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A metallurgical investigation into the friction welding of rapidly-solidified, dispersion-strengthened aluminum alloys

Koo, Hyung-Hoi, Ph.D.
The Ohio State University, 1991
A METALLURGICAL INVESTIGATION INTO THE FRICTION WELDING OF RAPIDLY-SOLIDIFIED, DISPERSION-STRENGTHENED ALUMINUM ALLOYS

DISSERTATION

Presented in Partial Fulfillment of the Requirements for the Degree of Doctor of Philosophy in the Graduate School of The Ohio State University

by

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Finally, I would like to thank my family for their love and support throughout my stay in the U.S.A..
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1) H. H. Koo, S. Krishnaswamy and W. A. Baeslack III,
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   temperature RS/PM Al-8.5Fe-1.3V-1.7Si alloy (AA-

2) H. H. Koo, S. Krishnaswamy and W. A. Baeslack III,
   "Solid phase welding of a rapidly-solidified
   dispersion-strengthened Al-Fe-V-Si alloy - FVS1212",
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FIELDS OF STUDY

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Studies in welding metallurgy, welding process and control, weldability and failure analysis, solid state welding, laser material processing, brazing and soldering, nondestructive evaluation, physical metallurgy, electron microscopy, solidification processing, corrosion engineering.
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The objective of this study was to investigate the similar- and dissimilar-alloy friction welding of dispersion-strengthened, elevated-temperature aluminum alloys. Results showed that both inertia- and linear-friction welding can be effectively utilized for the joining of state-of-the-art Al-Fe-X type alloys. Inertia-friction welding using a sufficiently high axial force minimized microstructural coarsening and, thereby, optimized mechanical properties of both similar- and dissimilar-alloy welds.

For the similar-alloys welds in Al-Fe-Si-V alloys, mechanical properties were dependent principally on the extent of dispersoid coarsening and redistribution (via non-homogenous deformation) across the weld interface region. Analytical-electron microscopy analysis determined that the fine, spherical $\text{Al}_{13}(\text{Fe,V})_3\text{Si}$ dispersoids and sub-micron alpha aluminum grains present in the original base metal microstructure were also exhibited across the weld heat-and-deformation zones. Inertia-friction welding using a high axial force promoted negligible dispersoid coarsening across the weld interface region. In contrast, linear-
friction welding (non-optimized) of Al-Fe-V-Si alloys resulted in a moderate coarsening of dispersoids due to a comparatively slower thermal cycle, but exhibited a more uniform microstructure radially across the weld interface. Relatively high joint efficiencies (exceeding 85%), but moderate to low ductilities, were exhibited by the similar-alloy welds. Hardness, tensile properties and fracture behavior correlated well with the microstructural observations.

Dissimilar-alloy inertia-friction welds between Al-9Fe-3Mo-1V and 2024-T351 exhibited macroscopic interface curvature with penetration of the 2024-T351 into the Al-9Fe-3Mo-1V. In contrast, dissimilar-alloy inertia-friction welds produced between Al-Fe-V-Si alloys and 2024-T351 exhibited relatively flat macroscopic interfaces. Although dispersoid coarsening in the Al-Fe-X alloys was minimal at the weld interface, appreciable metallurgical change occurred in the precipitation-strengthened 2024-T351 alloy. Despite these metallurgical changes, and occasional fracture along the weld interface, relatively high joint efficiencies comparable to those of similar alloy welds were measured.

Dissimilar-alloy inertia-friction welds produced between the dispersion-strengthened Al-Fe-Si-V alloys and titanium alloys exhibited marginal integrity and strength
due to the large strength differences between the alloys and enrichment of dispersoids at the interface due to the nonuniform deformation, both of which precluded effective bonding.

Elevated-temperature exposure experiments on similar-alloy inertia-friction welds in Al-Fe-V-Si at 425 °C showed minimal dispersoid coarsening and negligible degradation of mechanical properties up to 500 hours. Exposure above 500 °C resulted in the formation of undesirable acicular Al₁₃(Fe,V)₄ or Al₃(Fe,V) intermetallics distributed uniformly throughout the weld regions.
INTRODUCTION

Recent advances in rapid solidification and powder metallurgy (RS/PM) technology have led to the development of a new generation of high-strength, dispersion-strengthened aluminum alloys for elevated temperature applications up to 425 °C (Refs. 1-5). These alloys are based on hypereutectic Al-Fe base compositions with ternary and quaternary additions of transition and/or rare earth elements, such as Ce, Mo, Zr, Ti, V and Si (Ref. 2). These alloying elements exhibit low solubility and diffusivity in aluminum, and thereby promote the formation of extremely fine, thermally-stable dispersoids. These alloying elements further reduce lattice misfit and the surface energy of the dispersoids, thereby reducing their coarsening rate. The combination of these fine dispersoids and a fine alpha aluminum grain size (approximately 1 μm) provides high room-temperature strength and excellent elevated-temperature mechanical properties.

Al-Fe-V-Si and Al-Fe-Mo-V alloys represent two recently developed families of RS/PM, dispersion-strengthened aluminum alloys which show excellent room and elevated-temperature strength and stability. On a specific strength
basis, Allied-Signal Inc.'s Al-12 wt.% Fe-1 wt.% V-2 wt.% Si alloy (hereafter designated as FVS1212) exhibits strength comparable to Ti-6 wt.% Al-4 wt.% V (hereafter designated as Ti-6-4) up to 425 °C, as shown in Fig. 1 (Ref. 2). The coarsening rate observed at 425 °C in FVS1212 is two to three orders of magnitude lower than for earlier generation Al-Fe-Mo and Al-Fe-Ce alloys (Ref. 2). Consequently, the Al-Fe-V-Si base alloys are considered as potential replacements for titanium alloys and conventional ingot metallurgy (IM) aluminum alloys in many structural aerospace applications that operate at temperature up to 425 °C.

To advance the utilization of these new RS/PM Al-Fe-X alloys in structural applications, establishing effective joining methods is considered a necessity. Although a wide variety of fusion welding and solid-state welding techniques is available to join conventional ingot metallurgy aluminum alloys, the choice of welding processes to join the RS/PM Al-Fe-X alloys is limited due to the unique base metal microstructural characteristics resultant from rapid solidification processing. Ideally, the joining methods should be able to recreate or retain the microstructural characteristics exhibited by the base metal.
Fig. 1. Comparison of elevated-temperature tensile yield strengths for dispersion-strengthened RS/PM Al-Fe-V-Si alloys, IM-aluminum alloy 2014-T6 and Ti-6-4 (Ref. 2).
In recent years, several joining approaches have been applied to Al-Fe-X alloys. Pulsed Nd:YAG laser (Ref. 6) and electron beam (Ref. 7) welding were used to fusion-weld RS/PM Al-Fe-Mo alloys which contained a low hydrogen content (< 1 ml/100 g). The rapid thermal cycles associated with these high-energy density welding processes and the low hydrogen content exhibited by the base metal resulted in fine dispersoid microstructures free of hydrogen-induced porosity. Capacitor-discharge (CD) welding was also utilized to fusion weld an RS/PM Al-Fe-Ce alloy which contained a high hydrogen content (> 5 ml/100 g) (Ref. 8). The rapid thermal cycle experienced during this welding process, in combination with the simultaneous application of a high compressive stress, resulted in a spatially fine, high-strength microstructure and an absence of hydrogen-induced porosity.

An alternate approach for the joining of RS/PM Al-Fe-Ce and Al-Fe-Mo-V alloys which contain a high hydrogen content (> 5 ml/100 g) involved the application of solid-state welding processes (Refs. 9 and 10). In contrast to fusion welding, solid-state welding avoids melting and resolidification and, thereby, precludes the formation of weld defects such as porosity and solidification cracking. These methods also generally restrict the heat-and-deformation-affected zone (HDZ) to a relatively narrow
region. Preliminary results demonstrated that the solid-state inertia-friction welding process offers a strong potential for effectively joining RS/PM Al-Fe-X alloys.

Considering the potential utilization of RS/PM aluminum alloys in structural aerospace components in conjunction with conventional aluminum and titanium alloys, the dissimilar joining of the RS/PM Al-Fe-X alloys to IM aluminum and titanium alloys may also be required. The joining of dissimilar materials is usually accompanied by problems associated with differences in material physical and mechanical properties, such as melting point, thermal expansion coefficient and strength, and complex, often degrading metallurgical interactions. Traditionally, in an effort to reduce complex metallurgical reactions experienced during melting and solidification, solid-state joining processes have been widely used for dissimilar alloy joining. A previous study (Ref. 9) has, indeed, demonstrated the feasibility of inertia-friction welding Al-Fe-Ce to IM Al-2024.

Although the potential utility of friction welding for the similar and dissimilar alloy joining of dispersion-strengthened aluminum alloys has been demonstrated, these preliminary studies (Refs. 9 and 10) were limited in that microstructure characterization was limited principally to light microscopy. Considering the extremely fine spatial
size of the microstructural features in these alloys, it is very apparent that detailed microstructural characterization using analytical-electron microscopy is required to elegantly develop process/microstructure/property/fracture relationships.

Al-Fe-V-Si alloys, which represent the most recently developed RS/PM alloy family, exhibit superior room and elevated-temperature mechanical properties and there is a strong necessity to establish suitable joining methods for these alloys. The morphologically simple nature and excellent thermal stability of dispersoids present in these alloys make them ideal for studying microstructural evolution during the friction welding processes.

Finally, the recently-developed linear-friction welding process adds another dimension to conventional rotational friction welding by providing a capability to join non-axisymmetric components (Ref. 11). With the introduction of the newly developed linear-friction welding process, a comparison of performance characteristics and resulting microstructures between the conventional rotational-type inertia-friction welding and the linear-friction welding would be extremely valuable.

Considering the above, the current research was performed to extend previous preliminary friction welding studies into a more comprehensive investigation involving
the similar-alloy inertia-and linear-friction welding of state-of-the-art Al-Fe-V-Si alloys and the dissimilar-alloy friction welding of both Al-Fe-V-Si and Al-Fe-Mo-V alloys to IM aluminum and titanium alloys. Macro- and microstructural characteristics, room and elevated temperature mechanical properties and fracture behavior of the similar and dissimilar alloy welds were characterized in detail. Microstructure analysis included light, scanning-electron and analytical-electron microscopy. Utilizing these experimental results, detailed process/microstructure/mechanical property/fracture relationships were developed.
CHAPTER I

BACKGROUND AND LITERATURE REVIEW

1.1. Introduction

The purpose of this chapter is: 1) to review current status of RS/PM Al-Fe-X alloys and basic metallurgical principles related to these alloy systems; and 2) to provide general background information on the weldability of dispersion-strengthened Al-Fe-X alloys in order to establish the basis for the present investigation.

1.2. Metallurgical principles of dispersion-strengthened RS/PM Al-Fe-X alloys

1.2.1. Dispersion-strengthening in Al-Fe-X alloy system

Conventional aluminum alloys which are strengthened by precipitates generally show a decrease in strength around 150 °C, which precludes their use for elevated-temperature applications. The loss of strength at high temperature in precipitation-hardened alloys is mainly due to the rapid coarsening and loss of coherency accompanied with overaging phenomena. Dispersion-strengthening is achieved by placing non-shearable, stable dispersoid particles into an aluminum
matrix, thereby hindering dislocation motion. In order to obtain excellent room and elevated temperature mechanical properties, the alloy microstructure should have a high volume fraction of thermally-stable dispersoids. Excellent elevated-temperature mechanical properties of RS/PM Al-Fe-X alloys are mainly attributed to the presence of fine and evenly distributed stable dispersoids resulting from rapid solidification processing and subsequent thermomechanical processing. In addition, the fine, submicron-sized grains of the RS/PM alloys contribute to the strengthening of these alloys. In the following sections metallurgy and alloy design concepts of the RS/PM Al-Fe-X alloys will be reviewed.

1.2.2. Physical metallurgy of RS/PM Al-Fe-X alloys

1.2.2.1. Al-Fe binary system

Aluminum binary alloy systems can be grouped into three types - eutectic, peritectic and monotectic (Ref. 20). While eutectic and peritectic systems show complete solubility in the liquid state, monotectic systems exhibit a miscibility gap. The segregation effect due to the miscibility gap further contributes to the formation of inhomogeneous as-solidified microstructures. For this reason, the aluminum alloy systems utilized to achieve dispersion-strengthening are essentially limited to eutectic and/or peritectic systems. To ensure a high
volume fraction of dispersoids, a dispersoid-forming element should have the following requirements:

1) form an intermediate intermetallic phase;
2) exhibit high liquid solid solubility;
3) exhibit a low terminal solid solubility limit (TSSL).

Candidate alloying elements which meet these requirements are shown in Table 1.

Table 1: Diffusivity, liquid and solid solubility of alloying elements in aluminum (Ref. 2)

<table>
<thead>
<tr>
<th>Element</th>
<th>Alloy system</th>
<th>D at 618 °K (cm²/s)</th>
<th>Liquid solubility at 1400 °K (at. %)</th>
<th>Cₛ, max.</th>
</tr>
</thead>
<tbody>
<tr>
<td>Fe</td>
<td>eutectic</td>
<td>5.4x10⁻⁴</td>
<td>18.52</td>
<td>0.025</td>
</tr>
<tr>
<td>V</td>
<td>peritectic</td>
<td>7.4x10⁻¹⁵</td>
<td>0.85</td>
<td>0.32</td>
</tr>
<tr>
<td>Zr</td>
<td>peritectic</td>
<td>3.4x10⁻¹⁸</td>
<td>3.7</td>
<td>0.085</td>
</tr>
<tr>
<td>Cr</td>
<td>peritectic</td>
<td>1.0x10⁻¹²</td>
<td>8.4</td>
<td>0.40</td>
</tr>
<tr>
<td>Ce</td>
<td>eutectic</td>
<td>8.4x10⁻¹⁶</td>
<td>12.0</td>
<td>0.01</td>
</tr>
<tr>
<td>Mo</td>
<td>peritectic</td>
<td>2.4x10⁻¹⁴</td>
<td>0.86</td>
<td>0.056</td>
</tr>
<tr>
<td>Ti</td>
<td>peritectic</td>
<td>3.0x10⁻¹⁶</td>
<td>3.6</td>
<td>0.57</td>
</tr>
</tbody>
</table>

As shown in Table 1, Fe exhibits the highest liquid solubility with relatively low solid solubility, diffusivity and diffusivity flux (D x Cₛ, max). For this reason, most of the elevated temperature Al alloys which contain a high volume % dispersoids are based on the Al-Fe
system. Most of the Al-Fe base alloys are based on Al - (7 - 12) wt. % Fe.

The purpose of the rapid solidification process is to suppress the formation of coarse, equilibrium microstructures and to promote the formation of fine, nonequilibrium microstructures. The rapid solidification of hyper-eutectic Al-Fe alloys results in fine cellular alpha aluminum or a coupled eutectic microstructure rather than forming equilibrium phases composed of coarse primary AlFe intermetallic and a eutectic between alpha aluminum and AlFe. In addition to forming a fine microstructure, RS processing is intended to form beneficial metastable intermetallic phases instead of stable, acicular AlFe intermetallic phase, which reduces toughness of the alloys. As shown in Fig. 2, rapid quenching of a binary hypereutectic Al-Fe alloy promotes solidification to metastable AlFe instead of equilibrium AlFe (Ref.18). Subsequent thermomechanical processing promotes additional precipitation of dispersoids from the supersaturated alpha aluminum matrix. In comparison with binary Al-Fe alloys, the addition of ternary and/or quaternary alloying elements to the hypereutectic binary Al-Fe composition promotes the formation of more thermally-stable dispersoids and improves room- and elevated-temperature mechanical properties.
THE SYSTEM ALUMINUM–IRON

Fig. 2. Al–Fe binary phase diagram (Refs. 18 and 50).
1.2.2.2. Al-Fe-X alloys

Several approaches have been utilized to improve the elevated-temperature stability of Al-Fe base alloys: 1) provide additional binary strengthening dispersoids such as Al_3Zr (Ref. 13) and Al_{10}V (Ref. 14); 2) thermally stabilize the binary Al-Fe intermetallics by the addition of elements such as Mn, Ce, V, Mo and Si (Ref. 15, 16); 3) formulate ternary and quaternary intermetallics with a more symmetric lattice (Ref. 16, 17) and, thereby, reduce the lattice mismatch and the surface energy at the dispersoid/matrix interface.

There are basically four groups of Al-Fe-X alloys: 1) Al-Fe-A, 2) Al-Fe-B, 3) Al-Fe-B-B and 4) Al-Fe-Si-B with A as eutectic alloying elements (Ce, Gd) and B as peritectic alloying elements (Mo, V, Zr, Ti, Cr) (Ref. 20). Al-Fe-V-Si alloys belong to the fourth group and Al-Fe-Mo-V alloys belong to the third group. While ternary and quaternary alloying in other alloy families is intended to stabilize metastable the binary Al-Fe intermetallic phase(s), the addition of V in Al-Fe-V-Si alloys is intended to stabilize the intermetallic phase as well as to modify the crystal structure to cubic Al_{13}(Fe,V)_3Si (bcc). Currently, Al-Fe-Si alloys stabilized by the addition of V are reported to exhibit the most stable dispersoid phase, that being Al_{13}(Fe,V)_3Si (Ref. 21). The addition of these ternary
and/or quaternary alloying elements results in various metastable intermetallic dispersoids depending on the alloy compositions. The resulting microstructure contains 20 to 40 volume % dispersoids, as shown in Table 2.

Table 2: Size and volume fraction of intermetallic dispersoids in Al-Fe-X alloys (Ref. 19)

<table>
<thead>
<tr>
<th>Material</th>
<th>Intermetallic Dispersoids</th>
<th>Volume Fraction, %</th>
<th>Diameter, μm</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al-8Fe-7Ce</td>
<td>$\text{Al}_8\text{Fe}_5\text{Ce}$, $\text{Al}<em>3\text{Fe}$, $\text{Al}</em>{12}\text{Fe}_5\text{Ce}$</td>
<td>15.2</td>
<td>0.15</td>
</tr>
<tr>
<td>Al-8Fe-2Mo-1V</td>
<td>$\text{Al}<em>{12}\text{Fe} \left(\text{Mo},\text{V}\right)$, $\text{Al}</em>{12}\text{Fe}$, $\text{Al}_3\text{Fe}$</td>
<td>17.0</td>
<td>0.1 - 1</td>
</tr>
<tr>
<td>Al-10.5Fe-2.5V</td>
<td>above + $\text{Al}_{10}\text{V}$</td>
<td>27.3</td>
<td>0.2</td>
</tr>
<tr>
<td>Al-8Fe-1.4V-1.7Si</td>
<td>$\text{Al}_{12}\left(\text{Fe,V}\right)_3\text{Si}$</td>
<td>24.0</td>
<td>0.025 - 0.12</td>
</tr>
<tr>
<td>Al-12.4Fe-1.2V-2.3Si</td>
<td>$\text{Al}_{12}\left(\text{Fe,V}\right)_3\text{Si}$</td>
<td>36.0</td>
<td>0.025 - 0.12</td>
</tr>
</tbody>
</table>

Thermodynamic data and phase identification for the as-rapidly-solidified microstructures are very limited for RS Al-Fe-X alloys. Optimally-RS-processed Al-Fe-V-Si alloys contain only stable $\text{Al}_{13}\left(\text{Fe,V}\right)_3\text{Si}$ phase with an absence of $\text{Al}_3\text{Fe}$, metastable $\text{Al}_6\text{Fe}$ or other silicide phase shown in Fig. 3 (Ref. 21). The $\text{Al}_{13}\left(\text{Fe,V}\right)_3\text{Si}$ phase is reported to transform to $\text{Al}_3(\text{Fe,V})$ or $\text{Al}_{13}\left(\text{Fe,V}\right)_4$ phase above 500 °C (Ref. 24). The absence of transformation of
the metastable phases at the thermomechanical processing temperatures makes the consolidation and extrusion processes easier than other alloy families as there is no transformation to alternate, degrading phases. The \( \text{Al}_{13}(\text{Fe},\text{V})_3\text{Si} \) phase has a BCC structure \((I\bar{m}3, 138 \text{ atoms/unit cell})\) and is semicoherent with the FCC aluminum matrix, thereby reducing the lattice misfit (3.5 \%) and interface energy at the matrix/dispersoid interface (Ref. 21). The composition of this silicide phase varies over \( \text{Al}_{12.0-14.0}(\text{Fe},\text{V})_3\text{Si}_{0.9-1.28} \), where the Fe:V ratio can also be altered between 5:1 and 11.5:1 (Ref. 21). With the change of composition, the lattice parameter also changes from 1.2587 to 1.2620 nm, thereby changing the interfacial energy.

Phase identification for as-solidified Al-Fe-Mo-V alloys has not been reported. The intermetallic phase is reported to be \( \text{Al}_x(\text{Fe, Mo, V})_y \) type. After thermomechanical processing (TMP), the metastable phase transforms to a stable phase or alternate metastable phases. The intermetallic phases after TMP are reported to be \( \text{Al}_{12}\text{Fe}_3(\text{Mo, V}) \) (BCC), \( \text{Al}_3\text{Fe} \) (monoclinic) and \( \text{Al}_6\text{Fe} \) (orthorhombic) (Ref. 22)
Fig. 3. Al-Fe-V-Si phase diagram at 475 °C (Ref. 21)
1.3. Production and properties of RS/PM Al-Fe-X alloys

There are various ways to produce RS/PM products. Basically, the production process includes rapid solidification processing, degassing and consolidation of powder and secondary forming to the final products as shown Fig. 4.

The most important step is the rapid solidification process where the basic microstructure of the alloy originates (Ref. 49). The solidified microstructure is determined primarily by cooling rate and undercooling of the liquid as well as material composition. The obtainable undercooling is also dependent on RS process characteristics. It is claimed that the undercooling can also be achieved by the melt subdivision for the atomization process where melt is further divided into nucleant-free atomized droplets (Ref. 50). With sufficient undercooling, the RS microstructure becomes a fine cellular or microeutectic structure with a metastable intermetallic phase. In the Al-Fe-V-Si alloys, the rapidly-solidified structure is reported to be a fine microcellular structure containing a stable quaternary silicide phase while other alloy systems contain metastable $\text{Al}_x(\text{Fe},X)_y$-type dispersoids which experience further transformation during subsequent thermomechanical processing (Ref. 22).
Rapid Solidification Processing

Comminution

Degassing and Consolidation

Secondary Forming

Fig. 4. Diagram of RS/PM alloy processing.
The RSP techniques can be classified into basically two categories depending on production methods: 1) atomization techniques and 2) non-atomization techniques. In atomization techniques, droplet formation of molten liquid and solidification occur by interaction with high velocity fluids such as air, flue gas, inert gas, etc. Solidification occurs concurrently or subsequently to the droplet formation. The solidification is achieved predominantly by convective or conductive heat transfer to a high velocity fluid medium. Important characteristics of the powder particles are morphology, size, distribution, surface oxide and hydroxide and homogeneity of the particles. In general, fine, spherical, particles are preferred for satisfactory compaction and good metallurgical bonding (Ref. 49). However, extremely fine particles are difficult to handle, and certain alloys are pyrophoric. Also, increased surface area results in an increase in surface oxidation (Ref. 49). The cooling rate of an atomization technique depends on alloy type, process characteristics and cooling medium and typically ranges from $10^3$ to $10^7$ °C (Ref. 51).

In the non-atomization technique, a thin, continuous stream of melt is solidified by conductive or convective heat transfer to produce a thin ribbon (Ref. 52). Typical non-atomization processes are melt spinning, splat cooling,
planar flow casting, melt extraction and melt drag. The
ribbons are mechanically comminuted to a powder form for
compaction and secondary processing (i.e. extrusion,
rolling). Compared to the atomization processes in which
the cooling rate and, also, microstructure vary depending
on particle size, a uniform zone A structure is reported
for the planar flow casting (PFC) process (Ref. 52).

Al-Fe-V-Si alloys studied in this investigation
originated from Allied Signal Inc.'s planar flow casting
and comminution (PFC/C) process. Fig. 5 shows schematics
of a planar flow casting process and a melt spinning
process. The planar flow casting process was developed to
overcome the problems involved in scaling up the melt
spinning process with the additional advantage of uniform
thickness control. As shown, the planar flow casting
process involves positioning the nozzle from which the
liquid metal is ejected next to the rotating copper wheel.
This arrangement constrains the molten metal and enables
the production of uniform thickness ribbons in the air
(Ref. 89). By adjusting the nozzle-substrate gap, ribbon
thickness and cooling rate can be controlled. The cooling
rate on a production scale was reported to be above 10°
C/s. Ribbons produced by the PFC process are mechanically
comminuted to powder, vacuum degassed and hot pressed
followed by extrusion and secondary forming processes.
Fig. 5. Comparison of (a) free jet melt-spinning and (b) planar flow casting processes (Ref. 89).

Fig. 6. Schematic diagram of RSR process (Ref. 15).
Allied Signal Inc.’s recently-developed FVS1212 and FVS0812 Al-Fe-V-Si alloys have unique features in regard to the extremely fine dispersoids ranging from 0.025 to 0.12 \( \mu \text{m} \) in diameter due to the rapid cooling rate of the planar flow casting (PFC) process compared to other RS/PM Al-Fe-X alloys which exhibit dispersoids ranging from 0.1 to 0.2 \( \mu \text{m} \) in diameter (Ref. 2). Also, the volume % of dispersoids is much higher than in other RS/PM alloys. FVS1212 exhibits 37 volume % dispersoids and FVS0812 exhibits 24 volume % dispersoids (Ref. 2).

The RS/PM Al-8.7Fe-2.8Mo-1V alloy studied in this investigation originated as inert-gas atomized powder produced by Pratt and Whitney’s rapid-solidification-rate (RSR) process. Fig. 6 shows a schematic of the RSR process. A rapidly spinning disk produces fine, spherically shaped powder particles by centrifugal atomization. In conjunction with the rapidly spinning disk, a high velocity (100 m/s) helium gas jet is applied to induce convective cooling in an inert atmosphere. The liquid cooling rate was reported to range from \( 10^5 \) to \( 10^6 \) \( ^\circ \text{C/s} \), depending on particle size (Ref. 15). Degassing, consolidation, forming and secondary forming processes are similar to other RS/PM processing. Final product of the RS/PM Al-Fe-Mo-V alloy shows variety of different dispersoid types with coarser dispersoid particles.
exhibiting a diameter up to 2 μm (Ref. 22).

1.4. Strengthening mechanism in dispersion-strengthened RS/PM aluminum alloys

Strengthening of dispersion-strengthened aluminum alloys occurs by several mechanisms, including dispersion-strengthening, grain boundary strengthening, solid-solution strengthening and strengthening by alpha aluminum matrix itself. At low and intermediate temperature ranges, deformation occurs primarily by dislocation glide and strengthening mechanisms are basically such that block or inhibit the dislocation motion; such as solid solution, dispersoids and grain and subgrain boundaries. At elevated temperatures (>0.5 T_m), deformation by creep should be considered even at low stress ranges as well as deformation by glide. Also, coarsening of the second phase deteriorates both room and elevated temperature properties of the material.

Important factors which influence the mechanical properties of dispersion-strengthened Al-Fe-X alloys include (Ref. 23):

1) interparticle spacing, size, volume fraction and distribution of the dispersoids;
2) strength of the dispersoids;
3) surface energy and misfit of the dispersoids in the
4) solubility and diffusivity of dispersoid-forming elements;
5) alpha grain and subgrain size.

In the following sections, each strengthening mechanisms will be reviewed and correlated to these microstructural characteristics.

**1.4.1. Grain and subgrain boundary strengthening**

One of the advantages of RS/PM processing in producing structural materials is its ability to produce refined microstructural features, such as refined grains and second phase particles. The refined grains provide strength at low and intermediate temperatures by blocking dislocation glide across the grain boundaries, which is expressed by Hall-Petch relationship:

\[ \tau = \tau_0 + Kd^{\frac{1}{2}} \]

(Eq. 1)

where \( \tau \) is the shear strength, and \( \tau_0 \) and \( K \) are constants and \( d \) is the grain size. A similar relationship can be observed for subgrains or cells (Ref. 23). \( \tau_0 \) is the friction stress on a single dislocation and depends on the matrix hardening mechanism.

**1.4.2. Solid-solution strengthening**

Solid-solution hardening is basically caused by two
different effects; dislocation locking and interaction between moving dislocations and solute atoms. For the substitutional solute atoms as in aluminum alloys, the latter becomes a more important mechanism and two effects can be considered (Ref. 23); 1) the internal stress field caused by atomic size differences between solute and solvent atoms and 2) differences in moduli of solute and solvent atoms. The combined effect can be expressed:

$$\Delta \tau = (1/a)(da/dc) + (1/G)(dG/dC). \quad (\text{Eq. 2})$$

For the dispersion-strengthened RS/PM alloys, the solute concentration in the matrix is negligible after thermomechanical processing due to the inherently low solubility of the alloying elements.

1.4.3. Precipitation and dispersion strengthening

Strengthening due to the presence of second phase particles is basically determined by the coherency and distribution of the second phase particles (Ref. 23). When the second phase particles are coherent or semicoherent with a matrix, moving dislocations shear the particles as in the case of coherent precipitation-hardening. Additional energy is consumed in creating a new interface and the strengthening effect can be described by the following equation (Ref. 23).

$$\tau = c (f_v)^n r^n, \quad (\text{Eq. 3})$$
where: c is a alloy constant; f is a volume fraction of particles; r is radius of particles, m and n are constants.

When the second phase particles are incoherent and non-shearable, dislocations move by Orowan looping and the flow stress becomes (Ref. 23):

$$\tau = \tau_o + 0.8 \frac{G_b}{L},$$

(Eq. 4)

where L is the mean interparticle spacing and the other terms denote the conventional meaning. Even for coherent particles, when the particles are non-shearable, Orowan looping occurs and the same relationship can be applied. Assuming the particles to be evenly distributed and spherical with radius r, the above equation becomes

$$\tau = \tau_o + 0.8G_b(\frac{3f}{4\pi})^{1/3}(\frac{1}{r}).$$

(Eq. 5)

As shown in Eqs. 3 and 5, the shear strength increases for coherent precipitates and decreases for non-shearable dispersoids as the particle diameter increases. As a result, for incoherent or semicoherent dispersoid particles, fine particle sizes are required.

As explained above, a basic requirement for dispersion-strengthening is to provide a microstructure of fine, non-shearable, incoherent or semicoherent particles in a matrix. As the transition metal solute atoms in aluminum have limited solubility, normal ingot casting procedures result in coarse dispersoids. As a result, a rapid solidification and powder metallurgy consolidation of
the particulates are used to overcome this difficulty.

1.5. Coarsening mechanisms of second phase particles

1.5.1. Introduction

The excellent room- and elevated-temperature mechanical properties of dispersion-hardened RS/PM Al-Fe-X alloys are attributed to the stability of the dispersoids and their resistance to coarsening during elevated-temperature exposure.

The Al-Fe-V-Si alloy family shows excellent elevated-temperature stability and mechanical properties due to the excellent stability of silicide dispersoids and higher volume % of dispersoids compared with previously-developed dispersion-strengthened Al-Fe-X alloys. Despite the superior properties and stability of the silicides in Al-Fe-V-Si alloys, coarsening and phase transformation of this phase is unavoidable during a long-time exposure at elevated temperatures (above 500 °C).

Thermodynamic data for Al₃(Fe,V)₃Si, which is the only dispersoid present in Al-Fe-V-Si alloy at room temperature, is not fully established. A recent study by Franck et al. (Ref. 24) and Lee et al. (Ref. 25) showed that the silicide phase experiences phase transformation to coarse plates of Al₁₃Fe₄ above 427 °C, resulting brittle cleavage fracture. Below this temperature, the silicide phase is reported to
be very stable.

An important effect of coarsening of the second phase is degradation of mechanical strength due to an increase in interparticle spacing both for coherent and incoherent second phases. As there is no phase transformation of dispersoids in Al-Fe-V-Si alloys below 427°C, the volume fraction of Al13(Fe,V)3Si dispersoid is constant and the only factor affecting the particle spacing is the average size of the dispersoid particles. In other words, a fine particle size is necessary for a small interparticle spacing.

The driving force for the coarsening is the reduction in surface energy. Due to the Gibbs-Thomson effect, the solute concentration adjacent to a particle is given by (Ref. (26)):

$$C_r = C_0 \exp\left(\frac{2\gamma\Omega}{r\kappa T}\right)$$  \hspace{1cm} (Eq. 6)

where $C_0$ = equilibrium concentration of solute in the matrix;

$\gamma$ = surface energy at the particle/matrix interface;

$\Omega$ = atomic volume of solute in the particle;

$r$ = radius of the particle.

The concentration of solute at the interface is greater for the smaller particles. The concentration gradient between the differently-sized particles causes solute flux from the
smaller particles to the larger ones. As a result, the larger particles grow at the expenses of the smaller particles, which was qualitatively described by Ostwald first and bears his name as Ostwald ripening.

Modelling of coarsening kinetics has been applied in various cases (Refs. 26 - 29). A general equation which defines the average radius of particles after aging for a time, t, has the form:

$$r^t = A + B_n t,$$  \hspace{1cm} (Eq. 7)

where $A$ and $B_n$ are constants. The exponent $n$ is a generally an integer between 2 and 5 which depends on the growth mechanism ie., for interface reaction limited (IRL), $n=2$; for volume diffusion controlled (VDC), $n=3$; for grain boundary diffusion, $n=4$; dislocation pipe diffusion; $n=5$.

1.5.2. Coarsening by matrix diffusion

Initial modelling of the coarsening behavior was developed by Lifshitz and Slyozov (Ref. 27), and Wagner (Ref. 26)(LSW theory) based on the following assumptions:

1) a binary system with liquid particles in a liquid matrix of ideal solution;

2) spherical particles with isotropic interfaces;

3) Gibbs-Thomson or Thomson-Freundlich equations provide the solute concentration at the particle/matrix interface;

4) diffusion fields do not overlap;
5) concentration in the matrix is independent of time.

Summarizing the LSW theory:
for the VDC case,
\[ x^3 - x_0^3 = \left( \frac{8\gamma D}{9kT} \right) C_0 t; \quad \text{(Eq. 8)} \]
for the IRL case,
\[ x^2 - x_0^2 = \left( \frac{8}{9} \right) \left( \frac{M}{2} \right) C_0 t; \quad \text{(Eq. 9)} \]
where \( M \) is the interface mobility.

For the both cases, the distribution function \( f(p) \) is skewed to the smaller particle sizes. Experimental observations have generally shown a linear relationship between \( r^n \) and \( t \), however discrepancies in the proportional constant values and wider distribution have been observed. As mentioned above, the LSW theory assumes that the coarsening rate is dependent only on particle size but not on its surrounding. A large number of studies were performed to assess the influence of volume fraction of second phase particles on particle growth rate and size distribution for diffusion-controlled growth. Modification of the LSW theory was initially treated by Ardell (Ref. 30) and considered the diffusion geometry for the volume fraction of second phase for VDC and low volume fraction case. The results showed a wider particle size distribution and an increase in growth rate compared to the
original LSW theory. Unfortunately, the estimation of growth rate for low volume fractions was overestimated for higher volume fractions. Later Asimov (Ref. 31), Davies et al. (Ref. 32), Brailsford and Wynblatt (Ref. 33), and Voorhes and Glicksmann (Refs. 34, 35) tried to correct these discrepancies. Brailsford et al. (Ref. 33) included rate theory and continuity equations for the particle size distribution. The results were in accordance with measurements (Refs. 34, 35). Voorhes et al. (Refs. 34, 35) further treated this Ostwald ripening base upon a detailed solution to the multiparticle diffusion problem, which provides the best agreement with experimental data taken over a wider volume fraction range (0.05 - 0.5). Comparison of $K(f^\infty)/K(0)$ values for each prediction are shown in Fig. 7, where $K(0)$ denotes the constant from the LSW theory and $K(f^\infty)$ denotes the rate constant corrected for volume fraction. Also, the steady-state distribution function by Voorhes et al is given in Fig. 8.

Angers and Fine (Ref. 37) later pointed out that the chemical potential of the rate controlling species should be in equilibrium across the interface and modified the original Gibbs-Thomson equation to Equation 10.
Fig. 7. Comparison of $K(f_v)/K(0)$ values of different predictions for the coarsening by VDC (Ref. 34)

Fig. 8. Steady state distribution function for the coarsening by VDC. The numbers on the lines indicate volume fraction of second phase particles. (Ref. 34)
\[ C_{r_1}(\alpha) = C_{0_1}(\alpha) \exp\left(\frac{2\gamma\Omega_i(\beta)}{r_kT}\right) \]  \hspace{1cm} \text{(Eq. 10)}

where: \( C_{r_1}(\alpha) = \) concentration of i in \( \alpha \) equilibrium with \( \beta \) particle radius of \( r \)

\( C_{0_1}(\alpha) = \) concentration of i in \( \alpha \) for a planar interface

\( \Omega_i(\beta) = \) partial atomic volume of component i in \( \beta \)

Also, diffusion flux of component i equated with growth or shrinkage of the particle. The final results of Angers and Fines's modified equation for the proportional constant in VDC mechanism is expressed:

\[ K_3 = \frac{(8/9)\left[\gamma D_i \Omega_i(\beta)^2 C_{0_1}(\alpha)\right]}{kT x_1(\beta)} \]  \hspace{1cm} \text{(Eq. 11)}

where \( x_1(\beta) \) is the atomic fraction in particle \( \beta \).

Shiflet, Aaronson and Courtney (Ref. 36) analyzed coarsening kinetics for particles having partially coherent boundaries (SAC model). The SAC model incorporates both VDC and IRL coarsening by specifying of precipitate/matrix boundary areas for each process. The SAC model is based on ledge migration controlled coarsening. In this model, broad faces of the ledges are assumed coherent or semicoherent, and therefore immobile, while atomic attachment and detachment can readily occur at the risers of the ledges. In other words, the SAC model is well
suited to describe the coarsening kinetics of disk-shaped precipitates, where the growth kinetics are orientation dependent. For an ensemble of cubes, a linear relationship between the square of edge length and time was observed at the small particle size whereas, for a large particle size, the cube of the ledge length is proportional to time. The influence of the volume fraction of second phase upon coarsening kinetics was less for the ledge mechanism than for a VDC mechanism because of the shorter diffusion distance.

From the literature review (Ref. 37, 38, 39, 45), the coarsening mechanism for dispersoids in Al-Fe-X alloys is basically VDC with \( n = 3 \) superimposed by either static or deformation related short circuit diffusion mechanisms. Chen et al. (Ref.39) applied modified LSW of Brailsford et al. (Ref. 33) to analyze the coarsening in a Al-Zr-V alloy containing a low volume % of \( \text{Al}_3(\text{Zr},\text{V}) \) (5 vol.%) and reported VDC coarsening to be the dominant mechanism at 425 °C. Pentikakos et al. (Ref. 38) analyzed dispersoid coarsening in Al-Fe, Al-Zr and Al-Ni alloys and reported coarsening by grain boundary control at 500 °C and volume diffusion control at 600 °C. Angers et al. (Ref. 37, 45) analyzed coarsening of \( \text{Al}_{13}\text{Fe}_4 \) and \( \text{Al}_{10}\text{Fe}_2\text{Ce} \) dispersoids in Al-8.8 wt.% Fe-3.7 wt. % Ce, which contained 15 vol.% dispersoids. Comparison of the results with the calculated
values showed that grain boundary diffusion and dislocation pipe diffusion were predominant mechanisms below 425 °C and that volume diffusion was predominant at higher temperatures.

1.5.3. Coarsening by short circuit diffusion

Sources to provide short circuit diffusion include grain boundaries and dislocations.

1.5.3.1. Grain Boundary Diffusion

Speight (Ref. 40) and Kirchner (Ref. 28) treated coarsening by diffusion along grain boundaries based on dilute mixture model where the diffusion field and effect of volume fraction were not considered. Basic assumptions are: 1) particles are located in grain boundaries; 2) grain boundary energy and particle/matrix surface energy is in equilibrium; 3) particles take form of two symmetrical caps. The result is

\[ \Sigma_4 - \Sigma_0^4 = \frac{9}{32} \left( \frac{D_{gb} C_0 w y^2}{ABkT} \right) t, \]  

(Eq. 12)

where:

- \( D_{gb} \) = grain boundary diffusion coefficient;
- \( w \) = width of grain boundary;
- \( A = \frac{2-3m+m^3}{3} \) and \( m = \frac{\gamma_{gb}}{2\gamma} \);
- \( B = \frac{1}{2} \ln (n/f) \);
- \( n \) = number of particles on a grain boundary;
- \( f \) = fraction of area occupied by particles.

The size distribution takes a similar form to that
predicted by the LSW model.

1.5.3.2. Dislocation Pipe Diffusion

Krye (Ref. 29) treated coarsening of dispersed phase through dislocation pipe diffusion by the following expression:

\[ r^5 - r_0^5 = \left(\frac{4}{5}\right)^5 \left(\frac{5}{4\pi}\right) (nqD_p \theta C_0 / kT)t, \]  
(Eq. 13)

where \( D_p \) is a diffusion coefficient along dislocation lines, \( n \) is number of dislocations attached to each particle, \( q \) is a cross-sectional area of a dislocation core. When both the matrix volume diffusion and dislocation pipe diffusion contribute to coarsening together, the results may be expressed as the sum of the both equations.

1.5.4. Other Accelerated Coarsening Mechanisms

In addition to the previously described coarsening mechanisms, the coarsening of particles may be accelerated by creep, fatigue and other plastic deformation. Plastic deformation may affect the coarsening by enhancing the diffusion rates. Balluffi et al. (Refs. 41, 42, 43) suggested two major possible causes for this enhanced diffusion with plastic deformation as: 1) generation of excess point defects and 2) short-circuiting along static or moving dislocations or grain boundaries created by
An increase in vacancy concentration accompanied with plastic deformation is well known. Balluffi et al. (Ref. 41) estimated the upper limit of this enhanced diffusion due to the increase in vacancy concentration and concluded that the influence of deformation on diffusion rates through point defect model has very little effect.

In contrast, Ruoff et al.'s analysis (Ref. 42) of strain-enhanced diffusion by a dislocation and grain-boundary short circuiting model suggests an increase in bulk diffusion by these mechanisms. Enhanced diffusion through static dislocation pipe and grain boundary effect is only possible when the diffusion distance in the matrix is much larger than the spacing between the dislocation lines. For the moving dislocations, diffusing atoms can enter dislocation cores either by being overrun or by diffusing into the core and the short circuiting action can be more randomized. The same explanation can be applied to moving grain boundaries either by deformation or recrystallization. Based on the experiment, Laird et al. (Ref. 44) suggested a dislocation pump model for a fatigued Al-Ag alloy where a dislocation collects solute atoms from a GP zone and carry them to growing precipitates. Angers (Ref. 45) later proposed the accelerated diffusion by applied stress during fatigue and creep of materials at
elevated temperature with two effects: 1) a change in concentration of solute at the particle/matrix interface:

\[ C_r = C_0 \exp\left\{ \frac{(2\gamma/r + \sigma_{ij})\Omega}{kT} \right\}, \]  

(Eq. 14)

where \( \Omega \) is the partial atomic volume of solute atom. and 2) stress assisted diffusion produced by potential gradient:

\[ J = -D[\gamma C + (C\gamma V/kT)]. \]  

(Eq. 15)

However, mathematical treatment of the coarsening could not be obtained.

Holm et al. (Ref. 46) applied a grain switching concept to the coarsening of beta precipitates in Zr-2.5% Nb micro-duplex alloy based on Ashby et al.'s model (Ref. 47) to account for the coalescence of grain boundary particles. The mechanism is based on coalescence of grain boundary second phase due to grain switching by grain-boundary sliding and rotation with diffusional accommodation for materials of relatively small grain size. This kind of phenomena was also observed during superplastic deformation of a Cu-7 wt. % P alloy by Arrowood et al. (Ref. 48).

1.6. Review of joining dispersion-strengthened Al-Fe-X alloys

The basic requirement for the joining of RS/PM Al alloys is that the RS/PM processed microstructure be either
retained or recreated through the joining process in order to maintain room and elevated-temperature mechanical properties. Previous work has shown that this requirement can be achieved using both fusion and solid-phase welding process.

1.6.1. Fusion Welding Processes

Fusion welding processes are generally the first choice for the joining of structural materials due to their versatility in a wide range of applications. However, due to the metastable microstructure and inherently high gas content exhibited by RS/PM Al-Fe-X alloys, only fusion welding processes with high heating/cooling rates and steep temperature gradients such as electron beam (EBW) and laser welding (LW) can be applied to the alloys which have been rigorously degassed.

Krishnaswamy et al. (Ref. 6, 7) studied EB and Nd:Yag laser welding of an RS/PM Al-8Fe-2Mo alloy which was degassed down to 1 ml/100 g by producing and handling the powder in an inert gas followed by vacuum-degassing. The result of EB welding trials showed that unmelted intermetallic particles in the partially-melted zone (PMZ) served as nucleation sites for coarse, acicular intermetallic particles in the fusion zone (FZ) boundary. Coarsening of dispersoids in the HAZ was also observed.
Tensile failure occurred at the fusion boundary. Reduced energy input and higher cooling rate reduced coarsening of dispersoids both in the FZ and HAZ and resulted in improved joint efficiencies above 85%.

Pulsed Nd:Yag laser welding of Al-8Fe-2Mo was also studied by Krishnaswamy et al. (Ref. 7). Rapid cooling rates and steeper temperature gradients prevented coarsening of dispersoids in the PMZ and HAZ, and resulted in 100% tensile joint efficiencies. However, the shallower temperature gradient in the reheated HAZ of the FZ between the successive passes promoted the formation of acicular Al$_3$Fe dispersoids.

Baeslack et al. (Ref. 8) studied capacitor-discharge welding of the RS/PM Al-8Fe-4Ce alloy. The results showed that extremely rapid heating and cooling rates accompanied by the simultaneous application of high pressure suppressed the formation of hydrogen porosity and resulted in a rapidly-solidified structure in the fusion zone without any degradation of the surrounding base-metal microstructure. A subsequent TEM (Ref. 54) study of the weld showed interdendritic dispersoids finer than those present in the base metal.

In summary, principal limitations of fusion welding processes are: 1) they can be only applied to very low hydrogen content alloys; 2) heat input must be low to
promote high cooling rates to provide rapid-solidification rates in the FZ and to preclude coarsening of the dispersoids in the HAZ.

1.6.2. Solid-Phase Welding Processes

Most RS/PM Al-Fe-X alloys have relatively high hydrogen contents, resulting in the formation of hydrogen-induced porosity during fusion welding. Solid-phase welding processes, in which fusion and resolidification do not occur, provide an alternative for joining these materials. Various solid-phase welding processes can be considered; continuous-drive friction welding, inertia-friction welding, linear-friction welding, diffusion welding, flash welding, explosion welding etc..

Ananthananayaranan (Ref. 55) studied diffusion welding and transient-liquid phase (TLP) welding with an Ag interlayer for flue-gas atomized RS/PM Al-9Fe-4Ce. Contrary to the conventional aluminum alloys, diffusion welding could be performed up to 600 °C without a structural support due to the high elevated-temperature strength of the alloy. Large deformation of the contact area was required to produce quality welds suggesting that the welding was accomplished by the deformation rather than diffusion. The results showed that joint efficiencies up to 70 % could be obtained with a 50 % increase in joint
area. TLP (transient liquid phase) welding of the alloy with an Ag interlayer was not satisfactory due to non-uniform wetting by the Al-Ag eutectic of the RS/PM base metal and Kirkendall porosity formation in the base metal.

Hagey et al. (Refs. 9, 59) studied the inertia-friction welding of RS/PM Al-Fe-Ce alloys. The results showed the weld axial force to significantly influence the weld integrity, with an increased axial force eliminating the coarsened structure and dispersoids at the weld interface. A subsequent TEM study of the inertia-friction welding Al-8.7Fe-2.8Mo-1.0V alloy by Hou et al. (Ref. 10) showed the refinement of coarse base metal dispersoids rather than their coarsening due to their fracture and dispersion at the weld interface. This refinement promoted an increase in hardness in the weld interface region with the increase in weld axial force. At low weld axial force, nonuniform deformation caused dispersoid-lean and rich regions at the outer periphery of the welds. For both studies, the effects of hydrogen was negligible. Other than the above work, studies on the solid state welding of the RS/PM Al-Fe-X alloys (i.e. Al-Fe-V-Si) have not been reported.
1.6.3. Dissimilar-alloy welding between aluminum alloys and other alloys

1.6.3.1. Review of dissimilar-alloy welding

Dissimilar alloy welding has inherent problems due to differences of physical and metallurgical characteristics between alloys. Also, service characteristics of dissimilar alloy welds such as thermal fatigue, interdiffusion, galvanic corrosion etc. are also important concerns. Major joining problems originate from differences in strength, thermal expansion coefficient, melting point and intermetallic formation. Although dissimilar-alloy joints particularly in ferrous base alloys have been produced by fusion welding processes, a large number of the dissimilar welds can only be produced using solid-state welding processes due to the problems described above (Ref. 56).

A major objective of this review is to develop methods to join the RS/PM Al-Fe-X alloys to IM aluminum alloys and titanium alloys. Due to the large difference in melting point between aluminum and titanium alloys, and the formation of brittle intermetallic compounds, fusion welding of the dissimilar joints between the RS/PM Al-Fe-X and titanium alloys is precluded. Also, most of the IM aluminum alloys are heat-treatable with a high amount of alloying elements, fusion welding of the dissimilar welding
even between the RS/PM Al-Fe-X and IM Al alloys may result in a harmful intermetallic formation in addition to the structural degradation by the coarsening and transformation of dispersoids and precipitates.

1.6.3.2. Dissimilar welding between RS/PM Al-Fe-X alloys and IM aluminum alloys

Hagey (Ref. 59) studied dissimilar welding of the RS/PM Al-9Fe-4Ce to IM 2024-T351 by inertia-friction welding. The results showed extensive deformation of the soft IM 2024-T351 in the weld region and fine dynamically recrystallized alpha grains, which appeared featureless. In this recrystallized region, defects associated with intermetallic stringers were observed, which might have originated by melting or plastic deformation of intermetallics. Even with the slight deformation in the Al-Fe-Ce, mechanical mixing between the two alloys was observed. Coarsening of dispersoids near the weld interface was indicated. However, the limited spatial resolution of the light microscopy and scanning-electron microscopy (SEM) precluded the detailed microstructural characterization of this weld region. Fracture occurred in the Al-Fe-Ce side near the interface.
1.6.3.3. Dissimilar-alloy welding between RS/PM Al-Fe-X alloys and titanium alloys

As there are no previous studies on the dissimilar-alloy welding of RS/PM Al-Fe-X to titanium alloys, this section is rather devoted to a general review of the dissimilar-alloy welding characteristics of IM aluminum alloys to titanium alloys.

The basic problem with the joining of aluminum alloys to titanium alloys results from the formation of brittle Al₃Ti intermetallic. The conventional fusion welding of Ti/Al joints is very restricted to the special cases where the reaction between aluminum and titanium was minimized and most of the melting occurred in aluminum during braze welding (Ref. 60).

Diffusion welding of aluminum and titanium alloys has also been reported with and without insert layers (Ref. 61, 62, 63). Diffusion welds between IM-Al-2017 and CP Ti without any insert showed a noticeable increase in the joint strength at a welding temperature slightly above the solidus of Al-2017 (570 °C), meaning that the presence of liquid was required for weld formation. The presence of Al₃Ti intermetallics was observed and its reported thickness up to 10 μm did not influence tensile properties. Diffusion welding of pure aluminum and titanium with a silver interlayer showed that bonding was possible in the
range of 550 to 600 °C where a partial liquid phase formed and removed the surface aluminum oxide film. The above results are based on diffusion welding for 30 minutes. The possibility of applying the above diffusion welding of Al-Fe-X alloys depends on the behavior of dispersoids at the heating temperature as well as their interaction at the interface.

Iwamoto et al. (Ref. 64) used pressure welding to join an aluminum-precoated pure titanium to aluminum. Pure aluminum was precoated by dipping titanium into a molten pure aluminum bath for 15 seconds at 750 °C. The precoated titanium did not show any presence of intermetallics. The joint showed maximum strength when welded at 300 °C with 20 % aluminum deformation.

Friction welding of titanium alloys to aluminum alloys was also investigated by Nanba et al. (Ref. 65) and Shternin (Ref. 66). Nanba et al. (Ref. 65) studied friction welding of CP Ti to Al-6063-T83. After process parameter optimization, defect-free welds were produced. EPMA analysis across the joint showed no formation of intermetallics and negligible interdiffusion across the joint. Shternin's study with 3xxx and 5xxx Al alloys to VT5 Ti also showed the strength of the joint to be equal to the Al base metal.

Explosive welding of a transition joint of Al-Mg
alloys/Ti/mild steel was also studied by Tatsukawa (Ref. 67). Although melting phenomena and diffusion of Ti into the aluminum alloy side were observed, the joint was reported to be structurally sound.

Summarizing the above, the dissimilar-alloy welding of aluminum alloys to titanium using solid-phase welding processes is basically feasible after optimization of the process characteristics. For the dissimilar-alloy welding of RS/PM Al-Fe-X alloys to titanium alloys, a prime concern is the preservation or recreation of the RS microstructure as well as producing a high integrity weld between the two alloys.

1.6.3.4. Interdiffusion and intermetallic formation in Al/Ti diffusion couple

Interdiffusion between titanium (or titanium alloys) and aluminum (or aluminum alloys) was studied by several investigators (Ref. 68 - 71). The studies were related to the brazing of Ti with Al alloy fillers (Ref. 68) and kinetics of premade diffusion couples during isothermal annealing (Ref. 70, 71, 72). The major concern of the studies was the formation of intermetallic compounds at the interface and the kinetic behavior of the compound layer or layers. All of these experiments may be summed up by Van Loo et al's studies (Ref. 69, 70).
Van Loo et al (Ref. 69) studied the diffusion kinetics of Al and Ti, and Al and Ti-Al alloys using the diffusion couples made from hot dipping, cold pressing, diffusion bonding and vapor metallizing. The experiments were conducted at a relatively high temperature range from 580 to 640 °C. The diffusion couples of Ti and Al showed only growth of Al$_3$Ti at the interface even though the formation of other intermetallic compounds such as TiAl$_2$, TiAl, Ti$_2$Al$_5$ and Ti$_3$Al is thermodynamically possible. The basic reason for the absence of these other intermetallic compounds is the rapid diffusion of Al in Al$_3$Ti compared to the other intermetallics, resulting in negligible presence of the other intermetallics. The formation and growth of Al$_3$Ti occurred on the Ti side due to the higher diffusion rate of Al compared to Ti in Al$_3$Ti, forming Kirkendall porosity near the Al side. An increase in Al concentration in the titanium alloy increased the growth rate of the Al$_3$Ti layer. Additionally, the diffusion between Ti and various Ti-Al compounds were studied (Ref. 70). Intermediate phases were observed which were not observed for the Ti/Al couple but an arrangement as a Ti/Ti$_3$Al/Al only resulted in an Al$_3$Ti layer. Tardy et al (Ref. 71) performed a similar diffusion couple experiment on sputtered pure Al films on pure Ti substrates at temperatures ranging from 350 to 500 °C with similar results.
Tadashi et al (Ref. 68) studied the brazing of Ti at 680 °C using an Al-Si alloy with the Si content varying from 0.1 to 10 %. Up to 0.8 % Si, the formation of the Al₃Ti layer was decreased with an increase in Si content. Above 0.8 % Si, the thickness of intermetallic compound increased, forming Ti₃Al₂Si₂. The main reason for the reduced growth of Al₃Ti was reported to be the increase in activation energy of Al in the presence of Si in Al₃Ti.

Summarizing the above, Al₃Ti is the major intermetallic compound forming in the Al/Ti diffusion couple but uncertainty exists regarding the effects of alloying elements of Al-Fe-V-Si and Al-Fe-Mo-V alloys.

1.7. Friction Welding Processes

1.7.1. Introduction

According to the Welding Handbook (Ref. 72), "friction welding is a solid-state joining process that produces coalescence by the heat developed between two surfaces by mechanically induced rubbing motion." Welding occurs below the melting point of the material and friction between the two surfaces accompanied by an axial compressive force effectively removes softened materials and surface contaminants from the interface region.

Depending on the methods applying frictional motions, friction welding can be classified as rotational friction
welding and linear-friction welding (LFW). Rotational friction welding includes continuous-drive-friction welding (CDFW or simply "friction welding") and inertia-friction welding (IFW) depending on the methods of applying rotational motion. A comparison of the two processes is shown in Fig. 9.

1.7.2. Rotational-friction welding processes

Process characteristics of continuous-drive-friction welding (CDFW) and inertia-friction welding (IFW) are well documented in the literature (Ref. 72, 73). The basic difference between continuous-drive-friction welding and inertia-friction welding process is the method of applying the rotational motion. In CDFW, the work piece on the spindle side is rotated at a constant speed accompanied by a compressive axial force. The weld cycle duration can be controlled by preset time or axial shortening. In IFW the work piece is connected to a flywheel. The flywheel is accelerated to a preset speed and the stored inertial energy is dissipated through the friction at the interface.

Characteristics transient responses of the two processes are shown in Fig. 10. Basic differences between continuous-drive-friction welding and inertia-friction welding processes are the transient responses of rotation speed and torque.
Fig. 9. Comparison of inertia-friction welding (a, b) and linear-friction welding (c, d) processes.
Fig. 10. Comparison of transient responses of welding parameters in (a) continuous-drive-friction and (b) inertia-friction welding processes (Ref. 90).
The difference between the two processes can be also observed in behavior during the final upset stage; expulsion and axial shortening during the final upset stage in IFW is more pronounced due to the large peak in torque, while axial shortening is gradual in CDFW.

As only inertia-friction welding process were utilized in this study, the effects of process variables for the inertia-friction welding process are discussed. However, basic metallurgical phenomena can be assumed almost same with minor differences. Process variables in inertia-friction welding are initial-surface velocity, inertia, and axial pressure. Optimum conditions depend on material characteristics and dimensions; specific heat, thermal conductivity, flow stress, melting point, etc.. Other than process variables, surface cleanness is the most important factor for satisfactory welding. Little (Ref. 74) proposed a Q value, which is the energy required to weld a particular material as below:

\[ Q = (K \rho C)^{1/2} T_{mp}/10^4. \]  
(Eq. 16)

Oberle (Ref. 74) suggested required power values to calculate the lower critical speed based on material yield strength as shown in Table 3. However, these values are empirical especially for dissimilar alloy combinations. In general, the alloy combinations which form intermetallics are recommended to be welded at a lower
rotating speed, with high axial pressure to reduce the interface temperature and to improve mechanical interlocking.

Table 3. Specific power requirements and lower critical speeds for various materials (Ref. 74).

<table>
<thead>
<tr>
<th>Material</th>
<th>Specific Power ($10^6$ J/m²s)</th>
<th>Lower Critical Initial Speed (x10² m/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Lead</td>
<td>58</td>
<td>25</td>
</tr>
<tr>
<td>Stainless Steel</td>
<td>302</td>
<td>102</td>
</tr>
<tr>
<td>Aluminum</td>
<td>442</td>
<td>127</td>
</tr>
<tr>
<td>Tool Steel</td>
<td>500</td>
<td>140</td>
</tr>
<tr>
<td>Low Carbon Steel</td>
<td>546</td>
<td>178</td>
</tr>
<tr>
<td>Nickel</td>
<td>756</td>
<td>330</td>
</tr>
<tr>
<td>Titanium</td>
<td>930</td>
<td>381</td>
</tr>
<tr>
<td>Copper</td>
<td>1104</td>
<td>914</td>
</tr>
<tr>
<td>Molybdenum</td>
<td>1453</td>
<td>1016</td>
</tr>
<tr>
<td>Tungsten</td>
<td>2034</td>
<td>1270</td>
</tr>
</tbody>
</table>

The thermal history of the weld region in IFW was studied by Wang et al (Refs. 73, 91) for AISI 1020 steel and the results are shown in Fig. 11. The periphery of the weld was shown to experience the highest temperature due to increased friction with increased rotational movement. The heat generated at the outer interface conducts to the center region as well as in the longitudinal direction equilibrating the interface temperature. The softened
interface region is expelled out due to applied pressure with rotational motion.

With other parameters fixed, the applied pressure dictates the peak temperature and welding time; with an increase in pressure, peak temperature increases and weld time is reduced as shown in Fig. 12. The applied pressure is also affected by thermal gradient which varies during the major portion of the welding cycle. The applied pressure is uniform at the early stage of the weld cycle as the faying surfaces are brought to complete contact. The pressure decreases as the periphery temperature increases, initiating thermoplastic flow at the periphery and then it becomes uniform as the temperature distribution becomes more even at the later stage of the welding cycle.

1.7.3. Linear-friction welding process

TWI's recently developed linear-friction welding process utilizes the friction generated from a linear motion under a normal force instead of the rotational motion used in conventional rotational-friction welding processes (Ref. 11). A comparison of the two processes is shown in Fig. 9.
Fig. 11. Calculated interface temperature distribution in inertia-friction welding (Ref. 91).

Fig. 12. Effect of welding time on temperature distribution in inertia-friction welding (Ref. 91).
Similar to conventional rotational-friction welding, forging and upset force can be controlled. Instead of rotating speed, frequency and linear amplitude as well as axial force and upset can be controlled to change the welding behavior of joints. Flash is formed mainly in the direction of linear motion. A major advantage of this process is that it can be used to join non-axially-symmetric parts. Compared to rotational friction welding, relatively uniform heating is expected. As the process is recently developed, analysis of the welding behavior is not available but successful welding has been reported for various materials (Ref. 11).

1.7.4. Welding Mechanisms in Friction Welding

Several theories have been proposed to explain the welding mechanism involved in continuous-drive-friction and inertia-friction welding. Krye (Ref. 75) suggested microscopic melting as the bonding mechanism but evidence to support the theory was not obtained. Rao et al (Ref. 76) suggested mechanical interlocking of nascent surfaces, wedge formation and smearing of wedges to form a strong adhesion weld as a possible sequence of welding mechanism. Others suggested similar mechanisms (Ref. 77, 78), but all the explanations can be categorized as perfect contact of the two nascent surfaces. All these cases assumed
recrystallization accompanied with deformation and heating of the materials at the interface.

The other explanation of the welding mechanism is high temperature diffusion across the interface during the heating and forging stage of the welding either with or without recrystallization (Ref. 78). Even though it is very difficult to verify experimentally, it gives a good explanation for dissimilar-alloy welding with a hard/soft material combination, where negligible deformation is observed in the hard material. Szakacs (Ref. 79) further combined mechanical mixing and diffusion as a proposed welding mechanism. Whichever mechanism is applicable, the basic concept of the welding mechanism is perfect contact of two nascent surfaces to form a metallic bond across the interface. McMullan (Ref. 80) further explained the welding mechanism of dissimilar-alloy friction welds for soft/hard material combinations macroscopically. During the initial stage of welding, soft metal covers the hard material. The coating of soft metal on the hard metal forms a stagnant layer and consequently the plane of rubbing moves axially into the soft material.
1.7.5. Evaluation of friction welds

For the similar-alloy inertia-friction welding, mechanical tests, which have been conventionally used for the other similar-alloy welding process, such as tensile test, bend test, hardness test etc.

However, mechanical testing of dissimilar friction welds is a difficult problem as the values of conventional tests are dubious due to the restraint at the bond and occasional weaker strength of base metal compared to the bond strength (Ref. 81). The mechanical test should fulfill two requirements: 1) provide information about the quality of the welds and 2) provide information about the whole welded structure for design purposes. Conventional tests such as tensile tests can be utilized for the second purpose. Various tests have been tried to evaluate the bond quality but satisfactory tests were not found for these purposes. Available methods can be listed as the hammer bend test, controlled bend test, tensile test, impact test, torsion test, static shear test and impact shear test (Refs. 81, 82). Still, the basic problem arises from the fact that failure occurs slightly away from the interface in soft material for a hard/soft alloy combination. With the above uncertainty, conventional tensile testing still gives useful information on the joints.
CHAPTER II
OBJECTIVES

The overall objective of this research was to investigate the friction welding of state-of-the-art RS/PM dispersion-strengthened aluminum alloys both in similar-alloy and dissimilar-alloy combinations with conventional ingot metallurgy aluminum and titanium alloys. Specific objectives of this study are stated below.

1) Through process parameter variation and optimization, to determine the feasibility of producing high integrity friction welds of the following types: 1) similar-alloy inertia- and linear-friction welds in two Al-Fe-V-Si alloys which contain different dispersoid contents; 2) dissimilar-alloy inertia-welds between Al-Fe-V-Si and Al-Fe-Mo-V alloys and conventional aluminum alloy 2024-T351; 3) dissimilar-alloy inertia-friction welds between the RS/PM dispersion strengthened aluminum alloys and conventional titanium alloys.

2) Determine the influence of the friction welding

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process (inertia versus linear) and process parameters on microstructure evolution which occurs during welding through detailed microstructural characterization across the weld interface region.

3) Evaluate mechanical properties and fracture characteristics of the respective weld types and develop fundamental relationships between these characteristics and the weld structures both for the as-welded condition and, for similar-alloy welds, following exposure to elevated temperatures.

4) Investigate microstructure evolution during the friction welding process and subsequent thermal exposure and examine the potential for predicting the microstructural changes using particle coarsening models.
3.1. Introduction

The experimental program was comprised of three principal parts: 1) the generation and analysis of similar-alloy welds in dispersion-strengthened RS/PM aluminum alloys, 2) the generation and analysis of dissimilar-alloy welds between dispersion-strengthened RS/PM aluminum alloys and titanium alloys and 3) analysis of the weld structure following exposure to elevated-temperatures.

In the first and second parts, the experimental procedure consisted of base metal characterization followed by weld generation using inertia-friction welding (IFW) and linear-friction welding (LFW) with different, systematically varied combinations of welding parameters. Selected welds were subsequently evaluated using detailed microstructural characterization, mechanical testing and fractographic analysis. In the third part of the investigation, microstructure and hardness changes of welds subjected to thermal exposure were studied.
3.2. Materials

The materials evaluated in this study included RS/PM Al-11.7 wt.% Fe-1.2 wt.% V-2.4 wt.% Si (hereafter designated as FVS1212), Al-8.5 wt.% Fe-1.3 wt.% V-1.7 wt.% Si (hereafter designated as FVS0812), Al-8.7 wt.% Fe-2.8 wt.% Mo-1.0 wt.% V (hereafter designated as Al-Fe-Mo-V), Al 2024-T351 (Al-4.35 wt.% Cu-1.5 wt.% Mg-0.6 wt.% Mn), CP (commercially pure) Ti ASTM grade 4, Ti-6.0 wt.% Al-2.75 wt.% Sn-4.0 wt.% Zr-0.4 wt.% Mo-0.4 wt.% Si (Ti-1100) and Ti-6 wt.% Al-6 wt.% V-2 wt.% Sn (Ti-662). FVS1212 and FVS0812 were supplied by Allied-Signal Inc. as extruded bars 32 and 25 mm in diameter, respectively. The alloys were produced by a patented planar flow casting and comminution (PFC/C) process. Al-Fe-Mo-V was supplied as a 254 mm diameter extrusion produced by Pratt and Whitney’s "RSR" process. Cylindrical specimens 22 mm in diameter and 50 mm in length were machined from the extrusion with the longitudinal direction oriented perpendicular to the extrusion direction. The hydrogen contents of the alloys were reported to be greater than 1 ml/100 gm aluminum (Ref. 52).

Ti-1100 was supplied as upset forged pancakes 254 mm in diameter and 25 mm in thickness. Cylindrical specimens 25 or 22 mm in diameter and 50 mm in length were machined from the as-supplied pancakes in the diameter direction.
Ti-662 and CP Ti Grade 4 were obtained as commercially-available extruded round bar 25 mm in diameter. The rods were machined down to a diameter of 22 mm and sectioned into 50 mm lengths for inertia-friction welding. For linear-friction welding, rods 25 mm in diameter and 50 mm in length were used.

3.3. Welding Procedure

Most of the welds were produced using cylindrical bar specimens 22 mm in diameter except for linear-friction welds and some of the dissimilar-alloy welds between titanium and aluminum produced with restraint rings. For linear-friction welds, cylindrical bar specimens 25 mm in diameter was used. For dissimilar-alloy welds between aluminum alloys and titanium alloys with restraint rings, titanium alloy bar with a smaller diameter (19 mm) was welded to aluminum alloy bar 22 mm in diameter. All of the weld specimens were cleaned in methanol and the faying surfaces were dry-machined immediately prior to welding.

Inertia-friction welds were produced using MTI (Manufacturing Technologies Inc.) Model 120 and 180B inertia-friction welding systems. Linear-friction welds were produced using a large scale linear-friction welding machine located at The Welding Institute in Cambridge, England (Ref. 11). Inertia-friction welding parameters for
the welds in FVS1212 and FVS0812 alloys were chosen based on previous studies (Refs. 9 and 10) of inertia-friction welding of dispersion-strengthened RS/PM Al-Fe-Ce and Al-Fe-Mo-V alloys and extensive preliminary trials performed in this study. In the preliminary trials, symmetry, uniformity and continuity of the weld flash, upset and weld defects were evaluated. From the previous studies, weld axial force was found to be the most important factor in controlling the weld microstructure and defect formation. After identifying the optimum moment-of-inertia and rotational speed for the alloys, two different levels of axial force (high and low) were chosen to compare the weld structures and mechanical properties, with high axial force representing the optimum level. For linear-friction welding, the welding parameters were chosen based on the welding trials of conventional high strength aluminum alloys at The Welding Institute (Ref 11).

For the dissimilar-alloy welding of RS/PM aluminum alloys and titanium alloys, two different specimen arrangements were utilized; with and without a restraint ring on the aluminum side. For welding performed with a restraint ring, a high moment-of-inertia, axial force and a low rotating speed were used to obtain a forge weld type bond with a low interface temperature. For welding performed without a restraint ring, a low moment-of-inertia
and axial force and a high rotating speed were used to obtain a typical friction weld type bond with a high interface temperature. The restraint ring was applied to prevent excessive deformation and upset of the soft aluminum alloys when the high axial force was applied. For welding performed with a restraint ring, aluminum alloys 22 mm in diameter with a restraint ring were welded to the titanium alloys 19 mm in diameter. The stick-out of the aluminum alloy was 3 mm beyond the restraint ring. For the welding without a restraint ring, aluminum alloys and titanium alloy bars of the same diameter of 22 mm were used.

Welding parameters for each process/material combination are shown in Tables 4 through 7. Table 6 shows the selected parameter ranges of preliminary dissimilar-alloy welding trials between RS/PM aluminum alloys and titanium alloys and final parameter combinations are shown in Table 7.
Table 5: Inertia-friction welding parameters for aluminum alloy welds and axial displacement measurements

<table>
<thead>
<tr>
<th>Alloy&lt;sup&gt;a&lt;/sup&gt;/ Process&lt;sup&gt;b&lt;/sup&gt; Combination</th>
<th>Axial Force (KN)</th>
<th>Moment of Inertia (Kg-m²)</th>
<th>rpm</th>
<th>Axial Displacement (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1212/1212 IFW/L</td>
<td>55.8</td>
<td>0.17</td>
<td>5,000</td>
<td>4.5</td>
</tr>
<tr>
<td>1212/1212 IFW/H</td>
<td>83.7</td>
<td>0.17</td>
<td>5,000</td>
<td>8.9</td>
</tr>
<tr>
<td>1212/2024 IFW/L</td>
<td>83.7</td>
<td>0.17</td>
<td>3,500</td>
<td>2.4</td>
</tr>
<tr>
<td>1212/2024 IFW/H</td>
<td>139.5</td>
<td>0.17</td>
<td>3,500</td>
<td>7.1</td>
</tr>
<tr>
<td>0812/0812 IFW/L</td>
<td>55.8</td>
<td>0.17</td>
<td>5,000</td>
<td>2.7</td>
</tr>
<tr>
<td>0812/0812 IFW/H</td>
<td>83.7</td>
<td>0.17</td>
<td>5,000</td>
<td>10.2</td>
</tr>
<tr>
<td>FMV/2024 IFW/L</td>
<td>97.8</td>
<td>0.17</td>
<td>3,500</td>
<td>2.78</td>
</tr>
<tr>
<td>FMV/2024 IFW/M</td>
<td>126.6</td>
<td>0.17</td>
<td>3,500</td>
<td>5.96</td>
</tr>
<tr>
<td>FMV/2024 IFW/H</td>
<td>153.6</td>
<td>0.17</td>
<td>3,500</td>
<td>10.92</td>
</tr>
</tbody>
</table>

a. 1212, 0812 and FMV denote FVS1212, FVS0812 and Al-8.7 Fe-2.8 Mo-1.0 V, respectively.
b. L, M and H denote low, medium and high axial force respectively.
c. Axial displacement values are average values based on measurements from at least two weld specimens.
Table 6: Linear-friction welding parameters for RS/PM aluminum alloy welds and displacement measurements

<table>
<thead>
<tr>
<th>Alloy/Process</th>
<th>Friction Force (KN)</th>
<th>Upset Force (KN)</th>
<th>Frequency* (Hz)</th>
<th>Weld Time</th>
<th>Axial Displ. (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1212/1212** LFW</td>
<td>30</td>
<td>50</td>
<td>50</td>
<td>2.5</td>
<td>6.5</td>
</tr>
<tr>
<td>0812/0812 LFW</td>
<td>30</td>
<td>50</td>
<td>50</td>
<td>2.1</td>
<td>5.8</td>
</tr>
<tr>
<td>1212/2024 LFW</td>
<td>30</td>
<td>50</td>
<td>50</td>
<td>5.45</td>
<td>5.2</td>
</tr>
</tbody>
</table>

a. Frequency at amplitude of 2 mm.
b. Prebonding burnoff was set at 4 mm.

Table 7: Inertia-friction welding parameters for preliminary welding trials on dissimilar-alloy welds between RS/PM Al-Fe-Mo-V and Al-Fe-V-Si, and Ti alloys

<table>
<thead>
<tr>
<th>Moment of Inertia (Kg-m²)</th>
<th>RPM</th>
<th>Axial Force (KN)</th>
<th>Axial stress (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.05</td>
<td>675</td>
<td>28</td>
<td>102</td>
</tr>
<tr>
<td>0.17</td>
<td>825</td>
<td>56</td>
<td>203</td>
</tr>
<tr>
<td>0.82</td>
<td>950</td>
<td>84</td>
<td>305</td>
</tr>
<tr>
<td>0.89</td>
<td>1500</td>
<td>112</td>
<td>407</td>
</tr>
<tr>
<td>1.16</td>
<td>3500</td>
<td>140</td>
<td>509</td>
</tr>
<tr>
<td>2.18</td>
<td>5500</td>
<td>168</td>
<td>610</td>
</tr>
</tbody>
</table>

Some of the extreme cases such as low inertia, low RPM and low axial force, and high inertia, high rpm and high axial force were excluded, which were not practical due to excessive upset or minimal upset or obviously predictable lack of bond.
Table 9: Inertia-friction welding parameters for dissimilar-alloy welds between aluminum and titanium alloys and axial displacement measurements

<table>
<thead>
<tr>
<th>Alloy/Process</th>
<th>Axial Force (KN)</th>
<th>Moment of Inertia (Kg-m²)</th>
<th>rpm</th>
<th>Axial Displ. (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>2024/Ti-662 IFWLᵃ</td>
<td>157</td>
<td>1.16</td>
<td>950</td>
<td>1.0</td>
</tr>
<tr>
<td>2024/Ti-662 IFWM</td>
<td>310</td>
<td>1.16</td>
<td>950</td>
<td>2.2</td>
</tr>
<tr>
<td>2024/Ti-662 IFWH</td>
<td>462</td>
<td>1.16</td>
<td>950</td>
<td>5.8</td>
</tr>
<tr>
<td>FMV/Ti-1100 IFWLᵇ</td>
<td>210</td>
<td>0.89</td>
<td>950</td>
<td>2.5</td>
</tr>
<tr>
<td>FMV/Ti-1100 IFWH</td>
<td>210</td>
<td>1.75</td>
<td>675</td>
<td>5.5</td>
</tr>
<tr>
<td>FVS0812/Ti-1100 IFWLLLᶜ</td>
<td>84</td>
<td>0.17</td>
<td>3500</td>
<td>2.1</td>
</tr>
<tr>
<td>FVS0812/Ti-1100 IFWLLM</td>
<td>112</td>
<td>0.17</td>
<td>3500</td>
<td>4.8</td>
</tr>
<tr>
<td>FVS0812/Ti-1100 IFWLLH</td>
<td>140</td>
<td>0.17</td>
<td>3500</td>
<td>7.0</td>
</tr>
<tr>
<td>FVS0812/Ti-1100 IFWLHL</td>
<td>28</td>
<td>0.17</td>
<td>5500</td>
<td>1.3</td>
</tr>
<tr>
<td>FVS0812/Ti-1100 IFWLHH</td>
<td>56</td>
<td>0.17</td>
<td>5500</td>
<td>4.9</td>
</tr>
<tr>
<td>FVS0812/Ti-1100 IFWHLH</td>
<td>156</td>
<td>0.822</td>
<td>1500</td>
<td>12.6</td>
</tr>
<tr>
<td>FVS1212/CP Ti IFWLᵈ</td>
<td>112</td>
<td>0.17</td>
<td>3500</td>
<td>2.7</td>
</tr>
<tr>
<td>FVS1212/CP Ti IFWH</td>
<td>112</td>
<td>0.17</td>
<td>5000</td>
<td>8.9</td>
</tr>
</tbody>
</table>

a. 2024-T351 of 25.4 mm and Ti-662 of 22.4 mm in diameter with restraint ring.
b. Al-Fe-Mo-V of 22.4 mm and Ti-1100 of 19.1 mm in diameter with restraint ring.
c. and d. both sides are 22.4 mm in diameter.
d. Axial displacement values are average values based on minimum two measurements.
3.4. Microstructural characterization

Following welding, the weld axial displacement and flash uniformity were visually evaluated. Representative welds were sectioned axially, mounted in epoxy, mechanically ground down to 600 grit SiC paper, rough polished with 3 micron diamond compound and 0.3 micron alumina and final polished using a colloidal silica suspension. Following defect analysis of the as-polished specimen surfaces using light microscopy, microstructural features were revealed by etching with Keller's reagent.

Alternate specimen halves were carefully sectioned longitudinally at the axial centerline into thin slices 0.3 mm in thickness and mechanically thinned down to 0.125 mm. Disks 3 mm in diameter were punched from the foils in the unaffected base metal and from representative regions in the weld HDZ's and electrojet polished in one part nitric acid and three parts methanol at -30 °C. Thin foils were examined in a JEOL 200CX analytical-electron microscope equipped with Tracer-Northern TN-2000 energy-dispersive X-ray analysis (EDS) system. Also, electron microprobe analyses (EPMA) across the interfaces were performed for selected dissimilar alloy welds in a Cameca SX-50 Microprobe system.

3.5. Evaluation of mechanical properties

Knoop microhardness testing (150 gram load) was
performed across the weld region at the axial centerline and the outer periphery for correlation with microstructural characteristics. Transverse-weld oriented tensile test specimens (gage section: 25 mm length x 4.5 mm width x 1.5 mm thickness) were sectioned from the center and outer periphery of each weld type by electrodischarge machining (EDM) and tested as per ASTM E8-87 at an extension rate of 0.25 mm/min. Also, three-point guided bend testing was performed on axial half-sections to reveal fracture characteristics.

3.6. Evaluation of fracture surfaces

The fracture surfaces of the tensile and bend test specimens were examined using ETEC Autoscan and Hitachi S-510 scanning electron microscopes to determine the preferential fracture locations and fracture modes. In addition, representative portions of the fractured specimens were axially sectioned and the fracture locations were examined on a light microscope.

3.7. Evaluation of weld elevated-temperature exposure

Inertia-friction welds in FVS0812 were sectioned axially and encapsulated in pyrex tubes at 10⁻³ torr to avoid oxidation during thermal exposure. The encapsulated specimens were isothermally heat-treated in a box furnace
for times and temperatures of 100, 200, 300 and 500 hours at 425 °C and 100 hours at 500 and 573 °C. After thermal exposure, the specimens were metallographically prepared for microstructural characterization and hardness testing. Quantitative metallography was performed to determine the size of the dispersoids and alpha grains using projected images from TEM microscopy using diameter analysis (see Appendix). To minimize truncation effects near the surface regions of the specimens, the images were taken from relatively thicker regions compared to the dispersoid diameter. The overlapping of particle images was easily distinguishable and caused no difficulties in the analysis.
CHAPTER IV

RESULTS

4.1. Introduction

This chapter presents results obtained from four different sets of experiments: 1) the similar-alloy friction welding of RS/PM Al-Fe-V-Si alloys; 2) the dissimilar-alloy friction welding of RS/PM Al-Fe-V-Si and Al-Fe-Mo-V alloys to conventional IM 2024-T351 aluminum; 3) the dissimilar-alloy friction welding of RS/PM aluminum alloys to titanium alloys; and 4) the elevated-temperature exposure behavior of friction welds in FVS0812. Information presented in each respective section includes a detailed description of the base metal microstructures, welding conditions, weldment macro- and microstructural characteristics and fracture behavior. Analysis of these weldments characteristics and correlations are made in the subsequent Discussion section.
4.2. Similar-alloy welds in Al-Fe-V-Si alloys

4.2.1. Base Metal Characterization

4.2.1.1. FVS1212

The microstructure of the rapidly-solidified ribbon was not characterized in this study. However, previous studies (Refs.4,21) have shown that the extremely rapid cooling rates ($10^5$-$10^7 \, ^\circ\text{C/s}$) experienced during planar-flow casting promote an extremely fine, micro-cellular alpha aluminum structure supersaturated in Fe, Si and V, with extremely fine dispersoids present at cell boundaries. During subsequent consolidation and thermo-mechanical processing, these intercellular dispersoids coarsen and additional dispersoids form from the supersaturated alpha matrix.

The microstructure of the FVS1212 base metal oriented parallel to the extrusion direction is shown in black and white contrast in Figs. 13a and b, and in color contrast in Figs. 13c and d. Fig. 13a shows the overall microstructure at low magnification, with boundaries between the planar-flow cast and comminuted particles clearly observable. Regions of dark contrast in Fig. 13a, which contained coarser dispersoids, conversely exhibited light contrast in Fig. 13b at increased magnification. The white alpha aluminum matrix between the dispersoids was readily observable at high magnification. Regions of light
contrast in Fig. 13a, which contained extremely fine dispersoids, actually appeared dark grey at higher magnification in Fig. 13b. Occasionally, narrow white bands devoid of dispersoids were observed.

Color contrast light micrographs of the base metal microstructure more clearly revealed microstructural details (Figs. 13c and d). Regions of fine dispersoids exhibited a yellow contrast (large arrow in Fig. 13d), while regions of coarser dispersoids exhibited a blue contrast (small arrow in Fig. 13d). Regions containing mixtures of both coarse and fine dispersoids exhibited a dark brown contrast within a single ribbon particle. The microstructural variations within a single particle (as revealed by different colors) originated from differences in the cooling rate of the original ribbon, with the side contacting the quenching wheel experiencing appreciably higher cooling rates than the opposite free surface side. Higher cooling rates correspondingly promoted a finer as-solidified microstructure and ultimately a finer as-processed dispersoid/grain structure.
Fig. 13. Black-and-white (a,b) and color (c,d) light micrographs of as-extruded FVS1212 base metal. Large and small arrows in (b) and (d) indicate regions of fine and coarse dispersoids, respectively.
TEM analysis

TEM observation more clearly revealed microstructural characteristics of the FVS1212 base metal. Consistent with light microscopy observations, low magnification TEM bright-field micrographs revealed distinct boundaries between the original ribbon particles and exhibited noticeable differences in the size and distribution of the dispersoids throughout the microstructure.

TEM examination at an increased magnification showed three different regions in regard to the dispersoid size and distribution: those containing 1) fine dispersoids; 2) a mixture of fine and coarse dispersoids and 3) clusters of fine dispersoids. Figs. 14a and b show a region containing relatively uniform, fine-sized dispersoids. The fine dispersoids exhibited a nearly spherical morphology with diameters ranging from about 40 to 180 nm with an average diameter of 116 nm. Another representative region showed a bi-modal distribution of dispersoids consisting of fine dispersoids ranging from 60 to 100 nm in diameter and coarser dispersoids ranging from 60 to 300 nm in diameter (Figs. 14c and d). The average dispersoid diameter in this region was 170 nm.
Fig. 14. TEM bright-field micrographs of FVS1212 as-extruded base metal: (a,b) uniform distribution of fine dispersoids; (c,d) mixture of fine and coarse dispersoids; (e,f) clusters of fine dispersoids along alpha particle boundaries.
Alpha aluminum grains in these regions were equiaxed and ranged in diameter from 0.5 to 1 micron. TEM analysis of the regions which exhibited coarse dispersoids and a high proportion of alpha phase at low magnification (region indicated by small arrow in Fig. 13b) showed clusters of fine dispersoids (50 to 100 nm in diameter) located at alpha grain boundaries (Figs. 14c and d). The population density of dispersoids in this region was generally lower than the regions of uniform, fine dispersoid distribution (Fig. 14a) and the alpha grain size was also coarser, up to about 1.5 micron in diameter. It is suggested that the formation of these structures may be associated with slowly cooled regions on the free surface side of the PFC ribbons. Occasionally, dispersoid-free bands shown in Fig. 13b were observed, which may have originated from non-homogenous deformation during extrusion. Previous studies of these alloys (Refs. 1 - 4) and EDS analysis indicated the various dispersoids to be $\text{Al}_1\text{3(Fe,V)}_3\text{Si}$ type. Acicular dispersoids observed in previous studies of a rapidly-solidified Al-Fe-Mo-V alloy (Ref. 10) were not observed. This result confirms the previously observed influence of vanadium in stabilizing the cubic $\text{Al}_1\text{3(Fe,V)}_3\text{Si}$ over the hexagonal $\text{Al}_8\text{Fe}_2\text{Si}$ or monoclinic $\text{Al}_2\text{Fe}$ phases (Ref. 21).
4.2.1.2. FVS0812

Figures 15a and b show black-and-white and color contrast micrographs, respectively, of the base metal microstructure of FVS0812 oriented parallel to the extrusion direction. In general, the microstructure was similar to that of FVS1212 although the area of fine dispersoid structure was higher (64 % for FVS0812 versus 57 % for FVS1212). Considering the low melting temperature and smaller solidification range due to the lower Fe content, the undercooling required to achieve cellular solidification could be more easily obtained in FVS0812 than in FVS1212 (See Fig. 2).

TEM Analysis

Figs. 16a to 16e show TEM micrographs of the FVS0812 base metal. The low-magnification TEM bright-field micrographs in Figs. 16a and b show the boundary between ribbon particles and reveal differences in the dispersoid sizes and population densities on either side of this boundary. Examination of relatively uniform dispersoid structures at increased magnification in Figs. 16b and c, and at other locations in the base metal, showed the dispersoids to exhibit a nearly spherical shape with diameters ranging from about 20 to 100 nm in small dispersoid regions with an average diameter of 65 nm.
Fig. 15. Light micrographs of as-extruded FVS0812 base metal: (a) black-and-white contrast; (b) color contrast.
Fig. 16. TEM bright-field micrographs of FVS0812 extruded base metal: (a,b) low magnification micrographs showing variations in dispersoid size, population density and distribution and boundaries between ribbon particles (arrows); (c,d) uniform dispersoid structure in (a); (e) dispersoids at alpha grain boundaries in lower portion of (b); (f) dispersoid-free band (arrow).
Large dispersoids exhibited a spherical morphology with diameters ranging from 100 to 260 nm with an average diameter of 169 nm. A bimodal distribution similar to that observed in FVS1212 was also observed in this region but was less apparent. Dispersoids were located both within the alpha-aluminum matrix and at the boundaries of nearly equiaxed alpha aluminum grains. In the unaffected base metal, the alpha grain sizes ranged from about 0.5 to 1 micron in diameter. Occasional regions exhibited what appeared at lower magnification to be coarser dispersoids and a high proportion of alpha phase, as in the lower left hand portion of Fig. 16b. Analysis of these regions at increased magnification in Fig. 16e showed the presence of coarse dispersoids and clusters of finer dispersoids along alpha grain boundaries. Generally, the regions with agglomerates of dispersoids exhibited a lower population of dispersoids as compared to the regions of uniform dispersoid distribution (e.g., Figs. 16c and d) and a somewhat coarser alpha grain size, up to about 1.5 micron in diameter. EDS analysis of numerous dispersoids consistently indicated an approximately $\text{Al}_3(\text{Fe},\text{V})_3\text{Si}$ dispersoid type with similar results with FVS1212. As shown in Figure 16f, occasional dispersoid-free bands were also observed in the base metal microstructure. These regions may have originated from the presence of
dispersoid-free ribbon particles, or possibly from non-homogenous deformation during extrusion. No evidence of acicular dispersoids ($\text{Al}_3\text{Fe}$) or other phases was detected.

4.2.1.3. Quantitative analysis of dispersoid size and distribution in Al-Fe-V-Si alloys

Fig. 17 shows the normalized distribution curve of dispersoid size for the FVS0812 and FVS1212 base metal microstructures. As described above, both alloys exhibited two different regions showing different dispersoid size distributions. In both alloys, the regions containing coarser dispersoids showed a bimodal size distribution of dispersoids. The region of agglomerates of dispersoids exhibited a distribution similar to that of the fine dispersoid region. Average diameters of dispersoids are given in Table 8.

Although the size distribution was within the range predicted by Voohees et al (Ref. 34) (compare Fig. 8 with Fig. 17), the shape of the distribution curves deviated from prediction. The deviation is assumed to have originated from the rapid solidification which promotes a solidified microstructure comprised of microcellular solidification plus discrete silicide dispersoids formed at less rapidly cooled regions (Ref. 21).
Fig. 17. Normalized size distribution curve of dispersoids ($= \rho^2 h(\rho)$) vs. d for (a) FVS0812 and (b) FVS1212 base metals.
Table 8: Size distribution of dispersoids in Al-Fe-V-Si base metal

<table>
<thead>
<tr>
<th>Region</th>
<th>FVS0812</th>
<th></th>
<th>FVS1212</th>
<th></th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>small</td>
<td>large</td>
<td>small</td>
<td>large</td>
</tr>
<tr>
<td>Range of diameter, nm (d/d)</td>
<td>20 - 100 (0.31 - 1.69)</td>
<td>100 - 260 (0.59 - 1.54)</td>
<td>40 - 180 (0.34 - 1.55)</td>
<td>60 - 300 (0.35 - 1.76)</td>
</tr>
<tr>
<td>average diameter (nm)</td>
<td>65</td>
<td>169</td>
<td>116</td>
<td>170</td>
</tr>
<tr>
<td>volume ratio of the region</td>
<td>0.7</td>
<td>0.3</td>
<td>0.6</td>
<td>0.4</td>
</tr>
<tr>
<td>overall average diameter</td>
<td>96 nm</td>
<td></td>
<td>138 nm</td>
<td></td>
</tr>
</tbody>
</table>

Even after thermomechanical processing during the consolidation process, a steady-state distribution was not achieved (compare Fig 8 with Fig 17). In general, the size of the dispersoids in FVS1212 was larger than that of FVS0812 in each respective region.

4.2.2. Weld Zone Characterization

4.2.2.1. Macroscopic characterization

FVS1212

The as-welded inertia-friction (IF) and linear-friction (LF) welds in FVS1212 are shown in Fig. 18. Visual examination of the IF welds found the presence of smooth, symmetrically uniform flash around the entire weld.
circumference, with the flash and axial displacement increasing with an increase in axial force.

In contrast to the IF welds, similar-alloy LF welds exhibited flash formation principally in the direction of linear displacement. Axial displacement data for different alloy/parameter combinations are given in Table 4.

Examination of the as-polished surfaces of each specimen revealed no evidence of defects. Macrographs of the axially-sectioned weld specimens for different alloy/parameter combinations are shown in Figs. 19. The similar-alloy IF welds revealed that the entire weld heat-and-deformation zone (HDZ) typically exhibited an hourglass shape, with the HDZ width narrower for the high axial force welds (Figs. 19 a and b). The center of the HDZ's (hereafter referred as the inner-HDZ) exhibited featureless regions of light and/or dark contrast. In the low axial force (LAF) IF weld, a light-contrast center was surrounded by a region exhibiting a dark contrast, while only a region of dark contrast was apparent in the high axial force (HAF) IF weld. The width of the inner-HDZ remained relatively constant across the low-axial force IF weld, and gradually decreased in width toward the axial centerline in the high-axial force IF weld.
Fig. 18. As-welded macrographs of friction welds in FVS1212: (a) IF welds produced at low (left) and high (right) axial force; (b) LF weld.
Fig. 19. Macrographs of friction welds in FVS1212: (a) low axial force IF weld; (b) high axial force IF weld; and (c) LF weld; Arrows indicate axial centerline.
Macrographs of the axially-sectioned LF weld shown in Fig. 19c exhibited a similar feature but the inner-HDZ was much smaller than those of the IF welds. The "curling-back" of flash in the linear-friction welds was much more pronounced than in high axial force IF welds which exhibited an even greater volume of flash formation. The difference in flash geometry may be associated with more continuous upset and directional translation movement during linear-friction welding vs. inertia-friction welding, for which major upset and flash formation occur during the final stage of the weld sequence.

Visual examination of the inertia-friction and linear-friction welds in FVS0812 appeared almost the same as welds in FVS1212, with the difference being slightly larger upsets (10.2 mm for HAF IF weld, 2.7 mm for LAF IF weld). The total upset produced in the LFW was 5.8 mm, including a preset 4 mm burnoff displacement and the additional 1.8 mm upset experienced during the forge stage. These levels of flash formation and axial displacement were intermediate between those of IFW welds produced using low and high axial force levels.
Fig. 20. Light macrographs of inertia- and linear-friction welds produced in FVS0812: (a) LAF IFW; (b) HAF IFW; (c) LFW.
Macroscopic examination of the axially-sectioned weld specimens in FVS0812 in the as-polished condition found no evidence of lack-of-bonding defects. Etching of the specimens revealed approximately hourglass-shaped heat-and-deformation zones in both the IF welds and LF welds (Figs. 20a and b) showing similar features to those of the FVS1212 counterparts.

4.2.2.2. Light microscopy characterization

**FVS1212**

For the similar-alloy IF (Figs. 21 and 22) and LF (Fig. 23) welds, the base metal microstructure showed a continuous reorientation from parallel to perpendicular to the original base metal extrusion direction. Nearer to the center of the weld HDZ, the reoriented microstructure became increasingly "flattened" due to high axial compressive stresses. Comparing Figs. 21, 22 and 23, the width of the inner-HDZ was greater for the IF welds produced at both the high and low axial force as compared to the LF weld. In the inner-HDZ of the low axial force weld, microstructural variations which existed in the base metal became featureless due to severe mechanical mixing. As shown, the inner-HDZ was comprised of a grey central region and a white outer region (Note that the contrast of these regions are actually reverse of those observed at low magnification).
Fig. 21. Light micrographs and corresponding KHN hardness traverses for an IF weld produced in FVS1212 using low axial force: (a) axial centerline; (b) outer periphery.
Fig. 22. Light micrographs and corresponding KHN hardness traverses for an IF weld produced in FVS1212 using high axial force: (a) axial centerline; (b) outer periphery
Fig. 23. Light micrographs and corresponding KHN hardness traverses for a LF weld produced in FVS1212: (a) axial centerline; (b) outer periphery.
The color light micrograph in Fig. 24 effectively revealed the microstructural changes near the weld interface region. In the IF welds, the base metal microstructure became increasingly homogeneous and yellow in color at the boundary region between the inner-HDZ and outer-HDZ, showing an absence of blue-appearing coarse dispersoid regions (Fig. 24a). The center region of the homogeneous inner-HDZ which showed grey contrast in black-and-white micrographs appeared bluish-yellow or brown implying some coarsening of the dispersoid size and grain size. Considering color light micrographs (Figs. 13c and d) and TEM observation of the base metal (Fig. 14), the yellow color at the boundary region was due to the break-up of RS/PM particulates in the region, incomplete recrystallization and a textured grain structure with a higher dislocation density. The bluish-yellow appearance at the center region was apparently due to slight coarsening of the dispersoids and alpha recrystallization in this region.

The inner-HDZ at the outer periphery region of the low axial force IF weld exhibited a grey contrast central region and neighboring light contrast region similar to the axial center. However, narrow, white-appearing bands and adjacent dark-appearing regions were observed in the inner-HDZ at the interface (Fig. 21b). Examination at an
increased magnification indicated a very low dispersoid density in the light-appearing regions and a higher dispersoid density in the adjacent dark-appearing regions versus the surrounding inner-HDZ.

Figures 22a and b show microstructure traverses across the high axial force IF weld in FVS1212. In the inner-HDZ, the region of grey contrast was barely observable. The higher compressive stress expelled the homogenized and coarsened microstructure to the outer periphery, leaving only the fine textured structure as observed in the white-contrast region in the low axial force weld. The outer periphery region of the weld exhibited an appreciably wider HDZ which appeared comparable to the center of the low axial force weld. Dispersoid-lean bands were not observed in the high axial force weld. The absence of these structures in high axial force welds was attributed to the increased expulsion of the softened metal from the weld interface with the increase in weld axial force, which was indicated by the presence of the light-contrast dispersoid-lean band at the outer surface in the flash.

Figures 23a and b show microstructure traverses across the LF weld in FVS1212. Compared with the IF welds, the inner-HDZ was macroscopically homogeneous but comprised of several narrow, grey and white-contrast layers. The transition from the inner-HDZ to the outer-HDZ was more
abrupt than in the IF welds. The outer periphery region exhibited an asymmetric inner-HDZ, with one side showing white contrast and the other side showing grey contrast.

The color light micrograph of the LF weld (Fig. 24b) exhibited more details of the inner-HDZ showing light-yellow color at the boundary region with the inner-HDZ and streaks of brown or bluish-yellow at the center of the inner-HDZ, indicating dispersoid coarsening in this region.

**FVS0812**

Figures. 25, 26 and 27 show the microstructure traverse from the unaffected base metal to the center of the weld HDZ for low axial force and high axial force IF welds and the LF weld in FVS0812. The white appearing, dispersoid-lean band at the outer periphery of the low axial force IF weld was more pronounced and wider than that in the FVS1212 counterpart. The lower volume fraction of dispersoids in FVS0812 apparently promoted an increase in the dispersoid-free alpha aluminum than in FVS1212. The narrow grey- and white-contrast layers observed in the HAF IF FVS1212 weld were less pronounced in the FVS0812. As indicated in the macrographs, the width of the inner- and outer-HDZ's in the LF weld in FVS0812 was extremely narrow at the axial centerline and the base metal texture changed its orientation abruptly to the interface direction.
Fig. 24. Color light micrographs of the interface regions of friction welds in FVS1212: (a) IFW with low axial force; (b) LFW.
Fig. 25. Light micrographs and corresponding KHN hardness traverses for an inertia-friction weld produced in FVS0812 using low axial force: (a) axial centerline; (b) outer periphery.
Fig. 26. Light micrographs and corresponding KHN traverses for an inertia-friction weld produced in FVS0812 using high axial force: (a) axial centerline; (b) outer periphery.
Fig. 27. Light micrographs and corresponding KHN traverses of a linear-friction weld produced in FVS0812: (a) axial centerline; (b) outer periphery.
The light-microscopy features appeared almost the same as those of FVS1212. Differences were that the dispersoid-lean band at the outer periphery of low axial force weld and the homogenization inside the inner HDZ were more prominent than in the corresponding FVS1212 welds.

4.2.2.3. TEM Analysis

TEM bright-field observation of the inner-HDZ at the axial centerline in the similar-alloy IF weld produced with low axial force (Figs. 28a and b) showed a uniform distribution of dispersoids ranging in size from about 50 to 500 nm. This homogenization of the base metal microstructure appeared to originate from extensive mechanical mixing. The clusters of fine dispersoids shown in Figs. 14 e and f were not observed in this region, indicating their complete break-up and dispersion. The dispersoid sizes and size distribution in this region were generally comparable to those observed in the unaffected base metal, showing only a marginal increase in the range of dispersoid sizes. While the previous TEM analysis of inertia-friction welds in Al-Fe-Mo-V alloys (Ref. 11) showed the cleavage-type fracture of relatively coarse dispersoids (up to two microns in diameter) in the inner-HDZ, no evidence of dispersoid fracture was observed in
Fig. 28. TEM bright-field micrographs of an IF weld produced in FVS1212 using low axial force: (a,b) center of HDZ at axial centerline; (c - e) center of HDZ at outer periphery.
this study. This absence of the dispersoid fracture was attributed to the appreciably smaller original dispersoid size and also possibly to the higher fracture strength of the dispersoids.

The inner-HDZ of the low axial force weld adjacent to the homogenized center region exhibited alternating regions of light- and dark-appearing contrast at lower magnification. These observations were attributed to variations in the dispersoid population density and distribution; dark contrast for regions of a higher dispersoid population density and light contrast for regions of a lower population density. Regions containing clusters of fine dispersoids were also observed. These differences in dispersoid size and distribution were associated with variations in the original base metal microstructure which had been "flattened" but not completely homogenized in this region. The alpha aluminum grains in inner-HDZ regions showed a similar morphology and size to those observed in the base metal, being nearly equiaxed and ranging in diameter from about 0.25 to 1.0 microns. Considering the equiaxed nature and the size of the grains even with directional deformation in these regions, it is suggested that the grains were dynamically-recrystallized on weld cooling and that the growth was limited by the high volume fraction of dispersoids.
TEM bright-field micrographs in Figs. 28c-e show the region containing dispersoid-lean bands at the outer periphery of the IF welds produced using low axial force (white-appearing bands in Fig. 21b). In addition to a low dispersoid population density, this region contained coarser alpha grains up to several microns in diameter with a slightly elongated morphology. The low dispersoid density in this region apparently did not restrict the alpha grain growth. The occurrence of this microstructure may be explained by localized melting of the alpha phase at the weld interface, which is likely to occur at the outer periphery due to the higher rotational velocities and peak temperatures. However, analysis of the regions indicated no evidence of a resolidified microstructure. Alternatively, the microstructure could result from localized, nonuniform deformation in the HDZ. High tensile compressive stresses in combination with simultaneous shear stresses in this region may locally extrude soft alpha aluminum from the original base metal microstructure, resulting in dispersoid-lean regions. The presence of adjacent dark-contrast regions, which contained a high population density of dispersoids (Fig. 21b), may be explained by such nonuniform deformation. The coarsening of dispersoids in the dark-contrast regions was noticeable, with the diameter of dispersoids ranging from 50 to 500 nm.
TEM bright-field micrographs in Figs. 29a and b show the inner-HDZ at the axial centerline of the IF weld produced using high axial force. Considerable homogenization was indicated by the uniform distribution of a wide range of dispersoid sizes. The adjacent inner-HDZ exhibited regions of alternating dark and light-appearing contrast as in the weld produced using low axial force. Consistent with light microscopy observations, TEM observations of the weld zone at the outer periphery of the high axial force weld compared closely with the center of the weld produced at the low axial force (Figs. 29c and d).

Compared to the similar-alloy IF welds, TEM bright-field analysis of the homogeneous interface region at the center of the LF weld in FVS1212 (Figs. 30a and b) indicated a marked increase in dispersoid size, showing dispersoids ranging from 120 to 1,600 nm in major axis. While the finer dispersoids remained spherical, the coarser dispersoids experienced coarsening and elongation along the interface direction with an aspect ratio of approximately two. Despite this noticeable dispersoid coarsening, the alpha grain size in this region appeared equivalent to the base metal. Clusters of dispersoids observed in the base metal were not observed in this region, indicating their total break-up.
Fig. 29. TEM bright-field micrographs of an IF weld produced in FVS1212 using high axial force: (a,b) center of HDZ at axial centerline; (c,d) center of HDZ at outer periphery.
Fig. 30. TEM bright-field micrographs of a LF weld produced in FVS1212: (a,b) center of HDZ at axial centerline; (c,d) adjacent layered region; (e,f) center of HDZ at outer periphery.
Alternating layers observed in light microscopy (Fig. 23b) in the homogeneous inner-HDZ were found to result from a slight variation of dispersoid distribution. An inhomogeneous region outside of the central homogeneous region, where light-yellow contrast was observed in Fig. 24b, showed alternating deformed layers of fine dispersoids and deformed coarse dispersoids (Fig. 30c). Even though it was located at the boundary region, coarsening and deformation were noticeable, showing dispersoids ranging from 120 to 1350 nm in diameter in the major axis. The interface region at the outer periphery showed a very similar microstructure but with slightly less-deformed, lower aspect ratio dispersoids (Figs. 30d). The asymmetric nature of the interface in the region observed with light microscopy was not discernable via TEM observation. It was concluded that the contrast observed via light microscopy in this region was due to the difference in grain orientation rather than coarsening of dispersoids. The marked coarsening of dispersoids in the interface region originated from the slower thermal cycle experienced during the process and to a lesser extent of metal expulsion compared to the high-axial force IF weld. A higher axial force may be required to reduce the weld time and remove the dispersoid-coarsened region while achieving the same upset in the LF welds.
TEM analysis of the HDZ center in the low axial force IF weld in FVS0812 revealed a uniform distribution of dispersoids ranging in diameter from about 50 to 300 nm (Fig. 31). Consistent with light microscopy results, this homogenization appeared to result from the extensive mechanical mixing of base metal regions which originally contained varying distributions of relatively finer or coarser dispersoids (as shown in Figs. 31a and b). Despite the elevated temperatures experienced in this region during the weld thermal cycle (albeit for very short times), negligible evidence of dispersoid coarsening was observed.

Figures 31c and d show the near-HDZ region at the axial center of the weld produced using low axial force. Alternating regions exhibiting light and dark-appearing contrast in the lower magnification TEM micrograph (Fig. 31c) originated from variations in the dispersoid population density. Regions of dark contrast exhibited a higher dispersoid population density (right-hand side of Fig. 31d), while regions of light contrast (left-hand side of Fig. 31d) exhibited a lower dispersoid population density. These differences in contrast and dispersoid structure were likely associated with variations in the original base metal microstructure which in this region of the HDZ had not been completely homogenized.
Fig. 31. TEM bright-field micrographs at axial centerline of IF weld produced in FVS0812 using low axial force: (a,b) center of HDZ; (c,d) outer HDZ.
The morphology and size of the alpha aluminum grains in the central and outer-HDZ regions were similar to those exhibited by the base metal, being nearly equiaxed and ranging in diameter from about 0.25 to 1.0 microns, which is comparable to those of the base metal.

TEM bright-field micrographs in Figures 32a-c show the dispersoid-lean regions at the outer periphery of the low axial force IF weld, which showed a light-contrast band in light microscopy. In addition to a low dispersoid population density, these regions showed a slightly elongated alpha aluminum grain morphology and a coarser alpha grain size of up to several microns in diameter. Alternatively, these regions more likely resulted from localized, nonuniform deformation in the HDZ similar to the low axial force IF weld in FVS1212.

TEM bright-field micrographs in Figures 33a-c show the center of the HDZ at the axial centerline of the IF weld produced using high axial force. Although light microscopy analysis indicated incomplete homogenization of the base metal texture in this region considerable homogenization was, in fact, indicated by the presence of a wide range of dispersoid sizes within a highly localized region, as shown in Fig. 33c. The adjacent near-HDZ showed regions of alternating dark and light-appearing contrast as in the weld produced using low axial force.
Fig. 32. TEM bright-field micrographs at outer periphery of inertia-friction weld produced in FVS0812 using low axial force: (a-c) weld interface; (d,e) outer HDZ.
Fig. 33. TEM bright-field micrographs with different magnifications at center of HDZ along axial centerline of inertia-friction weld produced in FVS0812 using high axial force.
Consistent with light microscopy observations, TEM analysis observations of the weld zone at the outer periphery of the high axial force weld compared closely with the center of the weld produced at low axial force.

TEM bright-field analysis of the inner-HDZ at the axial centerline of the LF weld was consistent with light microscopy analysis, indicating homogenization of the base metal dispersoid microstructure due to intense mechanical mixing. As shown in Figs. 34a and b, in addition to homogenization of the base metal microstructure, TEM analysis revealed both dispersoid coarsening and morphological elongation, with a major axis length of up to 500 to 750 nm parallel to the interface direction. The alpha grain size in this region was comparable to that in the unaffected base metal (0.5 to 1.0 μm in diameter). For comparison, the dispersoid size in the LF weld was much larger than in the HAF IF weld, indicating a slower thermal cycle than in the IF weld. However, the dispersoid size in the LFW was smaller than in the outer periphery region of the LAF IF weld, which can be related to the higher peak temperature in the LAF IF weld.

At the outer edge of the inner-HDZ (regions of white contrast in light microscopy), dispersoid homogenization was also observed, but coarsening and homogenization were appreciably less than in the center of the inner-HDZ.
Fig. 34. TEM bright-field micrographs at axial centerline of LF weld produced in FVS0812: (a) center of inner-HDZ; (b) outer edge of inner HDZ.
Although the regions of dispersoid coarsening were wider, TEM analysis of the inner-HDZ at the weld outer periphery of the LFW appeared comparable to the inner HDZ at the axial centerline.

Summarizing, TEM observations of the FVS0812 friction welds (IFW’s and LFW) appeared similar to those of FVS1212. The most prominent differences were the finer dispersoid size and lesser coarsening of the dispersoids in the FVS0812 welds compared to the FVS1212 welds.

4.2.2.4. Quantitative analysis of dispersoid size and distribution of the welds in Al-Fe-V-Si alloys

Table 9 shows dispersoid size distributions for the FVS0812 and FVS1212 welds at different weld locations and Figure 35 shows normalized dispersoid size distribution curves for the weld interface regions at the axial centerlines for the IFW and LFW welds in FVS0812 and FVS1212.

As shown in the Table 9, the interface at the axial centerline of the high axial force IFW’s in both alloys showed negligible coarsening compared to the average diameter of the dispersoids in the base metals. During welding, only mixing and homogenization of the inhomogeneous base metal microstructure were observed in this region, while noticeable coarsening was observed at
the interface at the outer periphery. The alpha grain size in the interface region was slightly less than that of the base metal, measuring 0.45 μm for FVS0812 and 0.5 μm for FVS1212. Although a small proportion of the dispersoids were located at alpha grain boundaries, generally, the dispersoids were evenly distributed within the matrix. Quantitative measurement of dislocation density in this region could not be obtained but general observation showed it to be essentially equivalent to that of the base metal.

Table 9: Size distribution and average diameter of dispersoids at various locations

<table>
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<tr>
<th>Location</th>
<th>Diameter Range of Dispersoids (nm)</th>
<th>Average Diameter of Dispersoids (nm)</th>
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<tbody>
<tr>
<td>FVS0812 BM</td>
<td>20 - 260</td>
<td>96</td>
</tr>
<tr>
<td>FVS0812/IFWL/C</td>
<td>40 - 280</td>
<td>131</td>
</tr>
<tr>
<td>FVS0812/IFWL/OP</td>
<td>30 - 320</td>
<td>187</td>
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<tr>
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<td>30 - 260</td>
<td>97</td>
</tr>
<tr>
<td>FVS0812/IFWH/OP</td>
<td>40 - 280</td>
<td>135</td>
</tr>
<tr>
<td>FVS0812/LFW/C</td>
<td>30 - 440</td>
<td>141</td>
</tr>
<tr>
<td>FVS1212 BM</td>
<td>40 - 300</td>
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<tr>
<td>FVS1212/IFWL/C</td>
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</tr>
<tr>
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<td>282</td>
</tr>
<tr>
<td>FVS1212/IFWH/C</td>
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<td>191</td>
</tr>
<tr>
<td>FVS1212/LFW/C</td>
<td>30 - 530</td>
<td>256</td>
</tr>
</tbody>
</table>
Fig. 35. Normalized dispersoid size distribution curve for weld interface region at axial centerline of friction welds in FVS1212 and FVS0812.
for both alloys. The outer HDZ’s of both the IF and LF welds in FVS1212 and FVS0812 exhibited negligible dispersoid coarsening. Summarizing these observations, the interface at the axial centerline of the HAF IFW experienced negligible coarsening as it was "freshly exposed" to the interface region during the welding process. In contrast, more uniform heating in LFW resulted in a more uniform size distribution and coarsening across the entire interface except at the outer periphery. Greater coarsening at the interface of the LF welds was attributed to the longer welding time. Coarsening of dispersoids in the FVS1212 welds was greater than in the FVS0812 welds for both the IF and LF welds.

4.2.2.5. Hardness Testing

FVS1212

Knoop microhardness traverses for the similar-alloy FVS1212 welds are shown in Figs. 21, 22 and 23. Hardness values across the weld interface at the axial centerline for the low axial force IF weld (Fig. 21c) were equivalent to that of the base metal (KHN 183.5 in average, ranging from 170 to 210) outside of the inner HDZ, but decreased down to 150 KHN at the center of the inner HDZ. This decrease in hardness was attributed to the marginal increase in dispersoid coarseness and possibly the alpha
grain size. The hardness traverse at the outer periphery of this weld exhibited a similar trend, except at the dispersoid-lean, white-appearing regions which exhibited a very low hardness down to about 104 KHN. Hardness measurement at the grey region neighboring the dispersoid-lean region exhibited KHN 180, which was higher than the values at other regions in the inner HDZ. Even with the coarsening of the dispersoids in this region, an increase in local dispersoid population density apparently compensated for the coarsening effect.

Fig. 22c shows the KHN hardness traverse across the interface of the similar-alloy, high-axial force IF weld. Hardness values were more consistent with that of base metal, without any loss and an increase to a maximum hardness of 210 KHN in the inner-HDZ. This increased hardness was attributed to the essential absence of dispersoid or grain coarsening in this region, and the breakup and redistribution of dispersoid clusters. The increase in hardness at the boundary of the inner- and outer-HDZ was attributed to a change in dispersoid distribution and grain size, and possibly to residual cold work in the structure. A hardness traverse at the outer periphery of the high axial force weld exhibited a consistent hardness of about 175 KHN, which represented the lower end of the base metal hardness range. This slight
loss of hardness was attributed to the homogenized but slightly coarsened dispersoid structure in this region.

Knoop microhardness traverses across the interface of the similar alloy FVS1212 LF weld are shown in Fig. 23c. A noticeable hardness drop was observed within 200 μm of the interface at the axial center and within 500 μm of the outer periphery. The hardness drop at the interface region is explained by the presence of coarsened dispersoids in the region, as observed in Fig. 30b.

FVS0812

Knoop microhardness traverses across the low and high axial force inertia-friction welds corresponded well with the aforementioned microstructural characteristics. Local variations in the base metal dispersoid structures resulted in a range of hardness from about 150 to 165 KHN. As shown in Fig. 25c, hardness values across the weld zone at the axial centerline for the weld produced using low axial force were within this range of values measured for the unaffected base metal. These observations were consistent with the negligible evidence of dispersoid and alpha grain coarsening in these HDZ regions. As expected, the white-appearing, low-dispersoid density regions located at the outer periphery of the welds produced using a low axial force exhibited a hardness (107 KHN) appreciably below that
of the base metal and surrounding HDZ. Interestingly, the hardness value in the dispersoid-lean band was almost same as that of FVS1212, indicating those hardness values are closely related with dispersoid-free alpha aluminum matrix. As shown in Fig. 26c, the KHN hardness increased substantially at the center of the weld produced using a high axial force. At the outer periphery of this weld, the hardness was comparable to that of the base metal. As in the low-axial force weld, however, hardness measurements in this microstructurally homogenized HDZ region were more uniform than in the base metal, where variations were promoted by local differences in the dispersoid size and population density.

Microhardness traverses across the axial center line and outer periphery of the LF weld corresponded very closely with the microstructural analysis, indicating relatively little change in the outer-HDZ from the average base metal hardness of approximately KHN 150 - 155 as shown in Fig. 27c. At the boundary of the inner-HDZ, the hardness increased slightly up to KHN 155 - 160, which may be attributed to severe deformation in this region with negligible dispersoid coarsening. Within an extremely narrow region directly at the center of the inner-HDZ, where dispersoid coarsening was observed with TEM, the hardness decreased to KHN 135 - 140.
4.2.2.6. Tensile Testing

Tensile testing results for the FVS1212 and FVS0812 similar-alloy welds are shown in Table 10.

The similar alloy low axial force IF weld in FVS1212 exhibited a tensile strength of 463 MPa and a joint efficiency of 66%. The lower joint efficiency of the low axial force IF weld in FVS1212 resulted from the presence of dispersoid-lean bands in the interface region at the outer periphery. Consistent with the hardness test results, the similar alloy higher axial force IF weld exhibited a higher tensile strength of 489 MPa and a joint efficiency of 85%. Tensile fracture of both weld types occurred within the inner HDZ. The similar alloy LF weld in FVS1212 exhibited a joint efficiency of 78%, which was attributed to the greater coarsening of the dispersoids at the interface. Confirming the above, tensile fracture of the LF weld also occurred at the interface.

Relatively higher joint efficiencies were observed on all the FVS0812 welds compared to the FVS1212 welds. The results exhibited similar trends as the FVS1212 counterparts, showing the highest for the HAF IF weld but the differences were less apparent, which can be closely related to the small variations of hardness values across the weld regions of FVS0812. The higher joint efficiencies in the FVS0812 can be attributed to the lesser coarsening
of dispersoids in the weld region and inherently lower strength of FVS0812 base metal.

As shown in Table 10, tensile ductilities were relatively low as compared to the base metal, and were attributed to strain localization in the weld HDZ and reduced ductility of the "reorientated" HDZ microstructure.

Table 10: Room-temperature tensile properties of FVS1212 and FVS0812 friction welds

<table>
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<tr>
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<td>376</td>
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<td>IHDZ</td>
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<td>IHDZ</td>
</tr>
<tr>
<td>1212 BM</td>
<td></td>
<td>522</td>
<td>572</td>
<td>23.0</td>
<td></td>
<td></td>
</tr>
<tr>
<td>0812 BM</td>
<td></td>
<td>434</td>
<td>469</td>
<td>16.5</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

* Minimum two tests were performed on each test locations.
4.2.2.7. Fracture Analysis

FVS1212

Analysis of the fractured tensile specimens all showed failure to occur exclusively across the inner-HDZ. Also, guided bend test specimens showed fracture to occur at essentially the same locations as in the tensile tests.

SEM observation of the bend specimen fracture surfaces in FVS1212 are shown in Fig. 36. Fracture surfaces of the similar-alloy low axial force welds exhibited basically two different regions depending on the fracture locations; a smooth surface with fine shear tracks (Figs. 36 a and b) and a rough surface with coarse shear tracks (Figs. 36c and d). Comparing with the light microscopy results, the smooth surface region corresponded to the homogenized inner-HDZ region, where the mechanical texture and powder particle boundaries had totally disappeared and homogenized. Fine shear tracks are remnant of plastic deformation of powder particles and alignment of dispersoids in the rotation direction. The rough surface with coarse shear tracks corresponds to the region which experienced subsurface shear deformation but did not experience complete homogenization and break-up of particulate boundaries.
Fig. 36. SEM fractographs of bend test specimens of similar-alloy welds in FVS1212: (a,b) IFW produced with low axial force; (c,d) IFW produced with high axial force; (e,f) LFW.
Occasional steps in the tracks correspond to the severely elongated particle boundaries. Observation of the fracture surface at higher magnification showed typical dimple patterns with microvoid formation at the dispersoid/matrix interface and final rupture in the alpha aluminum matrix. Similar-alloy high axial force IF welds exhibited similar features as those of low axial force weld with a greater proportion of coarse shear tracks, as shown in Figs. 36c and d, which is due to the severe deformation and narrow recrystallized region as observed with the light microscopy.

The fracture surface of the LF weld exhibited a slightly different flow pattern at a low magnification (Fig. 36e). While the fracture surface of IF weld exhibited a spiral pattern, with narrow tracks, the fracture surface of the LF welds exhibited major flow principally in the linear displacement direction, with wide tracks. The wide track was attributed to the lower shear deformation in the LFW process, with particulates being deformed principally by the normal compressive stress. Observation at a higher magnification exhibited similar features as those of IF welds (Fig. 36f).

SEM observation of the fracture surfaces of tensile test specimens exhibited similar features as those of bend test specimens, as shown in Fig. 37.
Fig. 37. SEM fractographs of tensile test specimen of similar-alloy welds in FVS1212: (a,b) IFW produced with low axial force; (c,d) LFW.
Compared with IF welds (Figs. 37a and b), LF welds occasionally exhibited regions free of shear tracks as shown in Figs. 37c and d, as well as regions containing shear tracks.

**FVS0812**

In general, fractographic features of FVS0812 friction welds were comparable to those of their FVS1212 counterparts.

Fracture surfaces of the bend test specimens of the FVS0812 welds appeared identical to the counterparts of the tensile test specimens. One particular difference was observed at the outer periphery of the LAF IF weld bend test specimen shown in Figs. 38a and b. The region corresponds to the location of dispersoid-lean bands observed with light microscopy. As shown in Fig. 38a, the region exhibits two different features; a fibrous region and a flat region with a coarse dimple pattern. At increased magnification, the fibrous region appeared to correspond to the previously observed homogenized inner-HDZ (Fig. 38b). The coarse dimple pattern observed in dispersoid-lean band may be related to the ductile aluminum matrix free of dispersoids, where initial void formation sites were scarce and the voids grew until final plastic rupture.
Fig. 38. SEM fractographs at the outer periphery of the bend test specimen of the low axial force IF weld in FVS0812 at different magnifications.
The absence of this coarse dimple pattern in the LAF FVS1212 IF weld can be attributed to the relatively narrow dispersoid-lean band in the weld which was restrained by the stronger dispersoid-containing region.

4.3. Dissimilar-alloy welds between RS/PM Al-Fe-X alloys and IM 2024-T351

4.3.1. Base metal characterization

4.3.1.1. FVS1212

The FVS1212 alloy used in this experiment originated from the same heat used in the similar-alloy welding experiments and, therefore, microstructural characteristics were identical.

4.3.1.2. Al-9Fe-3Mo-1V

Light microscopy examination of the Al-9Fe-3Mo-1V base metal showed both dark- and light-etching regions containing fine-sized dispersoid particles (Fig. 39a). Examination at a higher magnification revealed (Fig. 39b) that the observed difference in contrast was due to variations in the size and distribution of dispersoid particles, with the dark-etching regions typically containing a higher population of fine-sized dispersoids and the light-etching regions containing fewer, coarse-sized dispersoids. In addition, particle boundaries between individual powder particles were clearly delineated.
Fig. 39. Light micrographs of RS/PM Al-9Fe-3Mo-1V base metal. Arrows in (b) indicate boundary between powder particles.
(arrow in Fig. 39b), suggesting possible inhomogeneous particulate deformation during extrusion.

TEM examination revealed three distinctive morphologies associated with the base metal dispersoid particles (Figs. 40a and b), viz., coarse spherical dispersoids ranging from 500 to 1,000 nm in diameter, fine acicular dispersoids ranging from 200 to 1000 nm in length and small spherical dispersoids ranging from 100 to 250 nm in diameter uniformly distributed throughout the alpha aluminum matrix. Previous electron microscopy and X-ray diffraction analysis of the Al-Fe-Mo-V alloy (Ref. 10) have indicated that the coarse spherical dispersoids were enriched in Al, Mo and V and were likely $\text{Al}_{12}\text{Fe}_3(\text{Mo,V})$. The small spherical dispersoids which were enriched in both Mo and V were likely Al$_6$Fe type, while the acicular particles were Al$_3$Fe type. EDS analysis performed in this study further corroborated chemistries for the corresponding dispersoids. TEM examination at higher magnification of select regions containing fine-sized particles exhibiting a darker contrast in Fig. 40a showed a near absence of acicular particles (Figs. 40c and d). These regions likely originated from fine-sized powder particles which underwent rapid solidification to an extremely fine-sized dendritic alpha aluminum and transformed during subsequent extrusion.
Fig. 40. TEM bright-field micrographs of Al-9Fe-3Mo-1V base metal: (a,b) predominant microstructure; (c,d) regions containing fine dispersoids.
Figures 41a and b show light micrographs of the 2024-T351 base metal oriented parallel to the extrusion direction. The microstructure exhibited both light and dark contrast, depending on the orientation of alpha grains or subgrains (Fig. 41a). Observation at a higher magnification showed agglomerations of various dispersoids and precipitates at grain boundaries and fine particles within the grains (Fig. 41b). TEM bright-field imaging revealed columnar intermetallic particles and fine intragranular second phase particles (Figs. 41c and d). Based on the morphology of these particles (Ref. 83) and EDS analysis, the small, lath-type precipitates were identified as precursor $S'(\text{Al}_2\text{CuMg})$ particles and the large, columnar particles were $\text{Al}_{29}\text{Cu}_2\text{Mn}_3$ dispersoids. Despite extreme care, the large-sized grain boundary dispersoid particles underwent preferential attack during twin-jet electropolishing and could not be effectively characterized.
Fig. 41. Light micrographs (a,b) and TEM bright-field micrographs (c,d) of 2024-T351 base metal.
4.3.2. Weld zone characterization

4.3.2.1. Macroscopic characterization

**FVS1212/2024-T351 combination**

The as-welded inertia-friction (IF) and linear-friction (LF) welds between FVS1212 and 2024-T351 are shown in Fig. 42. Consistent with the appreciably lower high-temperature strength of 2024-T351 relative to FVS1212, the dissimilar-alloy IF welds exhibited preferential flash formation in 2024-T351 versus the FVS1212. Dissimilar-alloy LF welds exhibited flash only in 2024-T351, with flash formation similar to that observed in the IF welds. The weld flash was also circumferentially more uniform than the similar-alloy LF welds. Axial displacement data for different alloy/parameter combinations are given in Table 4.

Examination of the as-polished surfaces of each specimen revealed no evidence of defects except at the very outer periphery of the interface region of the dissimilar-alloy LF weld. Macrographs of the axially-sectioned weld specimens for different alloy/parameter combinations are shown in Fig. 43. The dissimilar-alloy IF welds between FVS1212 and 2024-T351 exhibited an hour-glass shaped HDZ only in the 2024-T351 (Figs. 43a and b). The effect of the axial force was similar to the similar alloy IF welds.
Fig. 42. As-welded macrographs of (a) dissimilar-alloy IF welds produced with low (left) and high (right) axial force and (c) LF weld between FVS1212 and 2024-T351.
Fig. 43. Macrographs of dissimilar-alloy (a) IF welds produced with low (left) and high (right) axial force and LF welds between FVS1212 and 2024-T351.
Macrostructural feature of the dissimilar-alloy LF welds showed similar features to those of the dissimilar-alloy IF welds (Fig. 43c). The width of the outer-HDZ in the dissimilar alloy LF weld was slightly wider than that of the dissimilar-alloy IF welds.

**Al-Fe-Mo-V/2024-T351 combination**

Visual examination showed minimal extrusion of flash around the weld circumference in the Al-9Fe-3Mo-1V, with increased weld axial force promoting an increase in diameter in the Al-Fe-Mo-V near the weld interface. In contrast, appreciable plastic deformation in the form of extruded flash was observed in the 2024-T351, which is consistent with its relatively lower strength at elevated temperatures.

Macroscopic examination of the etched surfaces of the weld axial sections revealed that increased weld axial force promoted a progressive transition in the shape of the weld interface (Figs. 44a-c). While the low axial force weld exhibited a nearly straight interface, the high axial force weld exhibited a curved interface concave to Al-9Fe-3Mo-1V. Heavy etching of the welds revealed that the weld interface was bounded by a HDZ comprised of a featureless grey-etching region directly adjacent to the weld interface (inner-HDZ) and deformed regions farther from the weld interface (outer-HDZ).
Fig. 44. Light macrographs of IF welds between Al-9Fe-3Mo-1V (left side) and 2024-T351 (right side) produced at different axial force levels: (a) low axial force; (b) medium axial force; (c) high axial force. Arrows indicate axial centerline.
The inner-HDZ on either side of the weld interface exhibited a parallel shape at low axial force and transitioned to an hour-glass shape at higher axial force. The outer-HDZ in Al-9Fe-3Mo-1V showed only minimal evidence of macroscopic deformation in relation to the inner-HDZ in the 2024-T351.

A major difference in macroscopic features between the dissimilar-alloy welds of FVS1212/2024-T351 and Al-Fe-Mo-V/2024-T351 was the shape of the interface. The dissimilar-alloy welds of Al-Fe-Mo-V/2024-T351 showed a curved interface into the Al-Fe-Mo-V, while the dissimilar-alloy welds of FVS1212/2024-T351 showed a straight interface.

4.3.2.2. Light microscopy characterization

**FVS1212/2024-T351 combination**

Figures 45 - 47 show microstructures and corresponding hardness traverses from the unaffected base metal to the center of the weld HDZ at the axial center and outer periphery for each process/parameter combination.

For the dissimilar-alloy IF welds (Figs. 45, 46) and LF weld (Fig. 47), microstructural changes and reorientation of the base metal structure were observed only in the 2024-T351 due to the lower strength of the 2024-T351 relative to FVS1212 at elevated temperatures.
Fig. 45. Light micrographs and corresponding KHN traverses for IF weld produced between FVS1212 and 2024-T351 using low axial force: (a) axial center line; (b) outer periphery.
Fig. 46. Light micrographs and corresponding KHN traverses for IF weld produced between FVS1212 and 2024-T351 using high axial force: (a) axial centerline; (b) outer periphery.
Fig. 47. Light micrographs and corresponding KHN traverses for LF weld produced between FVS1212 and 2024-T351 weld: (a) axial centerline; (b) outer periphery.
At the outer HDZ's at the axial centerline, elongated alpha grains in the base metal gradually changed to near-equiaxed grains, and then to a severely laminated structure and finally to a featureless (dark contrast) microstructure within the inner-HDZ. At the outer periphery of each weld, elongated alpha grains gradually reoriented and flattened, with a similar dark contrast structure at the inner-HDZ. The width of the inner-HDZ was wider in the IF welds than in the LF weld, which was attributed to the larger subsurface shear deformation caused by the rotational motion. Higher axial force resulted in narrower inner-HDZ's in the IF welds. A small lack of bonding was observed at the outer periphery contacting the free surface in the LF weld. Mechanical mixing across the interface was not resolvable using optical microscopy.

Observation of the interface regions at increased magnification (Fig. 48) revealed more detailed microstructural features. While the inner-HDZ of the IF weld did not show any discrete features, the inner-HDZ of the LF weld clearly revealed recrystallized grains in this region. The larger recrystallized grain size in the LF weld was attributed to the slower thermal cycle compared with the IF welds. Slight deformation was observed in FVS-1212 contacting 2024-T351, but microstructural changes in the FVS1212 were not observed in either the IF or LF welds.
Fig. 48. Light micrographs of the interface regions in dissimilar-alloy type friction welds produced between FVS1212 and 2024-T351: (a) IF weld and (b) LF weld at axial centerline.
Al-Fe-Mo-V/2024-T351 combination

Light microscopy examination at higher magnification more clearly delineated the different regions within the HDZ's of these welds. The Al-9Fe-3Mo-1V showed a gradual transition from an as-extruded structure to a "flattened" structure parallel to the weld interface, and a progressively higher degree of deformation nearer to the weld interface (Figs. 49a and b). Comparatively, the degree of deformation in the Al-9Fe-3Mo-1V was higher at the center than at the outer periphery, although a similar trend was not readily discernible in the 2024-T351. The grey-etching inner-HDZ showed a homogenized, structurally-refined region on either side of the interface. The width of this refined region was wider in the 2024-T351 and narrower at the center in the Al-9Fe-3Mo-1V. Farther from the interface, the 2024-T351 exhibited a continuous re-orientation of the base metal grains from a direction along the specimen length to a direction nearly parallel to the weld interface in a similar manner as in the FVS1212/2024-T351 welds.
Fig. 49. Light micrographs and corresponding KHN hardness traverses for an IF weld produced between Al-Fe-Mo-V and 2024-T351 using low axial force: (a) axial centerline; (b) outer periphery.
Microstructure traverses across the axial section of the high axial force weld both along the weld centerline and outer periphery (Figs. 50a and b) revealed features similar to the low axial force weld. However, the widths of the grey-etching refined inner-HDZ and the adjacent outer-HDZ were narrower compared to the corresponding regions in the low axial force weld.

Observations at higher magnification of both high and low axial force welds (Fig. 51) revealed streaks of intermetallic particles along the deformation direction in 2024-T351 as observed in the dissimilar-alloy welds between FVS1212 and 2024-T351 except at the center of the high axial force weld. The central region of the interface of the low axial force weld also showed evidence of "mechanical mixing" of 2024-T351 into the Al-9Fe-3Mo-lV. In contrast, the central region of the interface of the high axial force weld exhibited a smooth transition and did not show any distinct evidence of mechanical mixing. At a higher magnification, the narrow, grey-etching band observed on the Al-9Fe-3Mo-1V side near the interface of the high axial force weld (Fig. 50a versus 51c) indicated that the etching response originated from severely deformed, fine-sized microstructures at these locations.
Fig. 50. Light micrographs and corresponding KHN hardness traverses for an IF weld produced between Al-Fe-Mo-V and 2024-T351 using high axial force: (a) axial center line; (b) outer periphery.
Fig. 51. Light micrographs of the interface regions of IF welds between Al-Fe-Mo-V and 2024-T351: (a) center and (b) outer periphery of LAF weld; (c) center and (d) outer periphery of HAF weld.
Light microscopy examination of the as-polished surface of the axial sections of these welds revealed that the linear discontinuities observed in 2024-T351 were pre-existing microcracks (Fig. 52a). Examination after etching with Keller’s reagent revealed that these microcracks were associated with grain boundary dispersoid (intermetallic) particles (Fig. 52b). These microcracks likely originated from the limited ductility of the intermetallic particles vis-a-vis the extent of severe plastic deformation observed at these locations. Comparing with the HAF weld between FVS1212 and 2024-T351, the occurrence of microcracks at the outer periphery of the HAF weld between Al-Fe-Mo-V and 2024-T351 were attributed to the inhomogeneous normal compressive stress distribution. Due to the strength reversal coupled with higher temperature at the outer periphery, the normal compressive stress was lower at the outer periphery and severe shear deformation at the outer periphery resulted in cracking.

Comparing the microstructures of the two types of dissimilar-alloy welds (FVS1212/2024-T351 and Al-Fe-Mo-V/2024-T351), the FVS1212/2024-T351 welds did not show any noticeable macroscopic deformation in the FVS1212 due to the higher strength of the RS alloy over the entire temperature range on heating.
Fig. 52. Light micrographs of weld defects in 2024-T351 at the outer periphery of high axial force IF weld: (a) as-polished; (b) etched outer periphery of HAF weld.
In contrast, the Al-Fe-Mo-V showed noticeable macroscopic deformation across the interface due to its lower strength at lower temperatures. Also, mechanical mixing between FVS1212 and 2024-T351 was minimal, showing an abrupt transition across the interfaces in contrast to the Al-Fe-Mo-V/2024-T351 welds, which showed greater mechanical mixing. Structural changes in 2024-T351 were essentially equivalent for the both alloy compositions.

4.3.2.3. TEM analysis

**FVS1212/2024-T351 combination**

TEM bright-field micrographs of the dissimilar-alloy IF weld produced using high axial force are shown in Fig. 53. Figure 53a shows the general features of the interface region at the axial centerline. Mechanical mixing between the two alloys is evident as penetrated layers of FVS1212 into the 2024-T351. Smearing of the microstructure between the two alloys in the layers was minimal. Even with the deformation and mechanical mixing of FVS1212 near the interface, deformation or growth of the dispersoids was not noticeable (Fig. 53b). Recrystallized 2024 grains were nearly equiaxed and the alignment of intermetallic particles was not severe (Fig. 53c). The presence of fine, S’ precipitates was not observed indicating that this phase was totally solutionized during the welding process.
Fig. 53. TEM bright-field micrographs of an IF weld produced between FVS1212 and 2024-T351 using high axial force: (a) general view at the interface; (b) FVS1212 near the interface; (c) 2024-T351 at the inner-HDZ.
The grain size of the recrystallized 2024-T351 ranged from 500 to 1000 nm. In the outer-HDZ of the 2024-T351, the microstructure gradually changed from a fully recrystallized structure to a cold-worked cell structure with entangled dislocations, and gradually changed to that of base metal nearer to the unaffected base metal. An overaged structure was not observed in this region, rather an absence of precipitates indicated solutionizing or reversion. TEM observation of the outer periphery region of the same weld showed essentially the same features. TEM observation of the low axial force IF weld revealed features slightly different from those of the high axial force IF weld, showing an abrupt transition across the interface with less mechanical mixing and a slightly larger recrystallized grain size (500 to 1500 nm).

TEM bright-field micrographs of the dissimilar-alloy LF welds are shown in Figs. 54a - d. As shown in Figs. 54a and b, mechanical mixing between the two alloys appeared more gradual, showing smearing of FVS1212 into 2024-T351 across the interface, which was attributed to the longer contacting time at the peak temperature without large shear deformation as in the IF welds. Dispersoids in the FVS1212 showed no evidence of coarsening or deformation (Fig. 54b).
Fig. 54. TEM bright-field micrographs of a LF weld produced between FVS1212 and 2024-T351: (a,b) general view of the interface; (c,d) 2024-T351 at the inner-HDZ.
Recrystallized 2024-T351 near the interface showed a wide range of alpha grain size distributions with the grain boundary intermetallics aligned semicontinuously along the interface direction (Figs. 54c and d), which was not observed in Fig. 53a. The grain size of recrystallized grains ranged from 500 to 1500 nm. Microstructural features in the outer-HDZ were similar to those of IF welds.

Al-Fe-Mo-V/2024-T351 combination

Figure 55 shows a TEM bright-field traverse across the weld interface at the center of the low axial force weld. Traversing from the Al-9Fe-3Mo-1V side toward the weld interface (left to right in top traverse in Figure 55), neighboring regions showed dark and light contrast depending principally on the local population of dispersoids. Dispersoid-rich regions typically exhibited a darker contrast while dispersoid-lean regions exhibited a lighter contrast. Interestingly, these contrast variations occurred as bands oriented parallel to the weld interface, indicating severe plastic deformation of the as-extruded powder particles near the weld interface. Bright and dark-field imaging of the Al-9Fe-3Mo-1V directly adjacent to the weld interface showed an extremely fine-grained recrystallized alpha aluminum matrix.
Fig. 55. TEM bright-field traverse across weld interface at the center of the low axial force weld. Large arrow indicates the interface between Al-Fe-Mo-V (left) and 2024-T351 (right) and small arrows indicate the regions of 2024-T351 in mechanically mixed region.
In comparison with the Al-9Fe-3Mo-1V base metal microstructure, the microstructural traverse across the Al-9Fe-3Mo-1V near the weld interface revealed a higher density of size-refined particles and a near absence of large spherical dispersoids. Much of the above structural change occurred within a narrow band of about 30 μm from the weld interface. As indicated by the small arrows in Fig. 55, regions near the weld interface in the Al-9Fe-3Mo-1V exhibited bands of 2024-T351. Near the weld interface (extreme right of bottom traverse in Fig. 55), 2024-T351 exhibited equiaxed alpha aluminum grains, indicating the occurrence of recrystallization. In the inner-HDZ of the 2024-T351 outside of this recrystallized zone (RXZ), a gradual microstructural transition from a well-defined, fine subgrain structure to a poorly defined large cell structure was observed.

Five distinct types of morphologically different dispersoid or precipitate particles in Al-9Fe-3Mo-1V were observed near the weld interface (Fig. 54): (1) large, spherical dispersoids approximately 1 μm in diameter; (2) small, spherical dispersoids 0.2 to 0.3 μm in diameter; (3) coarse, irregular shaped dispersoids 0.2 to 0.5 μm in size; (4) fine-sized, acicular dispersoids 0.1 to 0.2 μm in size; and (5) fine-sized, rod type particles less than 0.3 μm in size, which was observed in 2024-T351 base metal. EDS
semi-quantitative compositional analysis indicated that the large type 1 particles were essentially undeformed $\text{Al}_2\text{Fe}_3(\text{Mo, V})$ particles from the base metal; type 2 particles were from the region containing fine dispersoids identified as $\text{Al}_6\text{Fe}$; type 3 particles appeared to result from the fracture of type 1 particles arising from severe plastic deformation occurring in this region; likewise type 4 particles resulted from the fracture of $\text{Al}_3\text{Fe}$ acicular particles in the base metal. EDS analysis of the type 5 precipitates indicated that these are essentially $\text{Al}_{20}\text{Cu}_2\text{Mg}_3$ particles. Evidently, extensive plastic deformation at the weld interface fractured the coarse spherical particles in the Al-Fe-Mo-V base metal and dispersed them as finer-sized particles (types 2 to 4) thereby increasing the local population density of the dispersoids (Ref. 10), while type 5 particles resulted from local mechanical mixing with the 2024.

In comparison with Fig. 55, the TEM bright-field traverse across the interface of the low axial force weld near the outer periphery (Fig. 56) showed two significant differences. Although a random distribution of fine-sized particles together with coarse-sized particles was evident near the weld interface in Al-9Fe-3Mo-1V (large arrow in Fig. 56), the population density of intermetallics was appreciably lower compared to a similar location at the
weld axial centerline. Further, this location showed minimal evidence for mechanical mixing between the 2024-T351 and the Al-9Fe-3Mo-1V.

Figure 57 shows a TEM bright-field traverse across the weld interface at the center of high axial force weld. Away from the weld interface (not shown in Fig. 57), a striated microstructure consisting of dispersoid-rich regions with finer particles and dispersoid-lean regions with coarser particles was observed. The marginal variation in the widths of these alternating microstructural features related well to the difference in the magnitude of the weld axial force, with a higher weld axial force apparently increasing the extent of lateral deformation in the softened region near the weld interface. The inner-HDZ in Al-9Fe-3Mo-1V (large arrow in Fig. 57) exhibited a higher population density and a larger volume fraction of small dispersoids versus a corresponding location in the low axial force weld. TEM examination with minimum beam tilt also revealed an extremely fine-sized, recrystallized alpha aluminum matrix in the Al-9Fe-3Mo-1V. Fine-sized (1 μm) equiaxed alpha aluminum grains were also observed in the 2024-T351 (extreme right of Fig. 57) across the weld interface. 2024-T351 penetration into the Al-9Fe-3Mo-1V was barely noticeable compared to similar locations at the axial centerline of the low axial force weld.
Fig. 56. TEM bright-field traverse across weld interface near the outer periphery of the low axial force weld between Al-Fe-Mo-V (left) and 2024-T351 (right). Large arrow indicates the interface.
Fig. 57. TEM bright-field traverse across weld interface near the center of the high axial force weld between Al-Fe-Mo-V (left) and 2024-T351 (right). Large arrow indicates the interface.
Occasionally, dispersoid particles from the Al-9Fe-3Mo-1V alloy were observed in 2024-T351 across the weld interface, indicating mechanical mixing. Consistent with the higher weld axial force, the width of the RXZ in 2024-T351 was narrower than that in the low axial force weld. Likewise, the outer-HDZ exhibited several deformed alpha aluminum subgrains of a relatively smaller size.

4.3.2.4. EPMA across the weld interfaces

Figures 58 and 59 show composition profiles obtained by EPMA across the weld interfaces near the weld centerline and outer periphery for both the low and high axial force IF welds between Al-9Fe-3Mo-1V and 2024-T351. A marginal pick-up of Cu and Mg in the Al-9Fe-3Mo-1V and Fe in the 2024-T351 over a relatively short transition distance (approximately 5-10 μm) on either side of the weld interface is readily apparent. In comparison, the width of the transition zone was generally wider at the centerline than at the outer periphery region. The transition distance did not appear to vary with the weld axial force.
Fig. 58. EPMA traverses across low axial force weld produced between Al-Fe-Mo-V (right) and 2024-T351 (left). Straight lines in (a) and (b) are Cu K-alpha trace scans while (c) and (d) are results of point analysis.
Fig. 59. EPMA traverses across high axial force weld produced between Al-Fe-Mo-V (right) and 2024-T351 (left). Straight lines in (a) and (b) are Cu K-alpha trace scans while (c) and (d) are results of point analysis.
4.3.2.5. Hardness Testing

**FVS1212/2024-T351 combination**

Figs. 45 - 47 compare the hardness traverses in the dissimilar alloy IF welds and the LF weld. For both welds, the hardness in the FVS1212 remained constant across the region in accordance with TEM observations, which showed no structural changes. Both welds showed extensive hardness variations in the inner- and outer-HDZ’s in the 2024-T351. For the high axial force IF weld, hardness showed the lowest value (145 KHN vs. 177 in the base metal) at the recrystallized inner-HDZ, which gradually increased nearer to the base metal. At the boundary between the outer HDZ and BM, the hardness value reached a maximum (KHN 210). It is suggested that the hardness drop in this region is related to recrystallization and solutionizing or reversion of the S' precipitates. Even with the large deformation in this region, work-hardening was apparently relieved by the recrystallization and recovery processes. The peak hardness values at the HDZ/BM boundary are suggested to be due the combined effect of aging and work-hardening. The low axial force dissimilar alloy IF weld showed similar trends with slightly lower hardness values and wider regions of low hardness.

For the dissimilar-alloy LF weld, the HDZ region also
showed decreased hardness values with a minimum value of KHN 130 (Fig. 47c), which is attributed to the large grain size and comparatively slower cooling rate experienced during LF welding. Hardness variations for all of the welds were well in accordance with microscopic observations of the HDZ's.

**Al-Fe-Mo-V/2024-T351 combination**

Hardness evaluation of the base metals showed an average hardness of KHN 160 in the 2024-T351 and KHN 131 in the Al-9Fe-3Mo-1V. Consistent with the microstructural analysis, the light-etching regions in the Al-9Fe-3Mo-1V exhibited a relatively lower hardness (KHN 157 for a 10 g load) compared to the dark-etching regions (KHN 215 for a 10 g load). Figures 49c and 50c show microhardness traverses across the weld axial centerline and outer periphery for both the low and high axial force welds.

Compared to the Al-Fe-Mo-V alloy, the HDZ in 2024-T351 exhibited extensive hardness variations. In the low axial force weld, the outer periphery region exhibited a lower hardness than along the weld centerline. However, the observed hardness variations in the low axial force weld in 2024-T351 were minimal (-5%) across the HDZ, excepting for a marginal increase (+5%) in the severely deformed region. For the high axial force weld, hardness variations
in the HDZ of 2024-T351 were much larger (± 15 %) than that of the low axial force weld both at the center and outer periphery.

Appreciable hardness transitions were not observed in the Al-Fe-Mo-V alloy excepting for a marginal increase in the striated region near the interface of high axial force weld. Minor fluctuations in the hardness apparently resulted from the inhomogeneous nature of the base metal. Consistent with the microstructural analysis, a significant increase in hardness was observed in the Al-Fe-Mo-V near the weld interface for the high axial force weld (KHN 159 vs. 131 in the base metal).

4.3.2.6. Tensile Testing

Tensile testing results for the dissimilar-alloy welds between FVS1212 and 2024-T351, and between Al-Fe-Mo-V and 2024-T351 are shown in Table 11.

**FVS1212/2024-T351 combination**

Dissimilar-alloy welds exhibited a slightly lower strength than the 2024-T351 base metal and fractured at the inner HDZ in the 2024-T351. Some portion of the fracture occurred along the interface at the outer periphery region. Dissimilar-alloy IF welds produced with high axial force weld showed the highest joint efficiencies of 95 %, which were consistent with the hardness test results.
Table 11: Room-temperature tensile properties of dissimilar-alloy welds produced between FVS1212 and 2024-T351, and Al-Fe-Mo-V and 2024-T351

<table>
<thead>
<tr>
<th>Process</th>
<th>Spec. Loc.</th>
<th>YS (Mpa)</th>
<th>TS (Mpa)</th>
<th>EL. (%)</th>
<th>J.E. (%)</th>
<th>Fracture Loc.</th>
</tr>
</thead>
<tbody>
<tr>
<td>1212/2024,IFW/L</td>
<td>O.P.</td>
<td>299</td>
<td>385</td>
<td>6.0</td>
<td>82</td>
<td>IHDZ</td>
</tr>
<tr>
<td></td>
<td>Center</td>
<td>319</td>
<td>376</td>
<td>1.7</td>
<td>80</td>
<td>IHDZ</td>
</tr>
<tr>
<td>IFW/H</td>
<td>O.P.</td>
<td>343</td>
<td>445</td>
<td>4.2</td>
<td>95</td>
<td>IHDZ</td>
</tr>
<tr>
<td></td>
<td>Center</td>
<td>328</td>
<td>443</td>
<td>4.5</td>
<td>95</td>
<td>IHDZ</td>
</tr>
<tr>
<td>LFW</td>
<td>O.P.</td>
<td>305</td>
<td>393</td>
<td>2.3</td>
<td>85</td>
<td>IHDZ</td>
</tr>
<tr>
<td></td>
<td>Center</td>
<td>314</td>
<td>409</td>
<td>2.6</td>
<td>87</td>
<td>IHDZ</td>
</tr>
<tr>
<td>Al-Fe-Mo-V/2024,IFW/L</td>
<td>1/2 R</td>
<td>278</td>
<td>336</td>
<td>3.7</td>
<td>88</td>
<td>IHDZ</td>
</tr>
<tr>
<td>IFW/M</td>
<td>1/2 R</td>
<td>267</td>
<td>324</td>
<td>4.2</td>
<td>85</td>
<td>OHDZ</td>
</tr>
<tr>
<td>IFW/H</td>
<td>1/2 R</td>
<td>295</td>
<td>346</td>
<td>2.4</td>
<td>91</td>
<td>IHDZ</td>
</tr>
<tr>
<td>FVS1212</td>
<td>BM</td>
<td>522</td>
<td>572</td>
<td>23.0</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Al-9Fe-3Mo-V</td>
<td>BM</td>
<td>238</td>
<td>380</td>
<td>6.7</td>
<td></td>
<td></td>
</tr>
<tr>
<td>2024-T351</td>
<td>BM</td>
<td>324</td>
<td>469</td>
<td>20.0</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

* Minimum two tests were performed on each test locations.
  a. 1/2 R denotes half radius.

The low axial force IF weld showed a similar fracture location. Even with same minimum hardness values at the inner HDZ as those of high axial force IF weld, a wider softened region in the low axial force weld promoted a decreased joint efficiency of 80%. The dissimilar-alloy LF welds showed a joint efficiency of 85% for the same
reason, with fracture occurring in the 2024-T351 at the interface.

**Al-Fe-Mo-V/2024-T351**

In each case, the test specimens failed in the Al-Fe-Mo-V. Although the tensile test specimens showed marginal ductility, the tensile strength of the three types of welds varied between 323 and 346 MPa indicating a joint efficiency in excess of 85% relative to the Al-9Fe-3Mo-1V base alloy (380 MPa). All tensile fractures occurred away from the interface (5 mm from the interface for the low axial force weld and 25 mm for the medium axial force weld), except for the high axial force weld (0.2 mm from the interface), which was attributed to the severely laminated microstructure near the interface.

4.3.2.7. Fracture Analysis

**FVS1212/2024-T351 combination**

In order to characterize fracture across the entire interface region, three-point guided bend tests were performed on axial-half cut sections of the three types of welds with the flat diameter section in tension.

SEM fractographs of the bend test specimens of dissimilar-alloy welds are shown in Fig. 60. The IF weld produced with low axial force exhibited fracture at the interface with the transfer of FVS1212 metal across the
interface and with a major portion of fracture surface in the inner-HDZ of the 2024-T351 (Fig. 60a). At increased magnification, fine, elongated features were observed on the 2024-T351 side, which was caused by deformation of previous base metal grains and alignment of grain boundary intermetallics as stringers (Figs. 60b). The center region of the bend test specimen exhibited fracture on the FVS1212 side, which was attributed to the mechanical mixing at the region between the two alloys. The fracture surface of the IF weld produced with high axial force exhibited similar features as those of the low axial force IF weld, with slightly wider steps caused by higher compressive deformation of the 2024-T351 base metal. Fracture surfaces of the LF weld bend specimens exhibited similar features as those of the IF welds (Figs. 60c and d), however the shear track was more irregular and random due to less shear deformation in LF welding compared with IF welding.

EDAX analysis of the fracture surfaces in Fig. 60b revealed that the light-appearing regions are FVS1212 carrying a high percentage of Mg and Cu, which originated from the 2024-T351 (3.67 wt. % Mg and 2.79 wt. % Cu). The transfer of alloying elements was due to the combined effect of mechanical mixing and diffusion across the interface.
Fig. 60. SEM fractographs of bend test specimens of dissimilar-alloy welds produced between FVS1212 and 2024-T351: (a,b) outer periphery of IFW produced with low axial force; (c,d) outer periphery of LFW.
The dark regions in Fig. 60b showed the same composition as the base metal with a small increase in Fe content. It was concluded that the dark region is 2024-T351 with a slight transfer of dispersoids by mechanical mixing across the interface. Diffusional transfer was minimal due to the low diffusivity and solubility of Fe, V and Si in FVS1212, and the extremely rapid weld thermal cycle.

The fracture surface of the tensile test specimens from the low axial force IF welds exhibited similar features with a major portion of the fracture surface located in the 2024-T351 (Figs. 61a and b). EDAX analysis of the fracture surfaces showed similar results as in the bend test fracture surfaces, showing a light region as FVS1212 with Mg and Cu and dark regions as 2024-T351. Some portion of the fracture surface exhibited a trace of FVS1212 type dispersoids on the 2024-T351 fracture surface, implying that fracture occurred at the interface across thin layers of mechanical mixing (Fig. 61b) and some portion of the fracture propagated through the FVS1212. The fracture surface of the high axial force IF welds exhibited similar features as those of the low axial force weld with slightly wider steps caused by higher compressive deformation of the 2024-T351 base metal (Figs. 61c and d) and a higher proportion of FVS1212. The higher proportion of FVS1212 resulted from severe mechanical mixing with higher axial force.
Fig. 61. SEM fractographs of tensile test specimens of dissimilar-alloy welds produced between FVS1212 and 2024-T351: (a,b) center of IFW produced with low axial force; (c,d) with high axial force and (e,f) LFW.
Tensile fracture surfaces of the LF weld showed similar features as those of the IF welds, except for the deformation pattern of the fracture surface, where a random wave pattern was observed (Fig. 61e). A larger proportion of FVS1212 was observed on the fracture surface as shown in Fig. 61f, which was caused by smearing of FVS1212 into the 2024-T351.

Summarizing the above for dissimilar-alloy welds, all of the fracture surfaces were located in the inner-HDZ at the 2024-351 interface. Tensile test specimens showed a major proportion of fracture to occur in the inner-HDZ of the 2024-T351 interface and only occasionally to occur in the FVS1212 across the interface, of which the proportion depended on the welding processes and parameters. For the IF welds, an increase in axial pressure increased mechanical mixing across the interface and the tensile fracture surface of the high axial force IF weld showed a greater proportion of FVS1212 than the low axial force IF weld. For the LF weld, the smearing of FVS1212 into the 2024-T351 showed a higher proportion of FVS1212 than the high axial force IF weld.

Al-Fe-Mo-V/2024-T351 combination

In all welds, bend test fracture initiated at the flat bottom section and propagated to the top periphery region
following the interface curvature in the Al-Fe-Mo-V. The distance of fracture location from the interface showed a marginal decrease at higher axial forces: 0.25 mm in the low axial force weld versus 0.2 mm in the medium and high axial force welds.

Figures 62 and 63 show the results of the fractographic examination of the failed bend test specimens. At low magnification, the fracture surface of the low axial force weld exhibited a distinctive step at 2/3 radius from the center of the weld specimen (Fig. 62a). The fracture surface clearly revealed the outward flow of base metal grains in the Al-9Fe-3Mo-1V. Examination of the fracture surface at higher magnification showed a flat topography with relatively finer facets at the weld center and coarser facets near the periphery (Figs. 62b and d). Examination of these regions at increased magnification indicated fracture initiation by microvoid formation at dispersoid/alpha aluminum interfaces and final ductile rupture of alpha aluminum matrix (Figures 62c and e). A marginal increase in the size of the ductile dimples was also apparent in the outer periphery region versus the weld center (Figs. 62c and e).
Fig. 62. SEM fractographs of low axial force weld bend test specimen produced between Al-9Fe-3Mo-1V and 2024-T351: (a) general view at low magnification; (c,d) center; (d,e) outer periphery.
Fig. 63. SEM fractographs of high axial force weld bend test specimen produced between Al-9Fe-3Mo-1V and 2024-T351: (a) general view at low magnification; (c,d) center; (d,e) outer periphery.
In comparison with the fracture surface of the low axial force weld, the fracture surface of the high axial force weld did not exhibit a readily discernible step (Fig. 63a). However, closer examination did reveal the occurrence of a "faint" step nearer to the weld center. Examination at a higher magnification showed uniformly fine-sized flat facets both at the center and outer periphery (Fig. 63b and d) in sharp contrast with the appreciable size variations of the facets in these locations in the low axial force weld (Figures 62b and d). As in the low axial force weld, examination of these regions at increased magnification indicated fracture initiation by microvoid formation at dispersoid/alpha aluminum interfaces and final ductile rupture of alpha aluminum matrix with similar feature as shown in Fig. 63c.

The fracture surfaces of the tensile test specimens exhibited similar features as those of the bend test specimens.
4.4. Dissimilar-alloy welds between RS/PM Al-Fe-X alloys and titanium alloys

4.4.1. Preliminary dissimilar alloy welding study between IM aluminum and titanium alloys

Dissimilar-alloy welds between 2024-T351 and Ti-662 (Ti-6% Al-6% V-2% Sn) were produced and characterized as a preliminary experiment prior to the inertia-friction welding of RS/PM Al-Fe-X alloys to Ti alloys.

Basically two different approaches were utilized in the production of dissimilar-alloy welds between aluminum and titanium alloys. Welding mechanisms can be classified into two categories depending on the interface temperature: 1) adhesion of surface atoms, mechanical mixing and diffusion across the interface below the recrystallization temperature, 2) recrystallization across the interface above the recrystallization temperature.

With a high rotating speed and low inertia, no satisfactory welding could be obtained. For the first mechanism, applying a low rotating speed (950 rpm) and high inertia (1.16 Kg-m^2) with a resulting relatively longer welding time and a compressive stress above the UTS of 2024-T351 resulted in a satisfactory weld producing a joint efficiency of 100%. There was no observable macroscopic deformation in the Ti-662 while the 2024-T351 exhibited typical hour-glass-shape inner- and outer-HDZ's observed in
inertia-friction welding. Recrystallization was apparent at the inner HDZ in the 2024-T351 but not in the Ti-662. At an increased magnification with SEM microscopy, the Ti-662 showed slight deformation within 5 microns from the weld interface. Mechanical mixing between the two alloys was not detectable using light and SEM microscopy. Hardness traverses across the weld region exhibited little change from the base metal to weld region for the both alloys. It was concluded that deformation and work hardening during the final stage compensated for annealing effects during the welding process. Severe mechanical interlocking plus diffusion resulted in satisfactory welds and fracture occurred in the inner HDZ of the 2024-T351.

For the dissimilar-alloy welding of RS/PM Al-Fe-Mo-V to Ti-1100 and Ti-662, and Al-Fe-V-Si alloys to Ti-1100 with welding parameters given in Table 5, the welds surfaces exhibited minimal welding and shiny friction tracks after bend testing with minimal joint efficiency. A high volume % of dispersoids located directly at the weld interface are suspected to be responsible for this complete lack of welding in addition to the large strength difference between the alloys over the entire temperature range during the welding.
4.4.2. Dissimilar-alloy welds between Al-Fe-X alloys and Ti-1100

Experiments described above were performed with a weld geometry such that the diameter of titanium alloys was 3 mm smaller than the aluminum alloys with a restraint ring around the aluminum alloys to prevent bulging of the soft aluminum alloys. Even with this configuration, excessive deformation in the aluminum alloys was a problem. Considering the increased room and elevated-temperature strength of the Al-Fe-V-Si alloys, compared with IM 2024 or RS/PM Al-9Fe-3Mo-1V alloys, the same diameters (22.2 mm) were joined without a restraint ring.

Using a higher rotating speed (above 3500 rpm), metallic bonding was observed at the outer periphery. However, mechanical strength was minimal as measured by tensile testing.

4.4.3. Dissimilar-alloy welding between FVS1212 and CP Ti

As dissimilar-alloy welding between RS/PM Al-Fe-V-Si alloys and high hardness/strength titanium alloys were performed with limited success, FVS1212 (which has the highest hardness and strength) was welded to CP Ti to investigate the possibility of dissimilar-welding of RS/PM aluminum alloys to lower strength titanium alloys. Using the maximum weld axial force allowed by the welding system,
two different rotation speeds were chosen.

The welds exhibited smooth, axially-symmetric flash formation preferentially in the FVS1212. Macroscopic observation of the axially-sectioned welds showed almost the same features as those of dissimilar-alloy welds between other RS/PM aluminum alloys and titanium alloys, showing negligible change in the CP Ti, and a recrystallized inner HDZ and texture reorientation in outer-HDZ of FVS1212 as shown in Figs. 64a and b. The higher RPM weld exhibited slightly wider inner- and outer-HDZ's and the formation of more flash compared to the low RPM weld. Examination of the as-polished surfaces of axially sectioned specimens did not show any evidence of defects.

Microscopic observation of the interface showed four different regions as shown in Figs. 65a - d: 1) a light region in CP Ti, 2) a dark region in FVS1212, 3) a homogeneous inner-HDZ and 4) an outer-HDZ in FVS1212. The light-etching region in the CP Ti was confirmed to be a recrystallized region. Structural changes in the CP Ti outside of the recrystallized region were not observed. Occasionally, a slight change of contrast was observed on the FVS1212 side of the recrystallized CP Ti, indicating diffusion of Al into the Ti. However, evidence of Al,Ti intermetallic was not observed on optical microscopy scale.
Fig. 64. Light macrographs of dissimilar inertia-friction welds produced between FVS12121 and CP Ti: (a) low RPM and (b) high RPM.
Fig. 65. Light micrographs of dissimilar inertia-friction welds produced between FVS12121 and CP Ti: (a) center and (b) outer periphery of low RPM weld; (c) center and (d) outer periphery of high RPM weld.
The dark region in the FVS1212 was comprised of a high dispersoid density structure at the interface due to the inhomogeneous deformation as observed in similar-alloy IF welds in FVS1212. The width of the dark region increased at the outer periphery of the weld due to the more severe deformation in this region. Also, dispersoid-lean, white bands were observed accompanied by dark regions.

The high rotating speed resulted in a microstructure similar to that of the similar-alloy low axial force weld with same rotating speed, i.e. the high rotating speed resulted in more dispersoid-lean and rich bands due to the higher energy input and prolonged deformation. Considering the recrystallization in CP Ti, applying higher axial force may result in high quality welds with minimum energy input.

Microhardness traverses for the welds are shown in Figs. 66a and b. As shown in Fig. 66a, for the low RPM weld, a slight hardness decrease was observed in the inner-HDZ’s in the FVS1212, but was within the variation of hardness in the base metal. A slight decrease was also observed in the recrystallized region of the CP Ti. For the high RPM weld, similar trends were observed except at the dark band located at the outer periphery. The dark band, where the population density of the dispersoids increased due to nonuniform deformation, showed an extraordinary high hardness value (DPH 372).
Fig. 66. KHN traverses for low and high rpm inertia-friction welds produced between FVS1212 and CP Ti: (a) low rpm weld; (b) high rpm weld.
Tensile test results of the welds exhibited minimal strength (UTS values of 45 MPa for the high rpm weld and 100 MPa for the low rpm weld vs. 572 MPa of the base metal).

Summarizing, difficulties in achieving good bonds between RS/PM Al-Fe-X and titanium alloys by friction welding results from: 1) large strength differences between the alloys over the whole temperature range (especially at elevated temperatures); and 2) the enrichment of intermetallic dispersoids near the interface due to nonhomogeneous deformation in dispersoid-rich RS/PM aluminum alloys.

4.5. Elevated temperature exposure characteristics of the FVS0812 friction welds

Light microscopy observation of the base metal and weld region did not show any noticeable structural change for the isothermal heat treatment of high axial force weld in FVS1212 at 425 °C. The temperature was chosen as it is generally accepted as the maximum operation temperature for the alloy. DPH vs. exposure time for different weld regions are shown in Fig. 67. As shown minimal variation was observed in the base metal, at the weld interface and outer HDZ. Even with a loss of hardness, the inner HDZ at
the central axis still exhibited higher hardness than the other regions.

TEM bright-field micrographs for different exposure times are shown in Fig. 68. TEM observation of the interface regions revealed no change in grain size and a slight reduction of dispersoid population inside the matrix as compared to the as-welded interface (Fig. 68a vs. 68b).

![Graph showing DPH hardness vs. aging time in the base metal and interface regions of IF weld produced using high axial force in FVS0812.](image)

**Fig. 67.** DPH hardness vs. aging time in the base metal and interface regions of IF weld produced using high axial force in FVS0812.
Fig. 68. **TEM bright field micrographs of the inner-HDZ at the central axis for the FVS0812 IF weld (high axial force) aged at 425 °C:** (a) as-welded; (b) aged for 500 hrs; (c) fine dispersoid region and (d) coarse dispersoid region of the base metal exposed for 500 hours.
Minimal growth of grains were obviously due to the pinning effect of a large volume fraction of dispersoids. Some of the dispersoids were located at the corners of grains, but the dispersoid distribution was uniform across the grains and grain boundaries up to an exposure time of 500 hours. It is assumed that slight decrease in population density occurred both within the grains and at the grain boundaries accompanied by coarsening. TEM micrographs of the base metal (Fig. 11c and d) showed a similar trend but the degree of coarsening was slightly less than at the interface region. Quantitative measurements of dislocation density was not performed but multibeam bright-field imaging did show a slight reduction of dislocation densities at the interface and outer HDZ.

Size distributions of dispersoids for the isothermal heat treatment experiments are shown in Fig. 69 and the change in dispersoid size with exposure time is given in Fig. 70. Movement of distribution peaks and elimination of a bimodal distribution was observed, indicating that the size distribution was approaching steady state after 200 hrs.

To further study the coarsening effects, isothermal heat treatment was performed at 500 and 573 °C. Light microscopy analysis of the aged specimens revealed the formation of acicular dispersoids at 500 °C (Fig. 71a) and
noticeable coarsening of dispersoids and increase in the population density of the acicular dispersoids at 570 °C (Fig. 71b). The distribution of the acicular dispersoid was uniform across the base metal and weld interface region.

TEM observation clearly showed noticeable coarsening of dispersoids and the presence of acicular dispersoids as shown in Fig 72. The alpha grain size remained the same at 500 °C (Fig. 72a) but increased slightly to 0.55 μm at 575 °C (Fig. 72b). The agglomeration of dispersoids at grain boundaries and their coarsening were noticeable. The average diameter of dispersoids was 130 nm at 500 °C and 186 nm at 575 °C, excluding the acicular phase. The acicular dispersoids ranged from 1 to 8 μm in length at 575 °C and grew across several grains (Fig. 72c). E D S analysis of the acicular dispersoids showed 70.6 Al - 28.5 Fe - 0.3 V - 0.5 Si in at. %. Detailed diffraction analysis was not performed, however, comparing with previous diffraction studies (Refs. 24, 25, 85), morphology of dispersoids and EDS analysis, it was concluded as equilibrium Al₃(Fe,V) or Al₁₃(Fe,V)₄, which is C-centered monoclinic.
Fig. 69. Normalized distribution curve of dispersoids at the inner HDZ at the central axis for the FVS0812 IF weld (high axial force) aged at 425 °C.
Fig. 70. Average dispersoid size as a function of aging time at 425 °C for the FVS0812 IF welds (high axial force).
Fig. 71. Light micrographs of weld interface regions at the axial centerline in FVS0812 exposed for 1 hour at: (a) 500 °C; (b) 575 °C.
Fig. 72. TEM bright field micrographs of the inner-HDZ at the central axis for the FVS0812 IF weld (high axial force) exposed for 1 hour at: (a) 500 °C; (b) 575 °C; (c) acicular dispersoids produced at 575 °C.
Microhardness measurements at the weld interface, outer-HDZ and base metal showed that degradation of mechanical properties was negligible up to 500 °C (Fig. 73). However, the hardness rapidly decreased above 500 °C.

Fig. 73. DPH hardness vs. exposure time in the base metal and weld interface region in FVS0812 IF weld exposed for 100 hours.
5.1. Process, structure, properties and fracture relationships for the similar-alloy welds in FVS1212 and FVS0812

Inertia-friction welding experiments demonstrated that high joint efficiency welds in RS/PM aluminum alloys FVS1212 and FVS0812 are feasible without coarsening of the dispersoids and microstructure due to its rapid thermal cycle and the expulsion of heat- and deformation-affected metal out of the interface region. Weld axial force played an important role in the formation of the HDZ microstructure. The low axial force IF weld exhibited dispersoid-lean bands at the interface due to the inhomogeneous deformation resulting from simultaneous compressive and shear stresses at high temperature, while the high axial force IF welds showed an absence of defects and microstructural coarsening of the microstructure.

Linear-friction welds in FVS1212 and FVS0812 exhibited moderate coarsening of dispersoids at the interface region.
It is concluded that, for the welding parameters utilized in this study, the thermal cycle is relatively slow compared to the IF welds. However, uniform heat generation and deformation at the interface resulted in a uniform interface microstructure independent from radial location in the LF weld, while the inertia-friction welding resulted in a radially varying microstructure due to its rotational motion. Future optimization studies utilizing a higher axial force and resulting in a shorter welding time should reduce the coarsening effect by shortening the thermal cycle and increasing the expulsion of softened metals from the interface region.

Results of hardness testing and tensile testing corresponded well with microstructural observations. The most important factor in determining the mechanical properties of the welds was the coarsening of dispersoids and the alpha grain structure in the weld region, showing the highest strength and hardness for the high axial force IF welds where the microstructural coarsening was minimum. In addition, the distribution of dispersoids influenced the mechanical properties. Tensile and bend test fracture occurred at the interface region for all the similar-alloy welds and the features of both fracture surfaces were corresponded for each alloy/process combination. Fracture characteristics were also closely related with the
microstructural differences of each weld. The wide homogeneous inner-HDZ exhibited by low axial force IF welds resulted in a relatively smooth fracture surface with fine shear tracks, while the narrow striated microstructure in the high axial force IF welds resulted in a rough fracture surface with coarse shear tracks. Sometimes, fracture occurred in the dispersoid-lean band of low axial force IF welds, which confirms the low strength of the tensile test specimen in this region. Unidirectional flow in the LF welds parallel to the linear displacement direction resulted in directional fracture surface patterns. The narrow inner-HDZ of the LF weld showed similar features as those of the high axial force IF weld.

In general, both the FVS1212 and FVS0812 behaved in a similar manner during the friction welding. A higher joint efficiency observed in FVS0812 welds was attributed to the lower coarsening of dispersoids in the interface region of the FVS0812 welds. The same explanation can be applied to the LF welds compared to the IF welds for both alloys.

5.2. Comparison of dispersoid behavior in the similar-alloy welds of FVS1212 and FVS0812

In the previous study of inertia-friction welding of Al-9Fe-3Mo-1V (Ref. 10), the fracture of dispersoids, dispersion of the fractured dispersoids and presence of
three different types of intermetallic dispersoids \((\text{Al}_{12}\text{Fe}_3(\text{Mo},\text{V}), \text{Al}_6\text{Fe} \text{ and } \text{Al}_3\text{Fe})\) made it difficult to assess the behavior of dispersoids in response to the thermal and mechanical behavior of the inertia-friction welding process. Rather than coarsening, the fracturing of large \(\text{Al}_{12}\text{Fe}_3(\text{Mo},\text{V})\) dispersoid particles increased the population density of dispersoid particles and hardness in the weld interface region.

FVS1212 and FVS0812 exhibited only one type of dispersoid \((\text{Al}_{13}(\text{Fe},\text{V})_3\text{Si})\) of which the morphology is spherical. Due to the relatively small size of the dispersoids (100 nm), fracturing in the weld region was not observed for all of the combinations of weld processes/parameters/materials. Also, no transformation of dispersoids was observed in the friction welds. All the above facts simplified the analysis of dispersoid behavior during the different friction welding processes.

5.2.1. Comparison of dispersoid coarsening in the friction welding processes

For inertia-friction welding, a longer welding time for the low axial force welds resulted in more coarsening than in the high axial force welds. Negligible coarsening at the center of inner-HDZ at the central axis confirms the expulsion of softened materials from this region. Most
significant coarsening of the dispersoids was observed at the outer periphery of the low axial force IF weld and at the overall interface regions of the LF weld. Coarsening at the outer periphery of the low axial force IF weld was greatest, which is consistent with the fact that heat generation is highest and weld duration time is longest. Coarsening in the LF weld was mainly attributed to the low applied axial force in this trial. Also, as observed in the continuous drive friction welding, expulsion of softened material is continuous in LF welding, while the final increase in torque in inertia-friction welding results in a large final upset at the final stage, removing the already softened material to the periphery.

The outer periphery region of the low axial force inertia-friction welds represents the total coarsening during the welding sequence. Calculations of dispersoid coarsening in the FVS0812 inertia-friction welds were performed based on coarsening by static volume diffusion, grain boundary diffusion and dislocation pipe diffusion assuming the peak temperature of 655 °C with duration of 2 to 5 seconds. The results showed less than a 1% increase compared to the measured values at the outer periphery (see Section 5.4.1.). Considering this large discrepancy, coarsening observed in this experiment should be related to accelerated coarsening mechanisms. Enhanced diffusion by
moving dislocations and grain boundaries are definitely responsible for this accelerated coarsening. Also, considering the small grain size and recrystallization in this region, coarsening of grain boundary particles by coalescence may account for the coarsening. In addition, considering the high volume percentage of dispersoids, coalescence of the dispersoids inside matrix may be possible. Especially, extraordinarily large dispersoids observed in the LF weld in FVS1212 cannot be explained other than coalescence of already coarsened dispersoids. Unfortunately, any quantitative theoretical analysis were not available.

5.2.2. Comparison of dispersoid coarsening in FVS1212 and FVS0812

As the distribution of dispersoids had not achieved steady state, as evidenced by the bimodal base metal distribution, it was difficult to comparatively evaluate the coarsening kinetics of the two alloys. Assuming the initial conditions were almost the same and that the amount of coarsening does not depend on the initial dispersoid size, coarsening in FVS1212 welds was greater than that at corresponding locations in the FVS0812 welds. Two factors can be considered: 1) the volume fraction of dispersoids; and 2) the difference in lattice parameters and surface
energy. Lattice mismatch and surface energy values for both alloys are not available. An approximately 3.5 \% mismatch is reported (Ref. 21). A change in the lattice parameter of the \( \text{Al}_{12.0-14.0} (\text{Fe}, \text{V})_3 \text{Si}_{0.9-1.25} \) phase is reported to be less than 0.3 \% within this compositional variation in the silicide phase (Ref. 21). As a result, the difference in both the surface energy at the silicide/matrix interface and in \( K_j \) values in FVS1212 and FVS0812 should be small and the second argument cannot be applicable. Considering the large difference in volume fraction (24 \% for FVS0812 and 36 \% for FVS1212) and resulting difference in the rate constant \( (K_j(0.24)/K_j(0) = 2.2 \) and \( K_j(0.36)/K_j(0) = 2.5 \) (see Section 1.5), the difference in coarsening between the two alloys can be directly related to the VDC case, which can be indirectly related to other diffusion controlled coarsening mechanisms (Refs. 34, 35). As shown by Asimov (Ref. 31), for any diffusion controlled growth, \( K_n \) values are controlled by the diffusion fields of each particle. Also, the large volume fraction and severe deformation involved in the friction welding processes, coalescence of dispersoids in moving grain boundaries and also inside the matrix should be responsible for the coarsening of the large dispersoids.

Fig. 74 shows the tensile strengths of FVS1212, Al-9Fe-3Mo-1V and 2024-T351 as a function of temperature. As shown, a strength reversal is observed between 2024-T351 and Al-Fe-Mo-V alloys, while FVS1212 shows higher strength than 2024-T351. These two cases illustrate typical situations in dissimilar-alloy welding and the characteristics of both cases will be discussed in this section.

Dissimilar inertia-friction welding of the RS/PM Al-9Fe-3Mo-1V alloy with 2024-T351 involves a unique situation i.e., at room temperature 2024-T351 is stronger than Al-9Fe-3Mo-1V while at elevated-temperatures Al-9Fe-3Mo-1V exhibits a higher strength. Increases in temperature above 150 °C promoted a wider difference in the tensile strength between the two materials.

Considering that during inertia-friction welding, a larger amount of frictional heat is generated at the outer periphery than at the center (arising from a larger difference in relative speed in the outer periphery versus the center), a high peak temperature distribution occurs at the outer periphery versus the specimen center.
Fig. 74. Tensile strength vs. test temperature for RS/PM Al-9Fe-3Mo-1V, FVS1212 and 2024-T351 (Refs. 1, 22, 92).
Consequently, the strength difference between the two alloys and the extent of softening are larger near the periphery than at the center. The above reversal in strengths at elevated-temperatures and the extent of softening, together with the complex combination of axial and torsional forces at the weld interface promoted an extremely complex thermo-mechanical material processing condition during inertia-friction welding, which in turn leads to a variety of microstructural transitions across the weld interface. Specific macroscopic features of the welds, such as the occurrence of flash and appearance of the weld interface observed in this study, clearly demonstrate the complicated nature of the deformation conditions that occur during the inertia-friction welding process.

It is interesting to note that the weld axial pressures utilized in this investigation were greater than the 0.2% yield strength of Al-9Fe-3Mo-1V at room temperature (244 MPa). The weld axial pressures of the medium and high axial force welds also exceeded the room-temperature yield strength of 2024-T351. Despite utilization of axial forces greater than the room-temperature yield strength of the Al-9Fe-3Mo-1V alloy, the occurrence of weld flash principally in 2024-T351 indicated rapid softening and extrusion of 2024-T351 at elevated-
temperatures. In contrast to 2024-T351, the relatively higher strength of Al-9Fe-3Mo-1V at elevated-temperatures inhibited pronounced plastic deformation and retained much of the heat-affected Al-9Fe-3Mo-1V in the weld region.

Macroscopically, while most of the flash formed in IM-2024-T351, the extent of HDZ was larger in Al-9Fe-3Mo-1V. It appeared that rapid softening of 2024-T351 in the weld region removed the softened 2024-T351 resulting in a large amount of flash and narrower HDZ (Fig 44). On the contrary, a relatively higher strength of Al-9Fe-3Mo-1V at elevated-temperatures compared to that of 2024-T351 precluded extrusion of Al-9Fe-3Mo-1V as weld flash and retained much of the heat-affected Al-9Fe-3Mo-1V within the weld region. Also, the generation of a larger amount of frictional heat at the weld periphery than at the weld center resulted in a higher peak temperature distribution and larger extent of softened region versus at the weld center. As a result, strength differences between the two alloys were larger at the periphery than at the center. Due to the above complex conditions at the faying surfaces, 2024-T351 penetrated into Al-9Fe-3Mo-1V at the center and promoted a curved weld interface. This effect was more noticeable for the higher axial force due to the larger deformation in the Al-9Fe-3Mo-1V. Also, it should be noticed that the rapid removal of softened 2024-T351
exposes unaffected base metal near the interface increasing the load carrying capacity at the center compared to periphery.

Light and TEM microscopy analysis revealed significant mechanical mixing of 2024-T351 and Al-9Fe-3Mo-1V at the interface near the weld center of the low axial force weld. Rao et al (Ref. 76) have reported that mechanical mixing generally originates from "interlocking" of the initial asperity peaks at the specimen faying surfaces and subsequent formation of "wedges" and smearing of wedges. McMullan et al (Ref. 80) have proposed that during the initial stages of dissimilar (inertia-) friction welding, the soft material rubs with the surface of the hard material (interlocking of asperity peaks of soft and hard materials) and forms an initial coating (smearing) of soft material onto the harder one, causing the plane-of-rubbing to progressively move toward the soft material. The above mechanisms can be applied to the present case to explain both the appearance of the weld interface and mechanical mixing of 2024-T351 in Al-9Fe-3Mo-1V. During the initial stages of inertia-friction welding, the difference in strength between these two materials can be considered to be marginal at the center. Consequently, it is likely that wedge formation and smearing occurred both in the 2024-T351 and Al-Fe-Mo-V at the initial stage. In the high axial
force, the higher compressive stress at the interface region fully homogenized the layered structure in the later stage of the welding, while the low axial force weld retained the original layered structure. Near the outer periphery, a lack of mechanical mixing between the two alloys was basically due to the rapid softening of 2024-T351 and an increase in the strength difference between the two materials.

EDS and EPMA analysis across the weld interface showed compositional transitions over short distances (5-10 μm) on either side of the weld interface. The limited solid-solubility of iron in alpha aluminum (0.052 wt% maximum) and the superior elevated-temperature stability of the strengthening dispersoids in Al-9Fe-3Mo-1V likely precluded significant solid-state dissolution of Fe in the alpha aluminum matrix during the weld heating cycle. Further, considering the low solid-state diffusivity of iron in alpha aluminum, it appeared that the short-range transport of iron atoms across the weld interface occurred principally by mechanical mixing. Considering that Mg and Cu-rich precipitates exhibit an increasing solid-solubility in alpha aluminum at high temperatures (maximum solid-solubility of Cu in Al is 5.67 wt%, while that of Mg is 14.9 wt%), it is likely that these particles underwent progressive solid-state dissolution during the weld heating
cycle and were retained in solution during the subsequent rapid weld cooling cycle.

Depending on the local level of supersaturation of the alpha aluminum matrix and the weld cooling rate, the solid-state diffusion of Cu and Mg across the weld interface into the Al-9Fe-3Mo-1V could occur over microscopically larger distances. Mechanical mixing across the interface could also aid the above transport. However, considering that the width of the transition zone was nearly same for Cu, Mg and Fe, the transport of these elements was perhaps accomplished exclusively by mechanical mixing. Nominal variations in the widths of the transport zone at the weld center versus the periphery may be attributed to the differences in normal compressive stress and local temperature distribution during the inertia-friction weld cycle. The effect of axial force on the width of the transport zone was negligible.

Light and TEM metallographic characterization showed extensive microstructural changes in 2024-T351 and relatively minor changes in Al-9Fe-3Mo-1V. Consistent with the previous TEM investigation on the microstructure of the inertia-friction weld interface in Al-9Fe-3Mo-1V alloy (Ref. 10), the present investigation also showed fracture of coarse-sized base metal dispersoids in Al-9Fe-3Mo-1V alloy into fine-sized acicular and spherical particles. In
addition to a refined homogeneous structure, the near-HDZ in Al-9Fe-3Mo-1V exhibited recrystallized, fine-sized alpha aluminum grains. Hardness traverses indicated a marginal softening (hardness trough) in this region both in the low and high axial force welds. Longitudinal three-point guided bend testing indicated that weld integrity is essentially related to the interface between the inner-HDZ and the outer-HDZ in Al-9Fe-3Mo-1V. Preferential fracture initiation in this region during bend testing is attributed to a relatively smaller difference in hardness between this region and the outer-HDZ versus the weld interface. In comparison, transverse weld tensile testing showed that preferential failure was associated with the unaffected Al-9Fe-3Mo-1V farther from the weld interface. Despite fracture in the unaffected base metal, all the three welds exhibited a marginal loss in joint efficiency. The apparent discrepancy is attributed to two factors. While the narrow softened region in Al-9Fe-3Mo-1V produced a "constraint effect" and precluded failure in this location, the inhomogeneous deformation during extrusion of Al-9Fe-3Mo-1V alloy produced relatively weak inter-particulate bonding and promoted preferential failure by "delamination" along particulate boundaries.

For the dissimilar-alloy welds between FVS1212 and 2024-T351, most of the structural changes were observed in
2024-T351. Coarsening and other structural changes were negligible in FVS1212 for both the IF and LF welds, which was attributed to the fact that most of the applied energy was consumed in the flash formation in the softer 2024-T351. Mechanical mixing in the low axial force IF weld was minimal while the high axial force weld showed penetrated layers of FVS1212 in the 2024-T351. For the LF weld penetrated layers were not observed but smooth smearing of FVS1212 into the 2024-T351 was noticed, which can be related to the smaller shear deformation at the interface compared to the IF welds. Compared with the inertia-friction welds, the linear-friction welds resulted in the larger recrystallized grains in the inner-HDZ of the 2024-T351, which was not noticeable in the similar-alloy LF weld as the grain growth of the recrystallized grains were hindered by the high volume percentage of dispersoids. The grain growth can be related to the differences in heat generation location and subsurface shear deformation. In inertia-friction welding, most of the heat is generated at the outer periphery and the softened metals can be easily removed at the final upset stage. In the LF welding the heat is generated uniformly across the interface and the softened metal at the center cannot be easily removed. Also, the large final torque and severe subsurface shear deformation can increase the nucleation sites and refine
the resulting grain size.

Results of hardness and tensile testing agreed closely with microstructural observations. Even with the lowest hardness values in the inner-HDZ of the LF weld among the three types of welds, tensile strength of the LF weld exceeded that of the low axial force IF weld, which was attributed to the smooth smearing across the interface compared to the lack of mechanical mixing in the low axial force IF weld. Both the tensile and bend tests resulted in fracture initiation at the interface region in the 2024-T351. Depending on the mechanical mixing and smearing, the proportion of FVS1212 on the fracture surface changed showing a higher proportion of FVS1212 for the high axial force IF weld and the LF weld. As in the similar-alloy welds, the fracture surfaces of the dissimilar welds also showed flow patterns at the interface region, where IF welds showed a spiral shear patterns while LF weld showed an irregular wavy pattern.

5.4. Dissimilar-alloy welding between RS/PM Al alloys and Ti alloys

Successful inertia-friction welding of 2024-T351 to Ti-662 was possible utilizing a restraint ring around the softer 2024-T351 and by applying a high axial force. The fact that successful welds exhibited marginal deformation
in the Ti-662 up to 5 μm depth showed a necessity to form a nascent surface on the surface of harder titanium alloys.

Dissimilar-alloy inertia-friction welds between Al-9Fe-3Mo-1V, and Ti-1100 and Ti-662 alloys and between Al-Fe-V-Si alloys and Ti-1100 with a restraint ring did not produce high integrity welds using a range of welding parameter combinations.

FVS1212 alloy shows the highest room and elevated strength among the RS/PM alloys. Inertia-friction welding between FVS1212 and CP Ti was performed because the combination exhibited the smallest strength difference among the alloys chosen. Even with recrystallization and deformation in CP Ti, high integrity welds could not be obtained. A lack of weld strength was attributed to the presence of dispersoid-rich region caused by the severe non-uniform deformation in the inner-HDZ of FVS1212. The dispersoid-rich region was created by the same mechanism as in the similar-alloy inertia-friction welds. However, due to the presence of harder CP Ti at the interface, the harder dispersoid-rich region was concentrated at the interface while the softer dispersoid-lean region was removed from this region due to the compressive stress. It can be argued that the bond between intermetallic silicides and metallic titanium is weaker than the metallic bond between aluminum and titanium.
5.5. Behavior of friction welds on elevated temperature exposure

5.5.1. Comparison with theoretical analysis

From the literature review (Refs. 26 - 35), the analytical equation for coarsening becomes:

\[ X(t)^n - X(0)^n = K_n t \]  \hspace{1cm} (Eq. 17)

where \( n \) is an integer depending on the coarsening mechanism.

For VDC growth using Fines's modified equation (Ref. 37):

\[ K_3 = \frac{(8/9) (D_C \phi^2)}{(kT \gamma \phi)} \]  \hspace{1cm} (Eq. 18)

As \( \sigma C_0/X_i \) value for Fe is the lowest among the solute atoms, diffusion of Fe should be a rate limiting step.

From the literature:

\[ D_{fe} = 4.1 \times 10^{-13} \exp(-3300/RT) \text{ (m}^2/\text{s}) \]  \hspace{1cm} (Ref. 84);

\[ \gamma = 0.3 \text{ J} \]  \hspace{1cm} (Ref. 39);

\[ C_0 = 10^{-5} \text{ at } 425 \degree C, 0.000082 \text{ at } 500 \degree C, 0.000154 \text{ at } 575 \degree C, 0.0026 \text{ at } 655 \degree C \text{ (in atomic fraction)}; \]

\[ X_{fe} = 0.174 \text{ for } Al_{13}(Fe, V)_3Si \text{ in FVS0812}; \]

\[ \phi = 1.435 \times 10^{-29} \text{ m}^3/\text{Fe atom} \]

\[ K(\phi) = K(0.24) = 2.18 \]  \hspace{1cm} (Ref. 34, 35).

Then the equation becomes:

\[ K_3(0.24) = 4.03 \times 10^{-11} \exp(-16.2 \times 10^3/T) C_0/T. \]  \hspace{1cm} (Eq. 19)

For growth by grain boundary diffusion,

\[ K_4 = \frac{(9/64) (C_0 D_{gb} \gamma \phi^2)}{(XYkT)} \]  \hspace{1cm} (Eq. 20)

From the literature,
\[ D_{gb} = 5.0 \times 10^{-6} \exp\left(-58,000/RT\right) \text{ (Ref. 84)} \]

\[ w = 5 \times 10^{-10} \text{ m} \]

Taking \( \gamma_{gb} = 1 \text{ J/m}^2 \) and \( \gamma = 0.3 \text{ J/m}^2 \) (Ref. 37)

\[ X = 1.62 \]

From the measurements, assuming six dispersoids at each grain boundary with grain size of 4.5 \( \mu \text{m} \),

\[ Y = (1/2)(n/f) = 2.23 \]

and the equation becomes:

\[ K_4 = 6.13 \times 10^{-24} \exp\left(-6,976/T\right)C_0(T)/T. \] (Eq. 21)

For growth by dislocation pipe diffusion:

\[ K_5 = (4/5)^5(5/4\pi)(zqD_pq^2C_0)/kT. \] (Eq. 22)

From Gjostein (Ref. 86), \( D_p = 0.3 \times 10^{-4} \exp\left(-17.8T_m/T\right) \text{ (m}^2/\text{s}) \)

and taking \( T_m = 928 \text{ °K} \),

\[ D_p = 0.3 \times 10^{-4} \exp\left(-11,659/T\right) \]

\[ q = (5b)^2 = 6.4 \times 10^{-19} \text{ m}^2. \]

With other terms taken from the previous calculations,

\[ K_5 = 1.46 \times 10^{-30} \exp\left(-11,659/T\right)C_0(T)z/T \] (Eq. 23)

with \( z \) assumed to be 10.

From the measurements, 78 \% of dispersoids were located inside the alpha grains and 22 \% were located at grain boundaries. Calculated average radii were derived by summing up the radii from each calculation by compensating the distribution. Calculated values of each case and measured values are compared in Table 12.
Table 12: Comparison of measured and calculated dispersoid size for FVS0812 IFW isothermally exposed at 425 °C

<table>
<thead>
<tr>
<th>Time (hrs.)</th>
<th>VDC ((r_3))</th>
<th>gb diffusion ((r_4))</th>
<th>disl. pipe diffusion ((r_5))</th>
<th>(\bar{r}) (calculated)</th>
<th>(\bar{r}) (measured)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0</td>
<td>4.85</td>
<td>4.85</td>
<td>4.85</td>
<td>4.85</td>
<td>4.85</td>
</tr>
<tr>
<td>100</td>
<td>5.08</td>
<td>4.96</td>
<td>4.87</td>
<td>5.21</td>
<td>5.75</td>
</tr>
<tr>
<td>200</td>
<td>5.26</td>
<td>5.13</td>
<td>4.88</td>
<td>5.57</td>
<td>6.00</td>
</tr>
<tr>
<td>300</td>
<td>5.46</td>
<td>5.26</td>
<td>4.91</td>
<td>5.91</td>
<td>-</td>
</tr>
<tr>
<td>500</td>
<td>5.82</td>
<td>5.49</td>
<td>4.92</td>
<td>6.59</td>
<td>6.20</td>
</tr>
</tbody>
</table>

\(\bar{r}(\text{calculated}) = 0.78 (r_3 + r_5 - r_0) + 0.22 r_4\)

As shown, volume diffusion contributed most to coarsening followed by grain boundary diffusion. This result is reasonable considering the small grain size which connects the dispersoids. Initial rapid growth may be due to the bimodal distribution of dispersoids, which caused rapid dissolution of dispersoids smaller than critical radius. Also, a reduction of dislocation density and slight grain growth might have contributed.

For the heat treatment experiments above 500 °C, measured values were much smaller than predicted, which is due to the presence of acicular Al\(_{13}(\text{Fe,V})_4\) dispersoids.
5.5.2. Analysis of the behavior of friction welds on elevated temperature exposure

For the elevated temperature experiments of high axial force IF weld in FVS0812 at 425 °C, coarsening in the interface region was higher than in the base metal. The alpha grain size in the interface region was almost equal to that of base metal showing an average of 0.45 μm. The dislocation density in this region was also not higher than in the base metal.

From the LSW theory, the critical radius is given (Ref. 30): \[ r_c = \left( \frac{1}{C_m - C_0} \right) \left( \frac{2\sigma V_0 C_0}{RT} \right) \] (Eq. 24)

where \( r_c \) = critical radius;
\( C_m \) = concentration of solute in the matrix away from the particle;
\( C_0 \) = equilibrium concentration of solute for a planar particle in the matrix;
and other terms carry same meaning in the previous sections.

The particles smaller than \( r_c \) shrink by diffusion of solute atoms to the particles larger than \( r_c \). At steady state average radius \( r \) becomes \( r_c \). As \( r \) increases \( C_m \) approaches \( C_0 \) and \( r_c \) increases.

The base metal of RS/PM FVS0812 was composed of two different kinds of ribbon particles of even-sized
dispersoids in each particle. Approximately 70% of the particles were comprised of small dispersoids with average diameter of 65 nm and 30% of the particles were comprised of large dispersoids with average diameter of 169 nm. At steady state, the growth of dispersoids follows the general description explained in section 1.3. However, it should behave differently at non-steady state. In the base metal, with even-sized dispersoids, the distribution curve is located inside the steady state distribution curve. In this case, particle growth occurs by simply spreading the distribution curve, i.e. the particle smaller than \( r \) shrinks and the particles larger than \( r \) coarsens but the \( r \) does not change until steady state condition is achieved. As a result, the initial growth rate should be smaller than at steady state. On the contrary, when the two particles are mixed together, as in the interface region of the inertia-friction weld, the distribution curve lies outside the steady state distribution curve. In this case, smaller dispersoids outside the steady state curve dissolve and rapid coarsening is expected. The rapid coarsening at the interface can be explained by the above.

Even with the coarsening in the interface region, hardness changes in the interface region were stable after 200 hours, showing higher values than the base metal. Due to the counting of dispersoids from the projected image,
estimation of population density was not possible but it is concluded the higher hardness was attributed to the dispersion of agglomerated dispersoids in the base metal.

For the elevated temperature exposure experiment above 500 °C, the smaller coarsening rate than predicted value is attributed to the presence of $\text{Al}_3(\text{Fe,V})_4$ or $\text{Al}_3(\text{Fe,V})$ intermetallics, which could not be counted due to their scarce distribution. With the presence of these acicular dispersoids, the brittle structure may have no practical use.

Summarizing, the elevated temperature exposure experiment at 425 °C showed relatively smaller dispersoid coarsening and negligible degradation of mechanical properties up to 500 hours.
CHAPTER VI
CONCLUSIONS

Similar- and dissimilar-alloy friction welds in FVS1212, FVS0812, Al-9Fe-3Mo-1V, 2024-T351 and titanium alloys were produced using the inertia-friction and linear-friction welding processes. Relationships between the process variables, microstructures, mechanical properties and fracture behavior were determined. Specific conclusions of the work are described below.

1. Process characteristics
1.1. Both the inertia- and linear-friction welding processes were demonstrated to be effective for the similar-alloy welding of RS/PM Al-Fe-X alloys as well as the dissimilar-alloy welding of the RS/PM Al-Fe-X alloys to the IM aluminum 2024-T351.
1.2. Inertia-friction welds produced with optimum parameters exhibited less structural coarsening and degradation of mechanical properties for both the similar- and dissimilar-alloy welds compared to nominally-optimized linear-friction weld. However, linear-friction welds showed a more homogeneous microstruc-
ture radially across the weld interface.

2. Correlations between microstructure/mechanical properties/fracture characteristics in similar-alloy weld of FVS1212 and FVS0812

2.1. For the similar-alloys welds, mechanical properties were influenced principally by coarsening of dispersoids. Fracture of dispersoids was not observed and the original $\text{Al}_{12}(\text{Fe,V})_3\text{Si}$ dispersoid structure was retained during welding. The recrystallized alpha grain size in the weld zone was comparable to that of the base metal.

2.2. Dispersoid coarsening in the FVS1212 welds was greater than in the FVS0812 welds for the same welding parameters and was attributed to the higher volume percentage of dispersoids in the former alloy.

2.3. Dispersoid-lean bands were observed in the IF welds produced using low axial force for both the FVS1212 and FVS0812 welds due to the local inhomogeneous deformation at the weld interface.

2.4. Hardness, tensile strength and fracture behavior correlated well with microstructural observations, showing minimum hardness and tensile strength at the dispersoid-lean band in low axial force inertia-friction welds. Minimal coarsening in high axial force inertia-friction weld resulted in an increase in
hardness and higher tensile strength. Fracture occurred in the inner-HDZ of the welds, and fracture topographies correlated well with the flow pattern experienced by the weld interface during welding.

3. Correlations between microstructure/mechanical properties/fracture characteristics in dissimilar-alloy welds between Al-Fe-X alloys and IM-2024-T351

3.1. Dispersoid coarsening and alpha grain growth in RS/PM Al-Fe-X region was negligible. Fracture of large dispersoids in the Al-9Fe-3Mo-1V alloy was observed but absent in FVS1212.

3.2. A strength reversal on heating between the Al-9Fe-3Mo-1V and 2024-T351 resulted a unique interface curvature with an increase in axial force, while the higher strength of FVS1212 compared to 2024-T351 over the entire temperature range promoted a straight interface geometry.

3.3. Joint strength was strongly affected by the interface structure between the two materials and microstructural changes in the 2024-T351. For IF welds produced between Al-9Fe-3Mo-1V and 2024-T351, fracture occurred in the inner-HDZ of the weaker Al-9Fe-3Mo-1V, with an exception of outer periphery of low axial force weld where mechanical mixing was negligible. For the dissimilar alloy welds between FVS1212 and 2024-T351,
tensile fracture occurred primarily in weaker 2024-T351.

4. Dissimilar-alloy welding between dispersion-hardened RS/PM Al-Fe-X alloys and titanium alloys

Large strength differences and presence of a high volume percentage of intermetallic dispersoids preclude obtaining high integrity joints between the two alloy groups. Dispersoid-rich region contacting the titanium, which was originated from non-uniform deformation during welding, attributed to the above.

5. Behavior of friction welds on elevated temperature exposure

5.1. Elevated-temperature exposure experiments performed at 425 °C on FVS0812 IF welds showed minimal dispersoid coarsening and negligible degradation of mechanical properties up to 500 hours.

5.2 At 425 °C, the coarsening of the dispersoids occurred principally by the volume diffusion control mechanism, followed by grain boundary mechanism.

5.3. Exposure above 500 °C resulted in the formation of acicular $\text{Al}_3(\text{Fe},\text{V})_4$ or $\text{Al}_3(\text{Fe},\text{V})$ intermetallics distributed uniformly across weld region and base metal, which degraded mechanical properties.
CHAPTER VII
RECOMMENDATIONS

The following investigations are recommended for future research.

1. Weld parameter optimization

From the results of this investigation, minimizing the dispersoid coarsening by removing coarsened materials in the interface region and reducing the weld thermal cycle with higher axial force were found to be beneficial. As most of the RS/PM Al-Fe-X alloys are produced in the extruded form, the abrupt change of this extrusion texture with increase in axial force may result in harmful effect on toughness and elongation of the joints. The application of minimum energy and appropriate weld axial force will improve the tensile properties of the friction welds of extruded RS/PM Al-Fe-X alloys.

2. Dissimilar-alloy welding of RS/PM Al-Fe-X alloys to Ti alloys

The main difficulties in dissimilar-alloy welding
between RS/PM Al-Fe-X alloys and titanium alloys resulted from the large strength difference and the presence of a high volume percentage of intermetallic dispersoids at the interface. Considering that dissimilar-alloy welding between IM-2024-T351 and Ti-662 was possible using a restraint ring, strength difference can be overcome. The only remaining problem is the presence of intermetallic dispersoids. Applying an intermediate transition layer of ingot metallurgy aluminum alloys and restraint ring on the aluminum side may solve this difficulty.

3. Numerical analysis of weld thermal cycle and mechanical behavior and correlation to the weld microstructure.

Numerical analysis of the weld thermal cycle for continuous drive friction welding and inertia-friction welding were reported on well-known alloys such as mild steel. Data on the physical properties of Al-Fe-V-Si and Al-Fe-Mo-V alloys were not available during this investigation. Correlation between the thermal responses and microstructural changes in the weld region will further clarify the behavior of materials during the welding process. Numerical analysis (either by finite element method (FEM) or finite differential method (FDM)) of linear-friction welding is strongly recommended.
APPENDIX

Evaluation of dispersoid size distribution
by diameter analysis

General expression for the observed particle distribution in a projected image is (Ref. 88):

\[ N(d) = tN(D)_{\text{proj}} + \mu(d) - M(d) \]

where \( N(d) \) = number of separate circular images per unit area of projection plane with diameter \( d \pm d/2 \).

\( N(D) \) = number of spherical particles per unit volume with diameter \( D \pm D/2 \).

\( \mu(d) \) = number of circular intersections per unit area of two-dimensional sections with diameter \( d \pm d/2 \).

\( M(d) \) = number of circular images per unit area of projection plane with diameter \( d \pm d/2 \) lost by overlap.

\( t \) = thickness of the foil.

When \( t/D \gg 1 \), \( \mu(d) \) becomes negligible. Also, when the overlapped images are distinguishable, \( M(d) \) becomes 0. As a result, \( N(d) = tN(D)_{\text{proj}} \) for the above case. The measured value \( N(d) \) can be directly related to \( N(D) \). Even without thickness measurement, normalized \( N(d) \) values can be used for comparison of dispersoid distribution function.
REFERENCES


