Microstructure-Property Correlations in Friction Stir Welded Al6061-T6 Alloys

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Abstract
Friction Stir Welding (FSW) is a solid-state process that can be beneficially used for various transportation and defense applications. Understanding the microstructure evolution and properties of friction stir welded components is necessary to use this new process in critical structural applications. In this study, tensile properties and fatigue crack growth behavior of friction stir welded Al6061-T6 wrought aluminum alloys were investigated in the direction parallel to the friction stir welding direction. Fatigue crack growth tests were performed on compact tension specimens in ambient conditions for constant ΔK. The effects of critical FSW process parameters were also studied. The resulting microstructural changes, microhardness, and residual stresses were characterized and further correlated with the tensile properties and fatigue crack growth behavior of the Al6061-T6 material. The findings from these investigations will be presented and discussed.
1 Introduction

1.1 Problem Statement and Motivation

The author of this Major Qualifying Project is an Aerospace Engineering major interested in the materials and processes used in the construction of transportation vehicles. Advances in the materials used in the aerospace and transportation industries have greatly increased the prospects and possibilities within these fields. However, these advances have also carried implications for involved technologies: adequate joining techniques must be devised for these materials. Joints are often necessary in critical structural components that are subjected to both static and dynamic loads of high intensity, requiring maximum strength and durability. Friction Stir Welding (FSW) is a joining technique that emerged in the early 1990s with the ability to form high strength joints in many advanced transportation sector alloys.

The FSW process is entirely solid state, giving it many advantages over conventional welding processes. Conventional arc welding techniques all locally liquefy the workpiece resulting in large grain sizes and large residual stresses upon cooling, which greatly degrade material properties. In addition, arc welding often necessitates the use of a filler material which changes the joint composition and adds the weight of the filler used to the joint. Riveting requires the addition of mechanical fasteners and can also require parts to be designed with greater thicknesses around fastening areas to increase strength, both of which increase weight. It may also result in degraded tensile properties and fatique life in comparison to standard plate. These decreases and strength and increases in weight are critical in the aerospace and transportation industries where both these aspects of a component are critical.

In order to use FSW in critical components, a full understanding of the properties of friction stir welds when subjected to static and dynamic loads is necessary. One very common aluminum alloy that could benefit from use of FSW joints is the Al6061 alloy used in the transportation industry. This alloy is relatively inexpensive and has high strength and ductility. In industrial settings, this alloy is commonly MIG welded, laser welded, or riveted. However, these processes may be expensive and unreliable, degenerate macroscopic properties, or add weight. FSW offers a cost effective alternative, provided necessary research on FSW joints of Al6061 is performed.
1.2 Project Objectives

The objective of this project was to build the knowledge base regarding static and dynamic performance of friction stir welded Al6061 butt joints. Achievements in this area would contribute to the viability of using the FSW process in this alloy in the transportation industry, leading to significant improvements in component strength and cost of production over traditional joining methods used with the Al6061 alloy. This project also had a secondary objective of determining relationships between the static and dynamic performance of butt joints in the Al6061 alloy and the microstructural characteristics resulting from different sets of variables in the FSW process.

1.3 Approach to the Problem

In industrial applications, joints are subjected to compound forces and moments as well as a large variety of environmental conditions. In addition to complex loading scenarios, friction stir welded joints have highly variable microstructures with residual stresses that affect the performance of the material. Ultimately, the combination of the inherent material properties and static and dynamic loading can cause failure of the joint more rapidly than is expected.

If all loads and environmental conditions were combined in a single test, isolation of factors that determine failure would be difficult. In order to approximate the behavior of friction stir welds in actual applications, tests for both static and dynamic loading were used. In this way, data regarding the performance of the joint was collected for the two primary types of loading. This data permits greater insight into the behavior of friction stir welded components.

1.4 Achievements

This project investigates both static and dynamic properties of FSW in Al6061, as well as the relationship between the static and dynamic performance of the as-FSW alloy and the microstructure of the FSW joint. Determining the relationship between performance of the joint under static and dynamic loading and the microstructure of the joint is imperative in understanding methods for improving the process. Using information gathered from static and dynamic load testing of sample butt joints, links between microstructural features and processing parameters that affect both the static and dynamic performance of the joints are identified.
2 Literature Review

2.1 Friction Stir Welding (FSW)

Friction Stir Welding is a process in which a rotating tool is driven into a desired weld seam and traversed across the length of the seam to form a solid joint. No melting of the workpiece occurs in the FSW process. The mechanics behind FSW can be complicated, and require a balance of dynamic thermal and mechanical interactions as well as flow of solid metals. In the FSW process, maintaining tool rotational speed and position of the tool head in all three axes is critical in creating a weld with consistent characteristics (Frigaard, Grong, & Midling, 2001; Mishra & Ma, 2005). Since mills used in modern manufacturing are easily capable of producing the required output energy and maintaining the tool position to a high degree of precision, FSW can easily be instituted in most manufacturing facilities.

FSW tools have a consistent design including two primary components. Each tool is comprised of a tool shoulder that maintains contact with the workpiece surface during welding and a tool probe that penetrates the intended seam to the entire depth of the probe. Though tools designed for different applications may have slightly different tool probe shapes and tool shoulder shapes, all tools maintain this same two element design.

When the FSW tool is traversed along the weld seam, the tool shoulder and probe stir the material in the immediate area of the tool. In order for the FSW process to fuse the joint through the entire weld length, the FSW operation must maintain a high enough energy input per unit length traveled to drive the fusion process (Frigaard et al., 2001; Mishra & Ma, 2005). In order to adequately stir the material, FSW tools must be made out of material that is significantly harder and stronger than the material to be joined to maintain rigidity. Frictional interactions from tool rotation and the plastic flow of the surrounding material result in significant generation of thermal energy, raising the total process temperature. The combination of thermal energy and stirring action induced by the FSW tool are the driving forces behind fusion in the FSW process (Colegrove & Shercliff, 2005; Elangovan & Balasubramanian, 2008). Figure 2-1 depicts the FSW process.
When the FSW tool traverses a seam, the tool rotation causes a difference in relative velocity between both sides of the weld. The difference in relative velocity of the tool results in a directional plastic flow from one side of the tool to the other. The difference in plastic flow characteristics between the two sides causes different microstructures to form. Consequently, the two sides of the weld are described using different nomenclature. One side of the weld is denominated the advancing side, while the other side is termed the retreating side. The advancing side of the weld experiences a higher relative velocity in relation to the tool, while the retreating side experiences a lower relative velocity. This nomenclature is dependent upon the direction of tool rotation and travel; naming for a single process can be seen in Figure 2-1. (Mishra & Ma, 2005).

Friction stir welds have been produced in a wide variety of metals, all requiring different energy inputs and different types of tooling. The energy input per unit length in FSW is primarily a function of the variables of tool rotational speed and traverse speed, requiring that these welding parameters be modulated for each alloy to input sufficient energy to form a solid joint (Mishra & Ma, 2005). Welding dense, strong materials requires very high energy input and consequently higher power machinery. Because of the low density and strength of aluminum, FSW in aluminum requires a lower input energy than FSW in other common aerospace and transportation materials such as titanium and steel. The low strength of aluminum permits steel tooling to be used, further lowering the cost of implementing FSW manufacturing solutions for aluminum (Kwon, Saito, & Shigematsu, 2002).
Research in the mechanics of the FSW process has determined the method by which FSW causes fusion. As the tool rotates, it builds energy in the material immediately surrounding it and in the direction of the tool traverse. With roughly each full rotation, the tool builds enough energy to extrude a semi-circular shell of the base material from the front of the tool to the rear side of the tool. The entire weld is produced in this manner, meaning that the weld zone is essentially an extensive set of small extrusions (Mishra & Ma, 2005). Figure 2-2 shows the mechanics of the FSW process.

The shell extrusion process causes local strain significant enough to refine both primary and secondary phases in friction stir welds. Accordingly, this process can be used as a method to refine microstructures with problematic secondary phases and salvage material properties (Elangovan & Balasubramanian, 2008). Using the process in this manner is called Friction Stir Processing (FSP); it has been suggested that FSP could be used on castings to provide an increase in material properties across large areas of the part through dissolution of secondary phases and grain refinement. The size and dispersion of grains and secondary phases is determined extensively by the welding parameters, thus significant modulation of resultant microstructures is possible (Ma, Pilchak, Juhas, & Williams, 2008). The benefits of primary and secondary phase refinement also apply to welds produced using the FSW process.

There are several benefits to using FSW as opposed to other joining techniques. Generally, joining operations are expensive and can be complicated, but FSW promises to offer significant
cost savings over conventional welding techniques as a result of the simple mechanical nature of the process. Since FSW uses a non-consumable tool, filler material is not necessitated as in conventional MIG or TIG welding; this eliminates filler material cost per unit length weld and the associated troubles of dealing with feeding and storing filler for each alloy to be welded. In addition, FSW only consumes enough energy to drive a rotating tool through the base material, not to melt the workpiece. For aluminum alloys such as Al6061, this can be a considerable saving in energy consumption, estimated to a total cost savings of nearly 30% over conventional fusion welding (Heinz & Skrotzki, 2002; Kwon et al., 2002; Mishra & Ma, 2005; Pouget & Reynolds, 2008).

2.2 Microstructure of FSW
Regardless of the material in which a friction stir weld is performed, the resulting microstructure has three distinct zones that result from the welding process. The area of all three of these zones comprises what is commonly referred to as the Weld Affected Zone (WAZ). The first constituent of the WAZ is the Dynamically Recrystallized Zone (DXZ), also known as the weld nugget, which lies at the center of the weld along the weld seam. This zone is bordered on either side by the remaining two constituent zones, the Thermo Mechanically Affected Zone (TMAZ) immediately surrounding the DXZ, and the Heat Affected Zone (HAZ) surrounding the outside edges of the TMAZ (Mishra & Ma, 2005). All three constituents of the WAZ have distinct characteristics that will be described throughout Section 2.2.

2.2.1 Dynamically Recrystallized Zone (DXZ)
The DXZ is defined as the area that has direct interaction with the tool probe and is also referred to as the weld nugget. Dynamic recrystallization is the process by which extreme strain and elevated temperature cause recrystallization of material in the weld nugget as the tool passes through it, resulting in a dispersion of fine, equiaxed grains in this area. The DXZ is relatively small, and is characterized by a shape loosely resembling the FSW tool used. The zone is characteristic of all friction stir welds, and has several qualities that are significantly different from the surrounding microstructures. In the DXZ, the dynamically recrystallized grains are frequently an order of magnitude smaller than the grains of the base material (K. V. Jata & Semiatin, 2000; Mishra & Ma, 2005; Pouget & Reynolds, 2008). Figure 2-3 shows the recrystallized grains of the DXZ in the Al2024 alloy.
The final size of the grains in the DXZ is strongly dependent upon the thermal history of the weld nugget and degree of stirring action. Low stirring results in less dynamic recrystallization and larger grains, but higher temperatures from greater stirring also result in larger grains from growth of recrystallized grains. For each alloy, there is a minimum grain size that can be achieved through a balance of minimal thermal input but great enough stirring action. As a result, the two welding parameters of rotational rate and traverse speed primarily control the grain size in the DXZ for a given alloy (Heinz & Skrotzki, 2002; Kwon et al., 2002; Woo, Choo, Brown, Feng, & Liaw, 2006).

The rotational rate and traverse speed possible for a given alloy are determined by the necessity of producing a solid weld. If the energy input per distance traveled is too high, workpiece melting occurs. If the energy input per distance is too low, incomplete fusion occurs. In both cases, the process is a failure, limiting parameter selection to a specific range of rotational rates and traverse speeds. In precipitation strengthened aluminum alloys such as Al6061, the energy input into the weld nugget raises the temperature enough to dissolve strengthening particulates. However, the fine grain structure of the DXZ compensates in part for the loss of strength due to particulate dissolution (Heinz & Skrotzki, 2002; Kwon et al., 2002; Woo, Choo, Brown, Feng, & Liaw, 2006).

Another notable characteristic of the DXZ is the occurrence of the phenomenon known as “onion rings”. Onion rings are a feature of the microstructure depicted in Figure 2-4, where shell extrusions in the DXZ create a visible circular geometry that has a cross sectional view that is similar to a sliced onion. Examination of onion rings in the Al6061 and Al7075 alloys has
shown that these onion rings are a result of shell extrusions. Additionally, experiments indicate that because of the shell extrusion process, a much smaller and less continuous degree of stirring occurs in the DXZ than was originally theorized (Krishnan, 2002).

![Figure 2-4: Onion rings in the DXZ of the Al6061 alloy (Krishnan, 2002).](image)

### 2.2.2 Thermo Mechanically Affected Zone (TMAZ)

The TMAZ is a zone entirely unique to FSW and is characterized by severe plastic deformation of grains of the base material as well as exposure to elevated temperature from proximity to the DXZ. The grains in this zone have been plastically deformed from shear induced by tool rotation and traverse. The degree of plastic deformation in the TMAZ varies by proximity to the weld and depth in the joint. Grains have a higher degree of plastic deformation closer to the weld and nearer to the tool shoulder, tapering to grains that are less deformed further from the weld centerline (Kwon et al., 2002; Mishra & Ma, 2005). The zone boundaries between the TMAZ and HAZ can be hard to define, though a method has been developed for the definition of the TMAZ outer boundary based on the angular distortion of grains (Woo et al., 2006).

The elevated temperatures experienced in the TMAZ are significant enough to dissolve strengthening precipitates in areas close to the DXZ and coarsen strengthening precipitates close to the HAZ, causing significant decreases in strength. The exact line between where precipitates are dissolved and coarsened depends on the welding parameters, thus the resultant precipitate distribution is a function of the time-temperature history of the zone (Woo et al., 2006). The TMAZ also has significant differences in the size and sharpness of the transition zone from the DXZ on the advancing and retreating sides of the weld: on the advancing side, the transition is sharp; on the retreating side the TMAZ blends gradually into the DXZ (K. V. Jata & Semiatin,
The microstructures of the advancing and retreating sides of a friction stir weld can be seen in Figure 2-5.

Figure 2-5: a) shows the entire weld, b) the transition from unaffected base material to the DXZ on the retreating side, c) the DXZ, and d) the TMAZ to DXZ transition on the advancing side, all in the Al6061 alloy (Kwon et al., 2002).

2.2.3 Heat Affected Zone (HAZ)

In FSW, the HAZ is characterized by a microstructure that is not plastically deformed but is still affected by the thermal energy of the FSW process. Similar to precipitate coarsening in the TMAZ, in precipitation strengthened alloys the HAZ is characterized by overaging of precipitates, resulting in degradation of mechanical properties (K. V. Jata, Sankaran, & Ruschau, 2000; Schubert, Klassen, Zerner, Walz, & Sepold, 2001; Soundararajan, Zekovic, & Kovacevic, 2005; Zhang, 1999). The HAZ is defined by heat input to the workpiece, which is a function of the welding parameters. The welding parameters may vary significantly depending on the nature and intent of the process, resulting in a significant variation in corresponding HAZ width and properties (Kwon et al., 2002; Mishra & Ma, 2005).

Determining the boundary between the unaffected base material and the HAZ can be difficult even on a micrograph because the variation in properties between a large section of the HAZ and
the unaffected base material is very small. Measurements of the outer HAZ boundary necessitate the use of thermometers to mark temperature boundaries, adding unnecessary complication to the study. Generally, defining the outer HAZ boundary is unimportant because it is stronger than areas of the HAZ closer to the DXZ, especially in precipitation strengthened alloys (Genevois, 2005; Heinz & Skrotzki, 2002; Ulysse, 2002).

2.3 Microhardness Profiles of FSW

Microhardness profiles are excellent indicators of the changes in properties that occur across the WAZ. Microhardness reflects the state of strengthening precipitates within a material through measurement of surface hardness, which is determined by the base alloy and how it is precipitation strengthened. In FSW, microhardness profiles reflect the state of precipitates within the WAZ as well: since the alloy composition is fixed, changes in microhardness must result primarily from changes in precipitates and grain size (Lim, Kim, Lee, & Kim, 2004). Several microhardness plots for welds performed with different welding parameters can be seen for the Al6061-T6 alloy in Figure 2-6.

![Microhardness profiles and fracture locations for FSW in Al6061-T6 with welding parameters a) 1600RPM and 0.1mpm, b) 1600RPM and 0.4mpm, c) 2000RPM and 0.1mpm, and d) 2000RPM and 0.4mpm (Lim et al., 2004).](image-url)

Figure 2-6: Microhardness profiles and fracture locations for FSW in Al6061-T6 with welding parameters a) 1600RPM and 0.1mpm, b) 1600RPM and 0.4mpm, c) 2000RPM and 0.1mpm, and d) 2000RPM and 0.4mpm (Lim et al., 2004).
The significant variations shown in the microhardness plots in Figure 2-6 suggest equally large variations in the combinations of precipitate distributions and size as well as the width of the DXZ and WAZ. However, it is apparent that the ultimate low point of microhardness occurs somewhere between the TMAZ and HAZ for all four sets of welding parameters in Figure 2-6. It should be noted that a small increase in microhardness with respect to the TMAZ occurs in the DXZ for all four sets of welding parameters, resulting from the fine grain size in the DXZ. These microhardness profiles suggest that ultimate failure should occur somewhere between the TMAZ and HAZ for the Al6061 alloy when statically loaded (Lim et al., 2004; Liu, Fujii, Maeda, & Nogi, 2003).

2.4 Tensile Properties of FSW Butt Joints

In friction stir welds, the strength of any area of the weld is determined primarily by microhardness and severity of defects in that area. The microhardness profiles for the Al6061-T6 alloy in Figure 2-6 show a weld zone “soft spot”, where lowered microhardness and small defects from the welding procedure in the TMAZ and DXZ lead to lower tensile strengths within the WAZ (Aydın, Bayram, Uğuz, & Akay, 2009; Liu et al., 2003). Upon static loading, the lower strength of the WAZ results in plastic deformation in the TMAZ and DXZ, which increases stress concentrations in the area, ultimately leading to necking and rupture. In the Al6061-T6 alloy, this ultimate weak point and corresponding ultimate fracture point occurs in the boundary between the HAZ and the TMAZ, although the exact location varies depending on the welding parameters (Lim et al., 2004).

For each set of welding parameters, local tensile strength and local ductility vary widely. In turn, variations in these local conditions result in different bulk tensile strengths and bulk ductility. The lowest Ultimate Tensile Stress (UTS) found for welds in as-FSW Al6061-T6 was 66% of the base material strength, while the highest UTS found was over 80% of the base material strength. Ductility varied even more significantly, with the highest ductility found at 135% of that of the parent material, while the lowest ductility was only 65% of that of the parent material. This variety suggests that parameters can be tailored in order to impart desired weld characteristics (Lim et al., 2004).
Research has shown that samples fail at considerably lower strains than would be anticipated for the base material in large tensile specimens of precipitation strengthened alloys (Aydin et al., 2009; Lim et al., 2004; Liu et al., 2003). It has been theorized that this is a result of unequal strain distributions across the sample resulting from the heterogeneous properties of the weld. It is suggested that local strains may even reach close to the expected bulk strain in sections of the DXZ that were plastically deformed. This theory was validated using Electronic Speckle Pattern Interferometry (ESPI) during a tensile test to map local strains in the AA5083 alloy. Local strains were found to achieve values close to the expected base material value in areas of the WAZ, while the base material experienced only elastic strain. This revealed that FSW does not significantly alter local ductility, but instead affects bulk ductility through tensile strength in the WAZ, causing deformation to occur almost exclusively in the DXZ and TMAZ (Peel, Steuwer, Preuss, & Withers, 2003). Figure 2-8 shows local strain, local tensile strength, and fracture locations for tensile tests of the as-FSW AA5083 alloy.

Table 2-1: Tensile properties of Al6061-T6 for different sets of welding parameters (Lim et al., 2004).

<table>
<thead>
<tr>
<th>Rotating Speed (rpm)</th>
<th>Welding Speed (mmpm*)</th>
<th>Yield Strength (MPa)</th>
<th>Ultimate Tensile Strength (MPa)</th>
<th>Tensile Elongation (Pct)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1000</td>
<td>0.2</td>
<td>142</td>
<td>232</td>
<td>16.5</td>
</tr>
<tr>
<td></td>
<td>0.3</td>
<td>155</td>
<td>219</td>
<td>19.9</td>
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<tr>
<td>1400</td>
<td>0.2</td>
<td>144</td>
<td>231</td>
<td>15.6</td>
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<td>0.3</td>
<td>148</td>
<td>239</td>
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<td>0.1</td>
<td>135</td>
<td>207</td>
<td>10.6</td>
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<tr>
<td></td>
<td>0.2</td>
<td>135</td>
<td>219</td>
<td>14.1</td>
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<tr>
<td></td>
<td>0.3</td>
<td>146</td>
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<td>0.1</td>
<td>140</td>
<td>214</td>
<td>11.2</td>
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<td></td>
<td>0.2</td>
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<td></td>
<td>0.4</td>
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<td>0.1</td>
<td>143</td>
<td>218</td>
<td>9.5</td>
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<tr>
<td></td>
<td>0.2</td>
<td>145</td>
<td>234</td>
<td>11.1</td>
</tr>
<tr>
<td></td>
<td>0.3</td>
<td>149</td>
<td>238</td>
<td>13.9</td>
</tr>
<tr>
<td></td>
<td>0.4</td>
<td>155</td>
<td>250</td>
<td>15.1</td>
</tr>
<tr>
<td>Parent metal**</td>
<td>—</td>
<td>246</td>
<td>311</td>
<td>14.7</td>
</tr>
</tbody>
</table>

* mmpm = m/min.
** The tensile direction for the parent metal was parallel to the transverse direction.
Figure 2-7: For the AA5083 alloy, a) shows localized strain in the WAZ and b) shows yield strength in the WAZ, both as functions of lateral displacement from the weld centerline (figure adapted from Peel et al., 2003).

2.5 Residual Stress in FSW

Although the residual stress resulting from FSW is an order of magnitude smaller than that found in conventional fusion welding, it is still significant enough to cause quantifiable changes in macroscopic properties of strength and fatigue. For aluminum alloys used in the aerospace and transportation industries, fatigue crack growth is often of significant interest. In order to more accurately predict fatigue crack growth, there must be a comprehensive understanding of residual stresses in friction stir welded aluminum alloys (Mahoney, Rhodes, Flintoff, Bingel, & Spurling, 1998; Mishra & Ma, 2005).

For friction stir welded precipitation strengthened aluminum alloys, residual stress distributions take an “M” shape with tensile stresses peaking in the TMAZ and compressive stresses existing in the HAZ (Kwon et al., 2002; Mishra & Ma, 2005). An effective method used to measure residual stresses is the cut-compliance method, where sequential cuts are made through a sample and resultant displacements of the sample are measured. From these measurements, a residual stress distribution may be calculated for the sample tested, such as the one seen in Figure 2-8.
Figure 2-8: Residual stress distribution from the cut compliance method for a Compact Tension (CT) sample of the Al2024-T351 alloy for an as-FSW joint and a stress-relieved joint (Fratini, Pasta, & Reynolds, 2009).

The residual stress distribution in Figure 2-8 for an as-FSW joint promotes fatigue crack growth through the TMAZ and DXZ to a much greater extent than the stress relieved joint. In this case, tensile residual stresses exist throughout the TMAZ and DXZ, areas where grain distortion and small grains exist, increasing the rate at which fatigue cracks propagate. Although this residual stress profile was not performed on the Al6061 alloy, it has been suggested that precipitation strengthened aluminum alloys share similar residual stress distributions (Bussu & Irving, 2003; John, Jata, & Sadananda, 2003; Prime & Hill, 2002). Consequently, precipitation strengthened aluminum alloys suffer from tensile residual stresses in the same areas that have low microhardness, as seen in Figure 2-6 and Figure 2-8.

Fatigue crack growth is heavily influenced by residual stress. Fatigue cracks propagating in compressive stress fields grow more slowly than those through tensile residual stress fields or fields of no residual stress. This is due to compressive residual stresses forcing the crack tip closed, which then requires higher applied stresses to open the crack tip and propagate the fatigue crack. Conversely, tensile residual stresses hold the crack tip open, requiring a lower applied stress to propagate the fatigue crack. Unfortunately, tensile residual stresses exist in balance with compressive residual stresses; therefore damage tolerant design seeks to locate compressive residual stresses in areas with weak microstructures to resist fatigue crack growth. (Bussu & Irving, 2003; Fratini et al., 2009; Mishra & Ma, 2005).
2.6 Fatigue Crack Growth in FSW Butt Joints

In research performed on post-weld stress relieved plates of Al2024-T351, fatigue crack growth rates found through testing of Compact Tension (CT) samples were similar to those found in the base alloy, regardless of fatigue crack orientation and initiation site. This indicates that the post-weld residual stress distribution has a more significant effect on fatigue crack growth than microstructure. In as-FSW plate, however, fatigue crack growth rates were found to vary significantly depending upon orientation and location with respect to the weld (Bussu & Irving, 2003). Fatigue crack growth rates in this alloy were also not found to be symmetrical across the weld, indicating that the differences in microstructure and residual stress on the advancing side and retreating side of the weld affect fatigue crack growth in precipitation strengthened alloys (Bussu & Irving, 2003; Fratini et al., 2009; John et al., 2003).

Figure 2-9: Fatigue crack growth for several different weld locations in CT samples for the as-FSW Al2024-T351 alloy (Fratini et al., 2009).

In Figure 2-9, fatigue crack growth rate is displayed as a function of distance from weld centerline for three CTs with welds located at different positions in relation to the initial crack. Differences for fatigue crack growth rates between these three CTs suggest that residual stress and the microstructure of FSW affect these three different stages of crack growth in different manners (Fratini et al., 2009). Other studies in fatigue crack growth in precipitation strengthened aluminum alloys have focused on fatigue crack growth rates in individual weld zones. It has been found for the Al7050-T7451 alloy that the fatigue crack growth exhibits both lower and
higher thresholds than for that of the parent material in post-FSW plate at different R values (John et al., 2003). Fatigue crack growth curves for this alloy are presented in Figure 2-10.

![Fatigue crack growth curves](image)

**Figure 2-10: Fatigue crack growth rates for different R values in the base Al7050-T7451 alloy and in the post-FSW HAZ of this alloy for CT and MT specimens (John et al., 2003).**

The fatigue crack growth curves in Figure 2-10 suggest that the FSW operation alters fatigue crack growth rates and the fatigue crack growth threshold, or the applied stress ratio at which a fatigue crack will begin to propagate. The fatigue crack growth rate and fatigue threshold vary as a function of the parameter R, where R is defined as the ratio of the minimum applied stress divided by the maximum applied stress. The fatigue crack growth curves shown in Figure 2-10 imply that the fatigue crack growth rate will differ in each area of the weld zone as a function of the parameter R. In turn, this adds a significant degree of difficulty to analyzing the full fatigue behavior of any individual alloy, since the fatigue crack growth rates and fatigue crack growth threshold both vary as a function of R. To fully understand the fatigue behavior for each alloy, tests must be performed in all weld zones for multiple R values (John et al., 2003).
3 Methodology

3.1 Fixtures and Tooling

Fixtures were designed and machined at WPI to securely hold two sections of 10”x2” Al6061-T6 plate to be friction stir welded. The fixtures were made from an Al6061-T6511 extrusion with a separate base piece and top piece with inset pockets to firmly hold the welding samples. These fixtures minimized travel on all axes as well as quickly conducting heat away from the workpiece to maintain consistent time-temperature distributions in the weld. The base plate and upper plate can be seen in Figure 3-1.

A 6mm (0.2362in) welding tool and tool holder specifically designed for use in Al6061-T6 were acquired from Friction Stir Link Inc. When fixtured, the total weld length possible was 6” with a weld penetration of just over 6mm (minimal tool shoulder penetration). All welds were performed on a HAAS VM-3 mill in the WPI HAAS technical center. The tool chosen for the FSW operations of this project was a threaded pyramid tool made from hardened steel set into a steel tool holder. Figure 3-2 is a photograph of a larger version of this tool.
3.2 Rolled Al6061-T6 Properties and Microstructure

Al6061-T6 aluminum was used as the base material in this study. The material is readily available and was acquired from a local provider. Mass spectrometry was performed on the alloy purchased in order to pinpoint composition. Table 3-1 provides the composition detail. Table 3-2 describes the properties of the as-rolled Al6061-T6 alloy.

Table 3-1: Composition of Al6061-T6 stock material

<table>
<thead>
<tr>
<th>Element Detected</th>
<th>Mg</th>
<th>Si</th>
<th>Fe</th>
<th>Cu</th>
<th>Mn</th>
<th>Cr</th>
<th>Ti</th>
<th>Al</th>
</tr>
</thead>
<tbody>
<tr>
<td>% Composition</td>
<td>.9</td>
<td>.64</td>
<td>.38</td>
<td>.256</td>
<td>.033</td>
<td>.211</td>
<td>.018</td>
<td>balance</td>
</tr>
</tbody>
</table>

Table 3-2: Properties of Al6061-T6

<table>
<thead>
<tr>
<th>Property of Al6061-T6</th>
<th>Corresponding Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>Vickers Hardness</td>
<td>107</td>
</tr>
<tr>
<td>Ultimate Tensile Strength (UTS)</td>
<td>46.0ksi</td>
</tr>
<tr>
<td>Yield Strength</td>
<td>42.3psi</td>
</tr>
<tr>
<td>Elongation at break (rectangular prism)</td>
<td>17.0%</td>
</tr>
<tr>
<td>Modulus of elasticity (E)</td>
<td>9,400ksi</td>
</tr>
<tr>
<td>Solution Temperature</td>
<td>985˚F</td>
</tr>
<tr>
<td>Ageing</td>
<td>320˚F, held 18hrs</td>
</tr>
</tbody>
</table>

In rolled form, this alloy has a grain structure that varies significantly between the rolling direction and transverse direction. The grain size was found to be approximately 330μm in the...
rolling direction and 130μm in the transverse direction. The 3:1 aspect ratio of the grains is standard for the rolled alloy. An image of the microstructure of the base alloy can be found in Figure 3-3.

![Image of microstructure](image.jpg)

Figure 3-3: Al6061-T6 microstructure polished and etched in 3% Barker's Reagent for 80sec.

The two most common secondary phases that exist in this alloy are an $\alpha$-$\text{Al}_{12}(\text{FeMn})_3\text{Si}$ phase and a $\text{Mg}_2\text{Si}$ phase. The iron phase can be seen in Figure 3-4 by the gray, irregularly shaped inclusions in the microstructure. This phase appears in clusters and groupings throughout the base material. The $\text{Mg}_2\text{Si}$ phase is the smaller, rounded, and black secondary phase. The phases are described in Table 3-3.

<table>
<thead>
<tr>
<th>Phase</th>
<th>Color</th>
<th>Morphology</th>
</tr>
</thead>
<tbody>
<tr>
<td>$\alpha$-$\text{Al}_{12}(\text{FeMn})_3\text{Si}$</td>
<td>Gray</td>
<td>Rounded phase</td>
</tr>
<tr>
<td>$\text{Mg}_2\text{Si}$</td>
<td>Black</td>
<td>Small, round particles</td>
</tr>
</tbody>
</table>
Welds were created in 6” weld lengths in Al6061-T6 stock greater than 0.3” thick. In each weld, the base material was formed so that the rolling direction was perpendicular to the welding direction. Welds were created for three sets of parameters. The baseline parameters were chosen per recommendation of the tool manufacturer Friction Stir Link Inc. for their suggested feed and speed of 1000RPM and 2.0mm/s traverse for the welding tool used. This was selected as the baseline parameter because welds created using this set of feed and speed had no observable welding defects. The second set of parameters was chosen to reflect situations with a greater distance traveled per tool revolution of 1000RPM and 3.0mm/s traverse (lower energy input per distance traveled). The third set of parameters was chosen to produce a smaller distance traveled per tool revolution of 1500RPM and 2.0mm/s traverse (higher energy input per distance traveled). In all cases, loads on the VM-3 never exceeded 100%, indicating that the FSW operations were well within the capabilities of the machine, providing consistent and repeatable welds. The result of the FSW procedures was three sets of welds with individual time-temperature histories and differing resultant microstructures. All welding samples were named according to the order in which they were produced. The welding parameters that are paired with each sample are presented in Table 2-1.
From every 6” weld, the first 0.25” and the last 0.25” were determined to be transient entrance and exit moves of the welding tool. The transient entrance and exit regions were determined by evaluation of microstructure samples under an optical microscope. The transient weld effects were a result of a difference in heat transfer that occurred from changing the traverse speed of the welding tool in the tool entrance and tool exit regions. This created variation in the microstructure that rendered these sections of the sample unfit for use in testing.
3.4 Sample Extraction Locations

Samples for tests on weld response to static and dynamic loading, microstructure samples for imaging, and microhardness profile samples were taken from multiple areas of a cohesive 6” weld. In all cases, samples were taken from locations of the weld that did not include transient zones surrounding entrance and exit moves. From each 6” weld, roughly 5” was used to create both tensile samples and CT samples or both tensile samples and microstructure samples. Figure 3-5 shows the locations from which welds were extracted.

Figure 3-5: Sample extraction locations from welds.

3.5 Tensile Test Methodology

3.5.1 Tensile Test Samples

Tensile samples were cut from sections of the 6” weld outside of the transient zones as shown in Figure 3-5. Cohesive sections of weld were selected of 1” length in the tool traverse direction by the total 4” width to create the stock for tensile samples. These samples were designed and manufactured to conform to ASTM standards with a sample size of 0.8”x3.5”x0.1”. All samples were produced with minimal defects and conformed to specified dimensions with a tolerance of 0.01”. Figure 3-6 shows a dimensioned image of the tensile samples used in testing.
3.5.2  Testing Procedure
Tensile bars were made for all eight weld samples. Of these eight tensile bars, all were ruptured except for samples from weld B-2 and B-5. Weld B-2 was kept in an untested state for examination; weld B-5 was strained to a pre-rupture but post-necking strain in order to determine where initial deformations and cracks occurred in testing.

Tensile bars were tested on an INSTRON mechanical screw machine. All samples were checked visually during testing to ensure that no abnormal failures or sample slippage occurred. A set of properties including Ultimate Tensile Stress (UTS) and elongation at fracture were recorded by the control computer for each test. These values were then used to produce stress-strain curves for the material. Some values recorded by the computer involved slippage and measurement errors in the initial stages of tensile testing. Consequently, low stress values had to be corrected to account for measurement error.

3.6  Compact Tension Test Methodology

3.6.1  Compact Tension (CT) Test Samples
Compact tension samples were designed to conform to ASTM standards. Each sample was marked with crack growth lines every 0.1” to provide a visual check on crack growth during testing. The CT samples were cut from the center of welds B-5 and B-6 to the dimensions 2.4”x2.5”x0.2” with the weld running through the middle of the sample. The weld centerline was located 0.5” away from the Electronic Discharge Machining (EDM) crack tip. The low thickness (0.2”) of the specimens resulted from the small penetration of the FSW tool used
(0.23") and operations necessary to achieve microstructural consistency across the cross section of the test specimen. Figure 3-7 shows a schematic of the CT samples used.

3.6.2 CT Testing Procedure
An INSTRON 8800 servo-hydraulic machine was used in conjunction with software from Fracture Technology Associates to record data from the fatigue crack growth experiment performed. A Long Crack Growth (LCG) test was performed on this machine for one CT sample. The test performed was a Constant ΔK (CDK) test with ΔK=10.1MPa√m. A data sheet was used to record observations of fatigue crack growth as measured with the growth lines scratched into the sample. This provided a means to corroborate computer recorded data. Despite the low thickness of the specimen, no warping, buckling, or twisting was observed during, or after, testing. The findings from the CDK test are detailed in Chapter 5.
4 Microstructure and Microhardness of as-FSW Rolled Al6061-T6

The microstructures of welds produced for this project demonstrated all the characteristics of friction stir welds found in reviewed literature. All welds had a distinct DXZ, TMAZ, and HAZ that were clearly defined by the standard FSW geometries. Primary phases as well as secondary phases were redistributed during the friction stir welding process.

4.1 Baseline Welding Parameters: 1000RPM, 2.0mm/s

Figure 4-1 shows an image taken using optical microscopy of weld B-1 for baseline welding parameters 1000RPM and 2.0mm/s. The large difference in grain sizes between unaffected material and the DXZ is apparent even at extremely low magnification, as is the appearance of onion rings. As found in other studies, the advancing side of the weld is defined by sharply upturned grains in a thin TMAZ; the retreating side of the weld demonstrates a DXZ that more smoothly blends into the HAZ through an extended TMAZ (Kwon et al., 2002; Mishra & Ma, 2005). In Figure 4-1, the differences between the advancing and retreating sides of the weld are most notable due to a section of widened TMAZ and significant grain distortion. The differences in the TMAZ result from the combined influences of the tool shoulder and tool pin, which exaggerate the width and grain distortion of the retreating side weld geometry.

Figure 4-1: Microstructure of weld B-1 (1000RPM, 2.0mm/s), etched in 3% Barker's Reagent for 80sec.
Individual grain sizes in the weld nugget (DXZ) were measured using a calibrated optical microscope. Grain sizes were marked, tabulated, and averaged in a large area of the DXZ to calculate an average grain size. Although this method provides only a loose average, as grain sizes vary across the DXZ due to differences in the time-temperature history of each point, the grain size calculated is adequate for comparisons to the unaffected base material (Mishra & Ma, 2005). The grain size calculations indicate that the baseline weld produced an average grain size of 7.87μm, roughly 1/10th the size of the rolled grains in the unaffected base material. In addition to significant refinement of grains, the FSW operation also redistributed secondary iron phases from the strand-like geometry of the rolled base material to an equal dispersion throughout the DXZ. Secondary phases showed signs of refinement and redistribution in the DXZ and TMAZ for the baseline FSW parameters.

The band spacing of onion rings was also measured in order to verify that the FSW process was extruding a shell roughly every tool revolution. In this baseline weld, the band spacing was calculated to be approximately 140μm, which corresponds roughly to the linear distance traveled by the tool per rotation for this set of parameters. The theoretical value for this spacing would be 120μm, calculated by dividing the tool traverse rate by the tool rotational rate following the equation

\[
\text{Spacing (μm)} = \frac{\text{Traverse rate (mm/s)}}{\text{Rotational rate (RPM)/60}} \times 100
\]

The small deviation of 20μm may be due to viscous effects and inconsistencies in the welding process. This finding is consistent with those on onion rings in the literature (Krishnan, 2002). Figure 4-2 shows an image of the DXZ where band spacing is defined based on visual inspection of the banded geometry.
A microhardness profile was also taken for the baseline weld to determine degradation in properties on a localized scale. The microhardness profile for the baseline welding parameters shown in Figure 4-3 demonstrates that the DXZ is harder than the surrounding TMAZ and HAZ, and that the ultimate low point of microhardness for the baseline welds occurs in the HAZ bordering the TMAZ on the advancing side. The HAZ on the retreating side has a similarly low hardness. This microhardness profile takes the typical “W” shape for precipitation strengthened alloys described in the literature. The outer edges of this distribution would be expected to continue to increase in microhardness until reaching the unaffected base material value; at no point is the hardness greater than that of the unaffected base material. The two inner peaks correspond to areas of the outer DXZ that have been local maxima of microhardness in other studies. It is likely that the relatively high hardness of the DXZ is due to the small grain size, while the low hardness of the TMAZ is due to dissolution of strengthening precipitates and the low hardness of the HAZ is due to coarsening and clustering of precipitates (Lim et al., 2004; Liu, Fujii, Maeda, & Nogi, 2003).
Figure 4-3: Vickers Hardness as a function of distance from weld centerline for a baseline weld (1000RPM, 2.0mm/s).

4.2 Parameter S2, 1000RPM, 3.0mm/s

Figure 4-4 is an image of the microstructure of weld B-7 and reveals the typical structure of the DXZ and TMAZ, as in the case of the baseline welds. Even under the low magnification used in imaging Figure 4-4 (5x), it can be seen that significantly fewer onion rings exist than in weld B-1 (Figure 4-1), and that the spacing between these bands is greater. This indicates that the two different sets of welding parameters have created observably different microstructures which are expected to have different properties: the differences result from fewer shell extrusions per linear distance traveled by the tool (Krishnan, 2002). The larger spacing between onion rings produced a less clearly defined banded geometry in this sample than for the baseline welds.
A significantly thinner TMAZ along the retreating side of the weld in comparison to that of the baseline weld is also visible in Figure 4-4. The variation in properties of the TMAZ as a function of welding parameters has been previously suggested. The definition and thickness of the TMAZ is largely a function of the stirring actions of the FSW tool. As is evident in this sample, a weld with fewer revolutions per unit linear advance of the tool would have a smaller and less defined TMAZ (Mishra & Ma, 2005). Additionally, the secondary iron phases maintained characteristics similar to those of the baseline weld: changes in secondary phase size and dispersion were observed in the DXZ and TMAZ.

Measurements of grain size performed for weld B-7 revealed an average grain size of 17.0μm, significantly larger than the grain size of nearly 8μm observed in the baseline weld. The larger grain size found in weld B-7 than in baseline welds is consistent with the literature: lower stirring action (parameters similar to those for B-7) inhibits dynamic recrystallization due to insufficient energy input per unit length of tool traverse and results in larger grain size (Mishra & Ma, 2005). Band spacing in the sample from weld B-7 was measured at approximately 160μm, 20μm greater than that measured for the baseline weld. The observed band spacing for this set of welding parameters again deviates slightly from the theoretical band spacing prediction of 180μm.

The microhardness profile for weld B-7 (1000RPM, 3.0mm/s) shows a significantly flattened shape when compared to the profile found for the baseline weld. This finding is consistent with other microhardness profiles performed on welds of multiple combinations of feeds and speeds.
detailed in the literature. The flattening of the hardness profile is a result of the smaller thermal inputs and thermal gradients in this weld. These less intense thermal conditions do not coarsen or cluster precipitates as significantly as more intense thermal interactions. The microhardness profile maintains a relatively consistent shape with smaller changes in hardness across the TMAZ and HAZ, pictured in Figure 4-5 (Lim et al., 2004). Additionally, this microhardness profile suggests that a large section of the DXZ as well as areas in the HAZ on both sides of the weld have higher hardness than can be found in the unaffected base material. The literature suggests that this is a result of finer grains in the DXZ and compressive residual stresses in the HAZ (Bussu & Irving, 2003; John, Jata, & Sadananda, 2003; Prime & Hill, 2002). The ultimate low point of microhardness for this weld is in the HAZ of the advancing side, similar to that found in the baseline weld.

![Vickers Hardness (100g) vs. Location in Weld](image)

Figure 4-5: Vickers Hardness as a function of distance from weld centerline for weld parameter S2 (1000RPM, 3.0mm/s).

### 4.3 Parameter S3, 1500RPM, 2.0mm/s

The microstructure of weld B-8 (parameter S3) was characterized by a large TMAZ on the retreating side of the weld and very distinct onion rings in the DXZ. Both of these features are visible in Figure 4-6, which shows a panorama of the DXZ, TMAZ, and inner HAZ of weld B-8. These characteristics are a result of the rotational rate and traverse speed of this weld, which cause a higher energy input per unit distance traveled by the tool than for the baseline weld or weld of parameter S2. The higher energy input in this weld is manifested through greater stirring.
action by the tool. This broadens the TMAZ, extending its boundaries nearly 1000μm away from the DXZ in areas of the TMAZ closer to the tool shoulder for this weld. Large TMAZ thickness such as this has been observed in several other studies of FSW using different welding parameters (Lim et al., 2004; Liu, Fujii, Maeda, & Nogi, 2003). It is also evident from Figure 4-6 that the onion rings are significantly more numerous and thus more closely spaced in this weld than in either the baseline weld or the weld of parameter S2.

![Figure 4-6: Microstructure of weld B-8 (1500RPM, 2.0mm/s), etched in 3% Barker’s Reagent for 80sec.](image)

For weld B-8, the average grain size found in the DXZ was 18.0μm with a band spacing of 120μm, which is larger than the grain size observed in the baseline weld or in the weld of parameter S2. The larger grain size observed in this weld is a result of the welding parameters. The greater energy input per unit length traveled by the tool in parameter S3 elevates the temperature in the DXZ and the duration spent at the elevated temperature. The elevated temperature results in grain growth of the dynamically recrystallized grains in the DXZ (Mishra & Ma, 2005). The band spacing measured in this weld had the largest deviation from theoretical band spacing of the three sets of welding parameters tested, with a measured band spacing of 120μm compared to the theoretical spacing of only 80μm. It is possible that this large deviation is a result of weld inconsistency at the site of measurement or errors in the calculation of band
spacing resulting from using a two dimensional measurement of the three dimensional onion ring structure. As in the welds of the other parameters, refinement and redistribution of secondary phases was observed. An image of the secondary phase distribution can be seen in Figure 4-7, where unaffected secondary phases in the HAZ are pictured on the left and refined secondary phases in the TMAZ and DXZ are pictured on the right.

![Image of secondary phase distribution](image)

Figure 4-7: Refined and redistributed secondary phases in a polished microstructure sample of weld B-8.

A microhardness plot of this weld also evidences the typical “W” shaped hardness plot for precipitation strengthened alloys. Readings from the micrograph in Figure 4-8 suggest that the ultimate weak point in this weld is, as suggested in the literature, in the HAZ and also on the advancing side of the weld. The increased temperatures in the weld have likely caused more severe coarsening and clustering of strengthening precipitates throughout the WAZ, and consequently lowered the hardness curve for this weld (Lim et al., 2004; Liu, Fujii, Maeda, & Nogi, 2003). This micrograph again shows that the FSW operation created a higher hardness than can be found in the base material in several sections of this weld including the outer reaches of the DXZ and an area in the HAZ, although this could be due to microstructural phenomena. It can also be seen in Figure 4-8 that there is a significant difference between the hardness values observed on the advancing and retreating sides of the weld.
Figure 4-8: Vickers Hardness as a function of distance from weld centerline for weld parameter S3 (1500RPM, 2.0mm/s).
5 Analysis of Weld Response to Static and Dynamic Loading

Two very important measures of the performance of materials are their responses to static and dynamic loads. Especially in materials with highly variable microstructures, such as friction stir welds, the performance of the weld to static and dynamic loads may vary greatly from the performance expected of the unaffected base material. In order to determine these performance characteristics, testing of samples subject to both static and dynamic loads must be performed.

5.1 Analysis of Static Loading: Tensile Tests

All specimens tested necked in two different locations. These necks developed in the HAZ area on both sides of the DXZ. The occurrence of two necks is reasonable for the Al6061 alloy, since the microhardness profile reaches local minima of similar values in the HAZ on both the advancing and retreating sides of the weld. The minima of microhardness also imply that these areas of the resultant microstructure have similar tensile strengths (Lim et al., 2004). In order to identify connections between microstructural characteristics and bulk properties, tensile tests were performed on samples created from all three sets of welding parameters.

5.1.1 Tensile bar pulled to 7.25% elongation (pre-fracture)

Figure 5-1 shows the tensile bar created from weld B-5 of the baseline FSW parameters. In this sample, plastic deformation and necking is evidenced at two different cross sections of the weld. The ultimate strain at which this test was stopped was 7.25% elongation. In order to examine the deformations and possible crack initiation sites, a microstructure sample was made from this weld and imaged. A panorama of the sample can be seen in Figure 5-2.

Figure 5-1: Section of tensile bar from weld B-5, showing two separate necking points.
In Figure 5-2, initial deformations can be seen in two primary regions. The first region is severe deformation and necking in the DXZ-TMAZ-HAZ region of the advancing side of the weld. Here, significant deformations have occurred on both sides of the sample in the HAZ and on only one side of the sample in the DXZ and TMAZ. In the second region, a smaller degree of plastic deformation is evident only in the HAZ of the retreating side of the weld. For this sample, the large plastic deformations along the advancing side of the weld indicate that final failure would occur in this area. To further examine this area, samples were cut and examined using SEM. Two sites of interest are pictured in Figure 5-3.

In Figure 5-3 two different sets of slip bands are visible. Although both the main and side faces of the tensile bar were imaged on the advancing side of the weld, only the corner on the thinner face of the tensile sample showed significant flaws that indicate slip bands. Part a) of Figure 5-3 shows deformations on the corner of the sample that likely resulted from small initial flaws. Under stresses applied in testing, these small corner flaws caused stress intensification in the
immediate area. As a result of stress intensification, the flaw became the site of a slip plane that led to plastic deformation increasing the flaw size to that pictured. The slip bands pictured indicate that stress intensification at this point could result in crack formation.

Part b) of Figure 5-3 shows a dispersoid or secondary phase in the weld located near a corner of the sample. Similar to the slip bands pictured in part a), the particle indicates a section of disrupted corner material that results in a weaker cross section. The inconsistency in the material at this point could result in stress intensification around the particle upon static loading. Stress intensification around the particle could cause a cross section around this particle to become a slip plane. Initial deformation at this site could break the bond between the base material and the particle, causing further stress intensification and initiating a crack on this slip band.

5.1.2 Tensile tests of baseline welds
Tensile tests were performed on welds B-1, B-3, B-4, and B-6 of the baseline parameters. Of the four tests, those in B-1, B-4, and B-6 were successful; the test of B-3 recorded compressive strain at several points during loading, suggesting slippage or malfunction of the strain gauge or equipment, thereby invalidating the test. For the three successful tests, the average UTS recorded was 28.84 +/- 1.15ksi. This value is significantly lower than the value of 46ksi for the unaffected base material. The baseline welds fractured at a relatively low ductility with an elongation at fracture of 7.46 +/- 0.61%, which is considerably lower than the unaffected base material value of 17.0%.

The largest drop recorded for tensile properties, however, was in Yield Strength (YS). The recorded value for YS for the baseline welds, 14.17 +/- 1.76ksi, was only half that of the observed UTS. This indicates significant degradation of the material’s innate resistance to plastic deformation in a section of the WAZ. Stress-strain plots for the three successful tensile tests of the baseline parameters are presented in Figure 5-4.
Figure 5-4: Stress-strain plots for tensile samples from baseline welds of parameters 1000RPM and 2.0mm/s, a) for B-1, b) for B-4, and c) for B-6.
In all three tests, rupture occurred along the advancing side of the weld in the HAZ. The location of fracture corresponds to the ultimate low point of microhardness for this set of welding parameters, visible in the microhardness profile in Figure 4-3. In this case, the crack path followed the weld geometry on a slant paralleling the DXZ boundary, suggesting that the crack traveled through areas that have similar time-temperature histories and similar microhardness. When viewed before etching, no evidence was found that suggested the crack leading to ultimate failure was influenced by secondary phases present in the sample.

Figure 5-5: Fractured sample from weld B-1, etched in 3% Barker’s Reagent for 80sec, 50x magnification.

SEM analysis was performed on the fracture surface for the B-4 sample. This identified a ductile fracture where crack initiation appears to have occurred along the smaller face of the cross section. Evidence of fracture ductility can be found in the dimpling across the fracture surface that is pictured in the SEM fracture surface panorama in Figure 5-6. The crack initiation site was identified for this sample in the lower right corner of the panorama, a magnified view of which is shown in Figure 5-7. This figure supports the evidence for crack initiation in cross section corners suggested by Figure 5-3.
5.1.3 Tensile test of weld parameter S2
The tensile test performed on weld B-7 of parameter S2 recorded an ultimate tensile strength of 30.43ksi, roughly 2ksi greater than the average for baseline welds. While this value is lower than the value of 45ksi for the unaffected base material, the gain in strength over the baseline FSW parameters is significant. The ductility recorded at rupture was also higher than that recorded for the baseline weld at 8.17% elongation at fracture, although this value is still well below the unaffected base material value of 17.0% elongation at fracture.

As in samples from the baseline welds, YS suffered the greatest loss of all recorded properties. For the tensile test of B-7, the observed YS was only 16ksi. While this is a gain in comparison to the baseline weld, it is still less than half of the yield strength for the unaffected base material,
suggesting that the WAZ has an inherently low resistance to plastic deformation. It is likely that the slightly higher UTS, ductility, and YS are a result of the greater and more linear microhardness profile for this weld, which indicates that the properties of sample B-7 of weld parameter S2 are not as degraded as for the baseline weld. The stress-strain plot for the tensile test of B-7 is presented in Figure 5-8. This plot is consistent with expectations of ductile aluminum alloys aside from the low yield strength, and is comparable to other tensile work performed on the Al6061 alloy reported in the literature review (Lim et al., 2004; Liu, Fujii, Maeda, & Nogi, 2003).

![Weld B-7, σ vs. ε](image)

In this weld, rupture occurred again along a face paralleling the DXZ outer boundary. This rupture also occurred along the advancing side of the weld, as was the case for baseline samples. The rupture location corresponds to the point of lowest hardness recorded on the microhardness profile for this weld, displayed in Figure 4-5. This suggests that the tensile behavior for this weld is governed largely by the weld microhardness.
Figure 5-9: Fractured sample from weld B-7, etched in 3% Barker’s Reagent for 80sec, 50x magnification.

Scanning electron microscopy was used to image the fracture surface for the tensile sample from weld B-7. The SEM panorama in Figure 5-10 suggests that the fatal crack initiation site was on a corner of the weld. This crack ultimately led to ductile failure, as is evidenced by dimpling of the entire fracture surface similar to that seen for the baseline weld. The corner flaw can be seen in the upper left of Figure 5-10, and a magnified view of this corner is presented in Figure 5-11. In this magnification, it is evident that necking led to a stress concentration that caused crack propagation and rupture. This further reinforces the concepts of initial damage of corners leading to crack propagation and rupture, which is described in Section 5.1.1 detailing imaging of tensile specimen B-5.

Figure 5-10: SEM panorama of B-7 fracture surface, taken at 10x magnification.
5.1.4 Tensile test of weld parameter S3

For the weld of parameter S3 (sample from B-8) the UTS found was 29.02ksi and the elongation at fracture was 6.51%. The UTS for this weld falls in between that for the baseline parameters and parameter S2, while the ductility is the lowest from all three parameters tested. Although the microhardness profile in Figure 4-8 is suggestive of intermediate UTS, the ductility at rupture is not accounted for simply by the lowest magnitude of microhardness in the sample. Here, as was tested in previous research, the ductility is governed by localized strain distributions (Peel, Steuwer, Preuss, & Withers, 2003). The baseline weld minimum microhardness is lower than that of the weld of parameter S3, but the microhardness profile for parameter S3 has a significant consecutive series of low values not observed in the baseline weld or weld parameter S2. With many consecutive points near only 60VHN, there is a much more extended section of weakened WAZ in weld B-8 than the other welds. This section of low microhardness is the determining factor in the ductile properties of the weld, and explains why the lowest ductility observed for all three sets of welding parameters was a sample of parameter S3.

The YS for this weld was 16.5ksi, which is again very low in relation to the unaffected base material, but still the highest YS observed among all three sets of weld parameters. It should also be noted that this weld exhibited the most standard stress-strain curve of all welds, with clear transitions between regions of elastic and plastic deformation. Since the determination of
yield strength is based on a 0.02% elastic deformation, this sharp transition effectively raised the yield strength of the weld in comparison to baseline samples and samples of parameter S2 which exhibit small amounts of early plastic deformation that result in lower YS. The stress-strain curve for the tensile sample of B-8 illustrating these properties is recorded in Figure 5-12.

![Weld B-8, σ vs. ε](image)

**Figure 5-12: Stress-strain plot for weld parameter S3 (weld B-8).**

In Figure 5-13 a panorama of the ruptured tensile sample of weld B-8 is presented. It is evident upon inspection that fracture occurred in the HAZ of the retreating side of the weld. The fracture surface parallels the angle of the DXZ, suggesting that failure is predominantly controlled by the time-temperature history within the WAZ for this sample as well. Fracture in this location is not anticipated when viewing the microhardness profiles for sample B-8 (Figure 4-8). While fracture occurred in an area that is a local minimum of microhardness, it is not in the area of absolute minimum microhardness for the sample, indicating that the fracture may be influenced by other factors.
Figure 5-13: Fractured sample from weld B-8, etched in 3% Barker’s Reagent for 80sec, 50x magnification.

Figure 5-14 presents an image of the fractured sample from weld B-8. Clearly visible is the significant refinement of the secondary iron phase that occurred in the right side of the sample, which includes the DXZ and TMAZ. Along the fracture surface, the numerous secondary-phase-size gouges indicate that these secondary phases may have influenced the crack path or crack initiation in the HAZ. This implies that alterations of the secondary phase can be a determining factor in the location of ultimate failure in FSW in the Al6061-T6 alloy.

Figure 5-14: Side view of fracture surface in B-8 showing weld cross section, polished.
A SEM panorama of the fracture surface of the tensile sample from B-8 is presented in Figure 5-15. SEM imaging of the corners and sides of the welds indicates that cracks may have initiated in these areas. This fracture shows significant ductile characteristics determined by dimpling across the entire face, as was the case for previous samples. It is possible that the crack that propagated and led to ultimate rupture initiated in the upper right corner or along other features of the top side of the fracture surface shown in Figure 5-15. An image of a possible crack initiation site in the upper right corner is found in Figure 5-16. This site showed significant deformation around the corner as well as a three dimensional structure with sharp points.

Figure 5-15: SEM panorama of B-8 fracture surface, taken at 10x magnification.

Figure 5-16: SEM image of crack initiation site in B-7.
5.1.5 Summary of tensile testing

Table 5-1 is a summary of all tensile properties recorded from tensile testing. As mentioned in Sections 5.1.2 through 5.1.4, all of these values are significantly lower than those for the unaffected base material. Table 5-2 describes each property of the three sets of parameters in terms of percentage value of the property for the unaffected base material, facilitating comparison of deviations in properties. For the baseline parameters, only the average value was used to calculate percentage deviations.

<table>
<thead>
<tr>
<th>Property</th>
<th>Baseline (B-1,B-4,B-6)</th>
<th>Parameter 2 (B-7)</th>
<th>Parameter 3 (B-8)</th>
</tr>
</thead>
<tbody>
<tr>
<td>UTS</td>
<td>28.84 +/- 1.15ksi</td>
<td>30.43ksi</td>
<td>29.02ksi</td>
</tr>
<tr>
<td>YS</td>
<td>14.17 +/- 1.76ksi</td>
<td>16ksi</td>
<td>16.5ksi</td>
</tr>
<tr>
<td>E</td>
<td>8270.94 +/- 2038.42ksi</td>
<td>10567ksi</td>
<td>9560ksi</td>
</tr>
<tr>
<td>%el</td>
<td>7.46 +/- 0.61%</td>
<td>8.17%</td>
<td>6.51%</td>
</tr>
</tbody>
</table>

Table 5-2: Tensile properties as a percentage of unaffected base material values for the three different weld parameters

<table>
<thead>
<tr>
<th>Property</th>
<th>Baseline (B-1,B-4,B-6)</th>
<th>Parameter 2 (B-7)</th>
<th>Parameter 3 (B-8)</th>
</tr>
</thead>
<tbody>
<tr>
<td>UTS</td>
<td>63%</td>
<td>66%</td>
<td>63%</td>
</tr>
<tr>
<td>YS</td>
<td>33%</td>
<td>38%</td>
<td>39%</td>
</tr>
<tr>
<td>E</td>
<td>88%</td>
<td>112%</td>
<td>102%</td>
</tr>
<tr>
<td>%el</td>
<td>44%</td>
<td>48%</td>
<td>38%</td>
</tr>
</tbody>
</table>

Table 5-2 illustrates that the largest deviation from unaffected properties occurred for UTS, YS, and elastic modulus E in the baseline weld. The largest deviation in ductility occurred in the weld of parameter S3. For all properties, the weld of parameter S2 had the smallest or close to the smallest deviations from baseline properties. The deviations observed here are reflective of the microhardness profiles and microstructures of the three sets of welding parameters. The microhardness profiles and microstructures described in the prior sections on microstructure and tensile testing in Chapter 4 and Chapter 5 all suggest that the weld of parameter S2 would have the highest UTS and ductility, as was discovered through tensile testing.

While relative degradations are important in assessing the qualities of welds with respect to each other, it is also noteworthy that the deviations from accepted values for the unaffected base
material are still extremely large. For yield strength, all three sets of welding parameters suffered degradation of over 50% from the unaffected base material value. This large change in properties could have significant impact on the use of FSW in this alloy. The only property that did not suffer large degradation was E, with all three sets of welding parameters having deviations less than 15% from the unaffected base material value. The losses in UTS and ductility recorded were similar to those in published studies; however, no study in the reviewed literature has yet suggested changes in YS (Lim et al., 2004; Liu, Fujii, Maeda, & Nogi, 2003).

The differences between properties for the welds of the three different parameters can be explained by differences in the time temperature history of the weld and the totality of fusion in the DXZ and TMAZ. Each of the three sets of welding parameters was selected for testing because it had a significantly different input energy per unit length traveled. As a result, the time-temperature history and degree of stirring in each of these welds was also markedly different. The differences in the time-temperature history and degree of stirring action have been shown to cause corresponding differences in precipitate structures as well as in secondary phase structures in precipitation strengthened alloys (Lim et al., 2004). This study confirms these results. Subsequently, these differences were observed through microhardness profiles performed on welds of all three parameters. This data appears in Figure 4-3, Figure 4-5, and Figure 4-8.

Tensile testing revealed that tensile properties were primarily a function of microhardness curves. To a large extent, the observed flatness and ultimate low points of the microhardness curves for the three sets of welding parameters determined the tensile properties of all welds tested. In addition to microhardness, interactions from coarsened and clustered secondary phases also likely contributed to the mechanisms of ultimate failure of the samples tested. In all samples that were tensile tested, except for sample B-8, rupture occurred at the ultimate low point of microhardness for the weld. Distributions of the secondary iron phase in the ruptured sample from B-8 suggest that rupture at a local minimum of microhardness may have resulted in part from the presence of the secondary phase.
5.2 Analysis of Dynamic Loading: Constant ΔK Fatigue Crack Growth (FCG) Test

A Constant Delta K (CDK) Fatigue Crack Growth (FCG) test was performed on a baseline weld with the fatigue crack oriented to propagate through the WAZ. In this test, the stress intensity applied to the crack tip (ΔK) is kept constant so that the rate at which the crack propagates is a function of residual stresses and microstructure, not of a variable applied stress. During testing, both $K_{\text{residual}}$, the stress intensity at the crack tip inherent in the material due to residual stress, and $da/dN$, the fatigue crack growth rate, were recorded. The FCG rate $da/dN$ was used in this case to determine the material’s natural resistance to FCG through the WAZ. The values of $K_{\text{residual}}$ were used to calculate the residual stress distribution over the WAZ. The calculation was accomplished using a MATLAB model to determine a simple residual stress distribution for the sample tested using the $K_{\text{residual}}$ values recorded during testing. Figure 5-17 displays the results crack growth rate and the residual stress calculated in the CDK test.
In Figure 5-17, the crack length is measured from the tip of the EDM notch. Correspondingly, from crack length 0” to crack length 0.3”, the fatigue crack is propagating through the HAZ. From crack length 0.3” to 0.5”, the crack is propagating through the TMAZ and then through the DXZ to the weld centerline. In Figure 5-17, a large increase in da/dN and a large increase in the residual stress values were observed for all measured crack lengths greater than 0.6”. The deviation in measurements is due to the effects of plasticity in the crack as it becomes large enough to induce plastic deformation in the sample. As a result, values for da/dN and \( \sigma \) cannot be considered valid past a crack length of roughly 0.6” due to the interference of plastic deformation in testing measurements. Since the weld centerline was at 0.5” in this sample, the plastic effects invalidate the testing data that was gathered for the half of the WAZ beyond a
crack length of 0.5”. However, all testing results are still valid for the first half of the WAZ that the crack propagates through from crack length 0” to crack length 0.5”. As a result, only fatigue crack growth and residual stresses will be discussed for the first half of the WAZ the crack propagates through.

The FCG rate and residual stress distributions plotted in Figure 5-17 suggest that the fatigue crack initially propagates through an area of compressive residual stress from a crack length of 0” to a crack length of 0.1”. Compressive residual stresses in outer areas of the HAZ have also been identified in other research of precipitation strengthened aluminum alloys. As the residual stress changes from compressive residual stress to tensile residual stress around a crack length of 0.1”, the crack growth rate rapidly increases resulting from an increase in true stress applied to the crack tip. This has also been found in FCG experiments on other precipitation strengthened aluminum alloys (Bussu & Irving, 2003; Fratini et al., 2009; John et al., 2003).

After the rapid increase in crack growth rate around crack length 0.1”, the crack growth rate changes comparatively slowly. From a crack length of 0.1” to 0.3”, where the crack continues to propagate through the HAZ, the crack growth rate increases steadily, but at a slower rate. The gradual increase in crack growth rate across the HAZ is a function of the increase in residual stress across this zone. However, as the fatigue crack enters the TMAZ and DXZ starting at a crack length of 0.3”, the crack growth rate begins to decrease, even though tensile residual stress continues to increase. This indicates that the fine grained microstructure and higher hardness of the TMAZ and DXZ suppress the crack growth rate. Finally, at the weld centerline at 0.5”, an abrupt decrease in the crack growth rate occurs. This rapid decrease in crack growth rate may be due to the alignment of onion rings in the crack growth direction at the weld centerline or to hardness and other aspects of the microstructure at the weld centerline. At the centerline, the residual stress distribution is nearly at its maximum, with tensile residual stresses of roughly 2.5ksi.

In the early stages of crack propagation through the HAZ, an increase in tensile residual stress results in a continuously increasing crack growth rate. However, as the crack enters the TMAZ and DXZ, the grain size and change in microhardness suppress the crack growth rate despite the rising residual stress. These results from the CDK test indicate that both the microstructure and residual stress in the WAZ affect FCG in friction stir welds in the Al6061-T6 alloy.
6 Conclusions and Future Work

The analyses performed on the microstructure, microhardness, tensile properties, fatigue crack growth, and residual stress in the Al6061-T6 alloy demonstrate relationships between material properties and the response of welds to static and dynamic loading. Comparison of microstructure analysis, microhardness profiles, and sample responses to static loading for the three sets of welding parameters confirms that tensile properties are governed primarily by the microhardness profile but may be influenced by microstructural features such as secondary phase distributions. The results demonstrate that the microhardness profile of a sample is a result of the welding parameters selected, and therefore tensile properties are also determined by the welding parameters. In the analysis of sample response to dynamic loading, microstructural characteristics and residual stress in the WAZ changed fatigue crack growth properties. The residual stress distribution presented could also affect tensile properties.

Conclusions from the analysis of material response to static loading are:

- Microhardness profiles determine tensile properties:
  - UTS is a function of the minimum microhardness.
  - YS is also correlated to minimum microhardness.
  - Ductility is determined by continuous sections of low microhardness, not points.
  - Elastic modulus remains relatively unaffected by FSW.
- Fracture location may be affected by secondary phases.

Conclusions from the analysis of material response to FCG are:

- Compressive stresses in the outer reaches of the HAZ impede FCG.
- Tensile stresses in the inner HAZ increase the FCG rate.
- Though residual stress peaks in the DXZ, the FCG rate decreases in this area, possibly as a result of:
  - Fine grained microstructure in the DXZ.
  - Higher microhardness in the DXZ than in the TMAZ and HAZ.

In general, the coincidence of tensile residual stresses and low microhardness indicates that the inner HAZ is the critical area of the WAZ for applications of both static loading and dynamic
loading in cross-weld scenarios. The conclusions presented in this section may be used to develop methods to enhance specific properties for industrial applications. The selection of the welding parameters used fully determines the tensile behavior of the material. Accordingly, specific parameter combinations may be chosen for industrial applications in order to develop enhanced properties of frictions stir welds in the Al6061-T6 alloy.

There are several suggestions for future work to build upon this study. These include further tests and analysis of sample response to both static and dynamic loads:

- Perform tensile testing on a larger set of welding parameters with multiple tests for each set of welding parameters.
- Perform tensile tests on samples made entirely of the DXZ, in order to determine tensile and ductile properties of this weld zone alone.
- Analyze the residual stress distribution for several sets of welding parameters using the cut compliance method or another accurate method to determine how residual stresses change with welding parameters.
- Perform CDK FCG tests on multiple sets of welding parameters in cross weld orientation to determine how welding parameters affect FCG in Al6061-T6.
7 Distribution of Research Work

A significant set of results is presented in this work. The research behind the data presented was not performed solely by the author, however. Much of this work was in conjunction with, and some solely performed by, a graduate student at WPI, Brendan Chenelle. Shared research work performed by the author and Mr. Chenelle includes:

- Weld sample creation and FSW for all welds
- Microstructure sample creation, polishing and imaging presented in Ch. 4
- Imaging of the tensile samples, both with SEM and optical microscopy presented in Ch. 5

Work performed by the author includes:

- Tensile sample creation
- Tensile testing
- Analysis of tensile data
- Interpretation of all results

Work performed by Mr. Chenelle includes:

- Microhardness profiles for all welds
- Fatigue Crack Growth (FCG) tests and figures

The assistance and distribution of work is further detailed in the Acknowledgements in Ch. 8.
8 Acknowledgements

Several members of the WPI Faculty and several graduate students were instrumental in the completion of this project. This research is the first this author has performed not only in Materials Science, but in any engineering field, and the engagement of these individuals contributed to a dynamic adventure. The significant contributors to this work and their specific roles are listed below.

Professor Diana Lados acted as advisor throughout this project. This process was a long and challenging one for both Dr. Lados and the author, with many disappointments and difficult periods. In retrospect, this might have been expected for a student’s first experience in a relatively new subject. Dr. Lados’ patience, oversight, and persistent demand for the highest quality of results were the primary instruments in driving this work to completion. The author’s learning was immense throughout the course of this project, despite that frustration was also occasionally sufficient to cause the author to consider a career in a non-technical discipline. Many thanks are due to her for bringing the project to the finish line and for all her heartfelt help on many fronts, both in regards to this project and the author’s future in Materials Science.

Brendan Chenelle was pursuing a Masters degree in Mechanical Engineering and working on Friction Stir Welding with the WPI iMdc throughout the duration of this project. Brendan’s contributions to this work are innumerable. Most importantly, Brendan performed the Fatigue Crack Growth (FCG) tests and plots under great time pressure to enable the completion of this work. Brendan also imaged several of the panoramas and generated the microhardness profiles that are used in this study. His assistance in the interpretation of the data and images in this report was invaluable. The welds performed and the entirety of the sample creation process for microstructure samples was a joint effort. However, even the most detailed listing of his contributions would not adequately convey the enormous effect of his presence during this work. The author has learned as much, if not more, from Brendan about Materials Science than from the entirety of his formal education to date. Without Brendan, this project would not likely have survived its first stages. Beyond his continuous assistance, his abundant kindness and freely dispensed advice surpassed all reasonable expectation. The depth of gratitude to him is great and repayment impossible for his contributions.
Dean Richard Sisson provided assistance at several stages of this work. His reassuring comments and kind disposition encouraged the author to persist through the difficult stretches. Dr. Boquan Li and Dr. Libo Wang have earned thanks for their instruction and assistance in the use of WPI’s lab equipment. Torbjorn Bergstrom and Adam Sears provided continual assistance, often giving unusual advice in the operation of machines in WPI’s HAAS technical center, and always avoiding complaint when frequently asked to repeat it. Their presence and guidance contributed immensely to the author’s enjoyment during the hundreds of hours spent in the machine shop.

Two fellow students and friends contributed to this project and deserve mention and thanks. As a T.A. for Advanced Aerospace Materials, Chris Lammi, who received his Master’s degree in Mechanical Engineering during the project, piqued the author’s interest in Materials Science and contributed to the undertaking of this work. He also provided advice and support throughout its duration. Anastasios Gavras, a Ph.D. student at the time this project was performed, was always willing to laugh at the author’s futile attempts at one thing or another and offer instruction on how to better the work.

Again, the author wishes to offer the most sincere thanks to all.


9 References


