WELDABILITY OF DISSIMILAR 5000 AND 6000 SERIES ALUMINUM ALLOY COMBINATIONS

A Thesis

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By

Michele Marie Atwell, B.S.

*****

The Ohio State University

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Master’s Examination Committee:

Dr. John C. Lippold, Advisor

Dr. William A. Baeslack III

Approved by

[Signature]

Advisor

Graduate Program in

Welding Engineering
ABSTRACT

Aluminum because of its superior properties and high strength to weight ratio, is being substituted wherever possible weight savings can be achieved in new vehicles. Because of this increased use of aluminum in automobiles, dissimilar joints are becoming a necessary part of the overall structure of vehicles. Dissimilar aluminum combinations between 5xxx and 6xxx sheet alloys were the topic of this study. Gas tungsten arc welds (GTAW) and resistance spot welds (RSW) were used to characterize the weldability of these dissimilar combinations and the Sigmajig weldability test technique was used to quantify relative cracking susceptibility of each dissimilar combination.

A full matrix of sheet alloys 6111, 6022, 5754, and 5182 were studied. GTA welds between the 5xxx and 6xxx series showed signs of non-uniform mixing which were linked to differences in viscosity and surface tension. The only combination that could not be joined was that of 6111-6111. This combination experienced solidification cracking each time welding was attempted. 6022 did not experience solidification cracking, so it seemed to be very weldable from first inspection. During bend tests it was found that this alloy cracked preferentially in partially melted zone (PMZ) when highly restrained or joined to alloy 5754 or 5182. Natural aging for the dissimilar combinations was evaluated. The fusion zone was shown to age on the 6xxx series side with a decrease in hardness along the centerline and then another peak, reached before aging, for the 5xxx
series side. This decrease in hardness was attributed to an equiaxed zone that formed along the centerline in every combination. This formation of equiaxed dendrites could be due to the nature of solidification along the centerline (gradient and rate effects) and/or heterogeneous nucleation off of titanium particles. Feather crystals were found in combinations 5182 - 5182 and 5754 - 5182 because of their higher magnesium contents. No stress corrosion cracking was found when testing of dissimilar GTAW combinations in 0.5M NaCl solution.

Pickling of the samples before resistance spot welding (RSW) dramatically increased nugget size and consistency. Solidification cracking and porosity were still observed with the pickled material, but not to the degree that they were observed in the as-received material RSW. 6111 - 6022, 6111-6111, 6022 - 6022 formed a very fine equiaxed grain zone along the center of the nugget. This fine structure assisted in nearly eliminating porosity and solidification cracking.

From the base metal Sigmajig weldability tests, it was found that 5182 had the highest threshold stress value for cracking and 6022 had the lowest. It was interesting to note that in the GTA welding experiments 6022 was much more weldable than 6111, but in the Sigmajig tests 6111 was ranked higher than 6022 in it's resistance to crack formation. Dissimilar combinations were also tested and combination 5182 - 6111 at 50% - 50% dilution had the highest threshold stress value. 6022 was shown to have a degrading effect on the threshold stress values because of cracking in the partially melted zone (PMZ). When joined to another alloy, if 6022 was at a higher dilution rate (for instance 75% 6022 25% 5182), the threshold stress would increase slightly over lower dilution rates (for instance 25% 6022 75%5182), with cracking still located in the PMZ.
This was due to higher restraint of the weld pool when a 5xxx series was at the higher dilution because of the higher magnesium content in the weld pool.

Fractography of cracked SigmaJig test samples showed that cracking along the centerline was dendritic in nature. Fractography also revealed that the PMZ fractures were intergranular with a liquid film present along these grains.
To my parents, Gene and Lois, for their undying support, friendship, and unconditional love
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VITA

October 9, 1971  ...................... Born – Denver, Colorado

1994-1995  ......................... Associate Engineer
               Johnson Controls Inc.
               Plymouth, Michigan

1995  ................................. B.S. Metallurgical and Materials Engineering
               Colorado School of Mines
               Golden, Colorado

1996-Present  ....................... Graduate Research Associate
               The Ohio State University
               Columbus, Ohio

FIELDS OF STUDY

Major Field: Welding Engineering
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CHAPTER 1

INTRODUCTION

Government regulations are forcing the automotive industry to produce automobiles that have increased fuel efficiencies for today’s consumer. In contrast, the consumer wants automobiles with increased performance and additional luxury items, which can lead to a decrease in fuel efficiency. One way of achieving both these goals is by reducing the overall weight of a vehicle. Aluminum provides an excellent choice for reducing the overall weight and has been used by many automobile manufacturers. Aluminum has 1/3 the density of steel, excellent corrosion properties and tensile strengths up to 100 ksi. \(^1\) Due to these superior properties, aluminum is being introduced into more automotive components for both structural and non-structural applications.

With this broad introduction of aluminum into production, issues with joining have taken on renewed importance. Porosity and solidification cracking have been known to plague aluminum weldments. Joint efficiencies tend to be low because of the heat affected zones and soft fusion zones that form during welding. Work hardened alloys can never regain strength in these areas, unless the part is strained after welding. The heat-treatable alloys can recover strength in these areas depending on the condition in which they were welded, but it is generally only partial recovery in strength.
Aluminum alloys generally have excellent corrosion properties in the as-received condition. Welding, which changes the microstructure of the as-received material, can have a degrading effect on corrosion resistance. It has been demonstrated, however, that if the correct procedures for welding are followed, these problems can be minimized.

A number of automobile producers are creating aluminum intensive vehicles (AIV) like the production of the all aluminum Mercury Sable by Ford. With the production of these aluminum intensive vehicles and even under current manufacturing schemes with aluminum substitutions, a number of dissimilar joints between aluminum alloys will be required. This is due to the fact that different alloys are developed for different areas of an automobile. 5xxx sheet series are required for structural applications, such as the frame assembly, while 6xxx sheet series are used for body panels. Therefore, joints between these two alloys and many other possible combinations such as an aluminum extrusion to an aluminum casting will exist. Currently little information is available that addresses the topic of dissimilar aluminum joining.

This project is designed to develop a preliminary evaluation of dissimilar joining issues. Initially, gas tungsten arc welding (GTAW) and resistance spot welding (RSW) were investigated with sheet alloys of the 5xxx and 6xxx series. In the second phase of this project, the solidification cracking susceptibility of dissimilar combinations was studied.
CHAPTER 2

BACKGROUND

2.1. The Automotive Industry Driving Force

The government-mandated corporate average fuel economy (CAFE) standard states that new cars must meet a minimum of 27.5 miles per gallon (mpg), an approximate 8.5% improvement on the average fuel economy of 1990. The automobile manufacturers have been attempting to meet this standard by designing concept vehicles that are electrically driven and therefore zero emission output. Hybrid Electric Vehicles, powered by batteries with an alternate power unit for charging when needed, also have a decreased emission output compared to a vehicle which is powered solely by a combustion engine. While mass production of these vehicles will not begin for several years, a more economical solution based on existing technology is needed.

Raising engine efficiency, increasing drive train efficiency, reducing running resistance, and weight reduction are several ways to increase the fuel economy of vehicles within the next several production years. Combustion optimization could increase engine efficiency. Drive train efficiency could be increased by reducing driving loss and reducing running resistance could be controlled by
reducing rolling resistance and increasing aerodynamic performance. Reducing a
vehicles weight by 250kg could lead to a saving of fuel consumption by about 18%.\[^4\]

While increasing engine efficiency and drive train efficiency would assist in
decreasing fuel consumption by small amounts, producers feel that a dramatic change is
needed. Changing the overall weight of vehicles is what the government and consumer
have been wanting, decreased fuel emissions without losing luxury items, like air
conditioning.

In order for automobile producers to be able to reduce the overall weight of new
vehicles, lighter materials are being substituted for heavier ones. Plastics, metal matrix
composites, and aluminum are all considered viable alternatives. For structural and body
panel members, aluminum alloys have been found to reduce weight significantly without
degradation of safety. This is due, primarily, to the fact that aluminum has a high
strength to weight ratio and 1/3 the density of steel. Aluminum is also an attractive
alternative for it's excellent corrosion resistance, tensile strengths up to 100 ksi,\[^1\] good
fabricability and ease of recycling.\[^5\]

With this broad introduction of aluminum into automobiles, it seems reasonable
to expect many joints where two or more dissimilar aluminum alloys must be mated.
This could be in the form of a 6xxx sheet joined to a 5xxx sheet, an extrusion to a sheet,
or an extrusion to a casting. Since this is a relatively new subject to the automotive
industry, no technical articles directly focusing on weldability of dissimilar aluminum
alloy combinations have been published, to date. Therefore, literature relevant to the
physical and welding metallurgy of aluminum alloys used in this study, cracking theories
and weldability test techniques were reviewed.
2.2 Wrought Aluminum Alloy Designations

There are eight designations of wrought aluminum alloys. Each designation is made up of four numbers. The first digit indicates the major alloying element(s), which are listed in Table 2.1. The second digit if different from 0, indicates a modification from the originally registered alloy. The third and fourth digits identify the specific alloy. Following the alloy designation is typically a temper designation which describes the materials thermal and mechanical processing history and relates directly to how the alloy is strengthened.

<table>
<thead>
<tr>
<th>Alloy Series</th>
<th>Major Alloying Elements</th>
<th>Primarily Strengthened by:</th>
</tr>
</thead>
<tbody>
<tr>
<td>1xxx</td>
<td>None</td>
<td>Work-hardening</td>
</tr>
<tr>
<td>2xxx</td>
<td>Copper</td>
<td>Heat-treatment</td>
</tr>
<tr>
<td>3xxx</td>
<td>Manganese</td>
<td>Work-hardening</td>
</tr>
<tr>
<td>4xxx</td>
<td>Silicon</td>
<td>Work-hardening</td>
</tr>
<tr>
<td>5xxx</td>
<td>Magnesium</td>
<td>Work-hardening</td>
</tr>
<tr>
<td>6xxx</td>
<td>Magnesium + Silicon</td>
<td>Heat-treatable</td>
</tr>
<tr>
<td>7xxx</td>
<td>Zinc</td>
<td>Heat-treatable</td>
</tr>
<tr>
<td>8xxx</td>
<td>Other</td>
<td></td>
</tr>
</tbody>
</table>

(1) Can also be strengthened by work hardening.

Table 2.1 Aluminum alloy designation system.

There are two methods by which the alloying elements listed can be dissolved by the matrix. If the atoms of the alloying element are sufficiently small, they can fit into the spaces between the matrix atoms to form an interstitial solid solution. However, the only atoms small enough to do this are hydrogen, nitrogen, carbon and boron. The rest of the elements dissolve into the matrix by replacing an aluminum atom at a lattice point to
form a substitutional solid solution.\textsuperscript{[8]} The addition of these alloying elements results in distortion of the crystal lattice and increases it's internal energy and strength.

As can be seen from Table 2.1, aluminum alloys are divided up into two categories, either work-hardenable or heat treatable. The work-hardenable alloys are generally strengthened by solid solution strengthening of alloying elements and then further strengthened by cold working (strain hardening). When the material is cold worked, for instance, through rolling mills, the mechanical deformation of the metal structure causes crystallographic slip which in turn results in increased resistance to further strain and thus higher strength and lower ductility.\textsuperscript{[9]} The heat treatable alloys are strengthened by dissolving the alloying elements into solid solution and precipitating them throughout the matrix as strengthening particles. This occurs after a series of heat-treating steps are followed for each individual alloy system.

The main concern of this project was focusing on new aluminum alloys being developed for automotive use. It was determined that the 6xxx and 5xxx series sheet alloys were the most used aluminum alloys in the automotive industry and this project focuses on these series and will be discussed in further detail in the sections to follow.
2.3. Physical Metallurgy

2.3.1 5xxx Series

The 5xxx series aluminum alloys are strengthened primarily by strain-hardening with solid solution strengthening provided mainly by magnesium. These alloys typically have the highest strength of the work hardenable alloys.\textsuperscript{[10]} Figure 2.1 is a binary phase diagram for the aluminum – magnesium system.\textsuperscript{[11]} From this figure, it can be seen that magnesium has a solid solubility of 14.9 wt\% (C_{\text{max}}) at the eutectic temperature of 451°C (844°F). This solubility decreases rapidly with decreasing temperature to approximately 2.9 wt\% at 200°C (392°F). This alloy system differs from a precipitation hardenable alloy in that it is incapable of forming stable precipitates at room temperature. Although, magnesium has substantial solubility in aluminum, binary alloys do not show appreciable precipitate hardening characteristics with concentrations below 7 wt\%. Magnesium, however, does provide substitutional solid solution strengthening with good ductility as a result of cold working the material.\textsuperscript{[12]} The 5xxx series alloys also have very good corrosion resistance up to additions of magnesium of 5 wt\%. Alloying additions above this can result in the formation of continuous β (Mg_3Al_2) precipitates at the grain boundaries, which can lead to intergranular corrosion.\textsuperscript{[13]}

The 5xxx series can be work hardened to different degrees as shown in Table 2.2. The appropriate temper designation follows the alloy designation.\textsuperscript{[6]}
Figure 2.1 Aluminum – magnesium binary phase diagram.[11]

<table>
<thead>
<tr>
<th>Strain-Hardened Tempers</th>
</tr>
</thead>
<tbody>
<tr>
<td>H1</td>
</tr>
<tr>
<td>Strain hardened only. The degree of strain hardening is indicated by the second digit and varies from quarter hard (H12) to full-hard (H18).</td>
</tr>
<tr>
<td>H2</td>
</tr>
<tr>
<td>Strain-hardened and partially annealed. Tempers ranging from quarter hard to full hard obtained by partial annealing of cold-worked materials with strengths greater than desired. Tempers are H22, H24, H26, and H28.</td>
</tr>
<tr>
<td>H3</td>
</tr>
<tr>
<td>Strain-hardened and stabilized. Alloys are strain-hardened and then heated at a low temperature to increase ductility and stabilize mechanical properties. Tempers are H32, H34, H36, and H38.</td>
</tr>
</tbody>
</table>

Table 2.2 Strain-hardened tempers.[6]
2.3.2 6xxx Series

The 6xxx series aluminum alloys are primarily alloyed with magnesium and silicon. The additions are added in appropriate portions to form Mg₂Si, magnesium silicide. The solid solubility of Mg₂Si in aluminum is 1.85 wt% (C_{max}) at a eutectic temperature of 595°C (1100°F) as shown in Figure 2.2. This system has a steep decrease in solubility with decreasing temperature, as does the 5xxx series, but because of the silicon addition, stable precipitates form after proper heat treat cycles are conducted. Thus, this is a heat-treatable alloy.

Precipitation hardening heat treatments are designed to produce a uniform dispersion of fine hard coherent precipitates in a softer and more ductile matrix. The following sequence is needed in a precipitation hardened alloy:

1. Solutionize
2. Quench rapidly
3. Age

The material must be solution heat-treated to within the α region to dissolve any particles and to produce a homogenous single phase solution. After solutionizing, a super saturated solid solution is created by rapidly quenching the material. The material is unstable in this condition and precipitation will occur to lower the system's energy. If substantial precipitation occurs at room temperature, the alloy is said to naturally age. Most alloys, however, must be aged at some elevated temperature, below that of the solvus, in order for stable precipitation reactions to occur. This is artificially aging. The tempers that follow the alloy designation are listed in Table 2.3. The general sequence of precipitation in the Al-Mg-Si system is represented by:
SS $\rightarrow$ $\beta''$ $\rightarrow$ $\beta'$ $\rightarrow$ $\beta$ $\rightarrow$ $\beta$(Mg$_2$Si)

Super rods needles plates
Saturated Solution

With primary strengthening caused by precipitation of $\beta''$.$^{[15,16]}$

Figure 2.2 Aluminum – magnesium silicide binary phase diagram.$^{[6]}$
<table>
<thead>
<tr>
<th>W</th>
<th>Solution Treated</th>
</tr>
</thead>
<tbody>
<tr>
<td>T</td>
<td>Age-hardened</td>
</tr>
<tr>
<td>T1</td>
<td>Cooled from fabrication temperature and naturally aged*.</td>
</tr>
<tr>
<td>T2</td>
<td>Cooled from the fabrication temperature, cold worked, and naturally aged.</td>
</tr>
<tr>
<td>T3</td>
<td>Solution-treated, cold worked, and naturally aged.</td>
</tr>
<tr>
<td>T4</td>
<td>Solution-treated and naturally aged.</td>
</tr>
<tr>
<td>T5</td>
<td>Cooled from the fabrication temperature and artificially aged*.</td>
</tr>
<tr>
<td>T6</td>
<td>Solution-treated and artificially aged.</td>
</tr>
<tr>
<td>T7</td>
<td>Solution-treated and stabilized by overaging</td>
</tr>
<tr>
<td>T8</td>
<td>Solution treated, cold-worked and artificially aged.</td>
</tr>
<tr>
<td>T9</td>
<td>Solution-treated, artificially aged and cold worked.</td>
</tr>
<tr>
<td>T10</td>
<td>Cooled from fabrication temperature, cold worked, and artificially aged.</td>
</tr>
</tbody>
</table>

*During natural aging, the aluminum alloy is aged at room temperature.
*During artificial aging, the aluminum is aged at some temperature above room temperature.

Table 2.3 Heat-treatable tempers. [6]
2.4 Welding Metallurgy of Aluminum Alloys

2.4.1 Welding of the 5xxx Series

The absence of precipitate forming elements such as copper and silicon renders the 5xxx series very weldable as compared to the 6xxx. Additions of these alloying elements can lead to solidification cracking and liquation as will be shown later. The joint efficiencies, which can be described as the ultimate tensile strength (UTS) of the weld divided by the UTS of the base material, are generally higher than heat treatable alloys. This is due to the fact that there are no precipitate reactions occurring in the heat affected zone (HAZ) of the non-heat treatable alloys. However joint efficiency drops off slightly with increasing amounts of cold working. Figure 2.3 shows the regions of a weld in a non-heat-treatable alloy.

As shown there is a region of recrystallization, which forms, adjacent to the unaffected base material. This usually occurs at temperatures that are one third to one half that of the melting temperature. The recrystallization temperature will decrease due to increased amounts of cold work, lower working temperatures, smaller starting grain size, and longer hold times. Within this recrystallized zone, new grains nucleate and grow and are considered to be strain free, containing few dislocations only. Because the number of dislocations has dramatically decreased in this region, the strength is much lower than the surrounding unaffected base material.
Figure 2.3  A) Schematic of the HAZ of a 5xxx series alloy. B) Schematic of temperatures reached during welding. C) Schematic of the tensile strength. D) Binary-phase diagram of Al-Mg. [18]
The coarsening region is the region adjacent to the fusion zone. Within this region the recrystallized grains grow to effectively reduce the energy of the system by decreasing grain boundary area.

Reducing the heat input of the welding process, thus increasing the cooling rates can confine the width of these zones. The material therefore will not be exposed to high temperatures for prolonged periods of time.

The fusion zone of the 5xxx series welds is typically the weakest part of the joint. When a weld is stressed, failure will most likely occur in the fusion zone.

2.4.2 Welding of the 6xxx Series

Heat-treatable alloys, because of their high alloying content are susceptible to cracking during welding. Due to the precipitation sequence that is followed for optimal base material properties, these alloys can undergo a wide range of microstructural changes within the HAZ also. These issues are addressed in the following sections.

2.4.2.1 Cracking

There are two types of primary cracking in aluminum welds and specifically within the 6xxx series, solidification cracking and HAZ liquation cracking.

Weld solidification cracking is associated with the micro-segregation of solute and/or impurities that occurs along solidification grain boundaries and subgrain boundaries during the rapid freezing process of welding.\(^{[19]}\) As the alloy cools through the solidification range, \(T_L\) through \(T_S\), solute is rejected at the solid liquid interface and the liquid becomes enriched with solute for systems with \(k<1\). The rejected solute must
be redistributed in the liquid by diffusion according to Case 3 (macroscopic) solidification principles. However, the solidification process during welding is so rapid that the solute cannot re-distribute uniformly in the liquid by diffusion. As a result, the final liquid to solidify is enriched with solute, thereby lowering the melting point locally. This can lead to varying thicknesses of films, depending on the wetting characteristics of the system. When a critical stress is reached, these films can separate and form solidification cracks. If there is sufficient liquid available these cracks may be backfilled with this eutectic liquid. This phenomenon is called crack healing. When failure does occur, it usually occurs along solidification grain boundaries and the fracture surface is dendritic in nature.[20]

HAZ liquation cracking occurs in a region within the HAZ called the partially melted zone (PMZ). This PMZ is located adjacent to the fusion boundary. This zone is produced when solute or impurities intersect, penetrate and liquate grain boundaries during a weld cycle. Cracking forms when these films along the boundaries become too thick to support the stresses applied to the system. Liquation cracking in the PMZ is always intergranular.[17] As expected, higher heat inputs widen the PMZ and make it more prone to cracking.[13]

2.4.2.2 HAZ Degradation

Figure 2.4 shows schematically the changes that occur during welding of the HAZ of precipitation hardened alloys.[17] There are generally four zones of the HAZ as shown in Figure 2.4. Moving from left to right, each zone reaches a lower peak temperature than the previous zone. The overaged zone reaches the lowest peak
temperature, between $T_{\text{overaging}}$ and $T_{\text{solution}}$ and is located adjacent to the unaffected base material. Within this overaged zone, strengthening precipitates, $\beta''$, grow until the solvus is reached for this precipitate. Dissolution of $\beta''$ precipitates then occurs, which is followed by precipitation of a larger and essentially less effective strengthening phase, $\beta'$, the next phase in the transformation sequence.$^{[15,16,21]}$ The result of this transformation is a dramatic decrease in tensile strength. The next zone is a region of re-solutionizing. Once the solvus temperature for the system is reached, all precipitates dissolve. This provides some solid solution strengthening as shown in Figure 2.4. However, since there is no second phase present to pin the boundaries, the grains can coarsen readily. This coarsening causes the slight dip in strength. The last zone is the formation of the PMZ. This region has solute or impurity build up along grain boundaries, which occurs due to penetration or segregation mechanisms. Since this solute provides a second phase, grain growth slows and strength is recovered.
Figure 2.4  A) Schematic of the HAZ in a heat-treatable aluminum alloy.  B) Schematic of peak temperatures reached during welding.  C) Schematic of strength in HAZ.  D) Phase diagram of a heat treatable alloy.\(^\text{[17]}\)
2.5 Weldability Test Techniques

Weldability can be described as “the capacity of a material to be welded under the imposed fabrication conditions into a specific, suitably designed structure and to perform satisfactorily in the intended service”. However, the weldability of materials is a complicated matter and not easily quantified. As described earlier, due to the nature of the alloying elements, aluminum alloys experience solidification cracking in the FZ and/or liquation cracking in the PMZ during welding. One of the important considerations of weldability is the ability of the material to avoid cracking during weld fabrication. In order to avoid crack formation, one must understand under what circumstances cracks initiate, and then a solution, hypothetical or actual, to the cracking problem can be deduced.

There are numerous test techniques which characterize the cracking susceptibility of materials. Of these, only a few are designed to test thin sheet alloys, which is the principal concern for this project. There are four tests referenced frequently in the literature: Lambert test, modified circular patch test, Houldcroft test and the Sigmajig test. The first three tests are self-imposed restraint types, called representative tests. The Sigmajig test is a simulative test in which an augmented stress is applied.

2.5.1 Lambert Test

The Lambert test involves a thin strip of material securely fixtured in a seam welding unit. Autogenous GTA welds are made several passes wide side-by-side along the sample. Dye penetrate examination is used to inspect the sample and then
qualitatively assign a rating from no cracking to several cracks. Although this test has
proved to be good for characterizing the cracking sensitivity of materials qualitatively,
the process was found to be time consuming and operator dependent.[24]

2.5.2 Modified Circular Patch

The modified circular patch test was developed by S.A. David and J.J. Woodhouse in the late 1980's.[24] In this test, a specimen in the form of a disk 50 mm (2 in.) in diameter and 0.65 mm (0.025 in.) thick is inserted into the test fixture shown in
Figure 2.5. First, two circular autogenous GTA welds are made, at varying diameters, in
an inert atmosphere glove box. Secondly, the disc is removed from the fixture and
inverted, and then weld procedures were duplicated. The first welding procedure renders
the microstructure of the material susceptible to cracking. If cracking is observed here,
the second step is not repeated. The cracking index for this weldability test is qualitative.
If the welds do not show any signs of cracking after following the procedures listed
above, the material is classified as weldable. If the smaller diameter weld shows
evidence of cracking and the larger diameter does not, then the material is susceptible to
cracking. If both welds show evidence of cracking, then the material is classified as
highly susceptible to solidification cracking.
Figure 2.5 Schematic of the modified circular patch test.\textsuperscript{[24]}

20
2.5.3 Houldcroft Test

The Houldcroft test technique was developed in 1955 to quantitatively evaluate the solidification cracking susceptibility of aluminum alloys.²⁵ A schematic of the fixture is shown in Figure 2.6. The specimen is held flat against the workbench with a copper clamping strip, which assures the user that there are no external restraints acting on the system and that the specimen is held in place securely. In this test, a manual GTA weld is initiated at the high restraint side of the sample (shortest slot depth) and then moved toward the low restraint (deepest slot depth). The key of this test is to initiate a crack from the edge of the specimen and to have this crack terminate when stress becomes insufficient to continue the crack. A cracking index can be derived by dividing the total crack length by the original length of the specimen.

2.5.4 Sigmajig Test

The Sigmajig test was developed by Gene Goodwin at Oak Ridge National Labs in 1987.²⁶ A schematic and photograph of the set-up are shown in Figure 2.7 and 2.8. The fixture holds a 50 mm x 50 mm (2 in. x 2 in.) square specimen between hardened steel grips while a transverse stress, σ, (load) is applied to this sample. The load is applied by a pair of strain gaged bolts and maintained by stacks of Belleville washers in the load train. The washers provide an adjustable spring constant. After pre-loading, an autogenous GTA weld is produced along the centerline of the specimen. The test is repeated with a new specimen and higher pre-load until centerline cracking is initiated. The stress in which crack initiation is produced is the threshold stress level and is denoted by σ₉₅. Materials with a high threshold level of stress have better resistance to
solidification cracking. Figure 2.9 is a schematic of data from the Sigma jig test and how the threshold stress level is determined. This test was used extensively in this project and more details will be discussed in the sections to follow.

Threshold data for several materials are plotted in Figure 2.10.[27] As can be seen, aluminum alloys tend to have lower threshold stress values than most materials. Sigma jig weldability test data generated by Goodwin for several aluminum alloys is presented in Figure 2.11.[27]
Figure 2.6  A) Schematic of the Houldcroft test sample and fixturing.\textsuperscript{[17]}  B) Photograph of welded Houldcroft test specimen.\textsuperscript{[20,28]}
Figure 2.7 Schematic of the Sigmajig weldability test.
Figure 2.8 Photograph of the Sigmajig weldability test set-up.
Figure 2.9 Schematic of determination of threshold stress from Sigmajig data. Shown are two materials: 1) one which has a low threshold value and high cracking susceptibility 2) one which has a high threshold value and low cracking susceptibility.
Figure 2.10 Range of threshold stress values (σth) for different alloys relative to aluminum alloys. [27]
Sigmajig Cracking Stress Values
Aluminum Alloys

![Bar chart showing cracking stress values for various alloys.](image)

Figure 2.1 | Plot of threshold stress values ($G_{th}$) found during Sigmajig weldability testing of several aluminum alloys. [27]
CHAPTER 3

OBJECTIVES

This was a metallurgical study of the weldability of dissimilar aluminum alloy combinations. The weldability was evaluated through microstructural characterization, mechanical testing and cracking analysis. Detailed principal objectives of this research were as follows:

1. Develop a matrix of aluminum alloys intended for use in the automotive industry.

2. Determine the weldability characteristics using GTAW and RSW processes.

3. Quantify relative cracking susceptibility of each base material and dissimilar combinations of these materials using the Sigmajig test.

4. Investigate the mechanism of cracking by examining the fracture surface of cracks formed during the Sigmajig weldability test.
CHAPTER 4

EXPERIMENTAL PROCEDURE

4.1 Materials

Several producers of aluminum were contacted in reference to the developmental aluminum alloys that they were supplying to the automotive industry. The types and compositions of the alloys selected, based on this survey, are listed in Table 4.1. All materials arrived in sheet metal thickness and the actual thickness was dependent on supplier availability, as listed in Table 4.2. The nominal mechanical properties and respective suppliers are also listed in Table 4.2.

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Mg</th>
<th>Mn</th>
<th>Si</th>
<th>Cu</th>
<th>Fe</th>
<th>Other</th>
</tr>
</thead>
<tbody>
<tr>
<td>1100-H14</td>
<td>-</td>
<td>-</td>
<td>1.0Si+Fe</td>
<td>0.12</td>
<td>-</td>
<td></td>
</tr>
<tr>
<td>3003-H14</td>
<td>-</td>
<td>1.2</td>
<td>-</td>
<td>0.12</td>
<td>-</td>
<td></td>
</tr>
<tr>
<td>5052-H34</td>
<td>2.5</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>0.25 Cr</td>
</tr>
<tr>
<td>5754-0</td>
<td>2.6-3.6</td>
<td>&lt;0.5</td>
<td>&lt;0.4</td>
<td>&lt;0.1</td>
<td>&lt;0.4</td>
<td></td>
</tr>
<tr>
<td>5182-H16</td>
<td>4.0-5.0</td>
<td>0.2-0.5</td>
<td>&lt;0.2</td>
<td>&lt;0.15</td>
<td>&lt;0.35</td>
<td>0.25 Zn</td>
</tr>
<tr>
<td>6022-T4</td>
<td>0.2-0.7</td>
<td>0.02-0.1</td>
<td>0.8-1.5</td>
<td>0.02-0.1</td>
<td>0.05-0.2</td>
<td></td>
</tr>
<tr>
<td>6111-T4</td>
<td>0.5-1.0</td>
<td>0.015-0.45</td>
<td>0.7-1.1</td>
<td>0.5-0.9</td>
<td>&lt;0.4</td>
<td></td>
</tr>
</tbody>
</table>

Table 4.1 Composition ranges in wt% of the alloys selected for this project.[6,29]
<table>
<thead>
<tr>
<th>Alloy</th>
<th>Thickness</th>
<th>Yield Strength</th>
<th>Ultimate Tensile Strength</th>
<th>Elongation</th>
<th>Supplier</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>mm</td>
<td>in.</td>
<td>(ksi)</td>
<td>(ksi)</td>
<td>(%)</td>
</tr>
<tr>
<td>1100-H14</td>
<td>1.25</td>
<td>0.049</td>
<td>17</td>
<td>18</td>
<td>9</td>
</tr>
<tr>
<td>3003-H14</td>
<td>1.00</td>
<td>0.039</td>
<td>21</td>
<td>22</td>
<td>8</td>
</tr>
<tr>
<td>5052-H34</td>
<td>1.58</td>
<td>0.062</td>
<td>31</td>
<td>38</td>
<td>10</td>
</tr>
<tr>
<td>5754-0</td>
<td>1.02</td>
<td>0.040</td>
<td>14</td>
<td>32</td>
<td>26</td>
</tr>
<tr>
<td>5182-H16</td>
<td>1.07</td>
<td>0.042</td>
<td>47</td>
<td>55</td>
<td>6.5</td>
</tr>
<tr>
<td>6022-T4</td>
<td>1.04</td>
<td>0.041</td>
<td>22</td>
<td>37</td>
<td>26</td>
</tr>
<tr>
<td>6111-T4</td>
<td>0.97</td>
<td>0.038</td>
<td>22</td>
<td>42</td>
<td>26</td>
</tr>
</tbody>
</table>

Table 4.2 Alloy thicknesses, typical mechanical properties and respective suppliers. Mechanical properties are nominal values obtained from the producers and ASM handbook vol. 1.\textsuperscript{[30,31]}

The 1100, 3003, and 5052 alloys were used as control samples during later experiments and were not considered part of the experimental matrix of dissimilar combinations. The other four alloys are most frequently used in structural (5xxx) and panelized (6xxx) components of automobiles. The 6111-T4 and 5754-0, supplied by Alcan, were developed with a concurrent goal, to increasing weldability and formability. 5182-H16, is generally used for truck beds because of it’s increased corrosion resistance. 6022-T4 was developed with decreased copper and magnesium contents for increased weldability.\textsuperscript{[31]} Both 6111 and 6022 alloys are supplied to the automotive industry in the T4 temper. This softer temper allows more flexibility with forming operations and then during the paint cycle the materials are aged to a higher strength. Dent resistance is also increased as a result of age hardening during the paint cycle. Since all these alloys are relatively new to the automotive industry, they were chosen for this study.
Figure 4.1 Plan view of 1100-H14 base metal microstructure (Barker’s reagent).

Figure 4.2 Plan view of 3003-H14 base metal microstructure (Barker’s reagent).
Figure 4.3 Plan view of alloy 5052-H34 base metal microstructure (Barker's reagent).

Figure 4.4 Plan view of alloy 5754-0 base metal microstructure (Barker's reagent).
Figure 4.5 Plan view of alloy 5182-H16 base metal microstructure (Barker's reagent).

Figure 4.6 Plan view of alloy 6022-T4 base metal microstructure (Barker's reagent).
Figure 4.7 Plan view of alloy 6111-T4 base metal microstructure (Keller’s reagent).

<table>
<thead>
<tr>
<th>Combination Number</th>
<th>Materials Joined</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>6111</td>
</tr>
<tr>
<td>2</td>
<td>6111</td>
</tr>
<tr>
<td>3</td>
<td>6111</td>
</tr>
<tr>
<td>4</td>
<td>5754</td>
</tr>
<tr>
<td>5</td>
<td>5754</td>
</tr>
<tr>
<td>6</td>
<td>6022</td>
</tr>
<tr>
<td>7</td>
<td>6022</td>
</tr>
<tr>
<td>8</td>
<td>6111</td>
</tr>
<tr>
<td>9</td>
<td>5754</td>
</tr>
<tr>
<td>10</td>
<td>5182</td>
</tr>
</tbody>
</table>

Table 4.3 Combinations of selected aluminum alloys.
Base metal microstructures for all the alloys used in this study are shown in Figures 4.1 – 4.7. The matrix of dissimilar aluminum alloy combinations, compiled from the alloys listed above, is listed in Table 4.3. There are six dissimilar combinations along with the four similar combinations. The four similar combinations were used as standards to compare any changes when welding dissimilar combinations.

4.2 Welding Procedures

This was primarily a metallurgical study, with characterization of weldments the main priority. GTAW and RSW were selected because of their applicability to thin sections and/or use in the auto industry. Nominal process conditions were selected that produced acceptable welds. The welding procedures are described briefly in the following sections.

4.2.1 Gas Tungsten Arc Welding

This process is not used frequently in the automotive industry because of slow process speeds. However, the gas tungsten arc (GTA) welds were conducted to give an understanding of the weldability issues associated with welding dissimilar aluminum alloy combinations. A schematic of the GTA welding set-up is shown in Figure 4.8. The welding torch was moved using a Jetline sidebeam. This semi-automatic process was controlled by an Arc Length Control (ALC) and a carriage control, with a potentiometer, to control the travel speed. The constant current power supply employed was a Hobart Cyber Tig. The polarity selected for these experiments was direct current electrode
negative (DCEN), otherwise known as straight polarity. This process is one of the most useful and reliable methods for welding aluminum.\textsuperscript{[32,33]} DCEN produces narrow, deep welds with short arc lengths for aluminum alloys, which are easily maintained in the semi-automatic process used. No cleaning action is produced by this polarity, such as when using AC, and therefore the samples must be cleaned to remove any oxide layer before welding, which is described in later sections.

4.2.1.1 Shielding Gas

Helium, although more expensive than argon, was selected as the shielding gas. A hotter arc is provided by helium which allows increased travel speeds to be accommodated.\textsuperscript{[32,33]} Since welding was conducted on thin sheet material (1mm), full
penetration welds were achieved. Therefore, shielding was provided both by a primary and a back shielding. The primary gas flows through the torch and the backing gas, supplied by a separate helium bottle, flows through the copper backing bar and shields the root side of the weld.

Since helium does not provide a stable environment for arc initiation, high frequency (HF) was incorporated in assisting the initiation of the arc.\textsuperscript{[32,33]} After initiation the arc proceeds in a stable fashion.

4.2.1.2 Joint Configuration

The joint configuration used was a simple butt joint welded on a removable copper backing bar. The sheets were first sheared into 180 mm x 100 mm (7 in. x 4 in.) specimens, with the 100 mm side sheared such that the rolling direction was parallel to this side. The 180 mm length was chosen so that sufficient room would be allowed for two metallographic samples, two bend samples and start and end tabs. The 100 mm length was selected so that after welding a transverse cross section of the weld would be 200 mm which is needed for proper fixturing in the bend unit. Figure 4.9 is a schematic of this configuration.
Figure 4.9 Schematic of GTAW sample layout.

4.2.1.3 Pickling

The excellent corrosion resistance provided by aluminum alloys is due to the fact that they form a protective oxide layer. Unfortunately, this layer is very difficult to weld through, as it is a good insulator. In order to most efficiently weld these samples, this oxide layer must be removed. The samples were cleaned using a pickling procedure, which removes a thin layer of surface material. The procedure for pickling is as follows:

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1. A 5% sodium hydroxide solution was heated to 71°C (160°F).

2. The samples were placed in the solution for one minute.

3. The samples were removed and rinsed with water.

4. Samples were then placed in a nitric acid rinse for 30 seconds. This removes any smut that may have formed.

5. The samples were removed and rinsed for 30 seconds in a water bath.

6. The samples were allowed to air dry.

This process left the aluminum with a dull and clean finish. Welds were made within three hours of pickling to ensure minimum oxide re-formation.

4.2.1.4 GTA Welding Parameters

The welding parameters used for GTA welding are listed in Table 4.4. There is a current range listed, due to the different thicknesses of materials welded. The electrodes were selected based on the current used in welding. For example at 60 amps the 1 mm tungsten would be used and at the higher amps the 1.60 mm tungsten would be used.
<table>
<thead>
<tr>
<th>Current</th>
<th>60-100 Amps</th>
</tr>
</thead>
<tbody>
<tr>
<td>Voltage</td>
<td>12 Volts</td>
</tr>
<tr>
<td>Travel Speed</td>
<td>12.7 mm/s (30 ipm)</td>
</tr>
<tr>
<td>Shielding Gas Flow</td>
<td>40 cfh</td>
</tr>
<tr>
<td>Backing Gas Flow</td>
<td>30 cfh</td>
</tr>
<tr>
<td>Electrode Type</td>
<td>2% Thoriated Tungsten</td>
</tr>
<tr>
<td>Electrode Diameter</td>
<td>1.0 mm or 1.60mm (0.040 in. or 0.0625 in.)</td>
</tr>
<tr>
<td>Gas Cup Size</td>
<td>#12 (19 mm inner diameter)</td>
</tr>
</tbody>
</table>

Table 4.4 GTA welding parameters.

4.2.1 Resistance Spot Welding

Resistance spot welding (RSW) is used frequently in the automotive industry since it has extremely fast cycle times and relatively little metal preparation needed beforehand. The equipment used in this study was a 200 kVA single phase AC press type direct acting pedestal welding machine. The welding process was controlled by a Medar controller and monitored with a Miyachi Weld Checker. The welding head had an upper electrode which was moved vertically in a straight line. The lower electrode remained stationary. Truncated, copper zirconium electrodes with a 3/8 in. diameter face and a slightly radiused tip were used. Two series of welds were produced. As received material was welded first and then based on the results, a second set with pickled samples was conducted.
4.2.2.1 RSW Parameters

Parameters that were given in the Aluminum Associations Theory and Practice of Welding Aluminum\textsuperscript{36} book were used as starting parameters and revised as needed. The parameters used for each combination are listed in Table 4.5. These parameters were developed according to the set-up button diameter listed in Theory and Practice.\textsuperscript{36} To check for the correct button diameter, 2 50 mm x 50 mm (2 in. x 2 in.) samples were placed on top of one another and welded, Figure 4.10. The two samples are then pulled apart. The button diameter is the amount of material fused together. When the samples are pulled apart, the base metals will tear around the nugget if the parameters are set correctly. If the base metals do not tear (the nugget does not pull out), the parameters need to be adjusted. When a nugget does pull out, the diameter is measured and compared to the suggested diameter. Parameters can be fine-tuned to reach the desired nugget pull out diameter.
<table>
<thead>
<tr>
<th>Alloy Combination</th>
<th>Pickled Material</th>
<th></th>
<th>As-Received Material</th>
<th></th>
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</thead>
<tbody>
<tr>
<td></td>
<td>Secondary Current</td>
<td>Electrode Force</td>
<td>Secondary Current</td>
<td>Electrode Force</td>
</tr>
<tr>
<td></td>
<td>(kAmps)</td>
<td>(lbs)</td>
<td>(kAmps)</td>
<td>(lbs)</td>
</tr>
<tr>
<td>6111</td>
<td>27.2</td>
<td>700</td>
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<td>700</td>
</tr>
<tr>
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<td>700</td>
<td>19.0</td>
<td>700</td>
</tr>
<tr>
<td>5754</td>
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<td>700</td>
<td>20.7</td>
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<tr>
<td>5182</td>
<td>22.4</td>
<td>700</td>
<td>21.0</td>
<td>600</td>
</tr>
</tbody>
</table>

Table 4.5 RS welding parameters. Constant parameters are: 8 weld cycles, 60 squeeze cycles (1 second), and 30 hold cycles (0.5 second).

Figure 4.10 Schematic of RS welding set-up.
4.3 Sigmajig Weldability Experiments

The schematic for the Sigmajig test was previously shown in Figure 2.7. A transducer box connected to the strain gauged bolts, allowed the applied load on these bolts to be read in pounds. In order to be sure that this was the actual load, the strain gauged bolts and the transducer box were calibrated before any testing was conducted. Calibration was conducted using an Instron tensile machine and applying a known load to the bolt. If the transducer reads a different load than that applied, it can be adjusted to read the actual load applied.

The Sigmajig weldability test is based on the principle that a teardrop shaped weld pool can promote a solidification crack along the centerline. The Sigmajig test measures the formation of these cracks under incremental augmented stress levels. Therefore it is necessary to set the welding parameters so that a tear drop shaped weld pool forms every time for every base metal or combinations of base materials. The parameters for the autogenous GTAW used in the Sigmajig weldability test are: 50 amps, 14 volts, and 55 ipm. These parameters resulted in a tear drop shaped weld pool for every combination.

The augmented stress was increased sequentially by 50 lbs/bolt until cracking was initiated. When cracking was observed, the stress was decreased by increments of 10 lbs/bolt until cracking disappeared. The stress that causes crack initiation is described as the threshold stress value for cracking and this value was the value collected from the Sigmajig experiments. Samples were tested at 10 lbs/bolt above and below this threshold level and repeated three times to verify that this was actually the threshold stress value. The threshold values given are +/- 10lbs.

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4.3.1 Sample Preparation

Base metal weldability studies were conducted using the Sigmajig test on the four alloys chosen for this study. The control specimens, 1100, 3003, and 5052, were also tested to compare to previous test results from Goodwin.\textsuperscript{27} Approximately 30 specimens of 50 mm x 50 mm (2 in. x 2 in.) were sheared from each alloy. Since the alloys had a wide range of thicknesses, all the materials were pickled down to 0.97 mm (0.038 in). This is very important so that the same welding procedures can be followed resulting in the same heat input and bead geometry for each alloy.

In order to evaluate the weldability of dissimilar combinations using the Sigmajig test, the two dissimilar alloys must first be joined well enough to support the loading stress and with a bead width small enough to be consumed by the GTA weld that is used in this test. The laser welding process was used to do this. Because it was thought that the laser welding might introduce a factor of residual stress that could alter the test, similar combinations were also welded together and tested. These results were then compared to the base metal samples.

4.3.2 Laser Welds

Samples of 180 mm x 32 mm (7 in. x 1.25 in.) were sheared from the alloys collected. The samples were then pickled down to approximately 1 mm (0.040 in.) using the sodium hydroxide bath discussed earlier.
<table>
<thead>
<tr>
<th>Alloy Combination</th>
<th>Travel Speed</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>(lpm)</td>
</tr>
<tr>
<td>6111 6022</td>
<td>240</td>
</tr>
<tr>
<td>6111 5182</td>
<td>360</td>
</tr>
<tr>
<td>6111 5754</td>
<td>360</td>
</tr>
<tr>
<td>5754 5182</td>
<td>360</td>
</tr>
<tr>
<td>5754 6022</td>
<td>360</td>
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<tr>
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<td>6022 6022</td>
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<tr>
<td>6111 6111</td>
<td>240</td>
</tr>
<tr>
<td>5754 5754</td>
<td>360</td>
</tr>
<tr>
<td>5182 5182</td>
<td>360</td>
</tr>
</tbody>
</table>

Table 4.6 Laser welding parameters. Constant parameters are, 0.4 mm spot size at focus, 3.0 kW, 80 cfm helium primary and backing gas, pull angle of 15°.

Samples were placed in a butt joint configuration and then welded using a Nd:YAG 3.0 kW laser. The welding parameters are shown in Table 4.6. After welding, samples of 50 mm x 50 mm (2 in. x 2 in.) were sheared out of the laser welded specimens. Some of the start and stop tabs were saved to characterize the microstructures of laser welding dissimilar combinations. Before testing was conducted using the Sigmajig test, the samples were pickled down to 0.038”.

4.3.3 Dilution Effects

When welding the dissimilar combinations, dilution was varied from 50% - 50% to 75% - 25% and 25% - 75% by moving the sample in the fixture so that the GTAW electrode was offset to one side of the laser weld. It was important to make sure
that the laser weld was completely consumed by the GTA weld. Figure 4.11 shows the different dilution situations.

4.4 GTA Weld Analysis

As previously shown in Figure 4.9, the GTA welds were divided up into bend test and metallographic samples for analysis. The techniques used for each analysis are described in the following sections.

4.4.1 Bend Tests

The wrap around bend test was used to a) qualify the welding procedures used and b) to test weld ductility. A schematic of the bend unit used is shown in Figure 4.12. Both transverse and longitudinal bend tests were conducted according to AWS Structural Welding Code of Aluminum.[37] This standard lists the suggested die diameter for the 6xxx series as 16 mm (0.63 in.) diameter, 5182 has a suggested diameter of 6.35 mm (0.25 in.), and for 5754 the diameter suggested was 3.8 mm (0.15 in.). Since dissimilar combinations were being studied, it was decided that the smaller diameter of the two suggested diameters be used to test for bend ductility. The 3.8 mm diameter die could not be accommodated in the bend test unit used because insufficient fixturing would not allow the samples to be bent without “kinking”. The die of 6.35 mm diameter did not encounter this problem and so the 3.8 mm die was replaced by the 6.35 mm die. The only combination to be tested with the 16 mm die was the 6111-6022 combination. This combination was also tested using the 6.35 mm die so that all the results could be compared. Transverse and longitudinal face and root bends were conducted using a die
diameter of 6.35 mm for every combination. The strain produced by the bending can be calculated using the following formula:

$$
\varepsilon (\%) = \left( \frac{t}{2r + t} \right) \times 100
$$

The strain from the die diameter of 6.35 mm is 13% for these thin sheet alloys. The strain from the die diameter of 16 mm is only 5.7%.
Figure 4.11 Schematic of how dilution percentages varied with the movement of the sample in the Sigmajig test fixture.
Figure 4.12 Schematic of wrap around bend testing unit.[38]

4.4.2 Metallographic Samples

Plan (top view) and transverse metallographic samples were sectioned from the welds using an abrasive water-cooled cut-off wheel and wet ground using 320 and 600 grit silicon carbide paper. The samples were then polished using 9μm and 3μm diamond pastes. A final polish was produced with colloidal silica on the Vibromet for approximately one hour. The samples were ultrasonically cleaned between each grinding and polishing step with ethanol. Depending on the microstructure to be viewed, two etchants could be used, Keller’s and Barker’s reagents. Keller’s reagent is composed of 2.5% HNO₃, 1.5 HCl, 1% HF and 95% distilled water.[39] The sample is immersed in Keller’s reagent for approximately 10-15 seconds while keeping the sample in constant
motion while immersed. This etchant doesn’t reveal microstructures of the 5xxx series very well, but readily reveals those of the 6xxx. Barker’s reagent is 5% HBF₄ with 95% distilled water, and is an anodizing-electrolytic etchant. To use Barker’s reagent, a cathode is placed in the solution. With the anode in contact with the sample, both are immersed in the solution. To avoid pitting, the sample was taken out of the solution for a brief moment and then immersed in the solution which was repeated several times during the duration of approximately 90 seconds depending on the area of the sample to be etched. The voltage used was 20 v which, with maximum sample size of 25.4 mm, did not produce amperage over 0.5 amps. Since this process produces an anodic film on the surface, polarized lighting must be used to reveal the different structures of the sample. Macrostructures, microstructures, solidification patterns, fusion zone, HAZ, and base metals, were observed using optical light microscopy. Fractography, when cracking occurred, was analyzed using the scanning electron microscope in secondary electron imaging mode (SEM/SE). A Phillips XL-30 CP SEM was used with an accelerating voltage of 20 KV, with a beam current (spot size) of 5.0, and a working distance of 10 mm, which were found to produce high quality images of aluminum alloys.

4.4.3 Corrosion Tests

Corrosion tests were performed on the longitudinal face specimens, previously bend tested at 13% strain, of similar and dissimilar welds. To ensure that no cracks had formed due to the bend test, a visible dye penetrant inspection method was used. This is a four step process:
1) Clean samples with Cleaner/Remover and let dry for approximately 15 minutes,

2) Spray the visible dye penetrant on the bend surfaces and allow to set for 15 minutes,

3) Wipe of excess penetrant by spraying the Cleaner/Remover into a rag and wipe the sample,

4) Spray sample with Developer and allow to dry. Any cracks that may have been present were clearly seen. Rinse the samples in an acetone bath.

A Teflon bolt was inserted into the bend samples approximately 25.4 mm from the bend area, to ensure that the samples were kept at a constant strain at all times. The samples were then immersed in 0.5M NaCl solution for several weeks, with weekly inspections of the sample. A final inspection of the weld samples were made using the dye penetrate test technique as mentioned previously.

4.4.4 Natural Aging Experiments

Natural aging experiments were conducted on the GTA welds to show the effects of aging. Two samples of each combination were sectioned, transverse to the weld, using an abrasive water-cooled cut-off saw, ultrasonically cleaned in acetone and cold mounted immediately after welding. The samples were then ground through 600 grit paper and polished using the 9µm and 3µm diamond pastes. Approximately two hours after welding, a hardness traverse was made across the weld using a Vickers

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indenter on a microhardness machine (0.2 kg, 10 second load). After 2 months of aging, the samples were re-polished and another hardness profile was made across the weld.

Another set of GTA welds were made and superficial hardness readings taken approximately every hour for the first 12 hours and then every 6 hours for a couple of days and then once a day. This will show the dynamic natural aging response of welds.

4.5 RS Weld Analysis

RS welds were examined using metallographic samples from each weld, Figure 4.13. Each sample was sectioned, in the middle of the weld, using a water-cooled abrasive cut-off wheel and then wet ground with 600 grit carbide silica paper. The samples were then polished through 9μm and 3μm and final polished with colloidal silica on the Vibromet. Etching procedures listed in section 4.4.2 were followed here.

Inspection of the welds was conducted using optical light microscopy techniques. Welds were examined for cracks, porosity, and consistent nugget diameters.

Figure 4.13 Schematic of RS weld metallographic sample.
4.6 Analysis of the Sigmajig Weldability Test

Samples were visually inspected for cracking immediately following the test. Cracks, when formed, were then viewed using a SEM to characterize the cracking phenomenon that had taken place. If the crack that had formed was less than 100% of the sample, the crack needed to be open up to be analyzed. This was done by placing small slits at both ends of the sample and applying a tensile force to one side of the sample.

Metallographic examination was conducted on selected welds. The procedures followed are outlined in section 4.4.2.
CHAPTER 5

RESULTS & DISCUSSION

5.1 GTA W

GTA welding was performed on the similar and dissimilar combinations listed in Table 4.3. Several zones were formed upon welding and these are shown in Figure 5.1, while the formation of these zones was discussed earlier in Chapter 2. It was also observed that when autogenously welding the dissimilar combinations, non-uniform mixing in the fusion zone occurred in every combination between 5xxx and 6xxx series. Due to the different response of the 5xxx and 6xxx series to certain etching techniques, this non-uniform mixing is evident, which is shown in Figure 5.2. This can be attributed to the differences in viscosity and surface tension of 5xxx and 6xxx series. As shown in Figure 5.3, the surface tension decreases with increasing alloying content of magnesium. It can be assumed from this that the 5xxx series alloys have a lower surface tension (at approximately 600 dynes/cm) than that of the 6xxx series (at approximately 680 dynes/cm). Viscosity of these materials also changes as a result of alloying contents as shown in Figure 5.4. Viscosity, as shown in this figure, drops off with increasing amounts of magnesium. The 5xxx series viscosity could be approximated to be 1.07 centipoise where as for the 6xxx series it is approximated to be 1.12 centipoise.
Thermal expansion can also be affected by alloying content. As the magnesium alloying content increases the thermal expansion coefficient increases as well.\textsuperscript{40} Thus, the 5xxx series can be considered to have a slightly higher thermal expansion coefficient effect than the 6xxx for the same heat input.\textsuperscript{40,41} The difference in material thickness between 5754 (1.02 mm) and 6111 (0.97 mm) along with the thermal expansion differences lead to the uneven root side penetration of alloy 5754, as shown in Figure 5.2.

5.1.1 Solidification Structures

Solidification structures produced during GTA welding of aluminum alloys have been described by Ganaha et al.\textsuperscript{42} and will be referred to at various times throughout this section. Figure 5.5 shows a plan view of the solidification structure of 5182 welded autogenously to 5182. As can be seen from this figure, 5182 characteristically solidifies as columnar grains with large equiaxed grains at the centerline. This combination also exhibits a feathery grain (or feather crystal) formation which will be discussed in a later section. Combination 5754 – 5754 forms an even wider band of equiaxed grains at the centerline with smaller grain diameters than 5182. 6022 is the only alloy that didn’t exhibit equiaxed grains along the centerline because a lower heat input was used to join this particular combination. This combination solidified completely as columnar dendritic, with nucleation of a few new grains within the columnar grains, as shown in Figure 5.7. Figure 5.8 shows the combination of alloy 6111-6111 which formed the widest band of fine equiaxed grains, however this alloy could not be welded without solidification cracking along the centerline. Macrographs of the solidification structure of the dissimilar combinations are shown in Figures 5.9 – 5.14.
All combinations formed a wide band of equiaxed grains along the centerline with columnar growth up to this band formation. The microstructures of these dissimilar combinations are focused on in the following section.
Figure 5.1 Regions of a thin sheet autogenous GTA weld.

Figure 5.2 Representative macrostructure of a dissimilar weld between a 5xxx and 6xxx series alloys (Keller’s reagent).
Figure 5.3  Effects of alloying content on surface tension of 99.99% aluminum at 700°C in argon.\[^{40}\]

Figure 5.4  Effect of alloying additions on the viscosity of aluminum.\[^{40}\]
Figure 5.5 Plan view of a GTA weld between 5182 and 5182 (Barker’s reagent).

Figure 5.6 Plan view of a GTA weld between 5754 and 5754 (Barker’s reagent).
Figure 5.7 Plan view of a GTA weld between similar alloys 6022-6022 (Barker’s).

Figure 5.8 Plan view of a GTA weld between 6111 and 6111 (Keller’s reagent).
Figure 5.9 Plan view of a GTA weld between 6111-5754 (Barker’s reagent).

Figure 5.10 Plan view of a GTA weld between 6111-5182 (Barker’s reagent).
Figure 5.11 Plan view of a GTA weld between 6111 and 6022 (Barker’s reagent).

Figure 5.12 Plan view of a GTAW between 6022 and 5754 (Barker’s reagent).
Figure 5.13 Plan view of a GTA weld between 6022 and 5182 (Barker’s reagent).

Figure 5.14 Plan view of a GTA weld between 5182 and 5754 (Barker’s reagent).
5.1.2 Microstructural Characterization

Figures 5.15 – 5.20 show each dissimilar combination and the key areas that formed during autogenous GTA welding. The combination of 5754 – 6111 is shown in Figure 5.15. Within the 5754 HAZ there is no formation of a PMZ. Alloy 6111 however contains a very large PMZ. This combination exhibits an equiaxed grain zone along the centerline, with each grain averaging 200 μm in diameter. No problems were associated with this combination.

Figure 5.16 shows a weld between alloys 5182 and 5754. 5182 has considerable liquation in the PMZ. This liquation formation can be attributed to two factors: 1) 5182 has a much higher alloying content of magnesium then the other alloys and 2) the fact that the grain size is extremely small in this region due to the recrystallization of the cold worked grains. The increase in magnesium content will certainly increase the amount of segregation that can occur at the grain boundaries. The small grains also play a key role in this segregation by leaving a small distance for the diffusion of magnesium atoms. From the PMZ, growth was columnar to the equiaxed zone. This equiaxed zone contains large grains, on the order of 400 μm in diameter, which is the largest of the combinations joined. The 5754 has similar columnar growth to the center of the fusion zone where an equiaxed region is reached, but has no PMZ.

Alloy combination 6022 – 5182 GTAW is shown in Figure 5.17. As can be seen from the figure, alloy 6022 experiences severe PMZ cracking which then continues in the FZ. The PMZ extends into the HAZ the same distance as 6111, but contains a larger overall grain size. This has been linked to an increasing susceptibility to cracking in this region, which will be discussed in a later section. When joined to 5182, a heavily
strain hardened material, the restraint that is generated during welding is too much for the PMZ of the 6022 to compensate for and cracks form. The same phenomenon is shown with combination 6022 – 5754 in Figure 5.18. Even though 5754 is in the annealed condition, it provides a high enough restraint so that the liquid films along the grain boundaries in the PMZ of 6022 give way to the solidification induced strains. However, cracking is not as severe as in the previous case.

Figure 5.19 shows the weld between 6111 and 6022. 6111 experiences a wide region of eutectic formation along grain boundaries in the HAZ and also undergoes cracking within this PMZ. 6022 has a large PMZ formation, but with less eutectic formation and no visible cracking.

The last combination, shown in Figure 5.20, between 6111 and 5182 is similar to the combination of 6022 - 5182. 6111 undergoes partial cracking in the PMZ because of the increased restraint level produced by 5182. However, the cracking within combination 6111-5182 is not nearly as severe as the cracking seen in combination 6022 – 5182.
Figure 5.15 Plan view of GTAW between aluminum alloys 5754 and 6111 (Barker's reagent).
Figure 5.16 Plan view of GTAW between aluminum alloys 5182 and 5754 (Barker's reagent).
Figure 5.17 Plan view of GTAW between aluminum alloys 6022 and 5182 (Barker’s reagent).
Figure 5.18 Plan view of GTAW between aluminum alloys 5754 and 6022 (Barker’s reagent).
Figure 5.19 Plan view of GTAW between aluminum alloy 6111 and 6022 (Barker’s reagent).
Figure 5.20: Plan view of GTAW between aluminum alloys 6111 and 5182 (Barker's reagent). Equiaxed zone along the centerline.
5.1.3 Predictive Methods

5.1.3.1 Solidification Cracking

Many techniques can be used to predict the susceptibility to cracking of certain aluminum alloys. These techniques will be used here to link the cracking observed in GTAW to the theories of cracking.

The fraction eutectic \( f_e \) formed at the end of solidification along solidification grain boundaries and/or subgrain boundaries, can be calculated for each alloy and then based on these amounts cracking susceptibility can be ranked. Fraction eutectic calculations are shown in Table 5.1 and drawn from the Scheil equation developed for Case 2 solidification. Earlier research had focused on the fact that maximum cracking susceptibility occurred at \( C_{\text{smx}} \), because this is where the largest solidification temperature range existed. However, the data generated was for equilibrium solidification conditions where as the solidification that takes place in a weldment is far from equilibrium conditions. During welding, the actual composition for maximum cracking is lower than that of \( C_{\text{smx}} \), as shown in Figure 5.21.\(^{42}\) Dowd, Singer, and Jennings\(^{44, 45}\) developed plots of cracking susceptibility as a function of alloying additions based on non-equilibrium solidification. Fraction eutectic calculations, for binary alloys, at maximum cracking compositions are based on these diagrams shown in Figure 5.22. When comparing those values that have been calculated for \( f_e \) at maximum cracking compositions \(^{43}\) and for \( f_e \) for each alloy, it is easily seen that the magnesium in the 6xxx series doesn’t play a significant role, but the silicon content certainly does. Alloy 6022 is predicted to produce 6.63% silicon eutectic where as the \( f_{e-\text{Si}} \) at maximum cracking has been estimated at 3.9%. This alloy produces enough eutectic that during
solidification if a crack forms, it is highly likely that it would be backfilled (healed) with eutectic. Alloy 6111 is predicted to only produce 3.19% silicon eutectic in the last stages of solidification, which is very close to the amount needed for maximum cracking. This is probably the primary reason that during autogenous GTA welding of 6111, a solidification crack forms each and every time as shown in the previous section.

<table>
<thead>
<tr>
<th>Alloys</th>
<th>$f_{\text{eutectic}}$ (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Al-Mg</td>
</tr>
<tr>
<td>6022</td>
<td>0.088</td>
</tr>
<tr>
<td>6111</td>
<td>0.14</td>
</tr>
<tr>
<td>5754</td>
<td>1.45</td>
</tr>
<tr>
<td>5182</td>
<td>3.37</td>
</tr>
</tbody>
</table>

Scheil equation: $C_s = kC_0[1-f_s]^{k-1}$
Solved for $f_e$: $f_e = \left( \frac{C_e}{C_0} \right)^{1/(k-1)}$

Table 5.1 Calculation of fraction eutectic for binary alloys.
Figure 5.21 Schematic showing non-equilibrium and equilibrium compositions that yield maximum cracking. [43]

Figure 5.22 Cracking susceptibility versus alloying content for various binary alloys for non-equilibrium conditions. [44, 45]
Alloy 5754 was calculated to have 1.45% magnesium eutectic at the end of solidification. The \( f_{e-Mg} \), which produces maximum cracking for the Al – Mg system, is 2.4%. The eutectic produced in 5754 is lower than that needed for cracking and is probably why no cracking was seen in this alloy during the GTA welding experiments. Alloy 5182 has a higher formation of magnesium eutectic at 3.37%. This alloy doesn’t form any cracking as noted from the GTA welding experiments because of this substantial amount of eutectic formation, which aids in the healing of any cracks that do form.

A. Kostrivas and J. Lippold\(^{46}\) developed a plot for maximum cracking susceptibility during non-equilibrium solidification conditions by normalizing the composition with \( C_{\text{max}} \). This figure was redrawn to include the cracking susceptibility of silicon at various compositions. Table 5.2 shows the normalized compositions of the alloys under investigation, which were then superimposed on the chart as shown in Figure 5.23. Magnesium in 6111 has a slightly higher effect of increasing the susceptibility to cracking. Silicon in 6022 is more than the composition for maximum cracking, which could lead to crack healing. 6111 on the other hand is slightly below that of the composition for maximum cracking in an Al-Si alloy and therefore more likely to cause cracking. Alloy 6111, since it has the highest content of copper of the alloys, it is the only one that can be affected by copper eutectic formation. Alloys 5182 and 5754 are well past the maximum cracking composition of magnesium as seen in Figure 5.23.

Cracking susceptibility has also been developed based on varestraint data and plotted as total crack length (TCL) versus composition.\(^{47, 48}\) Figure 5.24 shows one of these figures, which plots TCL as a function of wt. % Mg and Cu. The alloys were once
again superimposed on this diagram showing that 6111 is more crack susceptible than
6022.

Based on these predictive methods it would seem reasonable to say that 6022 is
more resistant to solidification crack formation than 6111, but not by a tremendous
amount. 5754 and 5182 produce sufficient eutectic to promote crack healing if a crack
should form. 5182 has more eutectic formation at the end of solidification and therefore
more crack healing, and thus lower susceptibility to cracking. The predicted ranking of
these alloys in terms of their compositions are:

| Increasing weld solidification cracking susceptibility | 5182       |
|                                                 | 5754       |
|                                                 | 6022       |
|                                                 | 6111       |

This ranking was found to be true in the GTA welds, as the combination 6111-6111 could
not be welded without the formation of a centerline crack. These predictive methods and
rankings are based on solidification cracking, so only the fusion zone can be considered
here. The PMZ is considered in the next section.

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Normalized C_0 (C_0/C_{max})</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Mg</td>
</tr>
<tr>
<td>6022</td>
<td>0.04</td>
</tr>
<tr>
<td>6111</td>
<td>0.05</td>
</tr>
<tr>
<td>5754</td>
<td>0.2</td>
</tr>
<tr>
<td>5182</td>
<td>0.3</td>
</tr>
</tbody>
</table>

Table 5.2 Normalized compositions plotted on the graph in Figure 5.23.
Figure 5.23 Plot showing non-equilibrium solidification peak susceptibility to cracking as a function of composition.
Figure 5.24 Plot from varestraint data showing TCL as a function of alloying content of wt. % Mg and wt. % Cu.\textsuperscript{[47, 48]}
5.1.3.2 PMZ

Three factors that contribute to the partial melting of the grain boundaries in the HAZ are composition, grain size and solute segregation. As alloying additions increase, the amount of solute segregation can increase and therefore the potential for liquid film formation increases. However, the grain size of an alloy can have an effect on the amount of liqutation along these grain boundaries. It has been extensively researched\textsuperscript{[49]} that for alloys with the same base metal composition but different grain sizes in the HAZ, failure in the PMZ will preferentially occur in the alloy that has the larger grain size. This is due to the fact that less grain boundary area is available when larger grains are present and the solute segregation of these boundaries can become more concentrated with more liquid film development. Therefore, a base metal with coarser grains is expected to be more susceptible to crack formation in the PMZ.

This is believed to be the cause for the PMZ failures of 6022. Alloy 6022 has slightly larger grains within the PMZ than 6111 as shown in Figure 5.25. The wt. % Si in 6022 is much higher than that of 6111 and definitely contributes more solute to the matrix, which leads to an increased amount of solute segregation and wider films of liquid formation at the grain boundaries.
Figure 5.25 Micrographs showing the grain size variance between A) 6022 and B) 6111 (Barker’s reagent).
<table>
<thead>
<tr>
<th>Materials Joined</th>
<th>Transverse</th>
<th>Longitudinal</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Face</td>
<td>Root</td>
</tr>
<tr>
<td>6111</td>
<td>6022</td>
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</tr>
<tr>
<td>6111</td>
<td>5182</td>
<td>P</td>
</tr>
<tr>
<td>6111</td>
<td>5754</td>
<td>P</td>
</tr>
<tr>
<td>5754</td>
<td>5182</td>
<td>P</td>
</tr>
<tr>
<td>5754</td>
<td>6022</td>
<td>F</td>
</tr>
<tr>
<td>6022</td>
<td>5182</td>
<td>F</td>
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<tr>
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<td>6022</td>
<td>P</td>
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<tr>
<td>6111</td>
<td>6111</td>
<td>(I)</td>
</tr>
<tr>
<td>5754</td>
<td>5754</td>
<td>P</td>
</tr>
<tr>
<td>5182</td>
<td>5182</td>
<td>P</td>
</tr>
</tbody>
</table>

(1) No samples were tested due to solidification cracking.
P = Passed Test
F = Failed Test

Table 5.3 Bend ductility test results conducted at 13% strain.

5.1.4 Bend Test Results

The cracking described in the last section, correlates well with the data listed in Table 5.3. Out of the three samples tested of combination 6022-6111, one failed along the PMZ of 6111. 6022 – 5754 failed the root bend test with PMZ cracking on the 6022 side. The combination of 6022 – 5182 failed both root and face bend tests because of cracking in the PMZ of 6022. 6111 – 6111 could not be tested because of the severe solidification cracking experienced.
5.1.5 Equiaxed Grain Formation

As shown in Figures 5.5 – 5.14, equiaxed formation along the centerline is very common in aluminum alloys, especially at higher travel speeds and increased heat inputs. Notably, fine-grained equiaxed grains form in the 6xxx series very readily, while coarser equiaxed grains form in the 5xxx series. Aluminum alloys, as a result of the ingot processing, contain some level of titanium, generally less than 0.02 weight percent, which helps in grain refinement. \[50\] Heterogeneous nucleation of new grains can occur by growing off these titanium particles, commonly Al\textsubscript{3}Ti. \[51\] This formation of equiaxed grains is enhanced by higher heat inputs and welding speeds as Kou and Ganaha \textit{et al} showed. \[42, 51\] The increased heat inputs allow more time for nucleation and growth because of the decreased cooling rate. This effect was also seen in the current study as shown in Figure 5.28. As the travel speed was increased during GTA welding of combination 5182 – 5182, the formation of equiaxed grains increased. Ganaha \textit{et al} developed an equation for predicting equiaxed formation based on base metal composition. \[62\] This work concluded that titanium was the most potent element for nucleation of new equiaxed grains.

Basic solidification knowledge can also be used to describe this phenomenon. \[52\] Shown in Figure 5.26 is a schematic of thermal gradients versus various solidification rates. The solidification rates during a weld can be calculated by using the equation:

\[ R = V_w \cos \theta \]

Where R is the solidification rate, \( V_w \) is the welding velocity, and \( \theta \) is the angle between the welding direction and the direction of solidification. At the fusion boundary \( \theta \) is close to 90°, yielding a small R. The temperature gradient is steepest along the fusion
boundary, leading to the formation of a planar solidification front. As $\theta$ decreases and eventually reaches $0^\circ$ at the centerline, the solidification rate dramatically increases while the temperature gradient levels out. Along the centerline, solidification structures transform to equiaxed dendrites, as shown in Figure 5.27.\textsuperscript{[53]}

The nature of solidification as well as the heterogeneous nucleation off of titanium particles, contributes to the formation of the equiaxed zone along the centerline in aluminum alloys. However, the effect of solidification rate, or heat input, probably has the greatest impact on solidification structures. As will be discussed in the following section, travel speed can have a great effect on the equiaxed formation. At lower travel speeds, no equiaxed formation is observed, as shown with the combination 6022-6022 in Figure 5.7. This can be attributed to the fact that at the lower travel speeds, solidification rates dramatically drop off and are not sufficient to form an equiaxed zone. If heterogeneous nucleation was the primary determinant in the formation of this equiaxed zone travel speed should not affect this formation. The point can also be argued that with the faster cooling rates, sufficient time is not allowed for the heterogeneous nucleation and growth of this equiaxed zone off of the pre-existing titanium particles. No study was conducted to determine the major determinant for the formation of this equiaxed zone. It is assumed that they both contribute an uncertain amount of driving force for the formation of this equiaxed zone.
Figure 5.26 Schematic of solidification rates versus thermal gradients showing the nature of solidification.\textsuperscript{[52]}

Figure 5.27 Schematic representing differences in weld solidification due to changes in location in the weld and its effect on solidification morphology.\textsuperscript{[53]}
5.1.6 Feathery Grain Formation

A feather grain formation (or feather crystal) was found in a few combinations of the GTA welds. This structure can be described as the shape of an unfolded fan with a large radius of curvature. This feathery grain formation was most pronounced in the higher magnesium alloys, specifically 5182 as shown in Figure 5.5 and 5.28 A. In Figure 5.14, the combination of 5182 – 5754, this feathery formation was also noticed. However in the GTA weld of 5754 – 5754, no feathery formation appeared. These feathery grains have been studied by many authors and are believed to be the solid state formation of twins.\textsuperscript{[54,55]} Twinning occurs when a portion of a crystal assumes an alternate orientation that is related to the rest of the lattice in a symmetrical way. Lippold \textit{et al} proved that these were twinned boundaries with TEM analysis and also found that they have no effect on tensile or fatigue behavior.\textsuperscript{[54]} When twins form, they microscopically grow intermittently with a small inclination angle to each other.\textsuperscript{[55]} This is why macroscopically when observed, they look like unfolded fans. Nakagawa \textit{et al} concluded that this feathery crystal formation can be developed more readily at lower travel speeds, which is consistent with the observations made in this study, as shown in Figure 5.28. This work also stated that with the formation of these feathered grains in the weld metal, the tensile strength is not affected. However, it was found that the elongation of the weld metal dramatically decreased upon formation of these twins.
5.1.7 Travel Speed Effects

Experiments with travel speed variations were conducted on the combinations of 5182 – 5182 and 6111 – 6111. Figure 5.28 shows the effect of varying the travel speed on solidification grain structures with alloy combination 5182 - 5182. Micrograph A in this figure was made at 4.25 mm/s (10 ipm). A very dense and continuous feather grained formation formed along the centerline. Before the solid state transformation to a twinned formation, the solidification structure would have been completely columnar with no equiaxed grain formation. When the equiaxed grains form along the centerline, they tend to break up or lessen the degree to which the twins form.\textsuperscript{[55]} This phenomenon is seen in micrograph B and C of Figure 5.28. In these cases the travel speeds are such that an equiaxed grain formation forms along the centerline of the weld and therefore, twin formation decreases.

Lower travel speed welds were made with the combination of alloys 6111-6111, also. At 12.7 mm/s (30 ipm), travel speed which all the combinations were conducted at, the combination 6111 – 6111 produces a solidification crack along the entire length of the weld. It was postulated that decreasing the travel speed would have some effect on the cracking. This combination still cracked at the lower travel speed.
Figure 5.28 Solidification modes for different travel speeds for autogenously GTA welded combination 5182-5182. A) 4.2 mm/s. B) 12.7 mm/s. C) 23.3 mm/s.
5.1.8 Natural Aging Experiments

The aging response of the weld metal was measured using a superficial hardness indenter for all the combinations. Hardness measurements were taken immediately after welding and at varying intervals of time after that. This gave an understanding of the time that the 6xxx series took to naturally age after welding. Figure 5.29 shows this natural aging response for 6111 and 6022, respectively. Maximum natural aging occurs within 60 days after welding. However, the maximum hardness reached is lower than that of the original base material due to the dissolution of strengthening precipitates. The natural aging response of the dissimilar combinations was also observed and shown in Figure 5.30. The maximum hardness during natural aging is reached within 60 days as with the similar combinations, but the maximum hardness is shown to increase a small amount over the similar combinations due to the increased eutectic formation in the weld pool.

Micro-hardness was mapped out for each combination with a Vickers hardness traverse along the weld and base material. Every combination was recorded right after welding and then after the alloys had aged, approximately 2 months. A hardness traverse of welds conducted on base materials of 6111 and 5754 are shown in Figure 5.31. From these plots, it can be seen that 6111 naturally ages to 80 DPH, a gain of 15 DPH in the HAZ while in the FZ naturally ages to approximately 80 DPH a gain of 10 DPH. In this same figure it can be seen that 5754 doesn't naturally age as we would not expect it too since there are no precipitates in this alloy. Within the FZ this alloy experiences a hardening effect from the second phase, eutectic, that forms. Along the centerline though, there is a drop in hardness, which could be associated with the equiaxed
formation along the centerline. Figure 5.32 shows the hardness traverse for the dissimilar combination 6111-5754. From this plot, it can be seen that the 6111 alloy has age hardened to approximately 80 DPH in the HAZ and the 5754 hardness remained at approximately 60 DPH within the HAZ. There was a peak in hardness 10 mm from the centerline in alloy 5754 due to the recrystallization in this area. Within the FZ, it was observed that before aging 6111 and 5754 both showed characteristically their same hardness patterns as shown in Figure 5.31 up to the centerline of the FZ. After aging, the individual hardness traverses disappear, but with a characteristic hardness drop along the centerline, equiaxed zone.
Natural Aging of GTA Weld
6111-T4 B.O.P.

Figure 5.29 Natural aging curves of a GTA weld with alloy 6111 and 6022.
Natural Aging of GTA Weld
6111-5754

6022-5754

Figure 5.30 Natural aging curves of dissimilar combinations 6111-5754 and 6022-5754.
Figure 5.31 Hardness traverse across GTA welds for alloy 6111 and 5754 before and after natural aging.
Figure 5.32  Hardness traverse across GTA weld of combination 6111-5754 before and after welding.
5.1.9 Corrosion Testing

Localized pitting was seen on several samples including 6022 – 6022 and 6111 – 6022. Pitting took place approximately 5mm away from the fusion boundary of the 6022 side in both samples. 5182 – 5182 had pit formation along the fusion boundary. 6111 – 5754 pitted along both sides of the fusion boundary, with more severity on the 5754 side. It was noticed that the NaCl solution had precipitated salt crystals, which attached preferentially to the FB of the weld. No SCC was observed due to the galvanic cell set up by joining a 5xxx series to a 6xxx series.

5.1.10 GTAW Summary

Table 5.4 is a complete summary of the GTA welds of similar and dissimilar combinations. The only combination to experience severe solidification cracking during GTA welding was 6111 – 6111. 6022, although, cracked preferentially in the PMZ when joined to 5182 and 5754. This was noticed during the bend test of these combinations. The 5xxx series showed no signs of cracking during welding or the bend tests. There is non-uniform mixing in the fusion zone of the dissimilar combinations. The amount of mixing that takes place was thought to have an impact on the aging characteristics, but ultimately had little effect. From the original interface of the weld 6xxx series ages and the 5xxx series showed no signs of aging, indicating incomplete mixing had occurred. A drop in hardness was located along the centerline, which was believed to be associated with the equiaxed grain formation located along the centerline. The joining of these dissimilar combinations had no effect on stress corrosion cracking. Some alloys exhibited pitting in NaCl solution.
The formation of an equiaxed zone along the centerline was noted. This formation was believed to be due to solidification rates and/or heterogeneous nucleation off of titanium particles. Another formation seen in the higher magnesium alloys, mostly the 5xxx series, was that of a feathery grain structure. This formation was believed to be the formation of twins, which was easily stopped by the formation of equiaxed grains along the centerline. Travel speed was found to have an effect of the formation of equiaxed zone and therefore twin formation. At slower travel speeds, there is no equiaxed formation. Therefore, at the centerline a continuous formation of these feathery crystals or twins is displayed.
<table>
<thead>
<tr>
<th>Combinations Joined</th>
<th>General Observations</th>
<th>Cracking</th>
<th>Bend Ductility* $\varepsilon = 13%$</th>
<th>Corrosion</th>
<th>Metallographic Features</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Trans.</td>
<td>Long.</td>
<td>F</td>
<td>R</td>
<td>F</td>
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<tr>
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<td>P</td>
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<td>P</td>
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<td>P</td>
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<td>P</td>
</tr>
</tbody>
</table>

* F = Face Bend, R = Root Bend, P = Pass, F = Fail, NT = Not Tested due to severe solidification cracking.

Table 5.4 Summary of GTAW observations.
5.2 RSW

RS welding was conducted on the dissimilar and similar combinations listed in Table 4.3. As described in section 4.2.2, these experiments were conducted in two phases. The first phase was conducted on as-received material and the second on pickled material. Representative macrostructures of both phases are shown in Figure 5.33. The welds made on material in the as-received (unpickled) condition produced an erratic range of nugget diameters and occasionally no evidence of any nugget formation was found. Large porosity, 0.3 mm (0.012 in) diameter on average, was evident in 90% of the welds along with solidification cracks in several of the combinations. Copper contamination, found on the edges of the base material, caused by copper infiltration from the electrodes was also noticed, and shown in Figure 5.34.\textsuperscript{[55]} Watanabe et al\textsuperscript{[55]} described this phenomenon as “HAZ burning”. The RS welds made on pickled material showed a remarkable increase in process repeatability. The average variance in nugget diameter for certain combinations was only 1mm. No partial nuggets were formed. Some samples contained small porosity, with an average diameter of 0.15 mm (0.006in), and some solidification cracking. Thornton et al\textsuperscript{[57]} work states that the strength of a spot weld is dictated by nugget diameter, not the absence of porosity and cracking. In his work, he found that fatigue life was not affected by weld defects such as porosity or cracking. No mechanical tests other than the pull test, used in determining the nugget diameter, were conducted to confirm this theory.
Figure 5.33  RS welds between 6022 and 5754. A) Welded in the pickled condition. B) Welded in the as-received condition (Barker’s reagent).
Figure 5.34  Copper contamination in alloy 5754 from RS welding electrodes. Shown is top surface of a RS weld between 6022 and 5754 (Barker’s reagent).

Microstructures produced by RS welding similar combinations are shown in Figure 5.35 – 5.38. As shown in Figure 5.35, there are 4 regions of affected base material for the 5754-0 alloy. These regions can be divided up into a grain growth region (HAZ), columnar with heavy eutectic formation, columnar with light eutectic formation, and a large equiaxed grain region located in the center of the nugget. The heavy eutectic columnar region extends toward the center of the fusion zone for approximately 50μm. The light eutectic region extends 250 μm until the equiaxed grain formation is reached. These equiaxed grains are on the average of 75 μm in diameter, which is very large for these alloys.

Alloy 5182-H16 as RS welded is shown in Figure 5.36. For this base alloy, there are 5 regions of affected base metal that are shown. These regions can be described
as the HAZ, PMZ, columnar with heavy eutectic, columnar with light eutectic formation, and finally an equiaxed zone is formed. The PMZ extends for approximately 200 μm, forming large amounts of eutectic that surround the recrystallized grains. The columnar structure with heavy eutectic extends for 175μm, while the light eutectic columnar grains only extended 100 μm. The equiaxed grain formation is large similar to that of the 5754 with average grain size diameter of 70 μm.

As shown in Figure 5.36, there are 2 bands parallel to the rolling direction that form approximately 0.1 mm from the original interface, located within the HAZ region. These bands appear to be cracks that have formed along the prior elongated grain boundaries of this material and backfilled with liquid from the PMZ. Every RS combination that includes 5182 formed these bands. This may be a form of lamellar cracking, which is caused by low through thickness ductility of the parent material.\textsuperscript{[17]} The opening up of these lamellar cracks can be due to high through thickness stresses generated by the welding process. These stresses open up previous grain boundaries, which during strain hardening have formed continuous solute bands along these boundaries. This solute banding can lower the localized melting temperature so that upon cooling, this area is still liquid when other area have solidified and with the generated stresses, can tear open with least resistance.

Figure 5.37 shows a RS weld that was formed on 6111-T4 base material. Four distinct weld zones characterize this weld nugget. A HAZ, followed by the formation of a PMZ, dendritic growth followed by the formation of a very fine dispersion of equiaxed grains. A small PMZ is formed with eutectic melting along a width of three grains. The columnar solidification zone has a fine structure with intermittent equiaxed grains.
throughout. The zone in the center of the nugget consists of fine equiaxed grains averaging 10\(\mu\)m in diameter.

A RS weld with base material 6022-T4 is shown in Figure 5.38 and similar to the RSW of 6111. Four zones can be characterized as the HAZ, PMZ, columnar, and equiaxed zone. The PMZ only extends out into the HAZ one or two grains. The columnar solidification zone is a coarser structure than that seen in Figure 5.37 but still contains new intermittent grain growth. The equiaxed zone is a fine structure just as in Figure 5.37 with average grain diameters of 15\(\mu\)m.

Welds joining dissimilar combinations of these alloys lead to various modifications from those of the homogenous welds shown previously. These dissimilar RS welds are shown in Figures 5.39 though 5.46. A section taken from a RS weld made between alloys 6022 and 5754 is shown in Figure 5.39. There are still four zones produced within the weld area. HAZ (which may include a PMZ), a zone of columnar with heavy eutectic formation, a zone of columnar grains with finely distributed eutectic, and a zone of large equiaxed grains. This differs from the 5754 homogenous weld by the band of columnar grains with heavy eutectic formation being extended into the nugget 200\(\mu\)m. It can be seen from a higher magnification micrograph of this area, Figure 5.40, that the columnar bands, light and heavy eutectic, are about equal in size on the 5754 side of the weld. The 6022 side shows only a slight differentiation between the two bands of columnar grains in Figure 5.41. A much coarser equiaxed grain formation is located at the center of the nugget when compared to that of the homogenous 6022 weld.

A section of the microstructure produced from the RS weld of combination 6022-5182 is shown in Figure 5.42. In general, there was an immense amount of eutectic
within the weld nugget. Much more so than when these alloys were welded homogeneously. The equiaxed grain formation is slightly larger than that of the 6022 homogenous weld, while the grains are still smaller than that of 5182 homogenous weld. In the combination weld, the 6022 had a much smaller columnar zone. 5182 formed a columnar zone containing heavy amounts of eutectic, no columnar zone with small amounts of eutectic was formed, as with the homogenous 5182 weld.

The RS weld of combination 5182 – 5754 is shown in Figure 5.43. The 5754 side of this weld does not form the columnar dendritic region with light eutectic formation. It was also observed that the columnar region was confined to a very small region of the weld. 5182 side still contained a PMZ and the columnar growth region was also restricted to a small area due to the equiaxed zone formation. This combination formed a medium sized equiaxed zone with approximated diameters of 50 μm.

Combination 6111 – 5182 is shown in Figure 5.44. There is noticeable cracking within the PMZ and HAZ regions of alloy 6111. These cracks have been backfilled though with liquid from the FZ or PMZ. This combination produces a medium equiaxed zone slightly larger than in the homogenous weld of 6111.

Dissimilar combination of 6111 – 5754 is shown in Figure 5.45. The structure produced is similar to that of the 6111 – 5182 RS weld. A medium sized equiaxed zone is formed along the centerline of the nugget with columnar growth leading up to it.

The 6111 – 6022 combination is shown in Figure 5.46. Within the nugget is eutectic healing of porosity and small cracks, so that virtually no cracking or porosity exists. The equiaxed zone is made up of a very fine structure, approximately 15 μm in diameter.
It was found that the combinations with the fine equiaxed zone formation in the center region of the nugget had dramatically reduced porosity and solidification cracking that are normally formed during a resistance spot weld of aluminum alloys. The combinations; 6111-6111, 6111-6022, and 6022-6022 contained no porosity or solidification cracks because of the formation of this fine equiaxed structure. These alloys also produced a large amount of liquid, which aided in the healing of porosity and cracking when formed.
Figure 5.35 Regions of a RS weld made with aluminum alloy 5754-0 (Barker’s reagent).

Figure 5.36 Regions of a RS weld made with aluminum alloy 5182-H16 (Barker’s etch).
Figure 5.37  Base alloy 6111-T4 RS welded (Barker’s reagent).

Figure 5.38  Aluminum alloy 6022-T4 RS welded (Barker’s reagent).
Figure 5.39 Dissimilar RS weld between 6022 and 5754 (Barker’s reagent).

Figure 5.40 Increased magnification of 5754 side of 6022 - 5754 weld (Barker’s etch).
Figure 5.41  Increased magnification of 6022 side of a 6022 – 5754 RS weld (Barker's).

Figure 5.42  RS weld between 6022 and 5182. (Barker's reagent)
Figure 5.43 Section of a RS weld of combination between 5182-5754 (Barker's reagent).

Figure 5.44 Section of RS weld of combination between 6111-5182 (Barker's reagent).
Figure 5.45 Section of RS weld between 6111 and 5754 (Barker's reagent).

Figure 5.46 Section of RS weld between 6111 and 6022 (Barker's reagent)
5.2.1 RSW Summary

Every combination was able to be joined by the resistance spot welding (RSW) technique. Severe solidification cracking and porosity were evident in the welds made on as-received material. When the pickled material was welded, cracking and porosity were still an issue, but not to the extent as in the as-received material. Nugget consistency dramatically improved when welding pickled material. It was found that with the formation of a very fine equiaxed zone that solidification cracking and porosity were virtually eliminated. Copper contamination or "HAZ Burning" at the interface between the copper electrodes and the surface of the base material was noticed to decrease slightly with pickling of the base material. Table 5.5 summarizes these observations of the RS welds of dissimilar combinations.
<table>
<thead>
<tr>
<th>Alloy Combinations</th>
<th>Cracking</th>
<th>Porosity</th>
<th>Diameter of Equiaxed Zone</th>
<th>Crack Healing</th>
</tr>
</thead>
<tbody>
<tr>
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<td>X</td>
<td></td>
<td>Small</td>
<td>X</td>
</tr>
<tr>
<td>6111 - 5182</td>
<td>X</td>
<td></td>
<td>Small/Medium</td>
<td>X</td>
</tr>
<tr>
<td>6111 - 5754</td>
<td>X</td>
<td>X</td>
<td>Medium</td>
<td>X</td>
</tr>
<tr>
<td>5754 - 5182</td>
<td>X</td>
<td>X</td>
<td>Medium</td>
<td></td>
</tr>
<tr>
<td>5754 - 6022</td>
<td>X</td>
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<td>Large</td>
<td>X</td>
</tr>
<tr>
<td>6022 - 5182</td>
<td>X</td>
<td></td>
<td>Small/Medium</td>
<td>X</td>
</tr>
<tr>
<td>6022 - 6022</td>
<td></td>
<td></td>
<td>Small</td>
<td>X</td>
</tr>
<tr>
<td>6111 - 6111</td>
<td></td>
<td></td>
<td>Small</td>
<td>X</td>
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<td>5754 - 5754</td>
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<td></td>
<td>Large</td>
<td></td>
</tr>
<tr>
<td>5182 - 5182</td>
<td>X</td>
<td>X</td>
<td>Large</td>
<td></td>
</tr>
</tbody>
</table>

Table 5.5 Observations made during inspection of resistance spot welding of similar and dissimilar combinations of aluminum alloys.
5.3 Sigmajig Weldability Test Results

The selected alloys were tested using the Sigmajig weldability test to characterize individual alloy susceptibility to solidification cracking as a basis to compare this to dissimilar combinations.

5.3.1 Base Metals

Figure 5.47 shows the threshold stress for cracking ($\sigma_{th}$) for each individual alloy. These welds represent the bead-on-plate condition (BOP). This data was compared to the data found by Goodwin\textsuperscript{[27]} shown in Figure 2.11 for alloys 1100, 3003, and 5052 and found to show similar characteristics of cracking. Work done by Lippold \textit{et al.}\textsuperscript{[58]} has shown that strength of a base material can have a significant effect on the threshold stress levels obtained. To accommodate for this variance in strength between the alloys, the data shown in Figure 5.47 was normalized with each alloys ultimate tensile strength (UTS) and shown in Figure 5.48. Increased strength levels were shown to have decreased threshold values after normalization.

5.3.2 Laser Welding

Laser welds were used to join the dissimilar combinations to be tested using the Sigmajig test. Similar combinations (6111-6111 etc.) were also laser welded to evaluate the effects of residual stress from laser welding on the threshold stress. The threshold stress values for the laser welded similar combinations are compared to threshold stress levels of the base materials in Figure 5.49, demonstrating that the residual stress of the laser weld in not high enough to cause dramatic shifts in threshold stress levels. Laser
Sigmajig Threshold Values for Cracking
Base Materials

Figure 5.47 Plot of threshold stress values found for each base material during Sigmajig weldability testing experiment.
Sigmajig Threshold Values for Cracking
Normalized with BM UTS

Figure 5.48 Plot of threshold stress values found during Sigmajig weldability testing experiments normalized with each base materials ultimate tensile strength (UTS).
Sigmajig Threshold Values for Cracking
Effect of Laser Welds

Figure 5.49  Effects of laser welding, of similar combinations, on threshold stress values compared with threshold values obtained for base materials (no laser weld).
welding actually increased the threshold stress levels for both 6111 and 6022 alloys. The largest difference is 1 ksi, which can be considered insignificant.

Representative microstructures of the laser welds are shown in the figures to follow. Figure 5.50 shows a laser weld between 6111 and 5754. The HAZ in laser welds is very small and therefore confined to a very small region near the FB. The PMZ of 6111 only extends one or two grains, shown in Figure 5.51. 5754 does not form a PMZ. The solidification structure tends to be columnar and transforms to an equiaxed region along the centerline. All the laser welds between similar and dissimilar alloys produced the same structures. The 5182-5182 combination is the only combination that varied from the microstructure shown in Figure 5.50. Figure 5.52 shows the definite centerline formation where the columnar grains impinge on one another. In the top half of the weld, the distinct centerline is broken up by two equiaxed grains.

Cracking was observed in the 6022-6022, 6111-6111, and 6022-6111 laser welded combinations. This cracking, as shown in Figure 5.53, occurs along solidification grain boundaries. Despite these cracks, the samples for the Sigmajig testing could withstand the applied stress. It is interesting to note that, in Figure 5.49, the laser welds increased the threshold stress values of these alloys.

Large porous cavities were evident in several of the laser welds as shown in Figure 5.54. One factor that can attribute to this pore formation is the high cooling rates associated with laser welding. These rapid solidification times do not allow sufficient time for the entrapped atmospheric, shielding or hydrogen gases to evolve out of the weld metal as in GTAW and therefore become trapped within the weld. Many other factors are possible, but not investigated within this study.
Figure 5.50 Dissimilar laser weld between 6111 and 5754.

Figure 5.51 Increased magnification of 6111 side in Figure 5.37.
Figure 5.52 Laser weld of combination 5182-5182 base materials.

Figure 5.53 Cracking of a laser weld of combination 6022-6022 base materials.
Figure 5.54  *Entrapped* gas in a laser weld of 6111-6111 base materials.
5.3.3 Similar Combinations

Several macrostructures and microstructures of Sigmajig specimens are shown in Figure 5.55 – 5.62. Figure 5.55 shows the Sigmajig sample SN10 combination of 6111-6111. A very fine equiaxed structure is produced along the centerline. At the threshold stress the crack produced is within this equiaxed zone. A micrograph of the 6111 fusion boundary area, Figure 5.56, shows a very fine and densely populated non-columnar fine grained region with transformation to a fully columnar structure before a final transition to the equiaxed centerline as shown in Figure 5.55.

The 5754-5754 combination has the same formation of a fine equiaxed zone along the centerline and is shown in Figure 5.57. The solidification structure rapidly changes from columnar to equiaxed as shown in Figure 5.58. 5754-5754 cracked along the centerline region 100% of the sample, complete separation at the threshold level or above. No increment of stress could vary this percentage. Once the threshold stress value for cracking was reached, the specimen suffered complete separation. This phenomenon is not well understood and research in this area needs to be conducted to determine the cause of this “go or no-go” cracking. [39]

Figure 5.59 shows that the 6022-6022 combination produces an equiaxed region along the centerline as well, but this zone is not nearly as fine or as wide a zone as in 6111-6111 and 5754-5754. This combination doesn’t possess the non-columnar fine grained region as in 6111-6111. It has columnar structure until the centerline where it transforms to equiaxed as shown in Figure 5.60. Threshold stress levels cause cracking along the PMZ.
Figure 5.55 Sigma jig weldability test specimens of similar combination 6111-6111.
A) No cracking (sample SN10). B) Cracking (sample SN7).

Figure 5.56 Increased magnification micrograph of 6111 fusion boundary area.
Figure 5.57  Sigmaig weldability test specimens of similar combination 5754-5754.  
A) No cracking (SA47).  B) Cracking (SA66).

Figure 5.58  Increased magnification micrograph of 5754 fusion boundary area.
Figure 5.59 Sigmajig weldability test specimens of similar combination 6022-6022.
A) No cracking (sample SD19). B) Cracking (sample SD27).

Figure 5.60 Increased magnification micrograph of 6022 fusion boundary area.
Figure 5.61  Sigmajig weldability test specimens of similar combination 5182-5182.
A) No cracking (sample SC36).  B) Cracking (sample SC33).

Figure 5.62  Increased magnification micrograph showing 5182 fusion boundary area.
5182-5182 combination forms a narrow equiaxed zone with larger equiaxed grains when compared to the other alloys. This combination experiences an abrupt change from columnar to equiaxed solidification as shown in Figure 5.62, like the 5754-5754 combination. When threshold levels of stress are reached, cracking occurs along the centerline.

These abrupt changes in solidification also appear in base metal BOP micrographs (Figure 5.63), therefore there is no link between where the equiaxed zone forms and the previous laser weld location. 6111-6111 and 6022-6022 combinations do not experience these abrupt solidification changes.

5.3.4 Dissimilar Combinations

Sigmajig test samples for all dissimilar combinations are shown in Figures 5.64-5.69. When 6022 was joined to the other alloys, it cracked preferentially in the PMZ at low threshold stress levels. When it was joined to itself in a similar combination, Figure 5.59, cracking took place along the PMZ as well. From the GTAW experiments it would have been safe to assume that 6022 was more weldable than 6111, because 6111-6022 and 6022-6022 could be welded crack free. The combination 6111-6111 could not be welded without forming a centerline crack during GTAW experiments. 6111 when joined to 5182 or 5754, and tested with the Sigmajig cracked preferentially within the fusion zone, but along the 6111 side. Combination 5754-6111, shown in Figure 5.67, crack morphology changed from within the 6111 fusion zone to centerline cracking along the equiaxed region. 5182-5754, when Sigmajig tested, failed along the centerline with complete separation every time. No decreased increments of stress could lower the
cracking percentage as with the 5754-5754 combination. Table 5.6 lists the threshold stress values for each combination with 50%-50% dilution.

Figure 5.63  Fusion boundary of 5754 base metal Sigmajig weldability test. Showing that the abrupt change in solidification also appears in base metal specimens and not just with samples that contain a laser weld (Barker’s reagent).
Figure 5.64 Sigmajig weldability test specimen of 6022-5754. A) No cracking (sample SO5). B) Cracking (sample SO4).

Figure 5.65 Sigmajig weldability test specimen of 6022-5182. A) No cracking (sample SE4). B) Cracking (sample SE6).
Figure 5.66 Sigmajig weldability test specimens of combination 6022-6111. A) No cracking (sample SQ1). B) Cracking (sample SQ6).

Figure 5.67 Sigmajig weldability test specimens of combination 6111-5754. A) No cracking (sample SF2). B) Cracking (sample SF3).
Figure 5.68 SigmaJig weldability test specimens of combination 6111-5182. A) No cracking (sample SM16). B) Cracking (sample SM15).

Figure 5.69 SigmaJig weldability test specimens of combination 5754-5182. A) No cracking (sample SL8). B) Cracking (sample SL7).
<table>
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<tr>
<th>Alloy Combinations</th>
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</tr>
<tr>
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</tr>
<tr>
<td>6022 – 5182</td>
<td>5.8</td>
</tr>
</tbody>
</table>

Table 5.6  Threshold stress values for each dissimilar combination at 50% - 50% mixing.

As mentioned in the GTAW section (5.1), there is a lack of mixing when a 6xxx series and a 5xxx series are joined. This unequal mixing was also seen during the sigmajig experiments. Figures 5.70 and 5.71 show a weld between 6111 and 5754 at 50% - 50% dilution, with plan and transverse views. These samples were etched with Keller’s reagent to reveal the uneven mixing. The columnar zone growing off of the 5754 base metal has virtually no 6111 in Figure 5.70. The transverse section of this combination shows 2 distinct regions, a light region of 5754 and a dark region of 6111.

5.3.5 Dilution Effects

The actual dilution rates were measured to be within the range of 70-75% and 30%-25%. The combination of 6111-5754 was observed microscopically at 25% 5754 75% 6111 and 75% 5754 25% 6111 and these are shown in Figure 5.72. Micrograph A in this figure shows 25% 5754 75% 6111 dilution. This dilution rate has a very fine
Figure 5.70 Plan view of unequal mixing of 6111-5754 (50%-50%) Sigmajig weldability test specimen (Keller's reagent).

Figure 5.71 Transverse section of unequal mixing of 6111-5754 (50%-50%) Sigmajig weldability test specimen (Keller's reagent).
overall structure. 5754 has a small columnar zone, (400µm), which transforms into an equiaxed region. Mixing can be seen between the two alloys as the fine-grained region of 6111 can be seen near the fusion boundary of 5754.

Micrograph B is 50% - 50% mixing. The columnar zone formed from the 5754 base material was approximately 800µm, which is twice that of what formed in the previous micrograph. There is also an abrupt transformation from columnar solidification to the equiaxed zone, as seen previously in figure 5.63, on the 5754 side. The 6111 side is also as shown previously in Figure 5.56. A non-columnar fine-grained zone with transformation to a very narrow zone of columnar grains and then a final transformation to equiaxed grains.

Micrograph C is a micrograph of a sigmajet test specimen composed of 75% 5754 and 25% 6111. The fine grained region that 6111 produces is confined to a very small area, approximately 400µm. The equiaxed zone along the centerline contains larger grains and is broader than the previous two micrographs. In the particular area of this weld, the columnar grained zone that forms from the 5754 base material side has decreased in width. There is some cracking within the PMZ of 6111 which causes this dilution of 75% 5754 and 25% 6111 to have a lower threshold value than the other dilution rates.

The dilution effects on threshold stress values for all the dissimilar combinations are shown in Figure 5.74 - 5.77. It was found to be very common that when a 5xxx series is the higher dilution in a combination with a 6xxx series alloy, the threshold stress value for cracking would decrease because of cracking in the PMZ of the 6xxx series. Possibly a reason for this is more restraint is built up in the weld pool when
the 5xxx series is at 75% dilution than the PMZ of the 6xxx series can support. This leads to cracking in the PMZ. Micro-hardness traverses were conducted on combination 5182 – 6022 at 75% 5182 25% 6022 and 25% 5182 75% 6022. This hardness data is shown in Figure 5.73. As shown, the hardness of 75% 5182 25% 6022 is higher than that of the 25% 5182 75% 6022 weld, confirming that there is more residual stress within the 75% 5182 25% 6022 weld. This hardness increase can be correlated to an increase in strength due to the higher magnesium concentration within the weld metal. The solidification strains are then localized at the weakest part of the joint which is the PMZ of 6022 in 75% 5182 25% 6022 and cracking is initiated. All combinations with 6022 as a constituent suffered degradation in threshold stress values because of severe PMZ cracking of 6022. Higher dilution rates of 6022 give way to higher threshold cracking stress values. The threshold values, for each dissimilar combination, are listed in Tables 5.7 and 5.8.
<table>
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<th>Dilution</th>
<th>(\sigma_{th}(\text{ksi}))</th>
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<td>75%</td>
</tr>
<tr>
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<td>5182</td>
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</tr>
<tr>
<td>6022</td>
<td>6111</td>
</tr>
</tbody>
</table>

Table 5.7 Threshold stress values for dissimilar combinations at varying dilution rates.

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<th>Dilution</th>
<th>(\sigma_{th}(\text{ksi}))</th>
</tr>
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<td>6111</td>
</tr>
<tr>
<td>5754</td>
<td>6022</td>
</tr>
</tbody>
</table>

Table 5.8 Threshold stress values for dissimilar combinations at varying dilution rates.
Figure 5.72 SigmaJig weldability test specimens of combination 5754-6111 at varying dilution rates. A) 25% 5754 75% 6111. B) 50% 5754 50% 6111. C) 75% 5754 25% 6111 (Barker’s reagent).
Figure 5.73  Hardness traverses taken across Sigmajig test coupons of 75% 5182 25% 6022 and 25% 5182 75% 6022.
Sigmajig Threshold Cracking Stress ($\sigma_{th}$)

Effect of 5754 Dilution

* Area of failure indicated on right side of data bar

Figure 5.74 Plot of threshold stress values found during Sigmajig weldability testing experiments as a function of 5754 dilution.
Figure 5.75 Plot of threshold stress values found during Sigmajig weldability testing experiments as a function of 5182 dilution.
Figure 5.76 Plot of threshold stress values found during Sigmajig weldability testing experiments as a function of 6111 dilution.

* Area of failure indicated on right side of data bar
Sigmajig Threshold Cracking Stress ($\sigma_{th}$)
Effects of 6022 Dilution

* Area of failure indicated on right side of data bar

Figure 5.77 Plot of threshold stress values found during Sigmajig weldability testing experiments as a function of 6022 dilution.
5.3.6 Fractography

SEM/SE was used to analyze the fracture surfaces of several sig employees.

Figure 5.78 shows the centerline fracture surface of 6111 - 5754. The structure shown is dendritic with the formation of secondary arms with no evidence of mechanical (solid-state) fracture. Figure 5.79 shows the centerline fracture surface of combination 5754 - 5754. This is similar to the previous figure, in which the structure is dendritic in nature. All the combinations that cracked along the centerline possess dendritic morphologies as those shown. The combination 5182 - 5182 is the only one that didn’t follow this pattern. As shown in Figure 5.80 and 5.81 the surface is still dendritic, but it is very flat compared to that of the other alloys. This could be attributed to the large amounts of eutectic formation at the end of solidification with this alloy. The liquid film becomes so large that it hides the dendritic morphology. At low magnifications, the fracture surface appears to be along very large equiaxed grain boundaries. The other alloys show this equiaxed structure at low magnifications, but with much smaller grains as shown in Figure 5.83 A. In Figure 5.81 of alloy 5182, the liquid had solidified in parallel lines encompassing the equiaxed grains. This parallel formation was also seen in the fracture surface of combination 5754 - 5754 shown in Figure 5.82. The arrows point to an area that is believed to be a solidification grain boundary or solidification subgrain boundary which liquid has formed upon and solidified as parallel lines. This formation could just be an area of secondary dendritic arm growth too.

Figure 5.83 shows a centerline fracture surface of combination 6111-6111. Micrograph A in this figure shows the overall fracture surface at a low magnification,
which appears to be equiaxed. Micrograph B shows a dendritic region of the fracture surface with secondary arm growth. The last micrograph, C, shows a transition to a flat region at the crack arrest. The structure shown in this figure is flat with some area of mechanical failure, evidence that there was some solid-solid bridging at this stage of solidification.
Figure 5.78 SEM/SE micrograph of centerline fracture surface of Sigmajig weldability test specimen of combination 6111-5754.

Figure 5.79 SEM/SE micrograph of centerline fracture surface of Sigmajig weldability test specimen 5754-5754.
Figure 5.80  SEM/SE micrograph of centerline fracture surface of combination 5182-5182 Sigmajig test sample.

Figure 5.81  Increased magnification micrograph of Figure 5.80.
Figure 5.82  SEM/SE micrograph of centerline fracture of SigmaMig weldability test specimen of combination 5754-5754 (sample SA51). Arrows show an area that is believed to be a solidification grain boundary.
Figure 5.83  SEM/SE micrograph of centerline fracture surface of SigmaJig test specimen 6111-6111 (sampleSB63).  A) Low magnification.  B) Dendritic region.  C) Flat region.
Correlation to Other Data

Pumphrey and West\cite{60,61} have plotted ring casting data, which simulates a restrained weld, and this is shown in Figure 5.84. The alloys examined in this study were superimposed on Figure 5.84. As can be seen, 6111 and 6022 fall into the high cracking susceptibility category while 5754 and 5182 are located within the medium region. If tie lines are drawn between the alloys, dilution rates can be estimated. At 50%-50% dilution of 5182-6111, it can be seen that this weld metal composition lies in the medium susceptibility to cracking region. It could be assumed that the threshold stress found during sigmajig testing should increase from the 6111 base metal threshold stress. This in fact correlates well with the data found with the Sigmajig. The threshold stress for 6111 BM is 8.2 ksi, while the 6111-5182 at 50%-50% dilution has a threshold stress of 10.3 ksi. This effect is also seen with the dissimilar combination of 6111-5754.
Threshold stresses increase with increased percentages of 5754 within the weld metal, which correlate with Figure 5.84. According to Figure 5.84, 6022 should have the same response to dilution effects with the 5xxx series as 6111 did. The Sigmajig data did not show this. Threshold stress values increased as the amount of 6022 within the weld metal increased, which is reverse of what was observed with alloy 6111. This was explained in section 5.3.5 as being due to the higher restraint built up within the weld pool due to higher magnesium concentrations, which leads to cracking within the PMZ of 6022. The data correlates as long as PMZ cracking is avoided.
Figure 5.84  Ring casting data showing susceptibility to cracking for aluminum alloys.\cite{60,61}
5.3.8 Practical Implications

When autogenously GTA welding a simple head on plate (BOP) on alloy 6111, it was impossible to do so without a solidification crack forming along the centerline. When Sigmajig tests were conducted using the same parameters with an autogenous GTA BOP, the solidification cracking did not initiate until a threshold stress of 7.9 ksi was applied. From 0 load to the threshold stress load, there was no cracking. This means that the Sigmajig is fixtured in such a way that prevents cracking from forming for the initial pre-stresses. It is believed to be in the way the sample is torqued in the fixture which has some effect on the cracking. The samples were torqued down to 30 ft-lbs in between 2 steel hardened grips using hold down bolts. When GTA welding, the sample was clamped down to somewhat lower force (restrained less) than that in the Sigmajig and with this, cracking occurs.

It is proposed that a way could be developed for use in the automotive industry to hold the material at a constant load, thus stopping the initiation of crack formation. This theory is backed by Z. Feng et al. work.\(^{62}\) Feng et al found that the crack susceptible region, for the nickel base alloy tested, extends from the trailing edge of the weld pool, which is at approximately 1600K, to the region of 1300K. For this temperature range, stresses were plotted for Sigmajig test samples with three different pre-loads, shown in Figure 5.85. This plot shows that when the sample was prestressed at 68.9 MPa, the stresses seen in the crack susceptible region are all compressive. Thus no cracking was observed in this sample at this pre-load stress. With the other pre-loads, 0 and 178 MPa, the crack susceptible region experiences some degree of tensile stress, and thus the sample experiences cracking at these stress levels.
This theory goes against other theories that state less restraint will tend to reduce or prevent solidification cracking. Feng et al. works and the work conducted in the present study, indicate that, if applied appropriately, the restraining effect could be beneficial. Further work, however, needs to be conducted to insure proper placement of restraint by means of local heating and mechanical fixturing to suppress the development of tensile stress throughout the sample during the weld cycle.
Figure 5.85 Stress distribution in Sigmajig samples (autogenous GTA) of a nickel base material pre-stressed at 3 different loads. A) Schematic of 68.9 MPa prestress weld. B) Schematic of 172 MPa prestress weld.
5.3.9 Summary

The Sigmajet weldability test is a quantitative way of assessing the cracking susceptibility of different materials. In this study, the Sigmajet test was used to determine the variance in cracking susceptibility of dissimilar aluminum alloys. In order to study these dissimilar combinations, laser welds were produced to initially join the dissimilar combinations. Dilution rates were varied through the placement of each sample relative to the tungsten electrode. The effect of dilution on threshold values was very different from what was originally thought. Cracking occurred in the PMZ of 6022 when joined to any other alloy. Welds conducted with 6022 did not reflect weld metal behavior because of cracking in the PMZ, however, it was shown that dilution still had and effect on the PMZ cracking. At low dilution rates of 6022, for instance in 25% 6022 75% 5182 combination, more restraint is built up in the system, which leads to PMZ cracking in 6022 at lower threshold levels than that of combinations with higher dilution rates of 6022. The combination most resistant to cracking at 75% - 25% is 5754 - 5182 respectively. At 50% - 50%, the combination with the highest threshold value was 5182 - 6111, respectively. Again, at 25% - 75% the combination 5754 - 5182 had the highest threshold stress value. These values as well as other Sigmajet results with varying dilution rates can be seen in Figure 5.86. These threshold values are also listed in Table 5.6 - 5.8. The data collected was comparative to other techniques used to quantify cracking susceptibility, as long as cracking was along the centerline. Fractography of fractured samples revealed dendritic morphologies for centerline cracking, while the PMZ cracking appeared to be at liquated grain boundaries.
Summary of Sigmajig Cracking Stress

% Dilution of Italicized Alloys

% Dilution of Bold Alloys

Figure 5.86 Summary plot of threshold stress for dissimilar combinations at varying dilution rates.
CHAPTER 6

CONCLUSIONS

6.1 GTAW

1. All possible combinations were able to be joined together with the exception of 6111-6111. This alloy exhibited severe centerline cracking under the welding conditions used in this study.

2. 6022 when welded to 5182 and 5754 experienced cracking in the PMZ. This was found to be due in part to the large grains in the 6022 PMZ.

3. 6111 when welded to 6022 and 5182 experienced, to a small degree, PMZ cracking.

4. The welds that failed the bend test were combinations 6022-5182 and 6022-5754 which failed in the PMZ of 6022.

5. Non-uniform mixing was evident in dissimilar combinations due to differences in viscosity and surface tension.

6. The equiaxed grain formation along the centerline was determined to be caused by heterogeneous nucleation of new grains from titanium particles and greater solidification rates. This formation has been shown to result in less macrosegregation, avoid centerline cracking, and create less central porosity.

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7. A feathery grain formation was found along the centerline in higher magnesium alloys, 5182 and 5754.

8. Natural aging experiments showed that there was an increase in hardening of the weld metal to 75% of the base material hardness for alloy 6111 and to 84% of the 6022 base material.

9. No SCC was noticed in the dissimilar welds, but pitting did occur in samples 6022 - 6022, 6111 - 6022, 5182 - 5182, and 6111 - 5754.

6.2 RSW

1. All possible combinations studied were able to be joined with the specified nugget diameter of 5 mm (0.20 in.).

2. Pickling of the samples before welding resulted in a dramatic increase in nugget formation and consistency.

3. Copper contamination decreased when samples were pickled but still was considered a problem.

4. Solidification cracking and porosity were still a problem even when using the pickled sheet material.

5. 6111 – 6022, 6111-6111, and 6022-6022 formed a very fine equiaxed grain zone along the center of the nugget. This proved to be helpful in deterring the formation of porosity and solidification cracks.
6.3 Sigmajig Weldability Test

1. From base metal threshold stress values, it was found that 5182 was ranked the highest and 6022 the lowest in terms of resistance to solidification cracking.

2. After normalization of the data with each base metals UTS, 5754 was ranked the highest and 6022 was still the lowest.

3. The same unequal mixing was noted during the Sigmajig experiments as was in GTAW.

4. For the dissimilar combinations 5182-6111 at 50% - 50% dilution had the highest threshold stress value.

5. 6022 was shown to have a degrading effect on threshold values when in combination with another alloy because of the PMZ cracking experienced.

6. It was found with higher restraints the cracking formed in 6111-6111 was suppressed. It seems that if the right fixturing can be applied in industry, solidification cracking for some alloys could be stopped.

7. Fractography revealed dendritic morphologies for centerline cracking and liquated equiaxed grain boundaries for PMZ cracking.
LIST OF REFERENCES


41. Dr. Kazuhiro Nakata, Osaka University, Private Communication, 1997.


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