those of the other two investigated alloys. Also, at the strain of 0.6, where AA2017-O and AA2024-O experienced thermal softening, flow stress at AA2624-O started to increase to a peak value of 202 MPa at a strain with the value of 0.76 followed by excessive thermal softening. Strain rates in annealed samples are relatively much higher than those that are naturally or artificially aged due to their weaker mechanical properties and less resistance to deformation than the aged specimen.
Figure 4.37. shear stress-shear strain curves of all investigated alloys in Annealed (O) temper conditions using $\phi=12^\circ$.

4.5 Microstructural evolution of the test specimens after mechanical loading

After exposure to high strain rate loading, the deformed aluminum alloys were prepared for microstructural examinations. They were sectioned parallel to the axis of the hollow cylindrical samples. Microscopic examinations were made on the thin wall part in the middle of the sectioned torsional specimens, where the high strain-rate deformation occurred (Figure 4.38). Both optical and scanning electron microscopes were used to investigate polished and etched sections of the specimens deformed at the angle of twist ($\phi$) of $8^\circ$ and $12^\circ$.

Figure 4.38. Photograph and schematic showing sectioning of test specimens used for microstructural investigation.
4.5.1 Deformed AA2017 alloy

Figures 4.39(a) & (b) show the optical micrographs of the deformed AA2017-T451 aluminum alloy. It can be observed that for both $\phi=8^\circ$ and $\phi=12^\circ$, the second phase particles clustered together and form an elongated irregularly shaped particle in the aluminum-rich matrix. In the optical micrographs, the particles are observed to be more dispersed within the aluminum matrix in $\phi=8^\circ$, whereas in $\phi=12^\circ$, the particles are less dispersed. As the grains and grain boundaries cannot be detected, it is difficult to conclude whether the grains have elongated or not. Under the scanning electron microscope, clustering, and elongated shape of the particles can be more clearly seen (Figure 4-40(a) & (b)). The elongated shape is due to the alignment of second phase particles in the shear flow direction [1][71]. Observation of the shape of the second phase particles in specimens deformed using $\phi=12^\circ$ reveals that the coarse, second phase particles are more elongated compared to the specimens deformed with $\phi=8^\circ$. It could be due to the generation of higher strain rate in the loading with $\phi=12^\circ$, and more second phase particles cluster together as they aligned in the shear flow direction. In AA2017-T451, no localized deformation was observed, and the deformations appear to be relatively homogenous.

Figure 4.39. Optical micrographs of AA2017-T451 after torsional loading using (a) $\phi=8^\circ$ (b) $\phi=12^\circ$. 
Figure 4.40. SEM micrographs of AA2017-T451 after torsional loading using (a) $\phi=8^\circ$ (b) $\phi=12^\circ$.

Figures 4.41 and 4.41 includes the micrographs of artificially aged AA2017-T651 deformed using $\phi=8^\circ$ and $\phi=12^\circ$ after high strain rate torsional loading. In the optical micrographs of the AA2017-T651 in Figure 4.41(a) and 4.41(b), fewer particles align with the shear flow direction than the naturally aged sample. It could be due to more resistance to deformation in the artificially aged AA2017-T651 than the previous naturally aged AA2017-T451 alloy.
Figure 4.41. Optical micrographs of AA2017-T651 after torsional loading using (a) $\phi=8^\circ$ (b) $\phi=12^\circ$.

SEM micrograph of AA2017-T651 after deformation using $\phi=8^\circ$ (Figure 4.42(a)) shows the lower density of second phase particles and less clustering compared to the deformed specimens using $\phi=12^\circ$ (Figure 4.42(b)). For loading with $\phi=8^\circ$, second phase particles partially align with the shear flow direction whereas, for specimen deformed using $\phi=12^\circ$, second phase particles intensely cluster together and form very coarse clusters. These coarse particles do not exhibit elongated shapes and are not aligned in the shear flow direction. As a result, the formation of these coarse particles might be due to the generation of heat during torsional loading at a higher strain rate that increases the mobility of the second phase particles due to matrix softening with enhanced coalescence and clustering of the second phase articles [72].

Figures 4.43 and 4.44 show optical micrographs and SEM of the AA2017-O aluminum alloy after deformation. An optical micrograph of these samples (Figure 4.43) shows second phase particles spread all over the aluminum matrix. More intense clustering can be observed for specimen deformed using $\phi=12^\circ$ (Figure 4.43(b)). It could be due to thermal softening that contributes to increasing the mobility of the second phase particles and subsequently clustering of those particles.
Figure 4.42. SEM micrographs of AA2017-T651 after torsional loading using (a) $\phi=8^\circ$ (b) $\phi=12^\circ$.

Figure 4.43. Optical micrographs of AA2017-O after torsional loading using (a) $\phi=8^\circ$ (b) $\phi=12^\circ$.

The formation of the second phase particles along the grain boundaries can be observed in SEM micrographs of the deformed AA2017-O alloy after high strain-rate torsional loading using $\phi=12^\circ$ (specified with white arrows in Figure 4.44). These particles were observed mostly in white colors and have various irregular shapes. The composition of these particles within the grain boundaries
is unknown, and electron dispersive spectroscopy (EDS) is needed to detect them. No elongation was observed in the particles, and the particles with irregular shapes dispersed within the aluminum matrix and the grain boundaries. Along the grain boundaries, the HF-containing etching solution has attacked some of the bigger second phase particles, and they appeared as grooves (specified with white arrow in Figure 4.44) in the micrographs (Figure 4.44). No adiabatic shear band was observed in the micrograph of the AA2017-O sample, suggesting that the deformation is relatively homogenous.

![Figure 4.44. SEM micrographs of AA2017-O after torsional loading using ϕ=12°.](image)

4.5.2 Deformed AA2024 alloy

Micrographs of AA2024 aluminum alloys in temper conditions of T351, T651, O after mechanical deformation under high strain rate torsional loading are provided in Figures 4.45 to 4.50. Because all tested artificially aged AA2024-T651 failed during the high strain-rate loading at for ϕ=12°, more micrographs are provided in different parts of the broken samples. The results of the fractographic examination of these samples are also included.

Micrographs of AA2024-T351 in Figure 4.45 show intense clustering of the irregularly shaped particles. These particles seem to align in shear flow direction for both ϕ=8° and ϕ=12°. Second phase particles in AA2024-T351 alloy before loading were also coarse. However, before loading, clusters are spherical and pyramidal-shaped. After loading, the clusters have more irregular shapes.
Figure 4.45. SEM micrographs of AA2024-T351 after torsional loading using (a) $\phi=8^\circ$ (b) $\phi=12^\circ$.

Figure 4.46. Optical micrographs of AA2024-T651 after torsional loading using (a) $\phi=8^\circ$ (b) $\phi=12^\circ$.

Figures 4.46 and 4.47 shows the micrographs of artificially aged AA2024-T651 alloy after imposing torsional loading using $\phi=8^\circ$ and $\phi=12^\circ$. It can be observed from the optical micrographs (Figure 4.46(a) and 4.48(b)) that second phase particles are widely dispersed within the aluminum
matrix. The elongation of the second phase particles can not be confirmed. It could imply that AA2024 alloy in artificially aged conditions has a higher resistance to plastic deformation and a higher tendency to brittle failure than the naturally aged AA2024-351 alloy. That could be the reason for the failure of this sample using $\phi=12^\circ$. SEM investigations of the AA2024-T651 alloys (Figure 4.47(a) and 4.47(b)) indicate the clustering of the black and white second phase particles, some of which occur along the grain boundaries ($\phi=8^\circ$ and $\phi=12^\circ$).

![SEM micrographs of AA2024-T651 after torsional loading using (a) $\phi=8^\circ$ (b) $\phi=12^\circ$.](image)

Figure 4.47. SEM micrographs of AA2024-T651 after torsional loading using (a) $\phi=8^\circ$ (b) $\phi=12^\circ$.

AA2024-T651 alloy under loading condition with $\phi=12^\circ$, the intensity of particles clustering is much more than in the loading condition with $\phi=8^\circ$. It could be due to more heat generated during the application of load at a higher strain rate and more tendency for the second phase particles to cluster while they are more mobile. In the micrographs of the test specimens deformed using $\phi=12^\circ$ (Figure 4.47(b)), precipitation along grain boundaries was observed. This type of precipitation was observed more in the annealed samples, in which low strength was observed. Therefore, precipitates aligning along grain boundaries could be the reason for lower resistance to
deformation and fracture observed for AA2024-T651. Also, no grain elongation, or alignment of second phase particles in shear flow direction were observed for any of these two loading conditions.

The micrographs of the regions adjacent to the fracture surface of the fragmented AA2024-T651 alloy are taken using scanning electron microscopy. These micrographs are provided in Figure 4-48. It can be observed that small precipitates around the fracture aligned in the direction of the shear flow. These particles from the upper part shifted to the right, and the bottom section shifted to the left side. The grain boundaries are also partly visible, and they have elongated to the direction of the shear flow in both the top and bottom parts of the sample. There are also irregularly shaped white second phase particles around the fracture which are filled with voids. In a study on microstructure evolution of 7075 aluminum alloy after tensile test, Lin Hua [73] observed white particles filled with voids between them and suggested that they are brittle coarse S phase pieces that break into pieces during tensile deformation rather than elongation. Further from the fracture point, the microstructure of the second phase particles is different and they look like the micrographs in Figure 4.47.

Figure 4.48. SEM micrographs showing the (a) upper section (b) lower section of the broken AA2024-T651.
AA2024-T651 alloy under high strain rate torsional loading with the $\phi=12^\circ$ broke in the direction perpendicular to the rolling direction of the specimens. Fractography of one of these samples was conducted using scanning electron microscopy (SEM). In Figure 4.49, micrographs of the fracture surface are provided in different magnifications. Elongated dimples in the direction of shear flow were observed. Elongated dimple is characteristic of the ductile shear fracture. It can be concluded that the AA2024-T651, under a high strain rate torsional loading using $\phi=12^\circ$, experienced a ductile fracture.

![Fracture surface micrographs of AA2024-T651](image)

Figure 4.49. SEM micrographs of fracture surface of the AA2024-T651, showing ductile shear fracture.

Figure 4.50 shows the microstructural evolution in the AA2024-O alloy under torsional loading with the $\phi=12^\circ$. Formation of second phase particles along the grain boundaries can be observed. Second phase particles in AA2017-O subjected to high strain rate torsional loading also exhibited a similar pattern. However, the intensity of precipitation at the grain boundaries in the AA2024-O sample is higher than in the AA2017-O sample. This may be the explanation for the AA2024-O...
showing lower strength than AA2017-O (Figure 4.37) as it is suggested that grain boundary precipitation causes the weakening of an alloy [1].

Figure 4.50. SEM micrographs of AA2024-O after torsional loading using $\phi=12^\circ$.

4.5.3 Deformed AA2624 alloy

The typical micrographs of AA2624 aluminum alloy after torsional loading are provided in Figures 4.51 and 4.52. In AA2624-T351 alloy, finely dispersed second-phase particles within the aluminum matrix can be observed in the microstructure (Figure 4.51) after deformation using $\phi=8^\circ$ and $\phi=12^\circ$. The precipitates in both loading conditions did not align in the direction of the shear flow. That could be due to the higher resistance of this specimen against deformation. Also, no precipitation along grain boundaries was observed as in the annealed samples of AA2017 and AA2024 alloys. Some partial clustering of the second phase particles can be observed for both loading conditions ($\phi=8^\circ$ and $\phi=12^\circ$). However, compared to the other two aluminum alloys subjected to similar loading conditions, the size of these second-phase articles is much smaller in the AA2624-351 alloy.

Micrographs of AA2624 aluminum alloy in the artificially aged condition (T651) after high strain rate torsional loading at $\phi=12^\circ$ are provided in Figure 4.52(a). In the SEM micrographs of AA2624-T651, some partial clustering can be observed along some specific paths, indicated by white arrows in Figure 4.53(a). It could be due to the localized heat within the sample that has made the second phase particle more mobile, resulting in clustering at those regions.
Scanning electron microscopic investigation of the AA2624-O aluminum alloy after torsional loading using $\phi=12^\circ$ shows dispersed needle-like precipitates within the aluminum matrix (Figure 4.52(b)). In AA2624-O, grain boundaries are not as visible as they are in the other annealed aluminum alloys; therefore, it is hard to say if the grains are elongated or not. The formation of precipitates along grain boundaries can be observed within the matrix. However, these particles are finely dispersed along grain boundaries, whereas in AA2017-O and AA2024-O second phase particles clustered along the grain boundaries. The less clustering in the AA2624-O sample could be due to the lack of Si in the alloy, content which contributes to clustering [66]. Furthermore, bigger clusters were observed within the matrix of AA2624-O alloy. Hau [73] suggested that these coarse particles are S phase constituents ($Al_2CuMg$) that seem to be brittle and break into pieces along the tensile direction. In the micrographs shown in Figure 4.52(b) broken white clusters can be observed within the matrix (specified with white arrows). These brittle particles may also be
the reason for further increase in peak flow stress in AA20624-O compared to the AA2017-O and AA2024-O aluminum alloys (Figure 4.37).

Figure 4.52. SEM micrographs obtained for specimens of (a) AA2624-T651, and (b) AA2624-O alloys after torsional loading using $\phi=12^\circ$

4.5.4 Microstructure on transverse section

The observation of the microstructural evolution in the investigated aluminum alloys at different temper conditions under torsional loading at $\phi=8^\circ$ and $\phi=12^\circ$ angles of twist reveals that the deformation is relatively homogenous in all specimens and there is no sign of localized deformation within the matrix of the alloys. To further confirm no strain localization leading to the occurrence of adiabatic shear bands, some of the specimens were sectioned in the direction perpendicular to the axis on the cylindrical test specimens. Furthermore, the etching time was also increased by 2-3 seconds to see if the micrographs would exhibit more of the grain boundaries.
The microstructures on the transverse section of the AA2017-T451 subjected to high strain rate torsional loading (Figure 4.54(a)) also reveal no trace of strain localization leading to the formation of adiabatic shear bands. Deformations are relatively homogenous. Second phase particles had the same pattern of clustering as observed along the longitudinal section. The second phase particles clustered to form irregular shapes, and they align in the shear flow direction.

Micrographs obtained from scanning electron microscopy of the AA2024-O alloy after torsional loading on the transverse section (Figure 4.54(b)) show precipitation of second phase particles along grain boundaries as previously observed on the longitudinal section. No grain elongation was observed in this deformed sample, and the second phase particles did not align in the shear flow direction. After etching for a longer time, the grain boundaries for the AA2024-O sample became more visible. However, this procedure didn't make any difference on the visibility of the grain boundaries in aged samples.

SEM Micrographs of the AA2624-T651 under torsional loading with $\phi=12^\circ$ on the transverse section are provided in Figure 4.55. In this specimen, spherical second phase particles are mostly finely dispersed within the aluminum-rich matrix. Clustering is observed around a specific line indicated with a white arrow. This could be the initiation of strain localization that could lead to the formation of a shear band at that area due to high localized thermal softening. Clustering in
that area could also be due to the increase in the mobility of the second phase particles and their tendency to form clusters.

Figure 4.54. SEM micrographs of (a)AA2017-T451, (b) AA2024-O alloys after torsional loading using $\phi=12^\circ$

Figure 4.55. SEM micrographs of AA2624-T651 alloy after torsional loading using $\phi=12^\circ$
CHAPTER 5

CONCLUSIONS AND RECOMMENDATIONS

5.1 Summary

The dynamic mechanical responses of AA2017, AA2024, AA2624 aluminum alloys under high strain-rate torsional loading were investigated. Machined test specimens of each alloy were divided into three groups. One group was tested in the as-received naturally aged conditions (T351 or T451). One group was annealed to obtain O temper, while the last group was artificially aged to obtain T651 temper. The heat treatment processes were carried out to determine the effect of temper conditions on the dynamic mechanical response of the alloys under high strain-rate torsional loading. All the three groups of alloys were subjected to high strain-rate torsional loading using angles of twist ($\phi$) 4°, 8°, and 12°. Instrumented torsional split Hopkinson bar was used for the mechanical test. The results obtained from this test are presented in the form of shear stress-shear strain and strain hardening rate curves and discussed in this thesis. Results showed higher strength for the age-hardened alloys, whereas, in annealed specimens, higher ductility and less strength were observed. The AA2024 alloy in artificially aged condition experienced failure at $\phi=12^\circ$, while others did not.

Optical microscopy (OM) and scanning electron microscopy (SEM) were used to observe the morphology of the second phase particles in the alloys before and after the mechanical loading. The micrographs of aged AA2017 and AA2024 alloys exhibited clustering, forming elongated clusters that align in the shear flow direction. In the aged AA2624 alloy, fine second phase particles remain after mechanical loading. Electron backscatter diffraction (EBSD) was used to obtain more detailed information about the microstructure of the as-received alloys. Results obtained from the post-processing of the raw EBSD data are also presented and discussed. Energy dispersive spectroscopy (EDS) was conducted to detect the composition of the second phase particles observed in the micrographs.
5.2 Conclusions

The conclusions drawn from this project are:

1. The plastic deformation of AA2017, AA2024, and AA2624 aluminum alloys under high strain-rate torsional loading is relatively homogenous in the three investigated temper conditions. There is no evidence of shear strain localization leading to adiabatic shear banding.

2. At the highest strain rate ($\phi=12^\circ$), Al-Cu-Mg alloys (AA2024 and AA2624) exhibited higher strength than the Al-Cu alloy (AA2017) in the artificial age condition (T651).

3. In a naturally aged condition, AA2024 has the highest strength among all three alloys, and AA2624 has the lowest strength under the same loading condition.

4. Natural-aged AA2017 under high strain rate torsional loading using $\phi=12^\circ$ exhibited higher strength compared to the artificially aged sample. Furthermore, naturally aged AA2017 has the highest strength among all three temper conditions.

5. Artificially aged AA2024 alloy is more prone to fracture than the other two alloys. However, AA2024 has higher plasticity and strength in the naturally aged temper condition.

6. Observation of the fracture surface of the AA2024-T651 alloy revealed that the failure occurred by ductile fracture as elongated dimples were observed in the fractographs.

7. The strength of the AA2624 alloy increased after the artificial aging process without showing an decrease in ductility as observed for the AA2024 alloy.

5.3 Recommendations for future work

1. The structure and morphology of the ultra-fine precipitates which contribute to precipitation hardening should be analyzed before and after mechanical loading using transmission electron microscopy (TEM) to observe the response of these precipitates at high strain rates.

2. Observation of the deformed samples after mechanical loading using electron backscatter diffraction (EBSD) technique would exhibit more detailed information about the evolution of the grain structure, which can contribute to understanding the mechanism of deformation of the investigated aluminum alloy under high strain rate torsional behavior.
3. Since AA2624 is a newly-developed alloy, more studies on the damage mechanism of this alloy are needed under different loading and temper conditions.
References


[34] G. Owolabi, D. Odoh, A. Peterson, A. Odeshi, and H. Whitworth, “Measurement of the


[58] A. G. Odeshi and M. N. Bassim, “High strain-rate fracture and failure of a high strength


APPENDIX A

COPYRIGHT PERMISSIONS

Chapter 2, Figure 2.1.

This Agreement between University of Saskatchewan -- Nima Gharraie ("You") and Elsevier ("Elsevier") consists of your license details and the terms and conditions provided by Elsevier and Copyright Clearance Center.

License Number 5192180843385
License date Nov 18, 2021
Licensed Content Publisher Elsevier
Licensed Content Publication Acta Materialia
Licensed Content Title Precipitation processes in an Al-2.5Cu-1.5Mg (wt. %) alloy microalloyed with Ag and Si
Licensed Content Author K Raviprasad, C.R Hutchinson, T Sakurai, S.P Ringer
Licensed Content Date Oct 6, 2003
Licensed Content Volume 51
Licensed Content Issue 17
Licensed Content Pages 14
Start Page 5037
End Page 5050
Type of Use reuse in a thesis/dissertation
Portion figures/tables/illustrations
Number of figures/tables/illustrations 1
Format electronic
Are you the author of this Elsevier article? No
Will you be translating? No
Title Effect of temper conditions on torsional behavior of selected 2000 series Aluminum alloys at high strain rate
Institution name University of Saskatchewan
Expected presentation date Dec 2021
<table>
<thead>
<tr>
<th>Order reference number</th>
<th>1129</th>
</tr>
</thead>
<tbody>
<tr>
<td>Portions</td>
<td>Fig. 1. Aluminium-rich corner of the Al-Cu-Mg phase diagram indicating the phases present as a function of composition, after long term ageing at 190 °C.</td>
</tr>
<tr>
<td>Requestor Location</td>
<td>University of Saskatchewan 147 Candle Crescent</td>
</tr>
<tr>
<td>Publisher Tax ID</td>
<td>GB 494 6272 12</td>
</tr>
<tr>
<td>Total</td>
<td>0.00 USD</td>
</tr>
<tr>
<td>Terms and Conditions</td>
<td>Chapter 2, Figure 2.2.</td>
</tr>
</tbody>
</table>

This Agreement between University of Saskatchewan -- Nima Gharraie ("You") and Elsevier ("Elsevier") consists of your license details and the terms and conditions provided by Elsevier and Copyright Clearance Center.

<p>| License Number          | 5192181317205 |
| License date            | Nov 18, 2021 |
| Licensed Content Publisher | Elsevier |
| Licensed Content Publication | Materials Characterization |
| Licensed Content Title  | Microstructural Evolution and Age Hardening in Aluminium Alloys Atom Probe Field-Ion Microscopy and Transmission Electron Microscopy Studies |
| Licensed Content Author | S.P. Ringer, K. Hono |
| Licensed Content Date   | January–February 2000 |
| Licensed Content Volume | 44 |
| Licensed Content Issue  | 1-2 |
| Licensed Content Pages  | 31 |
| Start Page              | 101 |
| End Page                | 131 |
| Type of Use             | reuse in a thesis/dissertation |</p>
<table>
<thead>
<tr>
<th>Portion</th>
<th>figures/tables/illustrations</th>
</tr>
</thead>
<tbody>
<tr>
<td>Number of figures/tables/illustrations</td>
<td>1</td>
</tr>
<tr>
<td>Format</td>
<td>electronic</td>
</tr>
<tr>
<td>Are you the author of this Elsevier article?</td>
<td>No</td>
</tr>
<tr>
<td>Will you be translating?</td>
<td>No</td>
</tr>
<tr>
<td>Title</td>
<td>Effect of temper conditions on torsional behavior of selected 2000 series Aluminum alloys at high strain rate</td>
</tr>
<tr>
<td>Institution name</td>
<td>University of Saskatchewan</td>
</tr>
<tr>
<td>Expected presentation date</td>
<td>Dec 2021</td>
</tr>
<tr>
<td>Order reference number</td>
<td>1129</td>
</tr>
</tbody>
</table>

Portions

FIG. 1. Al-rich corner of the Al–Cu phase diagram showing the metastable solvus boundaries for GP zones, u₀ and u₉, together with the equilibrium solvus line for the u phase

University of Saskatchewan
147 Candle Crescent

Requestor Location
Saskatoon, SK S7K 5A1
Canada
Attn: University of Saskatchewan

Publisher Tax ID
GB 494 6272 12

Total
0.00 USD

Terms and Conditions

Chapter 2, Figure 2.6.

This Agreement between University of Saskatchewan -- Nima Gharraie ("You") and Elsevier ("Elsevier") consists of your license details and the terms and conditions provided by Elsevier and Copyright Clearance Center.

License Number
5192200809767

License date
Nov 18, 2021

Licensed Content Publisher
Elsevier
| Licensed Content Publication | International Journal of Solids and Structures |
| Licensed Content Title       | A failure criterion based on material instability |
| Licensed Content Author      | S.E. Schoenfeld, T.W. Wright |
| Licensed Content Date        | Jun 1, 2003 |
| Licensed Content Volume      | 40 |
| Licensed Content Issue       | 12 |
| Licensed Content Pages       | 17 |
| Start Page                   | 3021 |
| End Page                     | 3037 |
| Type of Use                  | reuse in a thesis/dissertation |
| Portion                      | figures/tables/illustrations |
| Number of figures/tables/illustrations | 1 |
| Format                       | electronic |
| Are you the author of this Elsevier article? | No |
| Will you be translating?     | No |
| Title                        | Effect of temper conditions on torsional behavior of selected 2000 series Aluminum alloys at high strain rate |
| Institution name             | University of Saskatchewan |
| Expected presentation date   | Dec 2021 |
| Portions                     | Fig. 1. Shear stress versus nominal shear strain for a typical work-hardening material during a torsion experiment. |
| University of Saskatchewan  | 147 Candle Crescent |
| Requestor Location           | Saskatoon, SK S7K 5A1 |
| Canada                       | Attn: University of Saskatchewan |
| Publisher Tax ID             | GB 494 6272 12 |
| Total                        | 0.00 USD |
| Terms and Conditions         | |
Chapter 2, Figure 2.8.

This Agreement between University of Saskatchewan -- Nima Gharraie ("You") and Elsevier ("Elsevier") consists of your license details and the terms and conditions provided by Elsevier and Copyright Clearance Center.

License Number 5192210977244
License date Nov 18, 2021
Licensed Content Publisher Elsevier
Licensed Content Publication Materials Science and Engineering: A
Licensed Content Title Shear strain localization in AA 2219-T8 aluminum alloy at high strain rates
Licensed Content Author G.M. Owolabi, D.T. Bolling, A.A. Tiamiyu, R. Abu, A.G. Odeshi, H.A. Whitworth
Licensed Content Date Feb 8, 2016
Licensed Content Volume 655
Licensed Content Issue n/a
Licensed Content Pages 9
Start Page 212
End Page 220
Type of Use reuse in a thesis/dissertation
Portion figures/tables/illustrations
Number of figures/tables/illustrations 1
Format electronic
Are you the author of this Elsevier article? No
Will you be translating? No
Title Effect of temper conditions on torsional behavior of selected 2000 series Aluminum alloys at high strain rate
Institution name University of Saskatchewan
Expected presentation date Dec 2021
Portions Fig. 7. Optical micrographs showing (a) deformed band (DB) (b) transformed band (TB) and (c) both DB and TB in an impacted specimen.
Requestor Location University of Saskatchewan

114
Chapter 2, Figure 2.9.

ELSEVIER LICENSE
TERMS AND CONDITIONS
Nov 19, 2021

This Agreement between University of Saskatchewan -- Nima Gharraie ("You") and Elsevier ("Elsevier") consists of your license details and the terms and conditions provided by Elsevier and Copyright Clearance Center.

License Number 5192230551771
License date Nov 18, 2021
Licensed Content Publisher Elsevier
Licensed Content Publication Materials Science and Engineering: A
Licensed Content Title Influence of {10-12} extension twinning on the flow behavior of AZ31 Mg alloy
Licensed Content Author Lan Jiang, John J. Jonas, Alan A. Luo, Anil K. Sachdev, Stéphane Godet
Licensed Content Date Feb 15, 2007
Licensed Content Volume 445
Licensed Content Issue n/a
Licensed Content Pages 8
Start Page 302
End Page 309
Type of Use reuse in a thesis/dissertation
Portion figures/tables/illustrations
Number of figures/tables/illustrations 1
Chapter 2, Figure 2.11.

ELSEVIER LICENSE
TERMS AND CONDITIONS
Nov 19, 2021

This Agreement between University of Saskatchewan -- Nima Gharraie ("You") and Elsevier ("Elsevier") consists of your license details and the terms and conditions provided by Elsevier and Copyright Clearance Center.

License Number 5192230551771
License date Nov 18, 2021
Licensed Content Publisher Elsevier
Title: High strain-rate fracture and failure of a high strength low alloy steel in compression
Institution name: University of Saskatchewan
Expected presentation date: Dec 2021
Order reference number: 1129

Portions:
Fig. 11. Fracture surface (SEM) along transformed shear band of a quench-hardened AISI 4340 steel specimen that was tempered at 205 °C showing (a) elongated dimples

Requestor Location: University of Saskatchewan
147 Candle Crescent
INTRODUCTION

1. The publisher for this copyrighted material is Elsevier. By clicking "accept" in connection with completing this licensing transaction, you agree that the following terms and conditions apply to this transaction (along with the Billing and Payment terms and conditions established by Copyright Clearance Center, Inc. ("CCC"), at the time that you opened your Rightslink account and that are available at any time at http://myaccount.copyright.com).

GENERAL TERMS

2. Elsevier hereby grants you permission to reproduce the aforementioned material subject to the terms and conditions indicated.

3. Acknowledgement: If any part of the material to be used (for example, figures) has appeared in our publication with credit or acknowledgement to another source, permission must also be sought from that source. If such permission is not obtained then that material may not be included in your publication/copies. Suitable acknowledgement to the source must be made, either as a footnote or in a reference list at the end of your publication, as follows:

"Reprinted from Publication title, Vol /edition number, Author(s), Title of article / title of chapter, Pages No., Copyright (Year), with permission from Elsevier [OR APPLICABLE SOCIETY COPYRIGHT OWNER]." Also Lancet special credit - "Reprinted from The Lancet, Vol. number, Author(s), Title of article, Pages No., Copyright (Year), with permission from Elsevier."

4. Reproduction of this material is confined to the purpose and/or media for which permission is hereby given.
5. Altering/Modifying Material: Not Permitted. However figures and illustrations may be altered/adapted minimally to serve your work. Any other abbreviations, additions, deletions and/or any other alterations shall be made only with prior written authorization of Elsevier Ltd. (Please contact Elsevier’s permissions helpdesk here). No modifications can be made to any Lancet figures/tables and they must be reproduced in full.

6. If the permission fee for the requested use of our material is waived in this instance, please be advised that your future requests for Elsevier materials may attract a fee.

7. Reservation of Rights: Publisher reserves all rights not specifically granted in the combination of (i) the license details provided by you and accepted in the course of this licensing transaction, (ii) these terms and conditions and (iii) CCC's Billing and Payment terms and conditions.

8. License Contingent Upon Payment: While you may exercise the rights licensed immediately upon issuance of the license at the end of the licensing process for the transaction, provided that you have disclosed complete and accurate details of your proposed use, no license is finally effective unless and until full payment is received from you (either by publisher or by CCC) as provided in CCC's Billing and Payment terms and conditions. If full payment is not received on a timely basis, then any license preliminarily granted shall be deemed automatically revoked and shall be void as if never granted. Further, in the event that you breach any of these terms and conditions or any of CCC's Billing and Payment terms and conditions, the license is automatically revoked and shall be void as if never granted. Use of materials as described in a revoked license, as well as any use of the materials beyond the scope of an unrevoked license, may constitute copyright infringement and publisher reserves the right to take any and all action to protect its copyright in the materials.

9. Warranties: Publisher makes no representations or warranties with respect to the licensed material.

10. Indemnity: You hereby indemnify and agree to hold harmless publisher and CCC, and their respective officers, directors, employees and agents, from and against any and all claims arising out of your use of the licensed material other than as specifically authorized pursuant to this license.
11. No Transfer of License: This license is personal to you and may not be sublicensed, assigned, or transferred by you to any other person without publisher's written permission.

12. No Amendment Except in Writing: This license may not be amended except in a writing signed by both parties (or, in the case of publisher, by CCC on publisher's behalf).

13. Objection to Contrary Terms: Publisher hereby objects to any terms contained in any purchase order, acknowledgment, check endorsement or other writing prepared by you, which terms are inconsistent with these terms and conditions or CCC's Billing and Payment terms and conditions. These terms and conditions, together with CCC's Billing and Payment terms and conditions (which are incorporated herein), comprise the entire agreement between you and publisher (and CCC) concerning this licensing transaction. In the event of any conflict between your obligations established by these terms and conditions and those established by CCC's Billing and Payment terms and conditions, these terms and conditions shall control.

14. Revocation: Elsevier or Copyright Clearance Center may deny the permissions described in this License at their sole discretion, for any reason or no reason, with a full refund payable to you. Notice of such denial will be made using the contact information provided by you. Failure to receive such notice will not alter or invalidate the denial. In no event will Elsevier or Copyright Clearance Center be responsible or liable for any costs, expenses or damage incurred by you as a result of a denial of your permission request, other than a refund of the amount(s) paid by you to Elsevier and/or Copyright Clearance Center for denied permissions.

**LIMITED LICENSE**

The following terms and conditions apply only to specific license types:

15. **Translation:** This permission is granted for non-exclusive world English rights only unless your license was granted for translation rights. If you licensed translation rights you may only translate this content into the languages you requested. A professional translator must perform all translations and reproduce the content word for word preserving the integrity of the article.

16. **Posting licensed content on any Website:** The following terms and conditions apply as follows: Licensing material from an Elsevier journal: All content posted to the web site must maintain the copyright information line on the bottom of each image; A hyper-text must be included to the Homepage of the journal from which
you are licensing at http://www.sciencedirect.com/science/journal/xxxxx or the Elsevier homepage for books at http://www.elsevier.com; Central Storage: This license does not include permission for a scanned version of the material to be stored in a central repository such as that provided by Heron/XanEdu.

Licensing material from an Elsevier book: A hyper-text link must be included to the Elsevier homepage at http://www.elsevier.com. All content posted to the web site must maintain the copyright information line on the bottom of each image.

**Posting licensed content on Electronic reserve:** In addition to the above the following clauses are applicable: The web site must be password-protected and made available only to bona fide students registered on a relevant course. This permission is granted for 1 year only. You may obtain a new license for future website posting.

17. **For journal authors:** the following clauses are applicable in addition to the above:

**Preprints:**

A preprint is an author's own write-up of research results and analysis, it has not been peer-reviewed, nor has it had any other value added to it by a publisher (such as formatting, copyright, technical enhancement etc.).

Authors can share their preprints anywhere at any time. Preprints should not be added to or enhanced in any way in order to appear more like, or to substitute for, the final versions of articles however authors can update their preprints on arXiv or RePEc with their Accepted Author Manuscript (see below).

If accepted for publication, we encourage authors to link from the preprint to their formal publication via its DOI. Millions of researchers have access to the formal publications on ScienceDirect, and so links will help users to find, access, cite and use the best available version. Please note that Cell Press, The Lancet and some society-owned have different preprint policies. Information on these policies is available on the journal homepage.

**Accepted Author Manuscripts:** An accepted author manuscript is the manuscript of an article that has been accepted for publication and which typically includes author-incorporated changes suggested during submission, peer review and editor-author communications.
Authors can share their accepted author manuscript:

- immediately
  - via their non-commercial person homepage or blog
  - by updating a preprint in arXiv or RePEc with the accepted manuscript
  - via their research institute or institutional repository for internal institutional uses or as part of an invitation-only research collaboration work-group
  - directly by providing copies to their students or to research collaborators for their personal use
  - for private scholarly sharing as part of an invitation-only work group on commercial sites with which Elsevier has an agreement
- After the embargo period
  - via non-commercial hosting platforms such as their institutional repository
  - via commercial sites with which Elsevier has an agreement

In all cases accepted manuscripts should:

- link to the formal publication via its DOI
- bear a CC-BY-NC-ND license - this is easy to do
- if aggregated with other manuscripts, for example in a repository or other site, be shared in alignment with our hosting policy not be added to or enhanced in any way to appear more like, or to substitute for, the published journal article.

**Published journal article (JPA):** A published journal article (PJA) is the definitive final record of published research that appears or will appear in the journal and embodies all value-adding publishing activities including peer review coordination, copy-editing, formatting, (if relevant) pagination and online enrichment.

Policies for sharing publishing journal articles differ for subscription and gold open access articles:

**Subscription Articles:** If you are an author, please share a link to your article rather than the full-text. Millions of researchers have access to the formal publications on ScienceDirect, and so links will help your users to find, access, cite, and use the best available version.
Theses and dissertations which contain embedded PJAs as part of the formal submission can be posted publicly by the awarding institution with DOI links back to the formal publications on ScienceDirect.

If you are affiliated with a library that subscribes to ScienceDirect you have additional private sharing rights for others' research accessed under that agreement. This includes use for classroom teaching and internal training at the institution (including use in course packs and courseware programs), and inclusion of the article for grant funding purposes.

**Gold Open Access Articles:** May be shared according to the author-selected end-user license and should contain a CrossMark logo, the end user license, and a DOI link to the formal publication on ScienceDirect.

Please refer to Elsevier's posting policy for further information.

18. **For book authors** the following clauses are applicable in addition to the above: Authors are permitted to place a brief summary of their work online only. You are not allowed to download and post the published electronic version of your chapter, nor may you scan the printed edition to create an electronic version. **Posting to a repository:** Authors are permitted to post a summary of their chapter only in their institution's repository.

19. **Thesis/Dissertation:** If your license is for use in a thesis/dissertation your thesis may be submitted to your institution in either print or electronic form. Should your thesis be published commercially, please reapply for permission. These requirements include permission for the Library and Archives of Canada to supply single copies, on demand, of the complete thesis and include permission for Proquest/UMI to supply single copies, on demand, of the complete thesis. Should your thesis be published commercially, please reapply for permission. Theses and dissertations which contain embedded PJAs as part of the formal submission can be posted publicly by the awarding institution with DOI links back to the formal publications on ScienceDirect.

**Elsevier Open Access Terms and Conditions**

You can publish open access with Elsevier in hundreds of open access journals or in nearly 2000 established subscription journals that support open access publishing. Permitted third party re-use of these open access articles is defined by
the author's choice of Creative Commons user license. See our open access license policy for more information.

Terms & Conditions applicable to all Open Access articles published with Elsevier:

Any reuse of the article must not represent the author as endorsing the adaptation of the article nor should the article be modified in such a way as to damage the author's honour or reputation. If any changes have been made, such changes must be clearly indicated.

The author(s) must be appropriately credited and we ask that you include the end user license and a DOI link to the formal publication on ScienceDirect.

If any part of the material to be used (for example, figures) has appeared in our publication with credit or acknowledgement to another source it is the responsibility of the user to ensure their reuse complies with the terms and conditions determined by the rights holder.

Additional Terms & Conditions applicable to each Creative Commons user license:

**CC BY:** The CC-BY license allows users to copy, to create extracts, abstracts and new works from the Article, to alter and revise the Article and to make commercial use of the Article (including reuse and/or resale of the Article by commercial entities), provided the user gives appropriate credit (with a link to the formal publication through the relevant DOI), provides a link to the license, indicates if changes were made and the licensor is not represented as endorsing the use made of the work. The full details of the license are available at [http://creativecommons.org/licenses/by/4.0](http://creativecommons.org/licenses/by/4.0).

**CC BY NC SA:** The CC BY-NC-SA license allows users to copy, to create extracts, abstracts and new works from the Article, to alter and revise the Article, provided this is not done for commercial purposes, and that the user gives appropriate credit (with a link to the formal publication through the relevant DOI), provides a link to the license, indicates if changes were made and the licensor is not represented as endorsing the use made of the work. Further, any new works must be made available on the same conditions. The full details of the license are available at [http://creativecommons.org/licenses/by-nc-sa/4.0](http://creativecommons.org/licenses/by-nc-sa/4.0).

**CC BY NC ND:** The CC BY-NC-ND license allows users to copy and distribute the Article, provided this is not done for commercial purposes and further does not
permit distribution of the Article if it is changed or edited in any way, and provided the user gives appropriate credit (with a link to the formal publication through the relevant DOI), provides a link to the license, and that the licensor is not represented as endorsing the use made of the work. The full details of the license are available at http://creativecommons.org/licenses/by-nc-nd/4.0. Any commercial reuse of Open Access articles published with a CC BY NC SA or CC BY NC ND license requires permission from Elsevier and will be subject to a fee.

Commercial reuse includes:

- Associating advertising with the full text of the Article
- Charging fees for document delivery or access
- Article aggregation
- Systematic distribution via e-mail lists or share buttons

Posting or linking by commercial companies for use by customers of those companies.

20. Other Conditions:

v1.10

Questions? customercare@copyright.com or +1-855-239-3415 (toll free in the US) or +1-978-646-2777.