Analysis of Microstructure Evolution and Mechanical Properties of
Linear Friction Welded Titanium alloys

by

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A thesis submitted to the Faculty of Graduate and Postdoctoral
Affairs in partial fulfillment of the requirements for the degree of

Master's in mechanical engineering

in

Master of Applied Science (M.A.Sc)

Carleton University

Ottawa, Ontario

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Abstract

Two titanium alloys, Ti-6Al-4V and Ti-6Al-2Sn-4Zr-2Mo were linear friction welded in similar and dissimilar combinations to analyze the effects of stress relief annealing on the mechanical and microstructural properties and to compare the results with as-welded state. For the selected set of weld parameters, high-quality welds were obtained with minimal defects. Scanning electron and optical microscopes were used to study the microstructures at various weld zones. Mechanical properties were determined by micro-indentation hardness, tensile, and fatigue testing. The formation of martensite at the weld zone was mainly the reason for the high strength of the weld in the as-welded condition. The local mechanical properties were determined by the digital image correlation (DIC) technique and showed the overall local strain concentration of the welds in the as-welded and stress relieved conditions. The fractographic analysis performed on both the tensile and fatigue samples using an electron microscope showed the characteristic features of ductile fracture.

Acknowledgments

Firstly, I would like to express my sincere gratitude to my advisor Prof. Abu Syed Kabir for the continuous support of my M.A.Sc. study and related research, for his patience, motivation, suggestions, and feedback. His guidance helped me in all the time of research and writing of this thesis. I could not have imagined having a better advisor and mentor for my M.A.Sc. study.

I would also like to thank Dr. Priti Wanjara of the National Research Council of Canada who provided me an opportunity to join their team, and who gave access to the laboratory and research facilities. Without their precious support, it would not be possible to conduct this research. My sincere thanks also goes to Dr. Javad Gholipour, Mr. Xavier Pelletier and Mr. Maxime Gurin at the NRC for their support. I am also thankful to many of my student colleagues for inspiring discussions and providing a pleasurable environment. In particular, I grateful to Mr. Peter Walker for his continuous support and suggestions.

Most importantly, I would like to thank my family; parents, my sister and all other members for encouraging me in all of my pursuits and inspiring me to follow my dreams. I would also like to dedicate this work in memory of my beloved grandparents who would have been more than proud to see me graduated.

Last but not least, thanks go to Almighty God for giving me the patience and energy to complete this work.

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Nomenclature

ASTM	American Society for Testing and Materials
AW	As-welded
BCC	Body-Centered Cubic
BLISK	Bladed integrated disks
BM	Base Metal
BSE	Backscattered Electron
DIC	Digital Correlation Imaging
DW	Dissimilar Weld
% EL.	Percentage of Elongation
HCP	Hexagonal Close Packing
LFW	Linear Friction Welding
PWHT	Post Weld Heat Treatment
SEM	Scanning Electron Microscopy
SE	Secondary Electron
SR	Stress relieved
SW	Similar Weld
SR	Stress relieved
TMAZ	Thermo Mechanically Affected Zone
UTS	Ultimate Tensile Strength
WCZ	Weld Center Zone
WI	Weld Interface
YS	Yield Strength

Chapter 1 : Introduction

Titanium and its alloys have long been utilized in the aerospace industry due to their low density, high specific strength, and excellent corrosion resistance. Properties of titanium can be tailored with an allotropic transformation from a low-temperature α phase (hexagonal close-packed structure) to a high-temperature β phase (body-centered cubic structure) at 882 °C [1]. This transformation occurs in most titanium alloys and widens the possibility of obtaining diverse microstructures through different thermal and thermomechanical heat treatments [2]. One of the major aerospace applications of titanium alloys is in the rotating parts of jet engine compressor blades, disks, and turbine parts that maximize efficiency and metallurgical stability at elevated temperatures [3].

Advanced designs of compressors are crucial for fulfilling the engine manufacturer's requirements. Traditionally, aircraft engines and industrial gas turbines use bladed compressor disks in a slotted central retainer with individual airfoils secured by nuts and bolts. BLISK is an improved design in which discs and blades are manufactured in one piece. The term "BLISK" is an acronym composed of the word's "blade" and "disk". BLISK is also referred to as an integrated bladed rotor (IBR), meaning blade roots and blade slots are no longer needed [4]. BLISK can be manufactured by machining a single forged part or by welding individual blades into a disk structure. Producing a BLISK from a cast monolithic billet is associated with very expensive CNC machining and a high buy to fly ratio. Therefore, the current research focuses on integrating blades and disk made of titanium alloys using linear friction welding (LFW) technology. Another major research

interest of LFW is focused on repairing or replacing the blades that suffer catastrophic failures during the flights (e.g. failure due to bird strikes, etc.) [4].

Weldability is a classic problem associated with titanium alloys. Joining of titanium alloys using conventional fusion welding techniques requires inert shielding gases to minimize the oxidation during the liquid phase processing, and the low thermal conductivity of titanium alloys requires the use of these shielding gases for an extended period of time. Fusion welding techniques such as laser or electron beam welding, consisting of rapid heating/melting and cooling by localized heat input, solve this issue to an extent. Studies conducted on gas tungsten arc welding (GTAW) of titanium alloys have reported ductility losses compared to the parent metal due to the coarsening of microstructure in the heat-affected zone (i.e. large prior- β grain size and coarse α - β lamella) [5-6].

Solid-state welding is a group of welding processes that achieve coalescence joining at temperatures significantly below the melting point of the base materials, without the addition of filler metal. LFW belongs to the category of solid-state welding in which two metal parts are joined by the intimate contact of a plasticized layer between the adjoining samples. Some of the practical advantages of LFW are i) safe technology in terms of automated operation, ii) lack of any special surface preparation, and iii) elimination of gaseous fumes [7]. Frictional heating and the applied force on the stationary workpiece joint the plasticized layer at the interface of the parts. Material removal occurs in the form of flash from the plasticized zone due to the combined activity of applied force and part movement [4]. Surface oxides and impurities are also removed along with the flash and this allows contact between parts to form a joint. The requirement of shielding

gases or a vacuum environment is eliminated as the material does not reach the melting temperature. This implies that solidification problems such as porosity, hot cracking, segregation, etc. associated with fusion welding techniques will be eliminated. Also, compared to the laser and electron beam welding process a narrow heat-affected zone (HAZ) is formed during LFW as a result of a high energy density cycle. Especially for titanium alloys having low thermal conductivity and extensive mechanical deformation by friction welding forces creates a narrow weld interface (WI) and thermomechanically affected zone (TMAZ) [5]. Hence, the application of linear friction welding to join the interface by solid-state processing is particularly attractive for titanium alloys.

The literature data published on LFW technology to date is primarily focused on popular $\alpha+\beta$ titanium alloys [5]. In this study, two industry-relevant titanium alloys, Ti-6Al-4V (Ti-64) and Ti-6Al-2Sn-4Zr-2Mo (Ti-6242), were linear friction welded in similar and dissimilar combinations for analyzing the effects of stress relief annealing on the mechanical and microstructural properties and compare the results with as-welded state. Ti-64 is an $\alpha+\beta$ alloy known as the ‘workhorse’ of the titanium industry since the 1950s due to its moderate to good weldability, relatively high strength, and fatigue properties [2]. The near- α Ti-6242 alloy was developed to improve oxidation resistance and creep property with the addition of 0.06 ~ 0.2 wt.% Si. Ti-6242 alloy serves as the structural component for aircraft parts exposed to elevated temperatures (<540 °C) such as gas turbine disks, blades, impellers, and seals, etc. [8].

The main objective of the research is to conduct a preliminary study to integrate a Ti-6242 disk with the typical Ti-64 alloy for the compressor blades designed for an operating regime of 430 to 520°C. The mechanical integrity of the welds was analyzed by

performing hardness, tensile, and fatigue tests. Previous research works [25-26, 45-46] on LFW of titanium alloys were mainly focused on the microstructural and mechanical properties of similar welded joints in the as-welded condition. For the first time the effects of stress relief annealing were performed on Ti-64 and Ti-6242 *similar welds* to study the global and local mechanical properties. Also, for the first time, the effect of stress relief annealing on the microstructures and mechanical properties of Ti-64 to Ti-6242 *dissimilar joints* was studied and compared with the as-welded condition. The digital image correlation (DIC) technique was used for the first time to study the local mechanical properties of linear friction welded similar and dissimilar titanium joints in both the as-welded and stress relieved conditions.

1.1 Thesis Arrangement

The contents of this thesis are arranged in seven chapters. Chapter 1 discusses the main objectives of this work, as well as a general introduction to the research.

Chapter 2 presents an up-to-date summary of the literature related to the research. The first section provides a background on the LFW technology and the welding process parameters. The second section of the review gives an overview of the titanium alloys and literature data published on the LFW of the titanium alloys subjected to analysis in this study.

Chapter 3 provides detailed information related to the experimental procedures, equipment and methodologies used to satisfy the research objectives.

Chapter 4 comprises of the microstructures, hardness, tensile and fatigue test results obtained for the similar and dissimilar weld combinations in the as-welded and stress relieved conditions.

Chapter 5 presents an overview of the results obtained in this study, as well as a comparative discussion based on the literature data published.

Chapter 6 concludes the findings of this research work.

Finally, Chapter 7 presents a summary of the contributions of this research to science.

Chapter 2 : Literature Survey

2.1 Linear Friction Welding

An American company, Caterpillar Tractor Co., patented a linear reciprocating machine for welding steel in 1969, although no further publications were reported on the topic during the following decade. TWI (The Welding Institute) designed and built an electromechanical equipment prototype in the early 1980s and demonstrated the viability of the linear friction welding technique for metals [4]. One of the main advantages of LFW compared to the other friction welding processes is that an axial symmetry is not required and therefore complex parts involving curves can also be manufactured [10]. Linear friction welding proved to be an ideal process for manufacturing BLISK and it has been adopted in companies including MTU Aero engines, Rolls Royce, Pratt and Whitney, and General Electric [9, 11]. For expensive and critical parts, friction welding is cost-effective despite the large upfront investment required [11]. Moreover, the introduction of this welding technique into other more conventional applications requires novel solutions that are still being developed, mainly to reduce the cost of equipment. Figure 2-1 illustrates a typical compressor disk and blade assembly using the LFW technique. Figure 2-2 shows the traditional slotted disk assembly and BLISK produced by LFW.

Figure 2-1 Disk and blade assembly

Figure 2-2 Illustrations of the mechanical attachment blade-disk (left side) and of a BLISK (right side) [4]

Studies of LFW have been conducted on a wider range of materials. For example, studies conducted on the Ni-based superalloys show that optimized process parameters and controlled crystal orientations can produce well-integrated welds [11-12]. In addition, intermetallics such as nickel aluminides [13] and gamma titanium aluminides [14-15] have been successfully welded using the LFW technique. Further, LFW can also be used to join dissimilar materials, Examples include Al to Cu [16-17], zircaloy to stainless steel [18], or polycrystalline superalloy to single crystal superalloy [19]. This allows optimizing the component parts separately. Al-based metal matrix composites have also been successfully linear friction welded [20-21].

2.1.1 Theory of Linear Friction Welding

The friction welding as explained by Vairis states that “ Friction welding is a solid-state joining of the materials by an intimate contact through a plasticized layer” [22]. The material at the interface is heated by the friction when one part under axial pressure is moved relative to the other. Most of the energy by friction at the interface and the

surrounding shearing material is dissipated in the form of heat. The microscopic local seizure and rupture due to dry friction between the surfaces result in the rapid formation of a plasticized layer. During this process the contact surfaces adhere, shear and fracture and also new junctions are created which are again subjected to the same cycle generating additional heat. The plasticized layer is effectively prevented from melting due to the nature of the process because fluid cannot transfer the stress needed to generate heat and therefore no additional heat is generated when the material approaches this state. In this stage the material acts as a lubricant instead of generating frictional heat and thus leads to a self-regulating process [23].

Friction welding can be classified into four main categories: stir, spin, rotary and linear. For the friction stir welding (FSW) a non-consumable tool is used that does not soften during the operation. The tool is pressed and moved along the interface between the components to achieve welding. The tool softens components around the interface and mixes softened material from both component interfaces to form bonding (Fig.2-3 (a)) [24].

In spin or orbital (OFW) friction welding material in the vicinity of the interface softens and moves outwards as a result of spinning and pressure. Once the interface has a homogeneous layer of soft material, sufficient force is applied to join the parts together. The orbital friction welding is characterized by its ability to weld non-circular parts and both parts are rotated in tandem but offset to each other to create the necessary friction (Fig.2-3 (b)). Rotary friction welding (RFW) is an inertia friction welding process in which energy required for joining components extracted from the stored energy of a flying wheel. One of the components is held stationary while the other is attached to a rotating flywheel.

Once the components are in contact, the flywheel's energy is converted into frictional heat and used for welding (Fig.2-3 (c)) [24].

Figure 2-3 Process schematic of (a) Friction stir welding, (b) Orbital friction welding, (c) Rotary friction welding [24]

In the linear friction welding (LFW) process, one part is held stationary while the other part reciprocates as illustrated in Figure 2-4. The motion between parts generates frictional heat and causes the material at the interface to plasticize and soften. The final stage consists of applying a forge force for fusing the two parts together. A uniform heat profile is generated by the LFW over the RFW process due to consistency in velocity and strain rate of parts relative to each other [25].

Figure 2-4 Schematic diagram of Linear friction welding

2.1.2 Phases of Linear Friction Welding

LFW is a self-regulating process, which means that specific interface conditions must be satisfied in order for the process to continue through its phases. The machine operations required for the LFW can be divided into six separate stages [27].

Part clamping: The tooling is designed for withstanding the forces generated during the process and for holding the parts firmly. Preparing the samples with precise edges, sides and a tight fit between the tooling is a vital part of the process. This means that the tooling is tailored in such a way to accommodate specific geometries of the sample.

Datum and retract: The clamped parts are brought together by a small compressive force to determine the location of the parts and to set the machine datum to zero. A small separation is left between the parts by retraction.

Conditioning phase: The oscillation of one of the components is increased and stabilized over a certain period of time (usually very fast) and the components are brought together in a predetermined time under a small force.

Frictional phase: The compressive force is increased to a set level and heat is generated at the interface. As a result of shearing motion between the parts and applied force, metal at the interface plastically deforms and flows out of the weld in the form of flash. This loss of material causes shortening also known as the burn-off. This phase ends when the predetermined loss of length or burn off distance is attained. The next stage of the process may also be initiated after the frictional phase continues for a predetermined time (burn-off time) or numbers of oscillations (burn-off cycles).

Forge phase: During this final stage, the amplitude is reduced to zero over a predetermined time period in order to ensure good alignment. A forge force is instantly applied to consolidate the joint. Applied forge force can be higher or the same as the frictional force.

Release phase: The welded parts are released from the clamp and removed from the machine [27].

Controlled loading is one of the parameters always used to carry out the process. For example, the load is controlled throughout the welding process when using a burn-off distance, but the burn-off is also monitored (although not controlled) and the next phase (forge phase) is triggered at a set burn-off distance. Similarly, the load is monitored throughout the welding process with burn-off time or cycles and the amount of time or cycles defines the transition to the next step. A schematic diagram of parameter traces obtained during the welding process is illustrated in Figure 2-5. The diagram defines a

number of input variables. Burn-off is defined as the length loss that occurs as the process goes on, while upset is the total length loss measured after the weld is produced [27].

Figure 2-5 Schematic diagram of the parameter traces obtained during linear friction welding process [27]

The friction phase of LFW can be further broken down into four distinct phases that are determined by the metal behavior [28]. The first phase is regarded as the initial phase in which columbic friction produces heat between rubbing surfaces (contact surfaces). As the heat increases, asperities between the surfaces soften and deform, increasing the true area of contact between the parts. The contact area increases to 100% of the cross-sectional zone during the second phase known as the transition phase. This transition is followed by an increase in the force required to oscillate the parts. In the third phase or known as the equilibrium phase, shear forces will reach a steady-state value and significant axial shortening occurs along with the generation of flash. The plastic zone develops progressively larger during this phase [24]. The final stage is known as the deceleration

phase or the forging phase in which the relative motion between the parts is ceased and aligned. The desired shortening is achieved by the end of this stage. Once the parts are aligned and brought to rest, axial pressure is increased to consolidate the weld [24]. Figure 2-6 illustrates the schematic representation of the four frictional phases during the LFW process.

Figure 2-6 Four phases of linear friction welding [4]

The slow-motion images of the four critical phases taken during a linear friction welding process conducted by TWI (the welding institute) are illustrated in Figure 2-7. The initial phase of the process shown in Figure 2-7 (a) and the deformation of asperities with an increased area of contact in the transition phase in Figure 2-7 (b). During phases 2 and 3 (Fig.2-7 (c)), the interface plasticizes and a significant amount of flash is generated and expelled outwards along the four sides. The final forge phase, which occurs after the required shortening is achieved is shown in Figure 2-7 (d), the weld is consolidated by applying a forge force [10].

Phase 1

Phases 2

Phases 3

Phase 4

Figure 2-7 Linear friction welding process [25]

2.1.3 Specific Power Input

Considering the self-regulating nature of the friction welding process and the phase changes being triggered by the physical changes, a sound weld (defect-free) is produced with a minimum specific power input value [23]. For Ti-64 linear friction welding, Varis and Frost [28] demonstrated that sound welds are produced when a specific power input parameter is exceeded. It was shown that the frequency, amplitude, and pressure have an effect on this parameter. In the present study, an amendment to the original expression proposed by Dalgaard et.al [5] is used as follows:

$$w \text{ (kW/m}^2\text{)} = \frac{F \times f \times \alpha}{2\pi \times A} \quad \text{(Equation 2.1)}$$

where the frictional force (F) measured in N, the amplitude of oscillation (α) is in mm, frequency of oscillation (f) is in Hertz, cross-sectional area (A) is in mm², and the unit of the power input parameter (w) is kW/m². The specific power input has an inverse relation with the cross-sectional area and directly proportionality to the product of effective velocity and the axial pressure applied. These parameters also influence the recrystallized β grains size formed in the weld center.

Figure 2-8 depicts the effects of frequency, amplitude, and pressure on the size of the recrystallized β grains in the weld center of Ti-64 similar welds. The size of the β grains at the weld center is quite important to quantify the weld quality [26]. This research work gives good insights into the influence of welding parameters on the interface temperature as the β grain growth directly depends on it. It was shown that an increase in frequency or pressure increases the size of prior β grains [26]. A study conducted by Attallah et al. [29] has shown that the prior β grain size actually decreases at high pressures, which has been

interpreted as a reduction in the peak welding temperature. This drop-in temperature was attributed to a large amount of flash expelled at high pressures causing large heat rejection and short welding times. Also, varying the frequencies and amplitudes revealed that sound welds (defect-free) are generated within a frequency range of 50-80 Hz and amplitudes between 2 to 3 mm [26] .

Figure 2-8 Effects of frequency, amplitude, and pressure on average prior β grain size in the weld zone of Ti-64 linear friction welds (Open markers represent the unsuccessful welds produced.) [26]

Figure 2-9 depicts the dependence of welding time on power input for different axial shortening lengths in Ti-64 welds. The graph shows an inverse relationship, as a higher power input generated by higher frequencies, amplitudes, or pressures requires a lower time to consolidate the weld. From the results, it can be concluded that a minimum value of specific power input is necessary for producing successful welds, this critical value of specific power input has not yet been modeled and is required to be empirically calculated for each material that needs to be welded. For Ti-64 similar welds this specific power input value is approximately about 2.4 kW, above which sound welds can be produced. This parametric study clearly demonstrates that sound welds can be obtained by modeling the critical value of specific power input for welding [28].

Figure 2-9 Dependence of weld success on specific power input; open markers designate unsuccessful welds (Ti-64 linear friction welds) [26]

Weld integrity also depends on the thickness of the TMAZ region and the amount of material being extruded [26]. Varis and Frost [28] showed that welds obtained with satisfactory bonding had expelled a considerable amount of flash from all sides due to the applied force and oscillatory movement. Minimizing the peak temperature at the interface is an important objective specifically for joining dissimilar materials (e.g. Al to Cu). The temperature at the interface can be controlled to an extent by varying the welding process parameters. The formation of intermetallics is usually a major obstacle. The purpose of minimizing the temperature of the welding is therefore to avoid the formation of detrimental intermetallic phases in the weld interface [27].

In summary, linear friction welding has many advantages over conventional welding processes. It is capable of producing high-quality joints for many engineering applications in a robust, repeatable and fast manner. Despite these advantages, the number of industrial applications are limited and generally confined to the aerospace industry. Most

of the articles published related to friction welding are primarily focused on titanium and its alloys although exceptions are [22] there. The friction welding process offers many advantages over fusion welding techniques by preventing solidification problems such as porosity, hot cracking, and segregation, etc. In most cases, a severe deformation at the weld center results in a refined microstructure that improves the strength compared to the base metal [26].

2.2 Titanium Alloys

Titanium alloys are used in various airframe sections, undercarriage parts, skins, engines and many other aircraft components that require high creep and corrosion resistance, high strength, high fracture toughness and resistance to fatigue. Compared to other lightweight materials such as Al and Mg alloys, titanium alloys show much better mechanical properties. Even though titanium exhibits good properties, applications of titanium are limited in industrial applications due to their relatively high cost. Component manufacturing is also an expensive process due to the high strength of titanium compared to other industry-relevant metals such as Al and steel. However, all these costs can be justified for components requiring high performance, especially in aerospace-related applications [23].

Pure titanium and most of the commercial titanium alloys undergo an allotropic phase transformation from low-temperature α phase (a hexagonal close-packed (HCP) structure) to high-temperature β phase (body-centered cubic (BCC) structure). The temperature where this transformation takes place is called β -transus temperature and it

varies with alloy addition. For commercially pure titanium the β -transus temperature is approximately 882 ± 2 °C, and for Ti-64 alloy this temperature is about 1000 °C [30, 39].

2.2.1 Alloying Elements

The alloying elements are classified according to the solubility of α and β phases as α -stabilizers, β -stabilizers, and neutral stabilizers. Elements that increase the β -transus temperature i.e. α phase is stable at higher temperature, are categorized as α stabilizers, examples include Al, Sn, Zr, O, N, and C. Aluminum is the most widely used element in Ti due to its large solubility in the α and β phases and its capability to increase the transformation temperature. Al, acts as a substitutional element and O, C and N act as the interstitial elements of this group. Figure 2-10 illustrates the classification of the alloying elements of titanium based on the influence of the β transus temperature.

Figure 2-10 a) Classification of the alloying elements of titanium according to their influence on the β -transus temperature [30]

β stabilizers lower the β transus temperature. They are classified into two subcategories: β -isomorphous and β -eutectoid elements. Elements such as V, Mo, Nb, Ta, and Re belong to the category of β -isomorphous with high solubility in β -Ti. β -eutectoid elements such as Fe, Mn, Cr, Co, Ni, Cu, Si, and H have restricted solubility in β -titanium and form intermetallic compounds by the eutectoid transformation of the beta phase. The hydrogen concentration in titanium is usually limited to a maximum of ~125-150 ppm to prevent embrittlement. There are also neutral elements (Zr, Sn, Si) that have a minor effect on the β transus temperature, however, they primarily strengthen the α phase [30, 39].

2.2.2 Alloy Classification

Titanium alloys can be classified as α , near- α , $\alpha+\beta$, metastable β , and β alloys based on the position in the β isomorphous phase diagram (Fig.2-11).

α -alloys: Four grades of commercially pure titanium belong to the α alloys category (Ti-O alloys with O varying between 0.18-0.40 wt.%). These alloys exclusively consist of α stabilizing elements or neutral stabilizers. Solid solution strengthening and cold working are usually applied due to their poor response to heat treatments [30-31].

Near- α alloys: These alloys are defined as having β stabilizers present up to 2 % wt. forming less than 10 % vol. of the β phase. For jet engine and rocket propulsion components operating up to 500-600 °C, near- α alloys are employed. The near- α titanium alloys typically possess high creep resistance with good fracture toughness and ductility. Ti-6242 is a good example of a near- α alloy used as a primary structural component in the hot section of aircraft turbine discs, turbines and sheet metal parts such as air burners [30-31].

$\alpha+\beta$ alloys: In these alloys, both α and β phases are stable at room temperature. $\alpha+\beta$ alloys are used for most aircraft structural applications. Ti-64 alloy is the most common example of an $\alpha+\beta$ alloy. Ti-64 has high creep resistance and toughness derived from the α -Ti combined with high strength and fatigue resistance that's derived from the β -Ti. A wider range of microstructures with different mechanical properties can be achieved using this alloy by various thermal and thermomechanical heat treatments [30].

Metastable β alloys: Metastable β alloys consist of 10-15 wt.% of β stabilizers and are characterized by good fatigue resistance, formability, toughness and high levels of strain hardening. These alloys can be further divided into high and heavily stabilized β alloys with regard to the high and low content of α respectively. Mechanical properties of metastable β alloys are highly influenced by thermal and thermomechanical processes due to the formation of complex precipitates.

The metastable β alloys are highly susceptible to thermal and thermomechanical heat treatments and the stable β alloys consist of about 30 wt.% β stabilizers and composed of a single β phase [30-32]. Figure 2-11 depicts a phase diagram for titanium alloys with the amount of α/β stabilizing elements in X and temperature in the Y-axis [33].

Figure 2-11 Classification of titanium alloys (phase diagram) [33]

The divisions between the α , near- α , $\alpha+\beta$, metastable β , β can also be understood by the level of the α and β stabilizers expressed as Al ([Al]eq.) and Mo ([Mo]eq.) equivalence (shown in equation 2.2 and 2.3). In order to avoid the embrittlement by precipitation of Ti_3Al , [Al]eq. is limited to 9 wt.%, while [Mo]eq. content of at least 10 wt.% is required to stabilize the β upon water quenching titanium alloys [31-32].

$$[\text{Al}]_{\text{eq.}} = [\text{Al}] + 0.33 [\text{Sn}] + 0.17 [\text{Zr}] + 10 [\text{O} + \text{C} + 2\text{N}] \quad (\text{Equation 2.2})$$

$$[\text{Mo}]_{\text{eq.}} = [\text{Mo}] + 0.2 [\text{Ta}] + 0.28 [\text{Nb}] + 0.4 [\text{W}] + 0.67 [\text{V}] + 1.25 [\text{Cr}] + 1.25 [\text{Ni}] + 1.7 [\text{Mn}] + 1.7 [\text{Co}] + 2.5 [\text{Fe}] \quad (\text{Equation 2.3})$$

2.2.3 Ti-6Al-2Sn-4Zr-2Mo (Ti-6242)

This alloy is suitable for relatively high-temperature applications and therefore is used for compressor disk/blades exposed to elevated temperatures. Mechanical properties

of this alloy are affected by the grain size, volume fraction and morphology of α and β phases, and distribution of the strengthening precipitates [34-35].

The final microstructure of the alloy depends upon the associated thermal and thermomechanical history [36]. Two important microstructures observed in conventional near- α alloys including Ti-6242 are the Widmanstätten and bimodal structures. Optical microscopy images of both microstructures observed in Ti-6242 are illustrated in Figure 2-12 (a) and (b) [37]. Generally, the Widmanstätten structure is formed by solution heat treatment above the β transus and bimodal structure forms below the β transus [38]. Studies conducted on the mechanical properties of the Widmanstätten structure show superior strength and optimum resistance against stress corrosion cracking. While, the bimodal microstructure has high yield strength and ductility. The thermomechanical process to achieve both bimodal and Widmanstätten microstructures is shown in Figure 2-13.

Figure 2-12 Optical images of Ti-6242 a) bimodal b) Widmanstätten [37]

Figure 2-13 Thermomechanical processing for a) bimodal b) Widmanstätten (I: Homogenization and primary hot rolling, II: secondary hot rolling, III: solution annealing and IV: aging). [37]

The α -stabilizing element, Al increases the tensile strength and creep strength while reducing the density of the alloy; Sn is used as a solid solution strengthener along with Al to improve the strength without causing embrittlement. Zr is also used as a solid solution strengthener in order to increase the strength at low and intermediate temperatures. Mo is the β stabilizer and it increases the heat treatment response of the alloys by refining the transformed β and also subsequently increases short-term high-temperature strength. At elevated temperatures, Si precipitates as silicides on dislocations and thereby hindering dislocation climbs. Dislocation climbs have a key role in the creep deformation mechanism. In addition, zirconium (4-5 wt.%) present in Ti-6242 alloy promotes uniform silicide formation by reducing the solubility for silicon. Therefore adding Si of (0.1-0.2 wt.%) with Zr extends the service temperature of Ti-6242 alloy from 450 °C to 540 °C [30-31].

2.2.4 Ti-6Al-4V (Ti-64)

Ti-64 is one of the very first titanium alloys developed and is by far the most popular titanium alloy. Now, more than 50% of all titanium alloys used are of this composition. The aerospace industry is the biggest consumer (more than 80 %) of Ti-64.

Ti-64 is an $\alpha+\beta$ alloy used in the aerospace industry, especially for engine and airframe components [8]. The manufacturing process for these applications involves machining oversized ingots, forgings and extrusions. Expenses for such conventional machining processes are very high. Linear friction welding technology reduces the amount of material required by joining small workpieces and then machining them to desired dimensions. This alloy is comprised of 6 % wt. Al and 4 % wt. V as alloying elements. Aluminum stabilizes the HCP α -phase while vanadium stabilizes the BCC β -phase at lower temperatures, and therefore Ti-64 exhibits a balance of both phases at room temperature.

The equilibrium microstructure of Ti-64 at room temperature consists of the α -phase with some retained β -phase. The β -transus defines the temperature above which the equilibrium microstructure will consist of only β -phase and this temperature is about 995 °C for this alloy. Different morphologies can be developed, depending on the temperature history of the alloy. The β -phase mainly transforms into a globular α -phase by slow cooling from above the β -transus temperature. Increasing the cooling rate accelerates the α nucleation rate in the β grain boundaries. The cooling rate determines the length and width of these α platelets formed. If quenched the β grain transforms fully or partially into a martensite α structure, denoted as α' [50]. Ahmed and Rack [60] reported that Ti-64 requires a minimum cooling rate of 410 °C/s for a complete martensite structure to form. A trace amount of β is always present in the martensite and attributed to the 4% V (β stabilizer) its presence is sufficient enough to drop the temperature for the end of the martensitic transformation (M_f) below room temperature.

To illustrate the formation of α upon cooling at a constant composition of 6 % of Al is shown in Figure 2-14. The darker regions (in the optical micrographs) represent the

β -phase that was left between the α -plates that have formed. The final microstructure consists of parallel plates of α delineated by the β phase between them. The etching reveals a high angle boundary existing between the α crystals. Such microstructural morphologies in which parallel plates formed having a relation with the phase which they have formed is known as the Widmanstatten structure [40].

Figure 2-14 Schematic illustration of the formation of a Widmanstatten structure in a Ti-64 [16]

The bimodal microstructure can be produced by a mill annealing process with balanced mechanical properties such as good fatigue strength and fatigue resistance by relying on the grain refinement during the process. This alloy is then cooled from either β phase and/or $\alpha+\beta$ region to form an α' martensite or Widmanstätten structure (Fig. 2-15).

This microstructure structure is transformed by subsequent working and low-temperature aging to form equiaxed α and lamellar transformed β grains [23].

Figure 2-15 Widmanstätten microstructure with α plates oriented according to β matrix (Ti-64) [41]

2.3 Linear Friction Welding of Titanium Alloys

2.3.1 Ti-6Al-4V

Linear friction welding studies conducted by Wanjara and Jahazi [26] on Ti-64 revealed that certain critical conditions need to be satisfied at the interface and adjacent regions for producing defect-free welds. Microstructural characterization at the weld center shows that this interface region is exposed to temperatures above β transus ($>995^{\circ}\text{C}$) combined with hot deformation and rapid cooling. The weld center consisted of a β grain structure that had a Widmanstätten transformation microstructure and the TMAZ region formed close to the weld interface was composed of highly deformed bimodal microstructure along with the broken β -phase particles. A similar study by Karadge et al. [42] reported fine lamellar α' martensite developed in the weld center and the TMAZ region consisted of heavily deformed primary α grains with some recrystallized α suggesting that

material did not exceed the β -transus in the TMAZ region. The α' martensite formed at the weld center was transformed into acicular $\alpha + \beta$ after the post-weld heat treatments (PWHT). Daymond and Bonner [43] and Preuss et al.[44] reported that Ti-64 and Ti-4Al-4Mo-2Sn-0.5Si (IMI550) exhibited large tensile residual stresses with a significant hydrostatic tensile stress component in the weld region and these stresses were relieved after the PWHT.

The hardness measurements performed by Wanjara and Jahazi on Ti-64 in the as-welded condition showed peak hardness values at the weld center compared to the base metal. This increase in hardness values attributed to the grain refinement and the phase transformations that occurred during the process [26]. Tensile results of the as-welded Ti-64 exhibited comparable results to that of a grade-5 Ti-64 alloy (as per ASTM specifications) since the fracture occurred in the base metal, indicating that the weld center and the surrounding TMAZ regions are stronger than the base metal region. Also, it was shown that the tensile mechanical properties of the Ti-64 welds depend on the LFW process parameters. Ti-64 samples welded using a frequency of 50 Hz, an amplitude of 2 mm, a pressure of 50-90 MPa, and a shortening of 2 mm showed excellent tensile properties that surpassed the base metal [26]. Fatigue tests performed by Flipo et al. [25] on the as-welded Ti-64 samples reported an average drop in performance of 50.4% for high cycle fatigue (HCF) compared to the base metal. Similarly, for low cycle fatigue (LCF), tests showed a drop-in performance of 27.9 %. The maximum stress amplitude applied in the study for HCF was 700 MPa and for LCF 540 MPa.

2.3.2 Ti-6Al-2Sn-4Zr-2Mo

Linear friction welding studies on Ti-6242 show comparable microstructural characteristics as reported in Ti-64 welds. A study conducted by García et al. [45] on the strength and fatigue properties of LFW Ti-6242 showed that the weld center was composed of α' martensite, as shown in Figure 2-16. The TMAZ region was characterized by highly thermomechanically deformed grains oriented in the weld direction. Similarly, Dorick et al. [46] reported intragranular precipitation of α' lathes in the weld center, which was transformed into acicular $\alpha + \beta$ microstructure after the PWHT.

Figure 2-16 Ti-6242 a) Heat affected zone containing weld center and TMAZ b) Acicular α' martensite weld center [45]

Hardness results showed an increase of 27% hardness in the weld center compared with the base metal [45]. Also, TMAZ hardness values were slightly lower than the weld center hardness. The tensile properties showed comparable results with the base metal. The fracture location was in the base metal far away from the weld center, suggesting that the weld center and TMAZ regions are stronger than the base metal. Additionally, DIC analysis performed on the as-welded Ti-6242 showed an early plastic activity in the weld center at yielding possibly due to the residual stresses present [45]. Post weld heat treatments have

shown that residual stresses are significantly reduced and improved mechanical properties [47]. Fatigue studies conducted on Ti-6242 showed HCF (10^5 cycles) for stress amplitudes ranging between 600 to 625 MPa. Fatigue Fractographic Analysis conducted for both the cases revealed characteristic features of a ductile material. The surfaces consisted of microcracks, striations, and smooth cleavages [45].

Chapter 3 : Experimental Procedure

3.1 Materials

Two titanium alloys, an $\alpha + \beta$ alloy, Ti-6Al-4V and a near- α Ti-6Al-2Sn-4Zr-2Mo were the subjects of this study. The optical microstructure of the as-received Ti-64 base metal is shown in Figure 3-1. The microstructure can be referred to as a duplex or bimodal microstructure, where light grains in the microstructure are the primary α and the dark phases are $\alpha + \beta$ lamellar colonies. Ti-64 alloy was received in mill annealed (hot rolled) condition and the as-received microstructural observations confirm the processing history of the material. Grains are oriented parallel to the rolling direction indicated by the arrow in the image.

Figure 3-1 Optical microscopy of Ti-64 base metal (as-received condition)

Ti-6242 was also received in mill annealed condition. As revealed by the optical microscopy (Fig. 3-2), Ti-6242 had a bimodal microstructure with primary α grains and transformed β grains, where transformed β is composed of randomly oriented α/β lamella.

Figure 3-2 Optical Microscopy of Ti-6242 base metal (as-received condition)

The nominal compositions (wt.%) of Ti-6242 and Ti-64 alloys as specified by the manufactures are shown in Table 3-1.

Table 3-1 Nominal compositions of starting materials (wt.%)

	Al	V	Mo	Si	Zr	Sn	Ti
Ti-64	5.5-6.75	3.5-4.5					Bal.
Ti-6242	5.5-6.5		1.8-2.2	0.6-.13	3.6-4.4	1.8-2.2	Bal.

Both Ti-64 and Ti-6242 alloys were received in the form of rectangular bars and these were sectioned to obtain weld coupons of size 12 mm in width (W), 24.5 mm in height (H) and 66 mm in length (L) (Fig.3-3 (a)). Prior to the LFW welding process coupons were machined to match the required dimensions. The contact surfaces of the coupons were prepared by grinding with an 800-grit SiC paper and thereafter cleaned using acetone to remove any contaminations. The sample geometry and oscillating direction of coupons during the LFW process are illustrated in Figure 3-3 (b).

Figure 3-3 (a) Sectioned coupon (b) Sample geometry and oscillation during LFW

3.2 Welding Equipment

In this work, an MTS linear friction welder was used for the welding process. The machine is hydraulically operated and consists of two main actuators. One of the actuators enables reciprocating motion while the other provides the upsetting force required. The actuator providing reciprocating motion is an in-plane device that oscillates the lower workpiece horizontal direction. The forge actuator transfers a downward force through the stationary top workpiece. The maximum tooling space available in the machine is about 13 mm x 26 mm in cross-section to be held in the clamping blocks one on top of the other and both the samples are manually loaded by the operator. The MTS software records all the parameters such as actuator position command and feedback, velocity, acceleration, load command and feedback at a rate of 1024 readings per second during the welding process. An image of the MTS linear friction welding machine used in this study is shown in Figure

3-4 (a) and the coupons clamped prior to welding in Figure 3-4 (b). Specifications of the MTS linear friction machine are given in Table 3-2. The forge actuator can generate a maximum load of 90 kN and a maximum friction force of 50 kN by the in-plane actuator.

Figure 3-4 (a) MTS Linear friction welding machine (b) Clamped coupons prior to welding

Table 3-2 MTS linear friction welder specifications

	Specifications
Forge Actuator:	
Forge Load	90KN (20,250lbs)
Displacement	$\pm 6\text{mm}$ (1/4in.)
In-Plane Actuator:	
Friction Force	50KN (11,2550lbs)
Displacement	$\pm 10\text{mm}$ (3/8in.)
Amplitude Range	$\pm 5\text{mm}$ (3/16in.)
Frequency Range	125Hz
Weld Duration Time	10sec

The welding parameters selected were based on the results reported by Wanjara and Jahazi [26] on the linear friction welding process development for Ti-64 similar weld to obtain quality welds (defect-free). Table 3-3 shows the linear friction welding parameters used in this study for the similar and dissimilar weld combinations. The macroscopic images of the welded coupons are shown in Figure 3-5, displaying the flash generated by the welding process.

Table 3-3 Welding Parameters of welds (SW: Similar Weld, DW: Dissimilar Weld)

Sample	Amp. (mm)	Freq. (Hz)	Pressure (MPa)	Upset Dist. (mm)	Specific Power Input (kW/mm²)
Ti64-Ti64 (SW)	2	50	90	2	4.87
Ti6242-Ti6242 (SW)	2	50	90	2	4.87
Ti64-Ti6242 DW	2	50	90	2	4.87

Figure 3-5 Linear friction welded samples (as-welded condition) (a) Ti-6242 (SW) (b) Ti-64 to Ti-6242 (DW) (c) Ti-64 (SW) (SW: Similar Weld, DW: Dissimilar Weld)

3.3 Heat Treatment

For the selected tensile, fatigue, and metallography samples, stress relief annealing heat treatments were performed to lower the residual stresses developed during LFW and possibly during the machining process while preparing the surfaces. Titanium alloys are very reactive to oxygen and therefore a vacuum furnace has been used for the stress relieving heat treatments. The furnace, as shown in Figure 3-6, consists of a long ceramic tube and thermocouples placed in the vicinity of the samples to verify the exact temperature.

Figure 3-6 Ceramic tube vacuum tube furnace

Prior to heat treatments, all samples were cleaned using acetone and dried to remove any contaminations present. The samples were placed in the middle of the ceramic tube for uniform heating. Samples were cooled to ambient room temperature under vacuum after the heat treatment cycle. Frankel et al. [47] showed that stress relief annealing applied for Ti-64 and Ti-6242 similar welds lowered the residual stresses by 50 and 100 MPa, respectively (measurements using high energy synchrotron X-ray diffraction). Stress relief annealing temperatures for similar and dissimilar weld combinations were selected based on this literature data (Table 3-4). For the dissimilar welds, there has not been any study so far, and hence an intermediate temperature between the similar welds was chosen. During

each heat treatment cycle, thermocouples showed a variation of ± 5 °C deviation from the desired temperature values and the samples were furnace cooled to room temperature at a rate of 0.06 °C/sec. The stress relieved samples were then cleaned using acetone prior to mechanical testing.

Table 3-4 Heat treatment cycle parameters (SW: Similar Weld, DW: Dissimilar Weld)

Joint	Heat Treatment (°C)	Time (hrs.)	Pressure (Pa)
Ti64-Ti64 (SW)	700	2	1.6×10^{-2}
Ti64-Ti6242 (DW)	750	2	1.6×10^{-2}
Ti6242-Ti6242 (SW)	800	2	1.6×10^{-2}

3.4 Metallography

The metallographic samples were taken from the center of the welded coupons. The polishing procedure developed for the samples is shown in Table 3-5. During polishing, the samples were carefully rinsed using water and drained at the end of each stage. These materials required an extensive and sensitive polishing procedure for obtaining a mirror finish. For the optical microscopic examination of samples, etching was performed using Kroll's reagent (2 ml hydrofluoric acid (HF) + 6 ml nitric acid (HNO₃) + 100 ml water (H₂O)). Etching time varied between 6-15 seconds according to the zones (base metal (BM)/weld center (WCZ)/thermomechanically affected (TMAZ)) that were to be revealed. Optical microscopy was carried out using an inverted optical microscope (Olympus GX71) equipped with AnalySIS Five digital image analysis software. Backscattered electron (BSE) imaging for the metallographic samples and secondary electron (SE) imaging for

fracture surfaces (tensile and fatigue) was performed using the Tescan Vega-II XMU scanning electron microscope (SEM).

Table 3-5 Grinding and polishing of metallographic samples

	Surface	Abrasive	Speed (rpm)	Load (N/sample)	Lubricant	Time (min)	Direction (rel. to cloth)
1	SiC Paper	220 grit	300	20	Water	0.5	Same
2	MD-Piano	1200 grit	300	25	Water	0.5-1	Same
3	MD-Largo	9 μm	150	30	DP-Blue	0.5-1	Same
4	MD-Chem	0.05 μm	150	30	OP-S	15	Opposite

*If the surface is not ready steps may be repeated accordingly.

3.5 Mechanical Testing

3.5.1 Microhardness Testing

The microhardness measurements were performed using DuraScan80 (Fig.3-7 (a)) with a fully automated testing cycle (focus, load, stage, and measure). A load of 500 g was used for the measurements with a dwell period of 15 seconds at an indent spacing of 0.1 mm. The line microhardness measurements were taken across the sample covering both BM and WCZ regions. For a selected area of $7.6 \times 2.6 \text{ mm}^2$ across the WCZ, a 3D hardness map was also generated for each set of the as-welded and stress relieved samples. The mapping for each sample set took about 5-6 hrs. to complete. Figure 3-7 (b) shows an example of the hardness measurements conducted on the Ti-64 similar welds.

Figure 3-7 (a) Microhardness testing machine setup (DuraScan 80) (b) Hardness measurements

3.5.2 Tensile Testing

For the tensile and fatigue tests, three of each fatigue and tensile samples were extracted from one set of the welded coupon using an Electrical Discharge Machining (EDM) machine. The schematics of the test samples used for EDM wire cutting are shown in Figure 3-8 (a) and (b). The average surface roughness throughout the samples after the final machining was about 5 μm .

The tensile samples with a 25 mm in gauge length (sub-size geometry), 6 mm in width, and 3.62 mm in thickness were machined in accordance with ASTM E8-16M specifications (Fig.3-8 (c)) [49]. The selected tensile samples in the as-welded and stress relieved conditions were tested at room temperature using a 250 kN MTS 810 material

testing system. The MTS machine is equipped with an Aramis 3D deformation measurement system (digital image correlation) for measuring the macroscopic strain distributions. Prior to the tensile tests, all the samples were painted using a high contrast random speckle pattern of black dots on a white background. The random speckles are generated by spraying black or white paint using a refined airbrush (black speckles on a white background or vice versa). The scale of the random speckles is critical and is typically very small, 1-100 microns [50]. The functionality of the Aramis system depends directly upon the quality of this pattern. Pattern quality was inspected prior to testing in order to ensure the strain recording along with the gauge length/region of interest (ROI). After quality evaluation of the coated pattern, tensile tests were performed with a crosshead speed of 0.05 mm/min with a strain acquisition rate of 3 frames/second. For the displacement measurements by the extensometer a reflective tape was attached to both ends of the gauge length on the back of the painted surface of the samples. After the tensile tests, broken samples were collected for the fractographic analysis. An example of the sample labeling used in this study is shown in Figure 3-9 (first set of welded coupons). The total number of tensile and fatigue samples extracted from Ti-64 and Ti-6242 (similar and dissimilar combinations) welded coupons are shown in Table 3-6.

Table 3-6 Total number of tensile and fatigue samples

Sample	Tensile	Fatigue
Ti-64 (SW)	24	24
Ti-6242 (SW)	21	21
Ti-64 - Ti-6242 (DW)	15	15

Figure 3-8 (a) EDM cuts of the as-welded coupon (b) After the EDM cutting (c) Tensile sample (d) Fatigue sample

Figure 3-9 EDM cut sample labeling (e.g., first set)

The experimental setup for performing a stereo correlation with a standard Digital Image Correlation (DIC) system is shown in Figure 3-10 (a) and (b). The system consisted of two 12 Megapixel digital cameras, to provide a 3D view of the sample, light sources were used to adjust the amount of light the sample is exposed to, a calibration panel to adjust the 3D view of the cameras and to synchronize them with each other, and a workstation with Aramis to post-process the results and analyze the data.

The DIC system with Aramis software is a non-contact optical technique capable of measuring full-field 2D or 3D surface deformations. DIC compares two digital images: one is the target image (specimen image in a reference state) and the other is the source image (specimen image after deformation). For the digital image correlation algorithm to work, an intensive calibration process of the two cameras is required to obtain accurate information about the placement and characteristics of the two cameras. The DIC algorithm searches for one-to-one correspondence of the points (pixels) in both images by matching all the pixel intensities in an area with a unique image pattern.

Figure 3-10 a) Schematic top view of the experimental setup for performing a stereo correlation with a standard DIC system [50] b) Sample, camera frame and load line aligned to gravity using a laser level (red laser lines)

3.5.3 Fatigue Testing

Fatigue samples with a standard gauge width of 6 mm and a thickness of 3.62 mm were machined in accordance with the ASTM E466-15 specifications (Fig.3-8 (d)) [51]. Prior to the tests, all fatigue samples were stress relieved to lower the residual stresses present. Fatigue tests were performed using a 250kN MTS 810 machine equipped with an

extensometer to measure the axial displacements. All fatigue tests were performed at ambient room temperature conditions. The top part of the MTS machine was held stationary while the lower part was subjected to vertical reciprocating motion and thus transferring a periodic cycle of tensile and compressive loads in the fatigue sample. Fatigue testing was performed by varying maximum stress amplitudes ranging from 300 to 1100 MPa, covering both low and high cycle fatigue with a stress ratio ($R=\sigma_{min}/\sigma_{max}$) of 0.1. Prior to the fatigue tests, a reflecting tap was attached to the ends of the sample just after the grip between the gauge length for the axial displacement measurement by the extensometer. Broken sample pieces were collected for the fractographic analysis. A brief summary of the experimental procedure followed in this study is shown in Figure 3-11.

Figure 3-11 Experimental Procedure Summary

Chapter 4 : Results

4.1 Ti-6Al-4V Similar Weld

A welded coupon of Ti-64 similar joint was shown previously in Figure 3-5 (c). There were no visible defects observed on the weld interface of the bulk material. All the similar welds of Ti-64 had a resemblance to this sample, with flash extruding along all four sides and the appearance of the flash remaining consistent. This similarity of the flash could be due to the fact that friction welding parameters did not vary. The flash layer comprised of plastically deformed material was found to have a longer length of flash along the oscillatory direction (parallel to sample height as compared to sample width). Additionally, flash along the sample height generally showed little evidence of ridges, while that along the sample width was usually comprised of a series of ripples due to the reciprocating motion in the direction of the height.

4.1.1 Microstructures

Figure 4-1 (a) and (b) show the optical microstructures (overview) of the as-welded and stress relieved samples, respectively. Regions of the metallographic sample are divided into the base metal (BM), thermomechanically affected zone (TMAZ), weld interface (WI) and weld center zone (WCZ) (for low magnification image). In the overview images, the magnified areas of these regions are labeled. The BM microstructure of Ti-64 shows primary α grains (light grains) and dark colonies of $\alpha+\beta$ (Fig.4-1 (c)). The optical microscopy of the stress relieved sample in the BM showed no significant differences in comparison with the as-received condition (Fig.4-1 (d)). TMAZ regions consisted of highly

deformed elongated α grains with intergranular β oriented in the weld direction (Fig.4-1 (e)). After stress relief annealing, minimal changes are observed in the TMAZ microstructures (Fig.4-1 (f)). The high magnification images in the WI suggests the formation of α' martensite, where the α' lathes are arranged as a brick-wall like structure showing the prior β boundaries indicating the formation of dynamically recrystallized fine β grains during the welding process (Fig.4-1 (g)). Due to the tempering of α' martensite during stress relief annealing, acicular $\alpha+\beta$ is formed at the WI (Fig.4-1 (h)).

Figure 4-1 Optical microstructures; as-welded (a, c, e, g) and stress relieved (b, d, f, h) Ti-64 (SW)

Figure 4-2 provides a more detailed characterization of the microstructure with BSE imaging using SEM. Similar to the optical microscopy images, the TMAZ regions of the stress relieved sample did not show any significant changes in microstructure compared to the as-welded condition (Fig.4-2 (a) and (b)). BSE images show clear indications of α' martensite in the WI of the as-welded sample (Fig.4-2 (c)) while the WI of the stress relieved sample consisted of acicular $\alpha+\beta$ formed by the transformation of α' during stress relief annealing (Fig.4-2 (d)). The increased β phase can be clearly observed in the BM microstructure of stress relieved sample compared to the as-welded condition (dark regions showing primary α and light regions, the transformed β) (Fig.4-2 (e) and (f)).

As-welded sample

Stress relieved sample

Figure 4-2 SEM (BSE) images of as-welded (a, c and e) and stress relieved (b, d and f) Ti-64 (SW)

4.1.2 Microhardness Test

The line hardness measurements performed across the WCZ of the as-welded sample are shown in Figure 4-3 (a) and the corresponding hardness profiles obtained in Figure 4-3 (b) with the X-axis showing the distance from the WI and the Y-axis, showing the hardness values. The average Vickers hardness across the BM ranged between 300-330 HV and the fluctuations in hardness measurements were minimal. In the TMAZ regions, the hardness values were slightly lower than the WI. A hardness map was also generated across a selected area covering both the BM and WCZ for a better understanding of the hardness distribution, as shown in Figure 4-4. The color bar showing the hardness distribution is shown next to the hardness map. Within the selected area of hardness mapping, a peak hardness value of 376 HV was measured in the WI.

Figure 4-3 Hardness profile of the as-welded Ti-64 sample

Figure 4-4 Hardness map of the as-welded Ti-64 sample

Similarly, hardness measurements were performed for the stress relieved sample and is shown in Figure 4-5. The average BM hardness was increased by 20-25 HV after stress relief annealing. Compared to the as-welded sample, the hardness distribution across the WCZ showed an opposite characteristic. However, hardness measurements in the TMAZ regions remain close to the results of the as-welded sample. The generated hardness map showed a maximum hardness value of 378 HV in the TMAZ region (Fig.4-6).

Based on the microstructural changes that occurred, the difference in the hardness distribution across the stress relieved sample can be explained. Due to the minimal changes in the TMAZ microstructure of the stress relieved sample, the hardness values remain close to the as-welded condition. While tempering of α' martensite during stress relief annealing resulted in the reduced WCZ hardness values. Furthermore, the BM hardness properties of the stress relieved sample improved due to the slow cooling used in this study, which allowed an ordered structure to be formed with an increased β phase, which is relatively a harder phase.

Figure 4-5 Hardness profiles of stress relieved Ti-64 sample

Figure 4-6 Hardness map of the stress relieved Ti-64 sample

4.1.3 Tensile Test

The room temperature tensile tests showed that the WCZ and the surrounding TMAZ regions of both the as-welded and stress relieved samples are stronger than the BM. Both the tensile samples failed in the BM away from the WCZ. The yield strengths of the as-welded samples ranged from 840 to 900 MPa, with a percent elongation of 13 to 15%. Tensile properties of the as-welded samples are comparable with a grade-5 Ti-64 alloy having a yield strength of about 880 MPa and UTS of 980 MPa, with a percent of elongation of 10 to 12% (as per ASTM specifications) [52].

The stress relieved samples exhibited higher values of UTS and yield strength compared to the as-welded samples at the expense of 2-3% of ductility. The UTS, yield strength and percent of elongation values obtained for the tested samples are shown in Table 4-1 and the corresponding engineering stress-strain curves in Figure 4-7.

Table 4-1 Tensile mechanical properties for linear friction welded Ti-64

Sample ID	UTS (MPa)	Yield (MPa)	EL. (%)
AW-2A	984	849	13.9
AW-3B	1003	905	14.8
AW-2C	999	845	14.8
SR-3A	1044	969	10.6
SR-6B	1060	915	11.8

Figure 4-7 Engineering stress-strain diagram for linear friction welded Ti-64

Local strain properties of the tensile samples were analyzed by the digital image correlation (DIC) technique. Figure 4-8 (a) shows the as-welded sample and its black and white speckles for performing the strain monitoring. The overlaid transparent green region depicts the region of interest (ROI) for the DIC. Figure 4-8 (b)-(d) shows strain maps for three critical stages (yield point, UTS and breakpoint) in the ROI during tensile testing. The macroscopic stress-strain curve along with the critical points corresponding to the strain maps is shown in Figure 4-8 (e). The WCZ and BM regions are marked beside the strain maps along with the color bars indicating the strain distribution. Information about macroscopic engineering stress and uniform strain for a uniaxial loading state up to necking can be achieved by conducting convectional tensile tests. Considering the heterogeneity of the welded sample, local mechanical properties need to be understood. An early plastic

activity observed in the as-welded sample near the WCZ at yielding and the macroscopic strain near this region reaches a maximum value of 1.03 percent. Also, the macroscopic engineering stress reaches UTS for a macroscopic strain of 6.92 percent. Afterward, localization takes place about 5 ± 2 mm away from the WCZ and at the breakpoint, a maximum local strain value of 53.53 percent is measured. The average strain measured in the WCZ at UTS is about 4.12 percent and it is lower than the average strain across the BM. There is no significant increase in strain values near this region between UTS and breakpoints.

Figure 4-8 (a) As-welded sample (with black and white speckle); (b)-(d) Strain maps; (e) Aramis stress-strain curve

Figure 4-9 shows the DIC results obtained for the stress relieved sample. The macroscopic strain reaches 0.7 percent at the yield near the WCZ (Fig.4-9 (b)). There is no significant increase in strain values near the WCZ relative to the strain maps of the as-welded sample at yield. This reduction in strain values attributed to the reduced residual stresses after stress relief annealing. After the strain localization, failure occurred 5 ± 1 mm away from the WCZ. At the breakpoint, a maximum strain of 28 percent measured in the BM region, which is lower than that of the as-welded sample due to reduced ductility (Fig.4-9 (d)). No significant changes in strain values between yield and breakpoints were observed near the WCZ, similar to the as-welded condition.

Figure 4-9 (a) Stress relieved sample (with black and white speckle); (b)-(d) Strain maps; (e) Aramis stress-strain curve

4.1.3.1 Fractographic Analysis

Secondary electron (SE) images of the tensile fracture surface for the as-welded and stress relieved samples are shown in Figure 4-10. The failure occurred in the BM region in both samples. The overview images of the samples (Fig.4-10 (a) and (b)) show no evidence of macropores, oxides, or other inclusions. High magnification images (Fig.4-10 (c) and (d)) of both surfaces confirm the ductile nature of the material. Dimples are present on the surfaces along with the areas of shear. Also, both the fracture surfaces consist of numerous micropores and smooth striations lines. These microvoids may open up due to an increased stress concentration and connects the crack propagation across the sample resulting in the final fracture.

Figure 4-10 SE images of (a, c) As-welded sample fracture surface; (b, d) Stress relieved fracture surface

4.1.4 Fatigue Test

Initial fatigue tests were performed using the as-welded samples and found that samples exhibited low cycle fatigue (LCF) failure for low-stress amplitudes. This failure can be possibly due to the internal residual stresses developed during LFW. In order to lower these internal residual stresses, all fatigue samples were stress relieved prior to testing (for stress relief annealing parameters are reported, in Table 3-4). Both the high cycle and low cycle fatigue tests were performed between the stress amplitudes ranging from 1100 to 300 MPa. Fatigue results of the tested samples are shown in Table 4-2.

Table 4-2 Fatigue test results; stress relieved Ti-64 (SW) samples

Sample ID	Parameter (Actual)				Results
	Stress (MPa)				Number of cycles to failure N_f
	Maximum σ_{max}	Minimum σ_{min}	Amplitude σ_{amp}	Average σ_{avg}	
3E	1100	110	495	605	10
4E	1050	105	472.5	577.5	420
2D	1000	101	449.5	550.5	2084
3D	900	91	404.5	495.5	5770
4D	800	81	395.5	440.5	11615
2F	800	81	395.5	440.5	11611
4F	700	70	315	385	14369
2E	600	61	269.5	330.5	27786
5D	600	61	269.5	330.5	25316
7E	500	51	224.5	275.5	133178
7D	450	45	202.5	247.5	258105
5F	300	30	135	165	834252

The fatigue testing process depends on two main parameters i.e., the amplitude and the microplastic deformation. The fatigue curve, also known as the Wöhler or S-N curve, represents the dependence of fatigue life on the number of life cycles. Figure 4-11 shows the S-N curve cycle obtained for stress relieved Ti-64 samples under symmetrical tension-compression loading and exhibits a continuous increase in the number of cycles ranging from 10 to 10^5 with decreasing stress amplitude/maximum stress values. Low cycle fatigue (LCF) failure of 10^2 to 10^4 cycles was achieved between stress amplitudes of 1100 to 600 MPa and a high cycle fatigue (HCF) failure of 10^5 cycles below 400 MPa.

Figure 4-11 S-N curve for stress relieved Ti-64 samples

Several sources suggest that bimodal or duplex microstructures generally ensure good fatigue life and endurance [53]. However, the mechanical response of this alloy is highly sensitive to the thermomechanical processing history, i.e., primary α to transformed β ratio, prior β grain size, α/β morphologies, and α grain size. Mentioned characteristics impact directly on the performance of the alloy directly, particularly for high cycle fatigue

(HCF) life. For example, Widmanstätten microstructure obtained by β recrystallization annealing exhibits maximum fracture toughness and fatigue crack growth resistance but this microstructure shows inferior fatigue performance. In order to prevent fatigue crack initiation, a bimodal or duplex microstructure is preferred. Progressive reduction of fatigue lifetime occurs during the transition to fine equiaxed, fine lamellar, coarse lamellar and coarse equiaxed structures, respectively [54].

4.1.4.1 Fractographic Analysis

Figure 4-12 (a) and (b) show overview images of the HCF and LCF fracture specimen surfaces, respectively. The HCF failure occurred in the vicinity of the WCZ and the LCF failure in the BM region close to the WCZ. High magnification images in the HCF sample's crack initiation site show numerous microcracks (Fig.4-12 (c)), while the final fracture site exhibits smooth striation lines and cleavages (Fig.4-12 (e)). Prior to failure, the LCF sample displayed a ductile necking characteristic and the surface consisted of numerous microvoids and smooth striation lines (Fig.4-12 (d) and (f)). The smooth striations found in the LCF sample's crack initiation site relative to the HCF sample are due to a shorter fracture time. On either surface, there were no traces of macropores, oxides, or any other inclusions detected.

Figure 4-12 SE images of (a, c, d) High cycle fatigue fracture surface; (b, e, f) Low cycle fatigue fracture surface

4.2 Ti-6Al-2Sn-4Zr-2Mo Similar Weld

Ti-6Al-2Sn-4Zr-2Mo (Ti-6242) coupons were linear friction welded with the same welding parameters used for Ti-64 similar welds. Similarly, the tensile and fatigue samples were sectioned from the welded coupons using an electron discharge machine (EDM) and then machined to match the required specifications. Similar to Ti-64 welds, the flash along the oscillating direction showed a higher number of ridges compared to the direction perpendicular to oscillation (previously shown in Fig.3-5 (a)). Prior to mechanical tests, all the fatigue samples and selected tensile samples were heat-treated under vacuum conditions for lowering residual stresses developed during welding (stress relief annealing parameters are reported, in Table 3-4).

4.2.1 Microstructure

Figure 4-13 (a) and (b) shows the overview images of the Ti-6242 as-welded and stress relieved specimens revealed by optical microscopy. The as-welded BM microstructure (Fig.4-13 (c)) indicates the presence of primary α and intergranular β grains, consisting of randomly oriented α/β lamella. The average grain sizes of β and primary α range between 20 - 30 μm and 30 - 50 μm , respectively. A significant reduction of intergranular α/β lamella is seen in the transformed β grains of the stress relieved sample compared to the as-received condition (Fig.4-13 (d)). TMAZ regions consisted of highly thermomechanically deformed α grains with intergranular β oriented in the weld direction (Fig.4-13 (e)). Also, in both the as-welded and stress relieved conditions TMAZ regions did not show any visible difference in microstructure (Fig.4-13 (f)). The WI of the as-welded sample shows α' martensite structure showing prior β boundaries indicating the formation of dynamically recrystallized fine β grains during the welding process (Fig.4-13 (g)). The WI consisted of $\alpha+\beta$ structure formed during stress relief annealing by the tempering of α' martensite (Fig.4-13 (h)).

Figure 4-13 Optical microstructures; as-welded (a, c, e, g) and stress relieved (b, d, f, h) Ti-6242 (SW)

From the BSE images shown in Figure 4-14, a better understanding of microstructures can be obtained. TMAZ regions of the as-welded and stress relieved samples did not reveal any significant microstructural changes similar to the optical

microscopy images (Fig. 4-14 (a) and (b)). The dark phase in the BM corresponds to the primary α phase and light regions to the transformed β phase. In the transformed β grains of both the as-received and stress relieved BM microstructure, randomly oriented α/β lamella are clearly seen. BSE images show a significant reduction of α/β lamella in the β grains compared to the as-received condition (Fig.4-14 (c) and (d)). Also, clear indications of α' martensite is observed in the WI of the as-welded sample (Fig.4-14 (e)), while the WI of the stress relieved sample shows an acicular $\alpha+\beta$ structure (Fig.4-14 (f)).

Figure 4-14 BSE images of as-welded (a, c, e) and stress relieved (b, d, f) Ti-6242 (SW)

4.2.2 Microhardness test

For as-welded Ti-6242, the microhardness profiles obtained show similar characteristics, as seen in Ti-64. The WI was found to be harder than the surrounding TMAZ regions. In order to better understand the hardness distribution, hardness maps were also generated for both the as-welded and stress relieved samples.

Multiple line hardness measurements were performed across the WCZ of the as-welded sample (Fig.4-15 (a)) and the corresponding hardness profiles plotted shown in Figure 4-15 (b). The as-received BM hardness values range from 320-360 HV. The BM hardness of the sample obtained is comparable to the Ti-6242 alloy with a Vickers hardness value of 333 HV in accordance with the ASTM specifications [55]. TMAZ hardness values were lower than the WI by 20-25 HV. Figure 4-16 shows the hardness map generated, with a maximum value of 404 HV measured in the WI within the selected area of hardness mapping. The color bar indicating the hardness distribution is shown next to the hardness map. The hardness results for the measured area range from 311 HV to 404 HV.

Figure 4-15 Hardness profile of the as-welded Ti-6242 sample

Figure 4-16 Hardness map of the as-welded Ti-6242 sample

The stress relieved sample showed similar hardness distribution characteristics as observed in the as-welded sample. There were no significant differences in the hardness measurements in the BM and TMAZ regions. However, the peak hardness values of WI

decreased by 20 HV. Figure 4-17 shows the hardness profiles obtained for the stress relieved sample and hardness map generated is shown in Figure 4-18.

Figure 4-17 Hardness profile of the stress relieved Ti-6242 sample

Figure 4-18 Hardness map of the stress relieved sample

The hardness distribution of the stress relieved and as-welded samples can be explained by the microstructural changes. TMAZ of the stress relieved sample showed minimal changes in the microstructure compared to the as-welded sample and therefore, comparable hardness values are obtained in this region. The tempering of martensite α' in the WI reduced the peak hardness values shown in the as-welded sample.

4.2.3 Tensile Test

Tensile tests were performed at room temperature for the as-welded and stress relieved samples. Test results of both samples show that the WCZ and TMAZ regions are stronger than the BM. Samples failed in the BM away from the WCZ. Figure 4-19 shows the tensile curves obtained for the tested samples and the corresponding ultimate tensile strength, yield strength, and percent elongation values shown in Table 4-3. The yield strength of the as-welded samples ranges between 1020 to 1030 MPa with a percent elongation of 13 to 16%. Results obtained for the as-welded samples are comparable to ASTM specifications for Ti-6242 alloy, with a yield strength of 990 MPa and UTS of 1010 MPa [55].

The yield strength of stress relieved samples ranges from 900 to 950 MPa and the percent elongation by 5 to 6% compared with the as-welded condition. Yield strength and UTS values were also reduced. Reduction in tensile properties of stress relieved samples is due to the weakening of interface β and the reduction of intergranular α/β lamella during stress relief annealing, as observed in the optical and SEM microscopy images [56]. However, the tensile properties obtained for the stress relieved samples were also comparable with the BM (as per ASTM specifications).

Figure 4-19 Engineering stress-strain diagram for linear friction welded Ti-6242

Table 4-3 Tensile results of linear friction welded Ti-6242

Sample ID	UTS (MPa)	Yield (MPa)	EL. (%)
AW-2B	1105	1028	13.7
AW-3B	1106	1026	16.4
AW-7B	1147	1032	14.8
SR-1A	1003	946	9.1
SR-3A	1044	969	10.6
SR-6B	1060	915	11.8

Figure 4-20 shows the DIC analysis results for the as-welded sample. Similar to the Ti-64 as-welded sample, an early plastic activity is observed near the WCZ and the macroscopic strain reaches a maximum value of 0.94 percent (Fig.4-20 (b)). There is no significant increase in strain in this region between the UTS and the breakpoint. The macroscopic stress reaches UTS for an engineering strain of 8.85 percent (Fig.4-20 (c)). The failure occurred 5 ± 2 mm away from the WCZ after the strain localization in the BM. A maximum strain of 41.5 percent measured at the breakpoint in the BM region (Fig.4-20 (d)).

Figure 4-20 (a) As-welded sample (with black and white speckle); (b)-(d) Strain maps; (e) Aramis stress-strain curve.

Local strain values in the stress relieved samples also exhibited similar characteristics, as observed in the stress relieved Ti-64 sample. There is no significant

increase in the strain near the WCZ at the yield point compared to the as-welded sample (Fig.4-21 (b)). However, strain increased slightly by 1.22 percent about 4 to 5 mm away from the WCZ. Later, stress is concentrated in the same area leading to the final fracture of the sample. Macroscopic strain distribution across the ROI at yield and UTS for both the as-welded and stress relieved samples are comparable (Fig.4-21 (c)). The macroscopic strain reaches a maximum value of 19.9 percent at the breakpoint (Fig. 4-21(d)). The overall macroscopic strain distribution across the ROI at the breakpoint is decreased by 20 percent compared to the as-welded sample.

Figure 4-21 (a) Stress relieved sample (with black and white speckle); (b)-(d) Strain maps; (e) Aramis stress-strain curve.

From the strain maps of the stress relieved sample, it can be concluded that stress relief annealing heat treatment reduced the average macroscopic strain distribution across

the ROI. This trend can also be confirmed by the reduced strain localization observed in the WCZ at yielding.

4.2.3.1 Fractographic Analysis

Figure 4-22 shows the SE images of the as-welded and stress relieved sample tensile fracture surfaces. As shown in the overview images, the crack spread across the surface from a common crack initiation site, resulting in the final fracture (Fig.4-22 (a) and (b)). The magnified images show microcracks and smooth cleavages at the crack initiation sites in both sample surfaces (Fig.4-22 (c) and (d)). In both fracture surfaces, striation lines formed during the loading cycle are also observed. The fractographic analysis reveals the characteristics of a ductile material. There was no sign of macropores, oxides, or any other inclusions on either fracture surfaces.

Figure 4-22 SE images of (a, c) as-welded sample fracture surface; (b, d) stress relieved fracture surface

4.2.4 Fatigue Test

As mentioned before, all fatigue samples were stress relieved prior to the tests to reduce residual stresses. Low and high cycle fatigue tests were performed at room temperature with stress amplitudes ranging from 1100 to 300 MPa. The results obtained are shown in Table 4-4 and in Figure 4-23 (the corresponding Wöhler or S-N curve). Similar to Ti-64 welds, LCF failure of 102 to 104 cycles was achieved with stress amplitudes ranging from 1100 to 600 MPa and the HCF failure of 105 below 400 MPa.

Table 4-4 Fatigue test results; Ti-6242 (SW) stress relieved samples

Sample ID	Parameter (Actual)				Results
	Stress (MPa)				Number of cycles to failure N_f
	Maximum σ_{max}	Minimum σ_{min}	Amplitude σ_{amp}	Average σ_{avg}	
2E	1100	110	495	605	456
4E	1050	105	472.5	577.5	1568
3E	900	90	405	495	4216
5F	800	80	360	440	6700
4D	700	70	315	385	12171
3F	600	60	270	330	15560
3D	500	50	225	275	33570
1E	400	40	180	220	90696
5D	300	30	135	165	1296576

Figure 4-23 S-N curve for stress relieved Ti-6242 samples

4.2.4.1 Fracture Analysis

Fatigue failure occurred in the BM near the WCZ during high and low cycle fatigue tests. The fracture surfaces of both HCF and LCF samples are shown in Figure 4-24. The magnified images of the crack initiation and final tensile loading sites are shown in the overview images (Fig.4-24 (a) and (b)). The propagation of the crack begin at the crack initiation sites and spreads throughout the surface resulting in the final fracture of both samples, as observed in the tensile specimens. Striations observed at the LCF crack initiation sites are smoother than the HCF fracture surface (Fig.4-24 (d) and (c)) due to shorter fracture time. Figures 4-24 (e) and (f) show the magnified images of the high and low cycle fatigue fracture surface at the final tensile loading sites. The final fracture sites

show smooth striations on both surfaces. On both the fracture surfaces, dimples and micropores are also observed. No trace of any macropore, oxides, or any other inclusions were detected on the surfaces.

Figure 4-24 SE images of (a, c, e) High cycle fatigue fracture surface; (b, d, f) Low cycle fatigue fracture surface

4.3 Ti-6Al-4V - Ti-6Al-2Sn-4Zr-2Mo Disimilar Weld

Coupons of Ti-64 and Ti-6242 alloys were linear friction welded (welding parameters are reported, in Table 3-3). The as-welded coupons exhibited similar flash morphologies, as seen in similar welds (Fig.3-5 (b)). The thermal and mechanical properties of the two alloys are shown in Table 4-5 according to the ASTM specifications (Table 4-5). The β transus temperatures of both alloys are almost the same. During the linear friction welding process, the maximum temperature surpasses the β transus at the WI.

Table 4-5 Important thermal and mechanical properties of Ti-64 and Ti-6242 [52, 55]

Properties	Ti-6Al-4V	Ti-6Al-2Sn-4Zr-2Mo
Hardness, Vickers	349 HV	333 HV
Tensile Strength, Ultimate	950 MPa	1010 MPa
Tensile Strength, Yield	880 MPa	990 MPa
Modulus of Elasticity	113.8 GPa	120 GPa
Beta Transus	980°C	990°C
Melting Point	1064-1660 °C	Max.1700°C

4.3.1 Microstructure

Microstructural analysis was performed using both optical and SEM imaging on the extracted metallographic samples. Figure 4-25 (a) and (b) show the optical microstructures (overview images) of the as-welded and stress relieved samples in different locations near the WCZ and BM regions. The TMAZ regions of both Ti-6242 (TMAZ-1)

and Ti-64 (TMAZ-2) sides in the as-welded sample consisted of highly deformed α grains and intergranular β oriented in the weld direction (Fig.4-25 (c) and (e)). The TMAZ regions on both sides of the stress relieved sample did not reveal any significant changes in this microstructure compared to the as-welded condition (Fig.4-25 (d) and (f)). From the optical microscopy images in the WCZ, it is clear that an interface is formed between the two alloys. Although both alloys have comparable thermal and mechanical properties, there are no signs of any diffusion across the WI. Both the alloys remained as separate regions in the WCZ. In the as-welded specimen, α' martensite formed during the LFW process can be observed on both sides of the WI (Fig.4-25 (g)).

The WI of the stress relieved samples exhibited combined characteristics as observed in the stress relieved samples of Ti-64 and Ti-6242 similar welds (Fig.4-25 (h)). Compared to the as-welded sample, the Ti-64 BM microstructure did not change much after annealing (Fig.4-25 (i) and (j)). However, the Ti-6242 BM shows a significant reduction in intergranular α/β lamella after stress relief annealing heat treatment (Fig.4-25 (k)) and (l)).

Figure 4-25 Optical microstructures of as-welded (a, c, e, g, i, k) and stress relieved (b, d, f, h, j, l) samples

Electron micrographs for the as-welded and stress relieved samples are shown in Figure 4-26. TMAZ microstructures of the as-welded and stress relieved samples did not show any visible differences in the BSE images as well (Fig.4-30 (a), (b) and (c),(d)). The WI formed between the two alloys is clearly visible for both conditions in the WCZ microstructure (Fig.4-30 (e) and (f)). The Ti-64 and Ti-6242 sides of the WI in the as-welded condition show clear indications of α' martensite formed, suggesting that the temperature surpasses the β transus in this region (Fig.4-26 (e)). Stress relief annealing

transformed the α' martensite on either side of the WI to an acicular $\alpha+\beta$ structure, as seen in the BSE image (Fig.4-30 (f)). Similar to optical microscopy images, there is a significant reduction of α/β lamella within the transformed β grains observed in the Ti-6242 BM after stress relief annealing (Fig.4-30 (g) and (h)). Comparing the Ti-64 BM microstructures, the stress relieved sample shows an increased β phase (Fig. 4-30 (i) and (j)).

Figure 4-26 SEM images of the as-welded (a, c, e, g, i) and stress relieved (b, d, f, h, j) samples

4.3.2 Hardness Test

The hardness results of the as-welded sample show the combined characteristics obtained in Ti-6242 and Ti-64 similar welds. The Ti-6242 BM hardness of the as-welded sample ranges from 330 to 360 HV. Lower hardness values were obtained in the Ti-64 BM compared to Ti-6242 BM, ranging from 310 to 330 HV. The hardness values obtained on both sides of the BM match the previous results obtained for similar welds. The WI hardness measurements were taken exactly at the interface between the two alloys, as shown in Figure 4-27 (a) and a median hardness of both sides was obtained in this region. From the hardness distribution, it is clear that the WI is harder than the surrounding TMAZ-1 (Ti-64) and TMAZ-2 (Ti-6242) regions, as with the previous cases. Also, TMAZ-2 hardness values were found to be higher than the TMAZ-1 region and remain comparable with the similar weld results. Figure 4-27 (a) and (b) shows the line hardness measurements performed and the corresponding profiles obtained for the as-welded sample, respectively. The hardness map generated (Fig.4-28) shows a maximum hardness value of 401 HV in the WI and a hardness range from 313 to 401 HV within the selected area of mapping.

Figure 4-27 Hardness profile of the as-welded DW sample

Figure 4-28 Hardness map of the as-welded DW sample

Hardness distribution across the WCZ of the stress relieved sample exhibits combined characteristics, as seen in the WCZ of the stress relieved similar welds. The average BM hardness of Ti-64 improved by 20 - 30 HV compared to the as-received condition. However, after stress-relief annealing, the Ti-6242 BM hardness did not show

any significant differences. The hardness values obtained in the TMAZ-1 and TMAZ-2 regions were also comparable to the as-welded sample. Similar to the as-welded sample, indent measurements in the WI of the stress relieved sample covers a half portion of Ti-64 and the other half on the Ti-6242 side. So that the hardness values obtained at this region range between 365 to 370 HV, which is exactly the median hardness on both sides of the WI. Figure 4-29 (a) and (b) show the line hardness measurements performed and the corresponding hardness profiles obtained for the stress relieved sample. The hardness map generated shows a maximum hardness value of 389 HV in the TMAZ-1 region (Fig.4-30) and the hardness values measured within the selected area of mapping ranges from 300 to 390 HV.

Figure 4-29 Hardness profile of the stress relieved DW sample

Figure 4-30 Hardness map of the stress relieved DW sample

The difference in the hardness distribution can be explained on the basis of the microstructural changes that occurred during stress relief annealing. The TMAZ of both the Ti-64 and Ti-6242 sides showed no significant changes in microstructures and therefore, the hardness values obtained were comparable to the as-welded sample. On both the Ti-64 and Ti-6242 sides of the WI regions, the tempering of α' martensite associated with heat treatment lowered the peak hardness values obtained in the as-welded sample. It is quite interesting to note the hardness variations in the BMs, where after the heat treatment, Ti-64 shows higher hardness than Ti-6242. This is mainly due to the increased β phase in Ti-64 BM after stress relief annealing that improved the hardness.

4.3.3 Tensile Test

Tensile tests were performed at room temperature on both the as-welded and stress relieved samples. The local strain properties were analyzed using DIC technology. Similar to the Ti-64 and Ti-6242 similar weld samples, samples failed 5 ± 2 mm away from the WCZ. Therefore, the WCZs of the dissimilar welds in both the as-welded and stress relieved conditions are stronger than the BM. Failures in both the as-welded and stress relieved samples occurred in Ti-64 BM. The BM hardness is higher in Ti-64 compared to the Ti-6242 BM after the heat treatment, but during tensile testing the samples still failed in the Ti-64 BM.

Figure 4-31 shows the engineering stress-strain curves obtained for the tested samples. The corresponding UTS, yield strength, and percentage of elongation data for both the as-welded and stress relieved samples are shown in Table 4-6. The UTS values of the as-welded samples range from 1000 to 1010 MPa with a percent elongation between 10% to 11%. The yielding of the samples occurred between 920 to 940 MPa. The tensile mechanical properties of the as-welded samples are in between the BM tensile properties of the two alloys as per the ASTM specifications (Table 4-5).

Like similar joints, the stress relieved dissimilar samples showed lower UTS and yield strengths compared to the as-welded condition. The yield strength of stress relieved samples ranges from 890 to 920 MPa. The ductility increased by 2 percent compared to the as-welded condition at the expense of strength values.

Figure 4-31 Engineering stress-strain diagram for linear friction welded Ti-64 and Ti-6242

Table 4-6 Tensile measurements of linear friction welded Ti-64 and Ti-6242

Sample ID	UTS (MPa)	Yield (MPa)	EL. (%)
AW-2B	1010	941	10.7
AW-3B	1006	928	10.6
SR-2C	977	894	11.7
SR-6B	974	912	12.1

Figure 4-32 shows the DIC results obtained for the as-welded sample. From the strain map at the yield point (Fig.4-32 (b)). It is clear that local strain distribution across the Ti-64 BM region is higher than in the Ti-6242 BM. An early plastic activity is observed

near the WCZ at yielding and the macroscopic strain reaches to ~ 0.92 percent. At the UTS, the strain in the Ti-64 BM region reached a maximum value of 3.79 percent, and the strain distribution across the WCZ increased slightly to ~ 2 percent (Fig.4-32 (c)). However, the overall local strain distribution across Ti-6242 BM between yield to breakpoint remains uniform. The strain map at the breakpoint shows the maximum strain concentration of 28.5 percent in the Ti-64 BM (Fig.4-32 (d)).

Figure 4-32 (a) As-welded sample (with black and white speckle); (b)-(d) Strain maps; (e) Aramis stress-strain curve

For the stress relieved sample at the yielding, there is no significant increase in the strain values near the WCZ compared to the as-welded sample (Fig.4-33 (b)). However, a

slight increase in strain values was observed 2-3 mm away from the WCZ in the Ti-64 BM. Later, strain is concentrated in the same area and reaches 5.20 percent at UTS (Fig.4-33 (c)). At the breakpoint, a maximum strain of 44.3 percent was measured (Fig.4-33 (d)). The higher macroscopic strain values at the breakpoint, compared to the similar welds attributed to increased ductility in the global tensile result. However, the strain distribution in the ROI of the as-welded and stress relieved samples are comparable between yield and UTS. The average strain distribution across the Ti-6242 BM regions remained uniform.

Figure 4-33 (a) Stress relieved sample (with black and white speckle); (b)-(d) Strain maps; (e) Aramis stress-strain curve.

4.3.3.1 Fractographic Analysis

Both the as-welded and stress relieved samples failed in the Ti-64 BM and fracture surfaces were studied by SEM analysis. Figure 4-34 shows the low and high magnification images of the fracture surfaces for the as-welded and stress relieved conditions. Micropores are present across the fracture surfaces. Dimples present on the surfaces confirm the ductile nature of the material. No trace of macropores, oxides or any other inclusions were detected on either surface.

Figure 4-34 SE images of (a, c) as-welded sample fracture surface; (b, d) stress relieved fracture surface

4.3.4 Fatigue Test

Prior to the room temperature fatigue tests, all the samples were stress relieved to lower the residual stresses present in the samples. Fatigue test results obtained for the tested

samples are shown in Table 4-7 and the Wöhler curve/S-N curve is plotted in Figure 4-35, with the number of cycles to failure on the X-axis and maximum stress (in MPa) applied to the Y-axis. The test results obtained for both low and high cycle fatigue failures are in between the results obtained for the Ti-64 and Ti-6242 similar welds, which shows the combined effect on the fatigue properties. LCF failure of 10^2 to 10^4 cycles was achieved with stress amplitudes ranging from 1100 to 600 MPa and an HCF failure of 10^5 below 400 MPa.

Table 4-7 Fatigue test results (stress relieved samples)

Sample ID	Parameter (Actual)				Results
	Stress (MPa)				Number of cycles to failure N_f
	Maximum σ_{max}	Minimum σ_{min}	Amplitude σ_{amp}	Average σ_{avg}	
6D	1100	110	495	605	51
5D	1000	100	450	550	2356
2D	900	90	405	495	4877
4D	800	80	360	440	6670
6E	700	70	315	385	12563
3D	600	60	270	330	19756
3F	500	50	225	275	43669
6F	450	45	202.5	247.5	70773
4F	400	40	180	220	95263
4E	300	30	135	165	1059126

Figure 4-35 S-N curve for stress relieved DW samples

4.3.4.1 Fractographic Analysis

Figure 4-36 shows the SE images of high and low cycle fatigue fracture surfaces. LCF failures occurred in Ti-64 BM near the WCZ and HCF failure occurred in the Ti-6242 BM in the vicinity of WCZ. Fracture locations shifted towards the Ti-6242 BM from the low to high cycle fatigue (Fig.4-36 (a) and (b)). This shift in the fracture location can be explained based on the fracture surface morphologies. HCF fracture surface shows crack propagation across the surface originating from a crack initiation site, as seen in the Ti-6242 similar welds. The magnified images in the HCF fracture surface at the crack initiation and final tensile loading sites (Fig.4-36 (c) and (e)) show striation lines and microcracks. Fatigue samples that failed at LCF exhibited necking, indicating large

accumulation of local strain at this stress level close to the material yield strength and the surface consisted of numerous micropores and smooth striations (Fig.4-36 (d) and (f)) as also previously observed in the Ti-64 similar welds. The striations observed in the final fracture site were smoother compared to the crack initiation area, indicating a short fracture time. On both surfaces, there was no trace of oxides, macropores, or any other inclusions.

Figure 4-36 SE images of (a, c, d) High cycle fatigue fracture surface; (b, e, f) Low cycle fatigue fracture surface

Chapter 5 : Discussions

So far, studies on linear friction welding have been conducted primarily on Ti-64 due to the unique properties of the material in manufacturing BLISK. Major focus of these studies was to characterize the mechanical and microstructural properties in the as-welded condition and also to develop optimum welding process parameters to obtain sound welds. However, there are still many areas of research that have not been addressed in the literature.

The more recently developed alloy Ti-6242 has not been studied in combination with Ti-64 using the linear friction welding technique. Therefore, the main objective was to investigate the LFW-ability of these two dissimilar alloys in terms of microstructures and mechanical properties in the as-welded and stress relieved conditions. It was realized that although stress relief annealing has been studied for similar Ti-64 and Ti-6242 LFWed joints, comparison of results with the as-welded conditions were never explained in terms of local mechanical properties. Thus, the objective was extended in this study and, the microstructures and mechanical properties (global and local) of Ti-6242 and Ti-64 similar welds are also investigated. The welding parameters for both similar and dissimilar welds were selected based on previous research work to obtain sound welds. As a near- α alloy, Ti-6242 behaved quite different from Ti-64 in some respects. However, both alloys are composed of primary α and transformed β phases. Also, the β transus temperatures of both alloys are comparable (Table 4-5). During linear friction welding, the temperature achieved at the interface exceeds the β transus and the material undergoes plastic deformation in the β phase. Some of the previous research work on Ti-64 and Ti-6242 similar welds can assist

in understanding the dissimilar linear friction welding between Ti-6242 and Ti-64 in terms of microstructural and mechanical properties.

5.1 Microstructures

Microstructural changes observed in different zones of the WCZ depend on the differences in temperature experienced. Finite element modeling of the LFW process has shown a temperature rise of about 650 to 750 °C at a distance of 1.6 mm away from the weld interface and more than 900 °C at the weld center [57]. Metallurgical evidence in the literature also suggests that temperature exceeded β transus and therefore martensite/acicular α is formed in the WI. Research conducted on Ti-alloy systems such as Ti-6Al-4V [26], Ti-6.5Al-1.9Zr-3.3Mo-0.25Si [58], Ti-6Al-2Sn-4Zr-2Mo-0.1Si and Ti-13.5Al-21.5Nb [15] have shown that the weld region is exposed to temperatures above β transus combined with hot deformation and rapid cooling. The narrow TMAZ region formed close to WI consisted of elongated α grains and metastable β phase. Post weld heat treatments resulted in the transformation of α' martensite to α and β products [15, 26].

Figure 5-1 shows the magnified SEM and optical images of the WCZ of Ti-64 similar weld in the as-welded and stress relieved conditions. The SEM and optical microscopy images of the WCZ of the as-welded sample suggest the formation of α' martensite where α' laths are rearranged as a brick wall-like structure. The formation of a very fine martensite structure can be seen in the SEM image as well. The TMAZ region consists of some heavily deformed primary α (appears fibrous and heavily distorted) and some recrystallized α can be seen suggesting that the material did not exceed the β transus

in this region (Fig.4-1 (d) and 4-2 (a)). During stress relief annealing α' tempered to acicular $\alpha+\beta$. Previous microstructural studies conducted on the linear friction welds of Ti-64 are in agreement with the results obtained in this study [26, 42, 48, 58].

Figure 5-1 SEM and Optical images of Ti-64 WCZ (a) as-welded condition (b) stress relieved condition

Figure 5-2 shows the magnified SEM and optical microscopy images in the WCZ for the Ti-6242 similar weld in the as-welded and stress annealed conditions. Clear indications of α' martensite can be seen in the as-welded sample. Stress relief annealing below the β transus for 2 hrs. permitted the transformation of α' to acicular $\alpha+\beta$ in the weld interface. The TMAZ region consisted of heavily deformed α and some recrystallized α grains (Fig.4-13 (d) and 4-14 (a)). Ballat-Duranda et al. [46] studied the microstructural changes during linear friction welding of Ti-6242 and showed that the weld interface underwent a complete $\alpha \rightarrow \beta$ transformation accompanied by the recrystallization of the prior- β grain and the intragranular precipitation of α' martensite. The post-weld heat treatments resulted in the transformation of $\alpha' \rightarrow \alpha+\beta$ [46]. Similarly, the presence of α'

martensite structure in the WCZ of the as-welded Ti-6242 was reported by García et al.[45]. The TMAZ characterized by heavily deformed grains with a partial $\alpha \rightarrow \beta$ transformation oriented in the weld direction. These microstructural observations reported on the Ti-6242 similar welds support the findings in the present study.

Figure 5-2 SEM and Optical images of Ti-6242 WCZ (a) as-welded condition (b) stress relieved condition

In the as-welded and stress relieved conditions, the microstructures of the dissimilar weld showed combined characteristics observed in the Ti-64 and Ti-6242 similar welds. Figure 5-3 shows the magnified optical and SEM images of WCZ. In both images, an interface between the alloys can be clearly observed. Although the properties of both alloys are comparable, there is no long-range inter-diffusion across the WI. This can be attributed to the fact that during the welding the total time where the temperature is above the β -transus temperature is very short, thus leading to only limited atomic diffusion. Solid-state diffusion is minimal at this temperature and also due to the fast cooling rate. Both sides of the WI in the as-welded condition show the presence of an α' martensite structure, which

was transformed into acicular $\alpha+\beta$ after stress relief annealing. The TMAZ regions consisted of highly deformed grains oriented in the weld direction (Fig. 4-26 (a) and (c)) and stress relief annealing showed minimal changes in the microstructure (Fig.4-26 (b) and (d)).

Figure 5-3 SEM and Optical images of DW at the WCZ (a) as-welded condition (b) stress relieved condition

5.2 Hardness

Figure 5-4 shows combined average hardness profiles of the similar and dissimilar welds in the as-welded and stress relieved conditions. The higher hardness values in the WI compared with the BM is mainly due to the stronger α' martensite formed during the process. The highly deformed α grains with intergranular β formed by the thermomechanical deformation of the interface material results in the lower hardness values of the TMAZ regions compared to the weld center.

Figure 5-4 Average Hardness Profiles

Post weld heat treatments on linear friction welded joints demonstrated a significant reduction in residual stresses. Frankel et.al [47] reported the influence of post-weld heat treatment on linear friction welded Ti-64 and Ti-6242 using high energy synchrotron X-ray diffraction. Conventional stress relief annealing heat treatment (700 °C for 2 hrs. followed by furnace cooling) of Ti-64 reduced the residual stresses about 50 MPa and Ti-6242 stress relieved at 800 °C shown a reduction of residual stress by 100 MPa [47]. Stress relief annealing temperatures used in this study were based on the above-mentioned research work and for the dissimilar welds, an intermediate temperature was chosen.

García et al. [45] reported an average BM microhardness of 330 HV and a 27 % higher hardness in the WCZ for linear friction welded Ti-6242, which supports the hardness measurements in this work. This increase in hardness values at the WCZ was due to the formation of metastable α' martensite phases. The fatigue and tensile tests confirmed this result as the failure occurred in the BM away from the WCZ. The hardness distribution

across the TMAZ regions was slightly lower than the WI hardness. Stress relief annealing lowered the peak hardness values in the WI due to tempering of the α' martensite and the TMAZ regions did not show any significant differences in hardness values due to the minimal changes that occurred in the microstructure.

Microhardness profiles of dissimilar weld in both as-welded and stress relieved conditions showed combined characteristics as observed in the as-welded Ti-64 and Ti-6242 similar welds, as expected. The hardness values recorded in the WI are the median of hardness values on either side. The TMAZ regions of both Ti-6242 (TMAZ-1) and Ti-64 (TMAZ-2) of the stress relieved sample did not show any significant difference in hardness values compared to the as-welded condition due to minimal changes in the microstructure after stress relief annealing heat treatment. The average BM hardness of Ti-64 improved by 20-25 HV due to the increased β phase while the Ti-6242 BM hardness was comparable with the as-received sample, as observed in the similar weld samples.

5.3 Tensile Mechanical Properties

The average tensile properties of the two alloys in the as-welded and stress relieved conditions are comparable with the BM properties as per the ASTM specifications. The average tensile properties of the tested samples are shown in Table 5-1 and the corresponding average curves in Figure 5-5. Tensile results published on LFW of Ti-64 have shown that the average tensile properties in the as-welded conditions were comparable with the BM tensile properties (as per ASTM specifications) [23, 26]. The welding process parameters influence the tensile mechanical properties. Wanjara and

Jazhai [26] reported a reduction of yield and UTS values from 1002 to 991 MPa and from 1078 to 1039 MPa, respectively by varying the frequency values between 15 to 70 Hz. They also reported that the yield and UTS values slightly increased by increasing the amplitude and axial shortening values. Their study suggests that sound welds are produced for specific input values greater than 2.4 kW [26]. The tensile mechanical properties obtained for Ti-64 similar welds in this study are in agreement with the literature. Stress relief annealing applied for lowering the residual stresses, increased the strength values at the expense of ductility. This deviation is attributed to the increased β phase after stress relief annealing that strengthens the BM.

Table 5-1 Average tensile properties

Sample ID	UTS (MPa)	Yield (MPa)	Elongation (%)
Ti-64-SW (AW)	995	866	14.5
Ti-64 SW (HT)	1027	930	11.9
Ti-6242 SW (AW)	1105	1027	15.0
Ti-6242 SW (HT)	1036	941	10.5
DW (AW)	1008	934	10.7
DW (HT)	966	862	12.2

Figure 5-5 Average tensile curves

García et al. [45] studied the tensile properties for Ti-6242 similar welds in the as-welded condition and reported comparable average tensile properties with this study. Also, the DIC analysis performed for evaluating the macroscopic deformation showed an early plastic activity at yielding near the WCZ, which is attributed to the presence of the residual stresses [45]. Stress relief annealing lowered the average tensile mechanical properties and can be attributed to the weakening of the interface β and reduction in intergranular α/β lamella during stress relief annealing heat treatment [56]. The tensile fracture surfaces consisted of microcracks, striations and cleavages and mostly dimples.

The average tensile results obtained for the dissimilar weld is the median of values obtained for the Ti-64 and Ti-6242 similar welds, showing the tailored effect on the tensile mechanical properties. For both the as-welded and the stress relieved conditions, failure occurred in the Ti-64 BM away from the WCZ, indicating that the WCZ is stronger than the BM. Comparing BM tensile properties of Ti-64 and Ti-6242, Ti-6242 alloy has higher values of yield and UTS, which explains the reason for failure in the Ti-64 BM. This is to be noted here that despite having a higher hardness in Ti-64 BM in stress relieved condition, fracture still took place in the Ti-64 BM, which seems to support the theory. Just to keep in mind that the hardness is a local property and is not generally been affected by any existing defects in the BM. On the other hand, tensile is a global property testing and is severely affected by any existing defects. Although, this is not supported by the fracture surface analysis, as no large or significant defects are visible. During welding the rolling directions for Ti-64 and Ti6242 were perpendicular to each other as shown in the schematics below (Fig.5-6). This is also visible in the optical microstructures of Fig 4-25. Tensile mechanical properties can also be affected by the rolling direction and the applied force direction. For Al-Li alloy sheet metal, it was shown that strength is the highest parallel to the rolling direction [61]. However, for Ti-64 and Ti-6242 dissimilar welds, the rolling direction did not show any effect on the tensile mechanical property as the material failed in the Ti-64 BM.

Figure 5-6 Rolling directions in Ti-64 and Ti-6242

The percent of elongation after stress relief annealing for the dissimilar joint is close to the stress-relieved Ti-64 similar weld. The increase in the percent of elongation may be due to the softening of samples that may have occurred during stress relief annealing heat treatment. The fractographic analysis shows comparable results with the Ti-64 similar welded fracture surfaces. Microvoids along with a few flat cleavages can be seen all over the fracture surface but mostly the dimples are predominant on the surface indicating a ductile failure mode. These microvoids formed at the path for crack propagation by opening up during tensile loading and finally results in fracture of the sample. The DIC test results show that average macroscopic strain values were reduced after stress relief annealing in Ti-64 and Ti-6242 similar welds. For the dissimilar weld in the stress relieved condition, exhibited higher macroscopic strain values at the failure compared with the as-welded sample, which is accounted by the increase in ductility.

5.4 Fatigue

Figure 5-7 shows the combined S-N curves obtained for the similar and dissimilar weld combinations (literature results on Ti-64 and Ti-6242 similar welds (as-welded condition) are also indicated). The S-N curve for dissimilar weld samples lies in between the S-N curves of Ti-64 and Ti-6242 similar welds as expected. The tests were performed

at 10 stress levels, with maximum amplitudes ranging from 1100 to 300 MPa, covering both high cycle (HCF) and low cycle (LCF) domains.

Figure 5-7 Fatigue Curves

Stinville et al. [59] studied the high and low cycle fatigue behavior of linear friction welded Ti-64 in the as-welded condition and reported that LCF failure occurred within a stress amplitude ranging from 1000 to 900 MPa. The fracture occurred in the BM close to the WCZ, which indicated that the WCZ was harder than the BM. Flipo et al. [25] performed a similar study on the Ti-64 in the as-welded condition and found that LCF failure occurred for stress amplitudes below 540 MPa and HCF for amplitudes above 700 MPa. Comparing fatigue results with these studies, the number of cycles to failure improved after stress relief annealing (Fig.5-7). Samples failed at LCF exhibited necking,

indicating large accumulation of local strain at this stress level close to material yield strength. The fracture surfaces consisted of micropores and smooth striations.

The fatigue results reported on the linear friction welded Ti-6242 (as-welded condition) similar welds by García et al. [45] suggest that a target life above 10^5 cycles obtained between stress amplitudes ranging from 600 to 625 MPa and the fracture locations were in the BM region close to the TMAZ region (Fig.5-7). Stress relief annealing applied in the present study lowered the fatigue properties of Ti-6242, which can be attributed to the weakening of the interface β and reduction in intergranular α/β lamella. Fracture surfaces showed cleavages, dimples, striations and microcracks. Striation lines observed to be smoother for low cycle fatigue fracture surfaces, indicating a shorter fracture time.

Fatigue results obtained for the dissimilar welds were in between the Ti-64 and Ti-6242 similar welds, which show the combined effect on the fatigue properties. HCF failure occurred in the Ti-6242 BM and LCF failure occurred in the Ti-64 BM close to the WCZ. This shift in the fracture location can be explained based on the fractographic analysis. For LCF failure in Ti-64 BM, a ductile necking characteristic is observed prior to fracture and the surface consists of numerous microvoids similar to Ti-64 welds. While the crack propagation in HCF failure originates from a crack initiation site, spreading all over the surface leading to a final fracture similar to Ti-6242 welds.

Chapter 6 : Conclusions

Ti-64 and Ti-6242 samples were linear friction welded (similar and dissimilar combinations) with the selected set of welding parameters (frequency of 50 Hz, an amplitude of 2 mm, a pressure of 90 MPa and an axial shortening of 2 mm) producing high-quality welds with excellent tensile mechanical properties. The following conclusions can be made from this study:

1. The as-received Ti-6242 BM microstructure is composed of primary α grains and transformed β where the transformed β consists of randomly orientated α/β lamella. The as-received Ti-64 BM consists of primary α and lamellar colonies of $\alpha+\beta$. The WI of both the similar and dissimilar welds surpassed the β transus temperatures (882 °C) and due to the fast cooling rate, α' martensite is formed in all the joints.
2. The TMAZ regions formed near the WI in both similar and dissimilar welds consisted of highly deformed primary α with intergranular β and some recrystallized α , suggesting that the material did not exceed the β transus in this region.
3. Stress relief annealing results in the tempering of α' martensite to acicular $\alpha+\beta$ in the WI of both similar and dissimilar welds at the WI. While TMAZ regions after stress relief annealing showed no significant difference in the microstructures compared to the TMAZ of the as-welded samples.
4. Hardness measurements across the WCZ of the similar and dissimilar welds in the as-welded condition showed peak hardness values in the WI. This difference in the hardness compared to BM is attributed to the phase transformation and grain refinement that

occurred by the hot deformation at high temperatures and rapid cooling during the LFW process. The relative increase in the hardness values in the TMAZ regions can be explained based on the grain refinement by the thermomechanical deformation.

5. Stress relief annealing lowered the peak hardness values in the WI by 15-20 percent in the Ti-64 similar weld and by 20 HV in the Ti-6242 similar weld. This reduction in hardness can be attributed to the tempering of α' martensite. The average BM hardness of Ti-64 increased by 20-25 HV due to the increased β phase as observed in SEM images. However, TMAZ hardness values remained the same as in the as-welded condition due to the insignificant changes in the microstructure. Dissimilar weld samples showed combined hardness characteristics of the similar welds.

6. The average tensile properties of the as-welded Ti-64 and Ti-6242 similar welds in this study conform with the ASTM standards for a Grade-5 Ti-64 and near- α Ti-6242 alloy, respectively. Stress relief annealing on Ti-64 similar weld samples improved strength (UTS and Yield) at the expense of percent of elongation by 2-3 percent. For Ti-6242 similar welds, tensile properties reduced compared to the as-welded condition. This reduction in tensile properties is attributed to the weakening of the interface resulting from β and depletion of interlamellar α/β within the transformed β grains. The average ductility of dissimilar welded samples improved by 1 percent after stress relief annealing.

7. DIC analysis showed that the overall macroscopic strain distribution across the ROI reduced after stress relief annealing in the Ti-64 and Ti-6242 similar welds compared to the as-welded condition. Higher strain values were recorded at the breakpoint in the stress relieved dissimilar weld, which is accounted by the improved ductility. Tensile fracture

surface analysis in both similar and dissimilar welds (as-welded and stress relieved conditions) confirms the characteristic properties of a ductile material.

8. Low cycle fatigue (LCF) failure of the samples occurred within a stress amplitude range of 1100 to 600 MPa and high cycle fatigue (HCF) failure occurred beyond 10⁵ cycles for stress amplitudes below 400 MPa. Stress relief annealing improved the fatigue properties of the Ti-64 similar welds, which is comparable to the as-welded results in the literature. However, the Ti-6242 similar welds showed a reduction in fatigue properties, which is attributed to the weakening of the interface β and depletion of interlamellar α/β within the transformed β grains.

Chapter 7 : Contribution to Science

1. For the first time, a comparison between the hardness and tensile properties of as-welded and stress relieved samples was performed for both Ti-64 and Ti-6242 similar welds.
2. DIC analysis was performed for linear friction welded Ti-64 and Ti-6242 (similar and dissimilar combinations) in both the as-welded and stress relieved conditions.
3. Microhardness, tensile, and fatigue properties have been studied for the first time for Ti-64 and Ti-6242 dissimilar LFW joints in the as-welded and stress relieved conditions. It was discovered that for dissimilar welds, a clear interface separating the two alloys was formed at the WCZ.

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