MICROSTRUCTURAL EVOLUTION IN FRICTION STIR WELDING OF

Ti-6Al-4V

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ABSTRACT

Friction stir welding (FSW) has been commercialized for certain applications of aluminum alloys. Ongoing research efforts are focusing on extending FSW to other high temperature materials such as steel, nickel alloys and titanium alloys. This study is a part of such an effort and looks into microstructural evolution in FSW of Ti-6Al-4V.

Friction stir welds were produced in both mill annealed and β annealed Ti-6Al-4V, followed by microstructural characterization, microhardness testing and texture analysis. Both the welds evaluated were interrupted welds, i.e. travel of the tool was stopped before the welding of the plates was complete. This facilitated the observation of microstructure around the point of tool removal. Microstructural characterization was done using SEM and texture analysis was done using OIM (Orientation Imaging Microscopy) technique. Residual stress analysis was conducted and compared to those in single pass fusion welds in Ti-6Al-4V.

Macroscopic examination revealed a symmetric stir zone where the tool had completely stirred the material and a thermomechanically affected zone (TMAZ) was observed between the stir zone and the base material. Macroscopic examination of the longitudinal sections revealed that a region similar to the stir zone existed in a small region ahead of the weld, near the point of tool removal. The joint penetration in the welds was incomplete.
The stir zone of both the welds consisted of colony $\alpha+\beta$ in small prior $\beta$ grains (~10 $\mu$m) in size. The stir zone of the mill annealed welds also consisted of discretely scattered equiaxed $\alpha$ particles. The microstructures indicate that recrystallization occurred above the $\beta$ transus and the dwell time above the $\beta$ transus was short. The TMAZ of both the welds consisted of deformed base material and a “microstructurally distinct” band. A microstructurally distinct HAZ was not observed in either of the welds. Some grain growth was observed at the top of the weld due to the frictional heating from the shoulder of the tool. An increase in hardness was observed in the stir zone of both the welds that could be attributed to the small grain size in the stir zone. Microhardness observation in the longitudinal sections revealed that hardness remained uniform within the bulk of the SZ, although some reduction in hardness occurred at the top of the weld due to grain growth.

Texture analysis revealed that the texture in the stir zone had changed due to recrystallization. The Burgers orientation relation between the $\alpha$ and $\beta$ phases was observed in the stir zone and indicating that the material was not deformed after the nucleation of the $\alpha$ phase. The randomness of the texture indicates that the cooling rate from the $\beta$ transus was relatively rapid. Some variant selection was also observed. Microtexture observation in the TMAZ revealed that the texture in the various regions of the TMAZ was different than either the base material or the stir zone. However, the deformed grains in the TMAZ of the mill annealed welds showed similar basal pole texture to the base material. This indicates some grain rotation mechanism about the basal poles in that region of the weld.
The magnitude of residual stresses in friction stir welds was less than that in single pass fusion welding and the nature of stresses was also different. The level of cold work in the stir zone of the friction stir welds was very low due to recrystallization. Residual stress analysis was not possible in the multipass fusion welds due to the large grain size in the fusion zone.

The fine grained microstructure and minimization of HAZ observed in the friction stir welds is highly desirable, compared to the coarse grained microstructure exhibited when fusion welding Ti-6Al-4V. The texture analysis provides valuable information about the deformation mechanism. The magnitude of residual stresses in FSW is less than that in single pass and multipass fusion welding. From a microstructure control standpoint, friction stir welding appears to be a promising welding technique for Ti-6Al-4V.
This thesis is dedicated to my family
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CHAPTER 1

INTRODUCTION

Friction stir welding (FSW) was patented in 1991 (1). Since then, it has evolved to become one of the most promising welding techniques for certain applications. Through extensive research on processing and metallurgical aspects of FSW aluminum alloys, it has matured to a stage where it can be successfully applied to commercial applications of aluminum alloys. FSW as a solid state joining process offers solutions to problems related to fusion welding of materials. As demonstrated for aluminum alloys, FSW produces a refined grain size and low residual stresses. High level of residual stresses and coarse β grain size are important issues during the welding of titanium alloys. Naturally, all these advantages have motivated people to extend FSW to high temperature alloys based on titanium, nickel and iron. However, extending friction stir welding to high melting point materials such as titanium alloys poses many challenges. Identifying suitable tool material and determining the tool wear and deformation characteristics is necessary for producing uniform and reproducible welds in titanium alloys. Robust welding systems have to be developed in which parameters and atmosphere can be easily controlled since titanium reacts easily at high temperatures and because processing
Ti-6Al-4V above or below the β transus temperature produces different microstructures and properties. Characterizing the weld microstructures and understanding the microstructural evolution is important for controlling the weld properties and improving weld performance. Development of suitable postweld heat treatments is necessary if as-welded properties are not satisfactory or the level of residual stresses generated is high. Since Ti-6Al-4V represents the most important structural titanium alloy used worldwide, this study focuses on characterizing the weld microstructures and understanding the microstructural evolution during friction stir welding of Ti-6Al-4V.

The goal of this research is to characterize the weld microstructure and understand microstructural evolution in FSW of Ti-6Al-4V as a foundation for future microstructure property relationship studies. FSW of Ti-6Al-4V was studied with two starting conditions; mill annealed and β annealed. Microstructural characterization using SEM was done to study the microstructure generated in the various regions of the weld and explain the underlying deformation mechanism. The microstructures generated were used to hypothesize metallurgical behavior relative to the β transus and the dwell time in the deformation temperature range. Processing above or below the β transus generates completely different microstructures and as such has a critical influence on two phase material such as Ti-6Al-4V. In Ti-6Al-4V, microstructural and texture evolution are not independent of each other and also the fact that texture has a great bearing on properties of anisotropic material such as Ti-6Al-4V (2). Thus micro and macro texture analysis was done using Electron Backscatter Diffraction technique and was correlated to the microstructural evolution. Residual stress analysis was carried out to evaluate the
magnitude of residual stresses. This analysis was important since residual stresses have an important bearing on fatigue properties and for any future study of post weld heat treatment of Ti-6Al-4V, knowledge of residual stresses is necessary. The level of stresses was compared to those generated in single pass fusion welds (using high penetration flux) and multipass fusion welds to evaluate the advantage of using friction stir welding in terms of residual stresses.
CHAPTER 2

LITERATURE REVIEW

2.1 Metallurgy of Ti-6Al-4V

Titanium is a transition metal with melting point of 1668°C and a useful temperature range for structural applications of 425-600°C (3). Pure titanium exists in a hexagonal closed packed (HCP) crystal lattice structure called the $\alpha$ phase up to 885°C. Above 885°C, it undergoes a phase transformation from hexagonal closed packed structure to a body centered cubic structure called the $\beta$ phase. The transformation temperature known as the ‘$\beta$ transus’, changes with the addition of alloying elements. The nature of this transformation depends on the thermo-mechanical treatment of the metal, composition, and component thickness. Depending on the relative amount and morphology of the phases, titanium alloys are classified as $\alpha$, near $\alpha$, $\alpha-\beta$ and metastable $\beta$ phase alloys. Ti-6Al-4V is an $\alpha-\beta$ alloy in which the main alloying elements are aluminum (6 wt %) and vanadium (4 wt %). Ti-6Al-4V represents the maximum tonnage of titanium alloys used worldwide because it has a good combination of strength, toughness and corrosion resistance. It has wide applications ranging from pressure vessels, aeroengine-turbine and compressor blades, to surgical implants and sporting goods.
2.1.1 Effect of Alloying Elements

Thermal dissociation of titanium tetraiodide produces the highest purity titanium called 'iodide titanium (4). Various alloying elements can be added to produce a material with predetermined properties. Alloying elements can be primarily classified as: 1) $\alpha$ stabilizing 2) $\beta$ stabilizing and 3) neutral.

$\alpha$ Stabilizers: These elements increase the $\alpha$-$\beta$ transformation temperature. The main $\alpha$ stabilizers are aluminum, tin, zirconium, nitrogen, carbon and oxygen. Nitrogen, carbon and oxygen dissolve interstitially in titanium and affect the mechanical properties (5). Addition of oxygen and nitrogen increases the strength at the expense of toughness (6). Addition of carbon up to 0.3wt% C increases the strength with a slight reduction in ductility (6). Addition of aluminum up to 8 wt% increases the strength without reduction in ductility. Above 8%wt, Al additions result in the formation of Ti$_3$Al which can cause embrittlement (6, 7). Tin is also used as an $\alpha$ stabilizer and solid solution strengthener.

$\beta$ Stabilizers: These elements stabilize the $\beta$ phase i.e., depress the $\alpha$–$\beta$ transformation temperature. Molybdenum, niobium, tantalum, vanadium and hydrogen are $\beta$ isomorphous elements; i.e., they are completely soluble in $\beta$ titanium and do not form any intermediate compound (8). Due to its small size, hydrogen has high solubility in titanium and precipitates as a hydride that significantly reduces the ductility (9). Titanium forms eutectoid systems with chromium, iron, copper, nickel, palladium, cobalt, manganese and certain other transition elements.
Neutral elements: Zirconium and hafnium are isomorphous with both the \( \alpha \) and \( \beta \) phases and do not affect the \( \alpha-\beta \) transformation temperature significantly.

2.1.2 Phase Transformations in Ti-6Al-4V

Ti-6Al-4V is an \( \alpha-\beta \) alloy in which the alloying additions are such that both the \( \alpha \) and \( \beta \) phases are present at room temperature. The mechanical properties of Ti-6Al-4V are related to the transformation kinetics and morphology of the phases present. The Figure 2.1 shows a Ti-6Al-4V ternary phase diagram. Aluminum stabilizes the \( \alpha \) phase and segregates to the \( \alpha \) phase and similarly vanadium being body centered cubic, stabilizes the \( \beta \) phase and segregates to the \( \beta \) phase. Isothermal sections of the Ti-6Al-4V ternary phase diagram are shown in the Figure 2.1 with the composition corresponding to 6%Al, 4%V shown by a dot. The dashed lines represent the tie lines used to determine the amount of alloying element in each phase and the relative percentage of each phase present. On cooling from 1000°C to 800 °C, the Al content of \( \alpha \) remains the same while the V content of \( \beta \) increases to about 14% at 800 °C. By using the lever rule, it can be seen in the phase diagram that the amount of \( \alpha \) increases upon cooling.

When this alloy is slowly cooled from the \( \beta \) region below the \( \beta \) transus (980°C), \( \alpha \) begins to form as plates with a crystallographic relationship to the \( \beta \) from which it forms (10-15). The basal plane of the \( \alpha \) plates (0001) is parallel to the \{110\} planes of the \( \beta \). It can be seen in the Figure 2.2 that the microstructure consists of parallel plates of \( \alpha \) with \( \beta \) between them. On cooling slowly, the plates thicken along a
Figure 2.1 Isothermal section of Ti-rich region of the Ti-Al-V phase diagram at 1000, 900, 800 °C (16)
crystallographic direction due to close atomic matching along that plane. This microstructural morphology consisting of parallel plates of $\alpha$ in a $\beta$ matrix is called Widmanstatten structure. The $\beta$ phase can also decompose by a martensitic reaction on rapid cooling. The martensitic product has a hexagonal closed packed (hcp) structure and is designated as $\alpha'$. Alloys containing higher concentration of elements such as Mo, Ta or Nb form an orthorhombic martensite which is designated as $\alpha''$. The martensitic plates also follow the same crystallographic relationship with the $\beta$ as the $\alpha$ plates. Increasing the alloying content depresses the martensite start ($M_s$) temperature below 25 $^\circ$C. Thus, some metastable $\beta$ can also be present after quenching in Ti-6Al-4V. Heterogeneous nucleation of grain boundary $\alpha$ is also observed at prior $\beta$ grain boundaries in these alloys.

The $\alpha$ phase globularizes due to recrystallization when mechanical deformation is superimposed on cooling (2). Depending on the working operation, $\alpha$ phase aspect ratios ranging from $\sim$30 in colony structure to $\sim$1 in recrystallized structures can be achieved. Thus, thermomechanical processing can be used to generate a variety of microstructures in Ti-6Al-4V. Table 2.1 shows a list of standard heat treatments used for Ti-6Al-4V and describes the corresponding microstructures generated. Table 2.2 shows the mechanical properties associated with the various heat treatments. It can be clearly seen that heat treatment has great impact on mechanical properties of Ti-6Al-4V.
Figure 2.2 Widmanstatten structure achieved by slow cooling from the $\beta$ transus (5)
<table>
<thead>
<tr>
<th>Heat treatment designation</th>
<th>Heat treatment cycle</th>
<th>Microstructure</th>
</tr>
</thead>
<tbody>
<tr>
<td>Duplex anneal (DA)</td>
<td>Solution treat at 50-75°C below T&lt;sub&gt;B&lt;/sub&gt;(a), air cool and age for 2-8 h at 540-675°C</td>
<td>Primary α, plus Widmanstätten α + β regions</td>
</tr>
<tr>
<td>Solution treat and age (STA)</td>
<td>Solution treat at ~40°C below T&lt;sub&gt;B&lt;/sub&gt;, water quench(b) and age for 2-8 h at 535-675°C</td>
<td>Primary α, plus tempered α' or a β + α mixture</td>
</tr>
<tr>
<td>Beta anneal (BA)</td>
<td>Solution treat at ~15°C above T&lt;sub&gt;B&lt;/sub&gt;, air cool and stabilize at 650-760°C for 2 h</td>
<td>Widmanstätten α + β colony microstructure</td>
</tr>
<tr>
<td>Beta quench (BQ)</td>
<td>Solution treat at ~15°C above T&lt;sub&gt;B&lt;/sub&gt;, water quench and temper at 650-760°C for 2 h</td>
<td>Tempered α'</td>
</tr>
<tr>
<td>Recrystallization anneal (RA)</td>
<td>925°C for 4 h, cool at 50°C/h to 760°C, air cool</td>
<td>Equiaxed α with β at grain-boundary triple points</td>
</tr>
<tr>
<td>Mill anneal</td>
<td>α + β hot work + anneal at 705°C for 30 min to several hours and air cool</td>
<td>Incompletely recrystallized α with a small volume fraction of small β particles</td>
</tr>
</tbody>
</table>

Table 2.1 Heat treatments for Ti-6Al-4V. (2)
<table>
<thead>
<tr>
<th>Condition</th>
<th>Yield strength, MPa</th>
<th>Tensile strength, MPa</th>
<th>Elongation, %</th>
<th>Reduction in area, %</th>
<th>Ref</th>
</tr>
</thead>
<tbody>
<tr>
<td>α + β forge + recrystallization anneal</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>(RA) (a)</td>
<td>711</td>
<td>876</td>
<td>12.4</td>
<td>36</td>
<td>80</td>
</tr>
<tr>
<td>α + β forge + mill anneal (minimum values)</td>
<td>828</td>
<td>897</td>
<td>10.0</td>
<td>25</td>
<td>82</td>
</tr>
<tr>
<td>α + β forge + STA (age 4 h at 594°C)</td>
<td>876</td>
<td>938</td>
<td>15.2</td>
<td>34</td>
<td>80</td>
</tr>
<tr>
<td>α + β forge + STOA (age 24 h at 594°C)</td>
<td>904</td>
<td>973</td>
<td>15.5</td>
<td>47</td>
<td>80</td>
</tr>
<tr>
<td>β forge + AC {BA} (+ 705°C/2 h/AC)</td>
<td>773</td>
<td>856</td>
<td>11.2</td>
<td>23</td>
<td>80</td>
</tr>
<tr>
<td>β forge + WQ {BQ} (+ 705°C/2 h/AC)</td>
<td>863</td>
<td>932</td>
<td>5.9</td>
<td>6</td>
<td>80</td>
</tr>
<tr>
<td>α + β forge DA</td>
<td>856</td>
<td>911</td>
<td>15.3</td>
<td>47</td>
<td>80</td>
</tr>
</tbody>
</table>

Table 2.2 Tensile properties of Ti-6Al-4V (2)
2.2 Welding Metallurgy of Ti-6Al-4V

Titanium has a strong chemical affinity for oxygen. The affinity for oxygen and the thickness of oxide layer increases with temperature. At temperatures above 500°C, Ti-6Al-4V becomes highly susceptible to embrittlement by oxygen, nitrogen and hydrogen. This has a detrimental effect on the mechanical properties (17,18). Therefore, careful cleaning procedures and an inert atmosphere are used to fusion weld Ti-6Al-4V. Processes using filler metal are usually performed with matching composition filler (10). Welding of Ti-6Al-4V by flux-based processes (submerged arc, flux-cored, electroslag) has not been proved to be economical due to the need for costly and high purity fluoride-based fluxes (10).

Ti-6Al-4V has good weldability compared to other structural titanium alloys. This is due to two reasons; the α' martensite formed is not as hard and brittle as that exhibited by other α-β or β alloys, and Ti-6Al-4V exhibits relatively low hardenability, which allows the formation of desirable Widmanstatten α and retained β structures at the high cooling rates exhibited by welds (4). Four microstructural characteristics should be considered in fusion welds: 1) β grain structure 2) macrosegregation 3) microsegregation and 4) formation of solidification defects (17, 18).

All Ti-6Al-4V fusion welds solidify as single phase β. The β grain structure depends on; 1) the thermal cycle in the near HAZ, because the growth and size of β grains in fusion zone depends on it due to epitaxial nucleation on the HAZ grains. 2) the shape of the weld pool, as it influences the competitive grain growth process. The β grains in fusion zone grow epitaxially on coarsened β grains in the near HAZ. According to the
principles of competitive growth developed by Savage, grains with preferred growth direction (100 in bcc) favorably oriented with the temperature gradient grow at the cost of less favorably oriented grains (19). This promotes the formation of a coarse, columnar fusion zone structure. In Ti-6Al-4V multipass welds, grains grow epitaxially across the total width of fusion zone. However, these grains then continue to grow across the fusion lines during subsequent weld passes without nucleating new grains and thus result in long columnar grains. Grain boundaries so formed provide a continuous crack path. In the centerline of thin sheet welds of Ti-6Al-4V, formation of single through thickness β grains is also observed. Thus several methods for grain refinement such as inoculation with heterogeneous nucleants (20, 21), surface nucleation induced by gas impingement (22), introduction of physical disturbance through techniques such as electromagnetic stirring (23, 24) and pulsed welding currents (25-29) have been tried.

The β grain growth in the near heat-affected zone occurs due to heat flow from the weld into the base material. The peak temperature in this region can range from the alloy solidus to the β transus. The width of this zone depends on heat input, i.e. it is very narrow for electron beam welding (EBW) and laser beam welding (LBW) and very wide in the case of gas tungsten arc welding (GTAW). Temperatures below the β transus are encountered in the far HAZ, which cause various solid-state transformations depending on the base metal microstructure. The presence of α phase at peak temperatures prevents grain growth as in fusion and heat-affected zone and thus improves ductility in this region. The solid state phase transformations taking place after solidification also have a great influence on the weld properties (30). The Figure 2.3 shows time-temperature-transformation diagram for Ti-6Al-4V.
Figure 2.3 TTT diagram for Ti-6Al-4V. Solution annealed at 1020 °C and quenched directly to reaction temperature (31)
Compared to other titanium alloys and structural materials, degradation of weld integrity due microsegregation and macrosegregation effects is not observed in Ti-6Al-4V fusion welds. Ti-6Al-4V fusion welds are also not susceptible to solidification cracking and liquation cracking in the HAZ. This is attributed to the limited segregation of elements and diffusional homogenization of alloying elements during weld cooling through the β phase field (30). However, under conditions of severe restraint, cracking can occur along the columnar β grain boundaries (32, 33). Porosity can also present a solidification problem in Ti-6Al-4V and also has a negative influence on fatigue (34).

Fusion welding of Ti-6Al-4V does not have an adverse effect on the strength in the fusion zone (17, 18). On the contrary, high yield strength is observed in the fusion zone due to the presence of martensite and a fine acicular microstructure (35). The strength observed varies with the cooling rate. Fast cooling produces fine microstructures and high strength while slow cooling coarsens the microstructure and makes intergranular slip easy and reduces strength (17). Reduction in strength is observed in the HAZ due to coarsening of the microstructure (36). Ti-6Al-4V fusion welds show low tensile ductility in the fusion zone due to the large prior β grain size, acicular and at least partially martensitic matrix (37). Ductility cannot be improved by altering the heat input, since increasing the heat input increases the β grain size and decreasing the heat input results in faster cooling and increases the aspect ratio of the α laths(38). Welding procedures that produce a lamellar microstructure and no grain boundary α, show very good fracture toughness in the weld metal (36). Structures with a large percentage of acicular α show good toughness. In the case of transgranular fracture, these platelets
with their high aspect ratios provide extended α-β interfaces for preferential crack
propagation (39). In the case of intergranular fracture, structures show high toughness
when the fracture path length is increased as the cracks follow prior β grain boundaries
(40). The thickness of grain boundary α is critical when transgranular fracture is
observed. Increasing the thickness of grain boundary α increases the toughness (41).
Thus, welding procedures and post-weld heat treatments should be chosen to achieve the
microstructure resulting in desired properties.

Most of the titanium alloys are postweld heat treated for the following reasons: 1)
relieving residual stresses 2) modifying the weld zone microstructure and 3) stabilizing
the microstructure for weld structural applications at elevated temperature (17). Postweld
heat treatment should be carefully chosen, since reduction in fracture toughness of weld
metal after PWHT has been observed in some cases (36).

2.3 Current Welding Techniques for Ti-6Al-4V

The various fusion welding processes commonly used for welding Ti-6Al-4V are gas-
tungsten arc welding (GTAW), gas-metal arc welding (GMAW), plasma arc welding
(PAW), laser-beam welding (LBW) and electron beam welding (EBW)(3). The details of
most widely used processes are discussed here.
2.3.1 Electron Beam Welding (EBW)

The EBW process is suitable for welding Ti-6Al-4V because high joint depths can be achieved due to the high energy density of the process (42). The high vacuum inside the chamber where the process is carried out shields the hot metal from contamination and prevents oxidation at elevated temperature (43). This process is used to weld Ti-6Al-4V plates ranging from 6 mm to 75 mm in thickness. Advantages of electron beam welding are high depth to width ratio, minimum joint preparation, narrow HAZ and low distortion. While the disadvantages are high capital equipment cost, high weld cooling rate and precise joint alignment.

2.3.2 Laser Beam Welding (LBW)

Similar to electron beam welding, laser beam welding is characterized by a low heat input and high energy density (44). The rapid solidification and quench associated with this process affects the microstructure and properties (45, 46). Residual stresses are very low due to the low heat input, distortion and shrinkage. LBW has limited applications for joining thick titanium plates since it requires very high power for joining thick plates.

2.3.3 Gas Tungsten Arc Welding (GTAW)

GTAW is the process in which the heat of welding is provided by an arc maintained between a nonconsummable tungsten electrode and the workpiece (3). GTAW welds can be produced manually and automatically. It is the most widely used process for joining Ti-6Al-4V, particularly in sheet form (47-49). GTAW can weld Ti-6Al-4V in
thicknesses from thousandths of an inch to more than several inches in single or multiple passes (50). A major drawback of the manual process is its low deposition rate. Welds in material up to 3 mm in thickness are autogenous and require no joint preparation, but for greater thickness a filler metal and joint preparation is required. The low deposition rate associated with the process requires increased number of passes per weld and causes greater distortion and chance of contamination. Various modifications have been introduced to improve productivity. Efforts have been made to improve the GTAW process be preheating the filler wire, use of ‘narrow groove’ and high current GTAW (50,51). The narrow groove technique in combination with the hot wire GTAW has been successfully employed to weld various titanium alloys. Efforts are also being made to apply keyhole GTAW for welding to titanium alloys (52). In keyhole GTAW, the peak arc pressure is increased and the process parameters are adjusted in such a way that a small opening is punctured through the root face of the weld. If the opening is kept small, the keyhole closes behind the arc and weld is completed. The opening releases the pressure within the arc and the process is found to be stable.

2.3.4 Gas Metal Arc Welding (GMAW)

GMAW is a process in which the heat of welding is provided by an arc maintained between a consumable electrode and the workpiece (3). The application of GMA welding is limited for Ti-6Al-4V alloys due to spattering and an unstable arc which reduces the weld quality and process efficiency (53-56).
2.3.5 Solid State Welding of Ti-6Al-4V

Solid state welding is defined as a group of processes that produce coalescence at temperature essentially below the melting point of the base material without the addition of a brazing filler metal (57).

Diffusion welding is a well-proven technique used to weld titanium alloys (58, 59). Solid state diffusion bonding of titanium involves the following steps: 1) development of intimate physical contact between the faying surfaces through deformation of surface roughness by yielding and creep at low pressures and high temperatures 2) formation of a metallic bond 3) diffusion across the faying surface and 4) grain growth across the original weld surface (1). It is necessary that clean surfaces be brought together for solid state joining since presence of oxide films on the surface results in low strength because oxide films are brittle and bonds between metals and oxides are weak (60). Titanium alloys are ideal for diffusion welding because of their high solubility for oxygen, which dissolves the surface oxides at high temperatures. The yield strength of Ti alloys decreases at high temperatures, thus facilitating solid state joining.

Friction welding is a group of solid-state processes that use intimate contact of a plastically deformed interface for joining materials (61). The suitable interface condition is generated by heat produced from relative motion of one component with other. To complete the weld, a forging force is used. There are three types of friction welding, depending on their geometry: rotary, orbital and linear. A notable limitation of rotary and orbital processes is that axial symmetry is required for at least one component to be welded (62). A variety of other solid state joining processes such as
flash-butt welding and explosive welding use a combination of heating of the material, pressure or relative mechanical motion to produce a joint (63).

2.4 Friction Stir Welding

Friction stir welding (FSW) was invented at The Welding Institute in 1991(1). It is defined as a solid state joining process, in which no melting occurs to join sheet and plate material (1). A rotating, non-consumable tool is plunged between the abutted edges of two plates and it travels the length of the joint leaving behind a high integrity solid-state weld. The Figure 2.4 shows a schematic illustration of friction stir welding. The tool has two distinct geometric features; a shoulder, which provides local frictional heating on the surface and a profiled pin that, protrudes from the shoulder and is responsible for the stirring action. The frictional heat evolved between the rotating shoulder and the top surface of the plate softens the material, and allows it to be plastically deformed or stirred. The extensive deformation allows the clean metal surfaces to come into contact and under the forging force and deformation, solid state bonding occurs. Other than microstructural evolution, current research on friction stir welding is focusing on modeling and analysis (64-67), damage tolerance of the welds (68-73), tool design, and welding parameters (74-77), etc.
Figure 2.4 Schematic showing friction stir welding and microstructurally distinct regions observed in a transverse section of the weld
2.4.1 Microstructural regions of a friction stir weld

There are three principle microstructural regions associated with friction stir welding; the stir zone (SZ) which is associated physically with the passage of the tool and where there has been massive plastic deformation; the thermo-mechanically affected zone (TMAZ) which is deformed but not associated physically with the passage of the tool and a heat affected zone (HAZ) where only heat is responsible for any changes if they occur. The weld has to accommodate a high degree of plastic deformation in the stir zone. It has been observed in aluminum alloys that the weld accommodates the high strain by a dynamic recrystallization mechanism (78). The TMAZ is a region of the weld that is between areas that are completely deformed (stir zone) and regions that are not plastically deformed (HAZ, base material). The change in structure in the TMAZ is due to the influence of heat and some plastic deformation. The HAZ is that region of the weld whose microstructure and properties have been altered by the heat of welding. The width and nature of HAZ depends on the peak welding temperature, time at the peak temperature and inherent properties of the material.

2.4.2 Current applications of friction stir welding

Compared to other solid state joining processes, relative motion of the work pieces is not needed so friction stir welding can be adapted for a wide range of applications and joint configurations like butt, lap, tee, edge and multiple lap joints (68). Low melting point alloys based on aluminum and copper are excellent candidates for friction stir welding because the temperatures required for plastic deformation can be achieved readily from the frictional heat generated by conventional steel tools. Friction stir
welding presents many advantages over conventional fusion welding processes such as elimination of cracking in fusion zone and heat affected zone, porosity, filler metals, shielding gases and weld preparation (77,79). In many cases, the mechanical properties of friction stir welds in the as-welded condition are comparable to those of the base materials and are equal to or superior to the properties of their fusion welded counterparts (80, 81). Friction stir welding of aluminum alloys has been successfully implemented in a variety of commercial applications. Aluminum structural components for aircraft, space vehicles, automotive and naval applications have been identified as those which can be manufactured economically using friction stir welding.

2.5 Texture in Ti-6Al-4V

2.5.1 Texture

Texture is a condition in which the crystal orientations in a polycrystalline aggregate is non random. Texture or preferred orientation can be developed during rolling, recrystallization, cold drawing, casting, electrodeposition, etc (82). A texture that reflects an average value obtained from large number of grains (typically thousands) is termed as macrotexture. Macrotecture measurement gives information on texture of the material over a large area, i.e., on the order of millimeters. This is critical for correlating the bulk properties to the texture. However measurement of macrotexture does not establish any direct connection between the microstructure and texture. The approach to texture, which deals with orientation statistics of a group of individual grains and also their spatial orientation, is known as microtexture (83). The knowledge of spatial location and the orientation of individual grains provide fused information about the crystallographic and
morphological aspects of the structure of a given material (83).

2.5.2 Representation of Texture

Pole figures give the relative intensity of orientation of a selected set of crystallographic planes or directions on an equal area projection with reference to the sample axis (84). If the grains have a completely random orientation, the poles will be distributed uniformly over the projection, as shown in the Fig 2.5 (a). If preferred orientation is present, the poles tend to cluster together as shown in the Fig 2.5 (b).

![Diagram](image)

Figure 2.5 (001) Pole figure showing random orientation in (a) and texture in (b) (82)
Figure 2.6 Euler angles with respect to the sample axis. (83)

Figure 2.7 Orientation map showing orientation of different grains.
OIM can be used to measure texture in bulk samples as well as small level. As shown in the Figure 2.6, Euler angles $\phi_1, \phi_2, \Phi$ are the angles by which the crystal needs to be rotated about the crystal axes $X, Y, Z$ respectively to bring it to the current orientation from an orientation in which all the sample and crystal axes are parallel. Complete orientation of any given crystal with respect to a fixed set of sample axes can be represented in terms of these Euler angles $\phi_1, \phi_2, \Phi$. The Figure 2.7 shows a typical orientation map. In orientation maps, regions with same Euler angles are mapped using the same color and showing correspondence with the microstructure (83). Thus, regions of different orientation can be distinguished.

2.5.3 Measurement of Texture Using Orientation Imaging Microscopy (OIM)

OIM can be used to measure texture in bulk samples. The specimen is placed in a scanning electron microscope (SEM) and tilted to an angle of 70° from the horizontal. The backscattered electrons released are captured on a phosphor screen. The sample is tilted to reduce the path length of the back-scattered electrons and thus improve the image contrast. The Figure 2.8 shows the geometry of backscattered diffraction. Background is measured over a larger area than the area of interest and is subtracted from the derived Kikuchi pattern to get the resultant Kikuchi pattern. With the help of software, orientation of the crystal at a given point is determined by analyzing the resultant Kikuchi pattern. OIM gives complete pole figures up to 90° from the center of the pole figure. Calibration and indexing of Kikuchi patterns depends on specimen to screen distance, specimen tilt and working distance in the camera. The system can be preprogrammed to determine the crystal orientations of large number of points in a single run. A beam scan in which the
stage remains fixed and the beam moves can be done to yield detailed orientation of a small scanned area (5μm² - 500 μm²). For larger areas (10 mm), stage scanning is done. In a stage scan, the beam remains stationary and the stage moves. OIM can give information about the orientation and misorientation between individual grains, two phases and other microstructural features. It can also give macrotexture information if stage scanning is done over a large area. It gives complete pole figures up to 90° from the center of the pole figure.

THE KIKUCHI DIFFRACTION PATTERN

![Diagram of Electron Backscatter Diffraction](image)

Figure 2.8 Electron Backscatter Diffraction (83)

2.5.4 Texture in Ti-6Al-4V

Evolution of texture in two-phase alloys such as Ti-6Al-4V is strongly dependent on morphology and volume fraction of the constituent phases (85). Both the α and β phases exhibit texture which is affected by a number of factors such as thermomechanical
processing, cooling rate, heat flow, stress fields, etc. Textures can be classified as
deformation or transformation textures, based on their origin. Since the base material for
friction stir welding was in the form of rolled sheet material, the literature survey focuses
on texture evolution in Ti-6Al-4V rolled sheet. As there is only one \{0001\} pole in an
hcp crystal, plotting the orientation of \{0001\} pole can represent the texture of the hcp \( \alpha \)
phase. However, complete description of texture requires determination of a second pole
figure. In Ti-6Al-4V, the prism pole (10-10) pole is usually used for this purpose,
because it is orthogonal to (0001) and it is an active slip plane in \( \alpha \) phase (16). The
Burgers orientation relationship, in which the close packed plane \{0001\}_\( \alpha \) is parallel to
closed packed \{110\}_\( \beta \), and the close packed direction \(<1\overline{1}20>\)_\( \alpha \) is parallel to the \(<\overline{1}1\overline{1}>\)_\( \beta \), is
exhibited between the \( \alpha \) and \( \beta \) phases (86). The Burgers orientation relationship between
\( \alpha \) and \( \beta \) does not hold true when the \( \alpha \) phase recrystallizes on working and or annealing
in the \( \alpha - \beta \) regime (87). The effect of recrystallization on the texture of \( \alpha \) phase is not
clear. One group of authors (87) who studied recrystallization of \( \alpha \) phase at 720 \( ^\circ \)C, have
observed a change in texture of \( \alpha \) phase due to recrystallization. In contrast, Lutjering
and coworkers (88) who studied deformation at 800-920 \( ^\circ \)C did not observe a change in
texture of \( \alpha \) phase due to recrystallization. It should be noted that rolling was used the
sole deformation mode by the above authors and the above results may not hold true for
other complex thermomechanical processes such as friction stir welding.
In general, two distinct basal textures are observed in rolled sheet material (2). The texture in which the basal poles are aligned with the sheet normal is referred to as a basal texture. The other texture, in which the basal poles lie in the plane of the sheet and are aligned with the transverse direction, is referred to as a basal transverse texture. A combination of these two textures can be obtained depending on thermomechanical processing schedule (temperature, time and strain profile) (2). The Figure 2.9 shows the different textures obtained by rolling Ti-6Al-4V at different temperatures. It can be seen that other than rolling at 900-930 °C, there is some texture in all the cases. If Ti-6Al-4V is annealed above the β transus and with no deformation, variant selection is not observed. However, if deformation is done above the β transus prior to β-α transformation, variant selection is observed (89). The selected variants have been linked to those of slip systems chosen among \{110\} <111> and \{112\} <111> which are strongly involved in the plastic deformation at high temperature prior to the transformation.

![Diagram of textures and temperature profiles]

Figure 2.9 Effect of rolling temperatures on texture in Ti-6Al-4V (2)
2.5.5 Significance of Texture in Ti-6Al-4V

Properties that are affected by texture are Young’s modulus, yield strength, Poisson’s ratio, ductility, toughness, and magnetic permeability. The study of texture is important in anisotropic materials such as titanium alloys. If a polycrystalline anisotropic material is textured, its anisotropy will cause a difference in properties along different crystallographic directions. If the material is randomly oriented, the properties will be averaged and anisotropy will be “cancelled out”. Numerous studies have shown the influence of texture on mechanical properties of Ti-6Al-4V (90-99). A change in texture in commercially rolled Ti-6Al-4V caused by subsequent heat treatment has been correlated with changes in tensile properties and fracture toughness (100, 101).

Deformation characteristics of Ti-6Al-4V sheets improve when \{0002\} pole shows edge-type texture (102). Studies have also shown significant effect of texture on smooth-bar fatigue life of Ti-6Al-4V (92, 94). Fatigue strength is greater when the stress axis coincides with the direction of a high density of basal poles.

2.7 Residual Stress in Ti-6Al-4V Welds

Residual stresses are stresses that exist in a body after all the external loads are removed. A weldment is subjected to complex and uneven thermal cycles. Fusion welds also undergo shrinkage and deformation during solidification and cooling. This produces complex stresses in the weld and surrounding region (3). The Figure 2.10 shows changes in temperature and the resulting stresses that occur during welding of bead-on-plate weld of a thin plate. This is a general representation of residual stress pattern observed in all type of fusion welds. The Figure 2.10(b & c) indicates the temperature gradients along
several cross sections through the weld bead path and the corresponding distribution of normal stress in the x-direction along the cross sections. It can be seen in the section D-D that after the weld cools, tensile stresses are produced in and around the weld which are balanced by compressive stresses away from the weld. Distribution of residual stresses in welds can be resolved into longitudinal and transverse residual stresses. Stresses that are parallel to the welding direction are called longitudinal residual stresses, designated by $\sigma_x$, and those that are transverse to the welding direction are called transverse residual stresses, designated by $\sigma_y$.

Robelloto and others (103) studied the magnitude and distribution of residual stresses in 2.5 mm thick Ti-6Al-4V (mill annealed) induced by gas tungsten arc welding. They used two passes on each side of panel and filler materials of similar composition to base material and made strain gage measurements to determine the magnitude of residual stresses. Their results are shown in the Figure 2.11. They found that longitudinal stresses ($\sigma_x$) in the weld were all tensile and transverse stresses ($\sigma_y$) were generally compressive. It was also found that these stresses were completely removed by a stress relief treatment of 788 °C for 15 minutes.
Figure 2.10 Stress map in the various regions of the weld (3)
Figure 2.11 Residual stresses in as-welded Ti-6Al-4V-perpendicular to weld (103)

They also found that residual stresses caused premature crack propagation through the weld compared to the stress relieved condition. Based on smooth unnotched specimens, the fatigue life of welded Ti-6Al-4V was increased by a factor of two on removing the residual stresses in the weld. Spraul and coworkers (104) studied transverse residual stresses in 180mm x 90mm x 20 mm Ti-6Al-4V produced by gas tungsten arc (GTA) and electron beam (EB) methods. They determined the magnitude of residual stresses using X-ray diffraction. Due to coarse grains in the middle of the weld, they had problems using X-ray diffraction to determine the residual stresses. It was found that the transverse residual stresses were compressive in the weld and tensile in the heat-affected zone. They also found that the level of residual stresses produced by low
heat input EB welding and high heat input GTA were the same, although zones affected by residual stresses were much wider in the case of GTAW than in EBW.
CHAPTER 3

EXPERIMENTAL PROCEDURE

3.1 Materials

Ti-6Al-4V was selected in two different conditions, mill annealed and β annealed. The plates of dimensions 24” x 6” x 0.236” (609 mm x 15 mm x 6 mm) were obtained from Timet Henderson. Table 3.1 shows the composition of Ti-6Al-4V. The β annealed plates were annealed at 1900 °F (1038 °C) for 30 minutes and air cooled. This was followed by stabilization annealing at 1350 °F (732 °C) for two hours and followed by air cooling. The mill annealed plates were annealed at 1333 °F (723 °C) for 0.5 hrs and air cooled. Both types of plates were then grit blasted and pickled. The pickling solution was 5% HF by volume and 35 % HNO₃ by volume in water.

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</tr>
</tbody>
</table>

Table 3.1 Chemical composition in wt% for Ti-6Al-4V used in this study
3.2 Welding Procedure

3.2.1 Friction Stir Welding

The joint design used was a simple square-groove butt joint with no root opening. The edges were milled with a ¼" (6 mm) carbide cutter. The plates were pickled and cleaned with acetone immediately prior to welding to remove grease or dirt.

The plates given to Edison Welding Institute for friction stir welding were 152 mm x 147 mm x 6 mm in size. The plates were welded on a 50 Hp Kearny-Trecker milling machine which is outfitted with rigid, a smooth steel backing plate and multiple toggle clamp hold-down system. The welding conditions are shown in Table 3.2. Argon gas shielding was used to prevent atmospheric contamination.

<table>
<thead>
<tr>
<th>Tool Type</th>
<th>Shoulder Plunge</th>
<th>Travel Speed</th>
<th>Tilt</th>
<th>RPM</th>
</tr>
</thead>
<tbody>
<tr>
<td>Commerically Pure Tungsten</td>
<td>0.005 inch (0.13 mm)</td>
<td>3.75 inch/min (1.6 mm/sec)</td>
<td>3.5°</td>
<td>275</td>
</tr>
</tbody>
</table>

Table 3.2 Welding conditions for friction stir welding
3.2.2 Single Pass Fusion Welding

Autogenous welding of the plates 152 mm x 102 mm in dimension was accomplished using Ti-71 penetration enhancing flux in a single pass. A 2% ceria-tungsten electrode of 2.4 mm diameter was used. The electrode was prepared with a 60° included angle with a stickout of 5/16” (8 mm). Cup size of #12 (22 mm) was used. The welding conditions were as shown in the Table 3.3.

<table>
<thead>
<tr>
<th>Torch Gas</th>
<th>Backing Gas</th>
<th>Trailing Gas</th>
<th>Welding Current</th>
<th>Welding Power</th>
<th>Travel Speed</th>
</tr>
</thead>
<tbody>
<tr>
<td>50/50 Ar/He</td>
<td>100% Ar@</td>
<td>100 % Ar @</td>
<td>165 amps</td>
<td>12 Volt</td>
<td>76 mm/min</td>
</tr>
<tr>
<td>@ 12 L/min</td>
<td>12 L/min</td>
<td>50 L/min</td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

Table 3.3 Welding procedure for single pass fusion welds
3.2.3 Multipass Gas Tungsten Arc Welding

The electrode, torch gas, backing gas, and trailing gas used for multipass welding were the same as those used for single pass welding. ERTi-5 filler material of diameter 0.035 "; heat number R 3330 and insert heat number D981038 was used. The mill annealed plates was welded using three passes and β annealed plates was welded using five passes. Table 3.4 shows the welding details of each pass. Pass number 1 is the root pass in both the mill annealed the β annealed welds. I_p and I_b represent the peak current and the base current respectively and T_p and T_b represent the time at the peak and base current respectively. The torch was oscillated and the different oscillation parameters are shown under the OSC columns. The GTAW was accomplished using rapid pulse and various rapid pulse parameters are shown in the Table 3.4.

3.3 Microstructural Characterization

Transverse and longitudinal sections of the weld were cut using EDM (Electrodeposition machine) and mounted in conducting bakelite. Mounted samples were ground using SiC paper through progressively finer grits (320, 400, and 600) and then polished with diamond compound of 6 μm and 1μm particulate size. Final polishing was done using colloidal silica. A standard Kroll's reagent (2 % HF, 4 % HNO₃ in water) was used for etching the samples. Scanning electron microscopy was used to study the microstructure evolution in the various regions of the welds and the base material. An XL-30 Field Emission Gun (FEG) scanning electron microscope manufactured by Philips was used to characterize the microstructures.
<table>
<thead>
<tr>
<th>Pass #</th>
<th>Ip/Tb</th>
<th>Tp &amp; TB</th>
<th>Voltage</th>
<th>Travel Speed</th>
<th>Wire Feed</th>
<th>OSC Width</th>
<th>OSC Dwel</th>
<th>Osc Speed</th>
<th>Rapid Pulse</th>
<th>Rapid Puls</th>
<th>Rapid Pulse</th>
</tr>
</thead>
<tbody>
<tr>
<td>MA 1</td>
<td>200/120</td>
<td>0.3</td>
<td>9</td>
<td>3</td>
<td>NA</td>
<td>NA</td>
<td>NA</td>
<td>NA</td>
<td>3000</td>
<td>75 %</td>
<td>75</td>
</tr>
<tr>
<td>MA 2</td>
<td>130/55</td>
<td>0.5</td>
<td>9.5</td>
<td>2.5</td>
<td>79</td>
<td>4</td>
<td>0.5</td>
<td>600</td>
<td>NA</td>
<td>NA</td>
<td>NA</td>
</tr>
<tr>
<td>MA3</td>
<td>130/55</td>
<td>0.5</td>
<td>9.5</td>
<td>3</td>
<td>39</td>
<td>4</td>
<td>0.5</td>
<td>600</td>
<td>NA</td>
<td>NA</td>
<td>NA</td>
</tr>
<tr>
<td>BA 1</td>
<td>145/85</td>
<td>0.3</td>
<td>9</td>
<td>3</td>
<td>NA</td>
<td>NA</td>
<td>NA</td>
<td>NA</td>
<td>3000</td>
<td>75%</td>
<td>75</td>
</tr>
<tr>
<td>BA2</td>
<td>145/55</td>
<td>0.3</td>
<td>9</td>
<td>3</td>
<td>NA</td>
<td>NA</td>
<td>NA</td>
<td>NA</td>
<td>3000</td>
<td>75%</td>
<td>75</td>
</tr>
<tr>
<td>BA 3</td>
<td>130/55</td>
<td>0.5</td>
<td>9.5</td>
<td>3</td>
<td>26</td>
<td>4</td>
<td>0.5</td>
<td>600</td>
<td>NA</td>
<td>NA</td>
<td>NA</td>
</tr>
<tr>
<td>BA 4</td>
<td>130/55</td>
<td>0.5</td>
<td>9.5</td>
<td>2.5</td>
<td>26</td>
<td>4</td>
<td>0.5</td>
<td>600</td>
<td>NA</td>
<td>NA</td>
<td>NA</td>
</tr>
<tr>
<td>BA 5</td>
<td>130/55</td>
<td>0.5</td>
<td>9.5</td>
<td>2.5</td>
<td>79</td>
<td>4</td>
<td>0.5</td>
<td>600</td>
<td>NA</td>
<td>NA</td>
<td>NA</td>
</tr>
</tbody>
</table>

Table 3.4 Welding parameters for multipass GTAW
Microhardness testing was used to study the hardness response in various regions of the weld. Microhardness measurements were made using a Vickers type indenter with a 500g load and a dwell time of 10 seconds. Microhardness traverses were made on transverse (perpendicular to the welding direction) and longitudinal sections (parallel to the welding direction) of the friction stir welds.

An XL 30 ESEM FEG scanning electron microscope with electron backscatter detector was used for orientation imaging studies. Software for orientation imaging developed by HKL Technologies was used. The samples used for studying macrotexture were not etched while those for microtexture studies were lightly etched.

3.4 Residual Stress Analysis

Residual stress measurements on the welded plate were obtained through collaboration with Lambda Research Inc. Measurements were obtained in the as-welded condition on the top face of the plates and at a depth of 0.25 mm from the top. The surface was electropolished to reveal the surface at depth of 0.25 mm. Stresses were measured parallel to welding direction and transverse to the welding direction. Measurements were done at 0.25 mm depth to avoid the effect of compressive residual stresses that are generated during grit blasting. X-ray diffractometers developed specifically for the measurement of subsurface residual stress and cold work distributions were used. The macroscopic residual stresses were determined using a conventional sine-squared-psi (21.3)/Cu K technique (105-107). The K1 peak breadth was calculated from the Pearson VII function fit used for peak location during macroscopic stress measurement (108). The peak breadth increases as the crystallite size is reduced and
microstrain increases with cold work during surface enhancement. The method of quantifying the degree of cold working of metals, by relating the x-ray diffraction peak broadening to the equivalent true plastic strain, is described in references 105 and 109. When the degree of cold work is taken to be equivalent amount of true plastic strain, the degree of cold work is then cumulative and independent of the mode of deformation (109). A previously generated calibration curve relating the (21.3) Cu K1 peak breadth to cold work expressed as the equivalent amount of true plastic strain for Ti-6Al-4V was used to determine the percent cold work in the welded plate.

Residual stress analysis was not possible in the multipass fusion welds due to the coarse grain size. Measurement of diffraction peak shifts is necessary to calculate the stress state in a polycrystalline material. It is possible to determine accurate values of the peak positions when the shape of the peaks does not vary much with the angle \( \psi \) (angle between the surface and the orientation between the planes). This cannot be verified in large grain sized materials for which the number of crystallites in the diffracting volume is not sufficient to lead to a diffracting peak with a regular shape. The shape of the peak strongly depends on the \( \psi \) angle and the sample position which leads to a large dispersion of the strain measurements (110).
CHAPTER 4

RESULTS AND DISCUSSION

4.1 Base Material Characterization

Scanning electron micrographs of the mill annealed and β annealed Ti-6Al-4V base materials are shown in the Figures 4.1 and 4.2. The mill annealed Ti-6Al-4V base material was a bimodal structure, with ~10 μm sized equiaxed α grains and some transformed β in the form of α lamellae. The β annealed base material consists of Widmanstatten α + β colonies in prior β grains of ~500 μm size. Grain boundary α is also observed in the β annealed base material.

The Figures 4.3 and 4.4 show the pole figure representations of the α and β phases in the mill annealed base material. All the pole figures in this study were recalculated from the orientation distribution function (ODF) that was determined from the individual orientation measurements assuming a half-width and cluster size of 5°. Figure 4.3 shows the pole figure representation of the α phase texture from a transverse section of the mill annealed base material. Similar pole figures are obtained for both of the welded plates, indicating that the texture in both the base material plates is the same. If the α phase texture would have been a basal one, density of poles would have been observed along
the short transverse (ST) direction and for a transverse texture, along the rolling plane (RP). As neither of the components is observed, it can be said that the α phase texture is neither basal nor transverse. The α phase texture is not very strong as the maximum intensity is ~ 4 times random only. The Figure 4.4 shows that texture of the β phase is also relatively weak. Recrystallization of the α phase that occurred after deformation imposed by the rolling process, modified the orientation of the α grains. Therefore the Burgers orientation relationship was not observed between the α and β phases.

The Figure 4.5 shows the pole figure representation of the α phase texture in the β annealed base material. Maximum intensity of the α and β phase textures is ~ 20 times random, indicating a sharp texture. The texture appears to be sharp due to the presence of large colonies with similar orientation within a colony. To get an idea of macrotexture, mapping was done over a large area (4mm x 4mm). This necessitated the use of large step size (5 μm). This caused poor indexing of the β phase, since the step size was larger than the width of thin β laths. Another scan with a smaller step size (0.18 μm) was done in a smaller area. This result can be seen in the Figure 4.6, and it can be seen that the Burgers orientation relationship exists between the α and β phases. Thus, it can be projected that the macrotexture of the β phase is similar to that of the α phase seen in the Figure 4.5. The α phase pole figures obtained for both the base material plates were same, indicating that the texture in both the plates is the same.
Figure 4.1 Photomicrograph of mill annealed Ti-6Al-4V base material

Figure 4.2 Photomicrograph of β annealed Ti-6Al-4V base material
Figure 4.3 Pole figure representation of $\alpha$ phase texture in transverse section of mill annealed Ti-6Al-4V base material.

Figure 4.4 Pole figure representation of $\beta$ phase texture in transverse section of mill annealed Ti-6Al-4V base material.
Figure 4.5 Pole figure representation of α phase texture in transverse section of β

Annealed Ti-6Al-4V base material
Figure 4.6 $\alpha$ and $\beta$ phase pole figures from a small area with small step size to verify
Burgers orientation relation
4.2 Friction Stir Weld Characterization

The Figures 4.7 and 4.8 show the macroscopic views of a transverse section of friction stir welds in both the starting conditions, i.e. mill annealed and β annealed. Macroscopic examination reveals that the stir zone is relatively symmetric for both the welds. This result is similar to that observed in the previous study on friction stir welding of mill annealed Ti-6Al-4V by Lienert and Jata (111). In contrast to this, some aluminum alloy friction stir welds observed are highly asymmetric. This could be due to the welding parameters, flow stresses, and thermal conductivity. In aluminum alloy friction stir welds, 6061 has lower flow stresses than 5454 and 2195 and it tends to produce relatively symmetric welds compared to 2195 and 5454 (112). However, titanium has higher flow stresses than aluminum alloys and it still produces symmetric welds. Titanium has poor thermal conductivity compared to aluminum and thus it will restrict the heating to a smaller area and consequently the thermally softened region will also be smaller and uniform. It has also been observed for aluminum alloys that slower welding speeds tend to produce symmetric welds (113). The thermomechanically affected zone (TMAZ) can be identified by a narrow region between the stir zone and base material that etches differently than either of them. The stir zone for both the welds does not extend to the bottom of the plates, indicating that the tool did not penetrate the plates completely in both the welds. The Figures 4.9 and 4.10 show the fracture surfaces at the weld centerline and the machined surface of the base material plates can be seen clearly at the bottom of weld. The top surfaces of the welds show slight concavity and some expelled metal (flash) at the edge of the stir zone. A small hole observed in the stir zone of the
mill annealed weld could be due to a pipe like defect which is created when the metal displaced behind the tool does not fill all the areas completely.

The Figures 4.11 and 4.12 show the macroscopic view of longitudinal sections of both the welds. The friction stir welds were interrupted welds, i.e. the travel of the tool was interrupted before it welded the plates completely. The hole shown in the figures is left by the tool when it was removed from the plates. The grooves observed in the hole were the thread marks left by the tool. It is observed that macrostructure similar to the stir zone exists in a small region ahead of the weld, near the point of tool removal. This could be due to the fact that the tool was rotated before removing and intense deformation was caused locally which generates conditions similar to those in the stir zone, resulting in similar macrostructure.
Figure 4.7 Photomacrograph of transverse section of mill annealed Ti-6Al-4V friction stir weld

Figure 4.8 Photomacrograph of transverse section of β annealed Ti-6Al-4V friction stir weld
Figure 4.9 Fracture surface at the weld centerline showing no joining at the root of the weld in the mill annealed FSW

Figure 4.10 Fracture surface at the weld centerline showing no joining at the root of the weld in the β annealed FSW
Figure 4.11 Photomacrograph of longitudinal section of mill annealed i-6Al-4V friction stir weld

Figure 4.12 Photomacrograph of longitudinal section of β annealed Ti-6Al-4V friction stir weld
4.2.1 Stir Zone

4.2.1.1 Metallographic Analysis

The microstructure of the mill annealed weld stir zone shown in the Figure 4.13 is bimodal, consisting of colony $\alpha + \beta$ and discretely scattered equiaxed $\alpha$ in prior $\beta$ grains of $\sim 10 \, \mu m$ in size. The presence of grain boundary $\alpha$ is also seen. The previous study by Lienert and Jata (111) observed similar microstructures. The microstructures observed in the stir zone suggest that the peak temperatures exceeded the $\beta$ transus. Intense deformation as in friction stir welding, if carried out below the $\beta$ transus would certainly result in globularization of the $\alpha$ phase. In contrast, lamellar $\alpha$ in small prior $\beta$ grains is observed which clearly suggests recrystallization in the $\beta$ phase. The fine grain size suggests that the dwell time above the $\beta$ transus was short and this prevented the $\beta$ grain growth. The presence of grain boundary $\alpha$ indicates that most of the deformation was carried out above the $\beta$ transus. If deformation takes place cooling through the $\beta$ transus, alternative sites for nucleation of $\alpha$ are provided and grain boundary $\alpha$ is absent (2). The presence of discretely scattered equiaxed $\alpha$ cannot be clearly explained. However it could be due to the fact that the low dwell time above the $\beta$ transus prevented all the equiaxed $\alpha$ grains from getting transformed.

The Figure 4.14 shows that the prior $\beta$ grain size in the $\beta$ annealed weld stir zone ($\sim 10 \, \mu m$) is considerably smaller than that of the base material grains ($\sim 500 \, \mu m$). The $\alpha + \beta$ colony microstructure observed in the stir zone is quite refined compared to that in the base material. The presence of grain boundary $\alpha$ is also observed. The microstructural
evolution in the β annealed welds seems to be similar to that in the mill annealed welds. Recrystallization occurred above the β transus and resulted in a lamellar microstructure in prior β grains ~10 μm in size. The presence of grain boundary α indicates that most of the deformation occurred above the β transus. The absence of discretely scattered equiaxed α strengthens the contention that the equiaxed α in the stir zone of mill annealed welds is the base material that did not transform.

Past studies on friction stir welding of aluminum indicate that the stir zone recrystallizes by a dynamic recrystallization mechanism. Since recrystallization in the present study occurs along with the deformation, it must be by some dynamic mechanism. However, understanding of deformation mechanisms in the β phase is not clear. Modeling of hot deformation mechanisms in the β phase field on the basis of room temperature microstructures is difficult since the β→α + β transformation removes the traces of dynamical microstructural features associated with the high temperature β deformation. Thus, indirect methods of modeling such as shapes of stress-strain curves, kinetic analysis, processing maps and hot ductility variations have been used to understand the deformation mechanism of the β phase (114). However, a clear understanding of the deformation mechanism of the β phase under a given set of conditions does not exist. Sheppard and Norley observed during hot torsion testing that the activation energy for mechanical working in the β region was equal to that for self diffusion in the β-phase and concluded that dynamic recovery was the operative deformation mechanism (115). They observed recrystallization at high strain rates and
considered it to be static. Prasad and coworkers conducted hot deformation studies above the β transus and proposed the β deformation mechanism to be continuous dynamic recrystallization on the basis of kinetic analysis, grain size variations, power dissipation, efficiency characteristics and hot ductility values (116). Seshacharyulu concluded the β deformation mechanism to be dynamic recrystallization due to the refined grain size and curved grain boundaries that were observed after deformation in the β regime (114).
Figure 4.13 Photomicrograph of mill annealed Ti-6Al-4V friction stir weld stir zone

Figure 4.14 Photomicrograph of β annealed Ti-6Al-4V friction stir weld stir zone
4.2.1.2 Texture Analysis

The Figures 4.15 - 4.18 show the pole figure orientations of $\alpha$ and $\beta$ phases in the stir zone of both the mill annealed and the $\beta$ annealed welds. It can be seen that the textures of the $\alpha$ and $\beta$ phases have changed in both the welds due to the thermomechanical processing. In general, recrystallization promotes a change in texture, which is attributed to the existence of oriented nuclei in the deformed grains, or to the oriented growth of the recrystallized grains (117). Coarsening of the prior $\beta$ grain size has not occurred, which indicates rapid cooling. Fast cooling rate results in undercooling, which causes domination of nucleation over growth. Thus more variants are nucleated to give weak texture in a particular direction (118)). Therefore, the $\alpha$ phase in both the welds does not exhibit very strong textures. A Burgers relationship can be observed between the $\alpha$ and $\beta$ phases, indicating that the complex thermomechanical processing occurred above the $\beta$ transus and very little deformation took place after the $\alpha$ phase was nucleated. Although the starting base material textures in the mill annealed and $\beta$ annealed welds were different, the textures in the stir zone of both the welds exhibit some similarity. This correlates with the similar microstructures observed in both the stir zones.

A total of 12 distinct $\alpha$ orientations may arise from an initial $\beta$ orientation during the $\beta \rightarrow \alpha$ transformation, according to the Burgers relationship and the symmetry of $\alpha$ and $\beta$ phases. Due to physical limitations, all the 12 inherited orientations are not exhibited at the scale of a grain. The grains with nearly the same orientation, transform into a limited set of variants by random selection. It has been observed that the $\beta \rightarrow \alpha$ transformation in Ti-6Al-4V proceeds without variant selection when the phase transformation occurs in
absence of deformation while variant selection is observed in the case of deformation preceding phase transformation (119). It can be seen in both the stir zones that the α phase preferentially picks a particular variant of the various possible orientations of the β phase, indicating variant selection. Variant selection is a complex phenomenon that depends on many factors, some of which are nature of deformation, stress fields and cooling rate. Although the starting base material textures in the mill annealed and β annealed welds were different, the textures in the stir zone of both the welds exhibit some similarity. This correlates with the similar microstructures observed in both the stir zones.
Figure 4.15 Pole figure representation of α phase texture in mill annealed Ti-6Al-4V

FSW stir zone

Figure 4.16 Pole figure representation of β phase texture in mill annealed Ti-6Al-4V

FSW stir zone
Figure 4.17 Pole figure representation of α phase texture in transverse section of β annealed Ti-6Al-4V FSW stir zone

Figure 4.18 Pole figure representation of β phase texture in transverse section of β annealed Ti-6Al-4V FSW stir weld
4.2.2 Thermomechanically Affected Zone (TMAZ)

4.2.2.1 Metallographic Analysis

Different regions of the mill annealed weld TMAZ can be seen in the Figure 4.19. The width of the TMAZ almost remains constant from root to the face of the weld, on both sides of the weld. The thermomechanically affected zone consists of deformed base material grains and 15-20 μm wide region that is microstructurally distinct compared to the base material and the stir zone. The deformed base material region of the TMAZ can be defined as the region that was deformed due to the deformation caused by the tool but where phase transformation did not take place. The thermomechanical processing caused a phase transformation in the microstructurally distinct band of the TMAZ. It can be seen in the Figure 4.20 that the band of microstructurally distinct region consists of fine equiaxed α particles, 1-2 μm in size and colony α +β. These equiaxed α could have been nucleated due to thermomechanical processing in the α +β region or could be transformed equiaxed α of the base material.

It can be seen in the Figure 4.21 that similar to the mill annealed welds, the β annealed weld TMAZ consists of deformed base material and a band of microstructurally distinct region. However, it is wider than the mill annealed weld TMAZ by approximately 20 μm. The deformed base material can be identified by α laths curving into the band of microstructurally distinct region. The microstructurally distinct region shown in the Figure 4.22 consists of fine equiaxed α particles, 1-2 μm in size, small α laths and colony α +β. The equiaxed α could be the α laths coming from below the surface. The small α laths can be the base material laths broken due to the deformation.
Figure 4.19 Photomicrograph of mill annealed Ti64 FSW thermomechanically affected zone (TMAZ)

Figure 4.20 Photomicrograph of microstructurally distinct region from the box in the Figure 4.19 in mill annealed Ti64 FSW TMAZ
Figure 4.21 Photomicrograph of β Annealed Ti-6Al-4V FSW TMAZ

Figure 4.22 TMAZ Photomicrograph of microstructurally distinct region from the box in the Figure 4.21 in β annealed Ti-6Al-4V FSW
4.2.2.2 Texture Analysis

Microtexture analysis was done for the different regions of the TMAZ in both the welds. The Figures 4.23 shows the orientation map and the corresponding α phase pole figure orientations of the deformed base material grains in the TMAZ of the mill annealed welds. It can be seen that the texture of the basal poles in this region is similar to that observed in the base material and texture of the other planes is not similar to that observed in the base material. This indicates that the deformation caused a rotation of the α grains along the basal planes. The highlighted region in the Figure 4.24 corresponds to the microstructurally distinct band in the TMAZ. The Figure 4.24 shows the texture of the α phase observed in this region and it can be seen that it is different than that observed in both the base material and the stir zone. This change in texture can be due to deformation of the equiaxed α of the base material in this region or due to recrystallization, that formed the equiaxed α. The same can be said about the region shown in the Figure 4.25 and its corresponding pole figure.

The Figure 4.26 shows the orientation map and the α phase pole figures of the deformed base material grains in the TMAZ of the β annealed welds. Although the deformed base material grains show some similarity to the base material texture, there are some additional texture elements observed which indicate that the deformation mechanism is more complex than rotation of the grains. The Figure 4.27 shows that although the intensity of the texture in the microstructurally distinct region is less than that in the deformed grains, it exhibits some similarity to the texture of the deformed base material grains. It can be observed in the Figure 4.28 that the TMAZ near
the stir zone exhibits a strong texture and the strong element of the texture is similar to that observed in the other areas of the TMAZ. As in the case of the mill annealed welds, the change in texture can be due to deformation of the base material or nucleation of equiaxed \( \alpha \).
Figure 4.23 Orientation map and pole figure representation of the α phase texture of the deformed grains in the mill annealed weld TMAZ
Figure 4.24 Orientation map and pole figure representation of the α phase texture of highlighted grains in the above orientation map
Figure 4.25 Orientation map and pole figure representation of the α phase texture of highlighted grains in the above orientation map
Figure 4.26 Orientation map and pole figure representation of the α phase texture of the deformed grains in the β annealed Ti-6Al-4V FSW Thermomechanically affected zone (TMAZ)
Figure 4.27 Orientation map and pole figure representation of the α phase texture of the highlighted grains in the above orientation map
Figure 4.28 Orientation map and pole figure representation of the α phase texture of the highlighted grains in the orientation map above.
4.2.3 Region Around Tool Removal

The Figures 4.29 and 4.30 show the microstructures observed in the various regions around the periphery of tool in the interrupted welds. Microstructure similar to that observed in the stir zone is observed in a small region ahead of the weld. There was rotation of the tool in the stationary position, before it was removed. This resulted in conditions similar to the stir zone in a small region ahead of the weld, which resulted in microstructure similar to the stir zone ahead of the weld. Grain growth is observed at the top of the plate, below the shoulder. The grain growth indicates that the friction of the tool generated heat and locally caused grain growth. A TMAZ similar to that observed in the transverse sections is also observed below the tool in both the welds.
Figure 4.29 Photomicrograph showing microstructures near point of tool retraction in mill annealed Ti-6Al-4V FSW
Figure 4.30 Photomicrograph showing microstructure near point of tool retraction in β annealed Ti-6Al-4V FSW
4.2.4 Heat Affected Zone (HAZ)

In contrast to FSWs observed in other aluminum alloy and Ti-6Al-4V friction stir welds by Lienert and Jata, a microstructurally distinct heat affected zone is not observed here. The Figures 4.31 and 4.32 show the microstructures close to the deformed grains of the TMAZ. It can be seen that the microstructures in the probable heat affected zone are similar to the base material and no phase transformation is observed. Substructure recovery could have occurred in the HAZ and further TEM studies are needed to provide insight in this regard.
Figure 4.31 Photomicrograph showing mill annealed weld HAZ

Figure 4.32 Photomicrograph showing β annealed weld HAZ
<table>
<thead>
<tr>
<th>Base Material</th>
<th>Stir Zone</th>
<th>TMAZ</th>
<th>HAZ</th>
<th>Point of Tool Removal</th>
</tr>
</thead>
<tbody>
<tr>
<td>Equiaxed α grains (~ 10μm) and colony α+β between them</td>
<td>Colony α+β, in prior β grains, 10 μm in size and grain boundary α present. Discretely scattered equiaxed α particles</td>
<td>Deformed equiaxed α grains and microstructurally distinct band consisting of small equiaxed α and small Colony α+β</td>
<td>No microstructurally distinct HAZ observed</td>
<td>Grain growth at the top of the weld, TMAZ below the tool, small stir zone ahead of the weld</td>
</tr>
</tbody>
</table>

Table 4.1 Summary of microstructures in mill annealed Ti-6Al-4V FSW

<table>
<thead>
<tr>
<th>Base Material</th>
<th>Stir Zone</th>
<th>TMAZ</th>
<th>HAZ</th>
<th>Point of Tool Removal</th>
</tr>
</thead>
<tbody>
<tr>
<td>colony α+β in large prior β grains, ~ 500μm in size</td>
<td>Colony α+β, in prior β grains, 10 μm in size and grain boundary α present</td>
<td>Deformed α laths microstructurally distinct band consisting of small equiaxed α and small Colony α+β</td>
<td>No microstructurally distinct HAZ observed</td>
<td>Grain growth at the top of the weld, TMAZ below the tool, small stir zone ahead of the weld</td>
</tr>
</tbody>
</table>

Table 4.2 Summary of microstructures in β annealed Ti-6Al-4V FSW

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4.2.5 Microhardness

Vickers hardness measurements were performed on transverse and longitudinal sections of the friction stir welds. Four traverses from the top to the bottom were taken across a transverse section of the weld. Three measurements were taken at each point across the weld. Figures 4.33 and 4.34 show contour plots of mean hardness plotted as distance from the centerline to the top of the weld. The region enclosed within the white arc, corresponds to the stir zone. Although not significantly different, the stir zone exhibits higher hardness compared to the base material in the mill annealed welds. The $\beta$ annealed weld stir zone also exhibits higher hardness than the base material. This could be due to fine grain size in the stir zone of both the welds, since grain size effects have a strong influence on the hardness in Ti alloys. The similar hardness levels observed in the mill annealed and $\beta$ annealed weld stir zones correlate well with their similar microstructures. One noticeable feature is the spike in hardness at the root of the weld joint, which is observed in both the welds.

Microhardness measurements were also done in longitudinal sections of the friction stir welds. The Figure 4.35 shows the hardness in the mill annealed weld longitudinal sections. An increase in hardness is observed in the stir zone as compared to the base material. An increase in hardness observed at the root joint in transverse sections is also visible here. Hardness in the stir zone seems to be uniform in the longitudinal direction. This could be related to the uniform microstructure observed throughout the stir zone except at the top of the weld, where the prior $\beta$ grain size is slightly larger. Correspondingly, the hardness values in the $\beta$ annealed stir zone at the top of the weld are
lower than the other regions of the weld. It can be seen in the Figure 4.36 that similar to the mill annealed welds, increase in hardness was observed in the stir zone of the β annealed welds. Increase in hardness at the root of the welds was spread over a wider area in the β annealed welds.
Figure 4.33 Mill annealed weld transverse section microhardness

Figure 4.34 β annealed weld transverse section microhardness
Figure 4.35 Mill annealed weld longitudinal section microhardness map

Figure 4.36 β Annealed weld longitudinal section microhardness map
4.3 Single Pass Fusion Welds

The Figures 4.37 and 4.38 show the macrographs of the single pass fusion welds. The welds appear similar to conventional fusion welds, with a fusion zone and a heat affected zone. The microstructures observed in the single pass fusion welds are similar to that observed in the previous studies on fusion welding of Ti-6Al-4V (17, 45, and 47). The Figures 4.39 and 4.40 show that the fusion zone of both the welds consists of colony \( \alpha + \beta \) in large \( \beta \) grains. The temperatures exceed the \( \beta \) transus in the HAZ near the fusion boundary in both the welds. It can be seen in the Figures 4.41 and 4.42 that the HAZ microstructure near the fusion boundary consists of colony \( \alpha + \beta \) in large prior \( \beta \) grains. The Figure 4.43 shows that undissolved equiaxed \( \alpha \) and colony \( \alpha + \beta \) is present in those regions of the mill annealed weld HAZ where the temperature did not exceed the \( \beta \) transus. The Figures 4.44 and 4.45 show that there is no significant change in the hardness in the fusion zone or the HAZ and hardness values obtained in these welds conform to those observed in Ti-6Al-4V fusion welds.
Figure 4.37 Macrograph of transverse section of mill annealed Ti-6Al-4V single pass fusion Weld

Figure 4.38 Macrograph of β annealed Ti-6Al-4V single pass fusion weld
Figure 4.39 Photomicrograph of mill annealed Ti-6Al-4V single pass weld fusion zone

Figure 4.40 Photomicrograph of β annealed Ti-6Al-4V single pass weld fusion zone
Figure 4.41 Photomicrograph of mill annealed Ti6Al-4V single pass fusion weld HAZ near fusion boundary

Figure 4.42 Photomicrograph of β annealed Ti6Al-4V single pass fusion weld HAZ near fusion boundary
Figure 4.43 Photomicrograph of mill annealed Ti-6Al-4V single pass fusion weld HAZ away from fusion boundary
Figure 4.44 Microhardness data from transverse section of single pass fusion weld of mill annealed Ti-6Al-4V

Figure 4.45 Microhardness data from transverse section of single pass fusion weld of β annealed Ti-6Al-4V
4.4 Residual Stress Analysis

The Figures 4.46-4.49 show the residual stress measurements and degree of cold work in the friction stir and single pass fusion welds. The curves labeled "parallel" refer to the stresses parallel to the welding direction and those labeled "perpendicular" refer to stresses transverse to the welding direction across the width of the plate (i.e., not in the short transverse direction of the plate). The curves labeled "0.01 in. depth parallel" and "0.01 in depth perpendicular" correspond to the residual stresses measured parallel and perpendicular to the welding direction after 0.25 mm of material was electropolished from the surface. In the as received condition, the plates were grit blasted, which introduced compressive stresses along with cold work on the surface. However, the compressive stresses and the cold work due to grit blasting are surface specific and extend to ~ 0.005-0.05 mm for Ti-Al-4V (120). Thus, stress measurement was also done at a depth of 0.25 mm to mitigate the effect of grit blasting and analyze the stresses introduced in the base material due to the welding process only. The grit blasting done on the base material has no effect on the stress generated in the stir zone, since grit blasting is only a surface effect and the material gets heavily deformed in the stir zone. Electropolishing was used to remove surface layer because it does not induce any residual stresses and preferential etching of grain boundaries is also avoided. Significant stress relaxation can occur on removing the surface layer. Closed-form solutions are available to correct the results obtained on the surfaces exposed by electropolishing for removal of the stressed layers above (121). The error of measurement was 5 ksi for friction stir welding, where the grain size was around 10 μm. The error of measurement
goes slightly higher in the case of single pass fusion welding because of the large grain size.

4.4.1 Friction Stir Welding

The Figure 4.46 shows the residual stresses in the mill annealed welds. It can be observed for the mill annealed welds that grit blasting resulted in significant compressive stresses on the base material surface, both parallel and perpendicular to the welding direction. Grit blasting also induced a high amount of cold work along with the compressive stresses on the surface, as can be seen in the figure. In the stir zone, the stresses in the advancing and retreating sides of the stir zone were symmetric. The stresses parallel to the welding direction were tensile in nature, while the stresses perpendicular to the welding direction were compressive in nature. The magnitude of stresses was higher at a depth of 0.25 mm compared to that at the surface. A similar gradient in residual stresses can be observed in the case of aluminum alloy friction stir welds (122). This is attributed to the nature of the process, in which the metal is churned upward and is restrained by the tool. The tool exerts downward force on the stir zone and the restraint increases with distance from the tool. This results in an increase in tensile stresses at increasing depths. No studies were done at further depths and the depth to which this gradient exists, remains the subject for future studies. It can also be seen that the level of cold work was significantly low in the stir zone even after intense stirring that occurs in the stir zone. This could be attributed to the recrystallization taking place in the stir zone. The actual stresses in the base material due to the welding process are observed by the stresses at 0.25 mm. The stresses in the stir zone were balanced by stresses in the
surrounding base material. The magnitude of these stresses was very low, 10 ksi compressive in the parallel direction and 2 ksi in the perpendicular direction.

The Figure 4.47 shows the residual stresses in the β annealed welds. Although these plates were grit blasted, the level of cold work and the resulting compressive stresses induced on the surface was very low. Similar to the mill annealed welds, the stresses parallel to the welding direction were tensile and those perpendicular to the welding direction were compressive. Although the stresses are quite symmetric in the stir zone, the gradient in the stresses in the stir zone were higher on the advancing side. This is in contrast to the gradient in the mill annealed welds. Similar to the mill annealed welds, recrystallization is observed in the stir zone which resulted in very low cold work in the stir zone. The stresses in the stir zone were balanced by those of opposite sign in the surrounding base material, as can be observed by the stresses at 0.25 mm depth surrounding the base material. Although the level of stresses on the surface is same in both the welds, the level of stresses at 0.25 mm was higher in the β annealed welds by 8-10 ksi.
Figure 4.46 Residual stress and cold work distribution in mill annealed Ti6Al-4V FSW
Figure 4.47 Residual stress and cold work distribution in β annealed Ti-6Al-4V FSW
4.4.2 Single Pass Fusion Welds

Residual stress analysis was carried out on single pass fusion welds at 0.25 mm depth to avoid the surface effect of grit blasting. The Figure 4.48 shows that the magnitude of residual stresses in the fusion zone of mill annealed welds was higher than in the stir zone of friction stir welds. The level of compressive stresses in the surrounding regions was also higher than that in the friction stir welds. It can be seen in the Figure 4.49 that the level of stresses in the fusion zone of β annealed welds is also higher than that in the stir zone of the β annealed welds. The compressive stresses in the surrounding base material of the β annealed weld fusion zone were lower than those in the mill annealed welds. The level of cold work in the fusion zone and the base material of both the welds is also very low.
Figure 4.48 Residual stress and cold work distribution in mill annealed single pass fusion weld
Figure 4.49 Residual stress distribution in β annealed single pass fusion weld
4.4.3 Comparison of Residual Stresses in FSW and GTAW

The Table 4.3 summarizes the results of residual stresses in the friction stir welds in this study. It can be concluded that the magnitude of residual stresses generated by friction stir welding is lower than single pass fusion welding and the pattern of residual stresses is also different.

In the previous studies on multipass fusion welding by Robelloto, the tensile stresses induced by multipass fusion welding were approximately equal to 60 ksi. These stresses were induced in welding 2.5 mm thick plates. Increasing the thickness of the plates increases the overall stiffness. The stiffer the weld the less it will distort. Therefore the level of residual stresses increases with thickness and it can be said that the level of stresses would be definitely higher than 60 ksi in fusion welding of 6 mm thick plates. However, FSW of 6 mm thick plates induces only 30-40 ksi of tensile stresses. Thus, it can be concluded that FSW induces less residual stresses than multipass fusion welding. The residual stresses generated by autogenous single pass fusion welding are lower than multipass fusion welding because the restraint generated is less due to the absence of filler material (123). Residual stresses generated in the single pass welds are also uniform through the thickness of the weld, unlike the multipass welds where the magnitude of stresses varies (124).
<table>
<thead>
<tr>
<th>Surface</th>
<th>Mill annealed BM</th>
<th>Mill annealed SZ</th>
<th>Cold work in SZ of MA weld</th>
<th>β annealed BM</th>
<th>β annealed SZ</th>
<th>Cold work in SZ of BA weld</th>
</tr>
</thead>
<tbody>
<tr>
<td>Surface</td>
<td>~ -70 ksi parallel and perpendicular</td>
<td>~ 20 ksi parallel, ~ -12 ksi perpendicular</td>
<td>&lt;5%</td>
<td>~ -8 ksi parallel and perpendicular</td>
<td>~ 28 ksi parallel, ~ -16 ksi perpendicular</td>
<td>&lt;5%</td>
</tr>
<tr>
<td>0.25 mm Depth</td>
<td>~ 4 ksi parallel and perpendicular</td>
<td>~ 40 ksi parallel, ~ -12 ksi perpendicular</td>
<td>~ 4 ksi parallel and perpendicular</td>
<td>~ 40 ksi parallel, ~ -16 ksi perpendicular</td>
<td>~ 40 ksi parallel, ~ -16 ksi perpendicular</td>
<td>~40 ksi parallel, ~ -16 ksi perpendicular</td>
</tr>
</tbody>
</table>

Table 4.3 Summary of residual stresses in friction stir welds

<table>
<thead>
<tr>
<th>0.25 mm Depth</th>
<th>Mill annealed BM</th>
<th>Mill annealed fusion zone</th>
<th>Cold work in FZ of MA welds</th>
<th>β annealed BM</th>
<th>β annealed SZ</th>
<th>Cold work in FZ of BA welds</th>
</tr>
</thead>
<tbody>
<tr>
<td>~ -50 ksi parallel</td>
<td>~ 50 ksi parallel</td>
<td>~ 50 ksi parallel</td>
<td>&lt;5%</td>
<td>~30 ksi parallel</td>
<td>~ 60 ksi parallel</td>
<td>&lt;5%</td>
</tr>
</tbody>
</table>

Table 4.4 Summary of residual stresses in single pass fusion welds
CHAPTER 5

CONCLUSIONS

5.1 Microstructural Characterization

1. Friction stir welding of Ti-6Al-4V in both the starting conditions; mill annealed and β annealed produces symmetric welds.

2. The welds consist of a stir zone and a thermomechanically affected zone (TMAZ). A microstructurally distinct HAZ is not observed in either of the welds.

3. The stir zone in both the conditions consists of colony α+β in small prior β grains (~10 μm), indicating recrystallization above the β transus.

4. The presence of grain boundary α in both the stir zones indicates that most of the deformation occurs above the β transus.

5. The presence of discretely scattered equiaxed α particles was observed in the stir zone of the mill annealed welds. Their origin cannot be clearly explained.

6. The thermomechanically affected zone (TMAZ) in both the welds, consists of deformed base material and a “microstructurally distinct” band consisting of equiaxed α and colony α+β. The equiaxed α present in this band could be newly nucleated equiaxed α or transformed base material equiaxed α grains.
7. The study of longitudinal sections reveals grain growth at the top of the weld local heating from the shoulder. A small region with microstructure similar to that in the stir zone is observed ahead of the weld, at the point of tool retraction.

8. An increase in hardness is observed in the stir zone of both the welds due to the refined grain size. This increase in hardness seems to be uniform throughout the longitudinal section.

9. The texture in the stir zone of both the welds changed due to recrystallization. The Burgers orientation relationship observed between the \(\alpha\) and the \(\beta\) phases in the stir zone of both the welds indicates that negligible deformation occurred after the \(\alpha\) phase was nucleated.

10. Microtexture analysis in the TMAZ reveals that the texture in the microstructurally distinct band in both the welds is different than the stir zone and the base material. The deformed grains in the TMAZ of the mill annealed welds have similar basal pole texture to the base material, indicating grain rotation along the basal poles.

5.2 Residual Stress Analysis

1. The stresses parallel to the welding direction are tensile in nature in both the friction stir welds.

2. The stresses in the friction stir welds are symmetric in the leading and the trailing side of the stir zone.

3. The level of cold work in the stir zone of the friction stir welds is very low due to recrystallization.

4. The magnitude of tensile residual stresses in the single pass fusion welds is slightly higher to that in friction stir welds.
5. Residual stress analysis was not possible in the multipass fusion welds due to the large grain size in the fusion zone but the level of residual stresses was determined to be higher than in FSW based on published data.
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