A STUDY OF MICROSTRUCTURE, TENSILE DEFORMATION, CYCLIC FATIGUE AND FINAL FRACTURE BEHAVIOR OF COMMERCIALLY PURE TITANIUM AND A TITANIUM ALLOY

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A STUDY OF MICROSTRUCTURE, TENSILE DEFORMATION, CYCLIC FATIGUE AND FINAL FRACTURE BEHAVIOR OF COMMERCIAL SUPPly PURE TITANIUM AND A TITANIUM ALLOY

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ABSTRACT

Rapid industrial growth and advances in the domains of engineering and related technologies during the last fifty years have led to the extensive use of traditional metals and their alloy counterparts. Titanium is one such metal which has gained wide popularity in the aerospace and defense related applications owing to a wide range of impressive mechanical properties like excellent specific strength ($\sigma_{UTS}/\rho$), stiffness, corrosion and erosion resistance, fracture toughness and capability to withstand significant temperature variations.

Two materials, namely commercial purity titanium (Grade 2), referred to henceforth as Ti-CP (Grade 2) and the “work-horse” alloy Ti-6Al-4V have been chosen for this research study. The intrinsic influence of material composition and test specimen orientation on the tensile and fatigue behavior for both Ti-CP (Grade 2) and Ti-6Al-4V have been discussed. Samples of both Ti-CP (Grade 2) and Ti-6Al-4V were prepared from the as-provided plate stock along both the longitudinal and transverse orientations. The specimens were then deformed to failure in uniaxial tension for the tensile tests and cyclically deformed at different values of maximum stress at constant load ratio of 0.1 for the high cycle fatigue tests. The microstructure, tensile properties, resultant fracture behavior of the two materials is presented in the light of results obtained from the uniaxial tensile tests. The conjoint influence of intrinsic microstructural features, nature
of loading and specimen properties on the tensile properties is discussed. Also, the macroscopic fracture mode, the intrinsic features on the fatigue fracture surface and the role of applied stress-microstructural feature interactions in governing failure for the cyclic fatigue properties for both the materials under study Ti-CP (Grade 2) and the “work-horse” alloy Ti-6Al-4V have been discussed in detail.

Careful study of the microstructure for Ti-CP (Grade 2) material at a low magnification revealed the primary alpha (α) grains to be intermingled with small pockets of beta (β) grains. Observation at the higher allowable magnifications of the optical microscope revealed very fine alpha (α) phase lamellae located within the beta (β) grain. The microhardness and macrohardness measurements were consistent through the sheet specimen for Ti-CP (Grade 2) and slightly lower compared to Ti-6Al-4V. However, the macrohardness was marginally higher than the microhardness resulting from the presence of a large volume fraction of the soft alpha phase. The hardness values when plotted reveal marginal spatial variability. Tensile fracture of Ti-CP (Grade 2) was at an inclination to the far field tensile stress axis for both longitudinal and transverse orientations. The overload region revealed a combination of fine microscopic cracks, microscopic voids of varying size and randomly distributed through the surface, and a large population of shallow dimples, features reminiscent of locally brittle and ductile failure mechanisms. The maximum stress (σ_{maximum}) versus fatigue life (N_f) characteristics shown by this material is quite different from those non-ferrous metals that exhibit a well-defined endurance limit. When compared at equal values of maximum stress at a load ratio of 0.1, the fatigue life of the transverse specimen is noticeably greater than the longitudinal counterpart. At equivalent values of maximum elastic strain,
the transverse specimens revealed noticeably improved fatigue life as compared one-on-one to the longitudinal counterparts.

Careful observations of the Ti-6Al-4V alloy microstructure over a range of magnifications spanning very low to high magnification revealed a duplex microstructure consisting of the near equiaxed alpha (α) and transformed beta (β) phases. The primary near equiaxed shaped alpha (α) grains (light in color) was well distributed in a lamellar matrix with transformed beta (dark in color). The microhardness and macrohardness values recorded for the Ti-6Al-4V alloy reveal it to be harder than the commercially pure (Grade 2) material. However, for the Ti-6Al-4V alloy the microhardness is noticeably higher than the corresponding macrohardness value that can be ascribed to the presence of a population of processing-related artifacts and the hard beta-phase. Tensile fracture of the Ti-6Al-4V alloy was macroscopically rough and essentially normal to the far field stress axis for the longitudinal orientation and cup-and-cone morphology for the transverse orientation. However, microscopically, the surface was rough and covered with a population of macroscopic and fine microscopic cracks, voids of varying size, a population of shallow dimples of varying size and shape, features reminiscent of locally brittle and ductile failure mechanisms. When compared at equal values of maximum stress at a load ratio of 0.1, there is a marginal to no influence of microstructure on high cycle fatigue life of both orientations of the alloy.
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CHAPTER I

INTRODUCTION

Industrialization coupled with rapid advances in the field of science and technology have resulted in the development and use of a wide range of metals, their alloy composite counterparts i.e., metal matrix composites, in various applications and industries. Over the past few decades, titanium and its alloys have emerged as a preferred choice of use, owing to their excellent, mechanical properties like, high specific strength ($\sigma_{UTS}/\rho$), stiffness, commendable corrosion and erosion resistance, fracture toughness and the ability to withstand huge temperature variations [1,2]. Continuous research in the field of material sciences have established titanium as a preferred choice for use in many performance-critical industries like aerospace and defense-related applications [3-4].

Pure titanium metal has a melting point of 1675°C and an atomic weight of 47.9 [5]. The density of pure titanium metal is 4.5 g/cm$^3$, which is about sixty-percent of steel [2, 5]. Furthermore, pure titanium and its alloys are essentially non- magnetic and offer good heat transfer capability. The excellent specific strength [$\sigma_{UTS}/\rho$] of pure titanium and its alloys at both high temperature (in excess of 590°C) and low temperature (less than -253°C) enables it to be the preferred metal that is often chosen and used as an ultra-high speed metal for aircraft structural components and even components of the space shuttle [3, 5]. The high melting point of titanium metal enables it to be preferred choice for use in turbine engines [6].
Further, since the titanium metal has excellent superplastic properties, it can be easily deformed to the tune of two-thousand percent without experiencing appreciable necking and/or cracking when heated to a temperature of 925°C during the super-plastic forming process [6]. Also, titanium and its alloy counterparts are nonmagnetic and have a lower linear coefficient of expansion and lower thermal conductivity than the widely chosen and used family of steel and the alloys of aluminum [7]. In more recent years, the automotive industry has increased its use of the titanium metal because of its performance at elevated temperatures coupled with good formability. Noticeably, the metal has minimal degradation and high oxidation resistance during long term service or extended service at elevated temperatures [3, 4].

Essentially, the high cost of producing the titanium metal limited its selection and use to those applications that either required high performance or where life-cycle cost analysis justified its selection and use [8]. It was the aerospace and defense industries that stimulated the initial development of titanium in both commercially pure (CP) form and as alloys in the early 1950’s [9, 10]. From a scientific perspective, titanium is categorized as a polymorphic metal because at room temperature [T = 25°C], it has a hexagonal close-packed (HCP) crystal structure, which is referred to as the alpha phase [11]. However, at temperature above 882°C, an allotropic phase transformation occurs to form the body-centered cubic (bcc) crystal structure, which is referred to as the beta phase [12-14]. Unalloyed titanium is generally referred to as commercially pure (CP) titanium metal, which has noticeably low strength when compared one-on-one to alloys, but is the most corrosion resistant version of the metal [13, 14], making it a preferred choice for military aircraft stationed on ships. Interstitial elements, such as oxygen and nitrogen, are
beneficial since they tend to contribute to strengthening of the commercially pure
titanium and its different grades [15]. Since commercially pure titanium has excellent
corrosion resistance, it is used on-board kitchens, toilets and de-icing equipment.
Helicopters use titanium alloys in highly stressed components like rotor head and rotor
mast. In space vehicles, titanium alloys are used for the fuel and satellite tanks due to its
light weight, high strength and long term chemical compatibility [1, 4]. The automotive
industry was attracted to titanium alloys for its light weight, high specific strength (σ/ρ),
high elastic energy absorption capacity and excellent corrosion resistance [4]. It is used in
exhaust systems, valves, valve cups, connecting rod, turbochargers, suspension springs,
etc [5]. The Marine and energy industries require materials having
(a) High corrosion resistance,
(b) Wide range of strength and performance characteristics under static, cyclic and
dynamic loading conditions,
(c) Cold resistance in a temperature range of -50°C,
(d) High erosion resistance and fire resistance, which they found the alloys of titanium to
meet all of the requirements [6, 16].
Biomedical applications rely on biocompatibility of titanium alloys. Hip and knee-joint
prostheses and other permanent implants like casing for cardiac pacemakers, bond
fixtures, orthodontic and dental implants are made from the titanium metal [2, 7, 16].
Though titanium and its alloys are expensive, it cannot be compromised on account of its
varied properties, low maintenance and long life usage. The higher cost of titanium can
be ascribed to the metal’s strong affinity for oxygen, creating challenges both during
extrusion and downstream processing. This limitation has engineered a considerable
amount of scientific and technological interest in developing potentially viable and economically affordable manufacturing methods that aid in reducing the cost of the product.

The primary objective of this research project was to study the performance of built-up welded beams of commercially pure (Grade 2) titanium and a common titanium alloy Ti-6Al-4V under both static and fatigue loading conditions. The secondary objectives were:

(i) To study the conjoint influence of intrinsic microstructural features, nature of loading and specimen properties on the tensile properties and resultant fracture behavior of the alloys of titanium, Ti-CP (Grade 2) and the “work-horse” alloy Ti-6Al-4V in the light of results obtained from the uniaxial tensile tests.

(ii) To understand the macroscopic fracture mode, the intrinsic features on the fatigue fracture surface and the role of applied stress-microstructural feature interactions in governing failure for the cyclic fatigue properties for both the alloys of titanium, Ti-CP (Grade 2) and Ti-6Al-4V.

(iii) To evaluate of static strength and fatigue behavior of commercially pure titanium (Grade 2) and Ti-6Al-4V built-up beams to produce high strength, low weight, corrosion resistant and low cost structural beams for the purpose of applications in the defense sector.

(iv) To experimentally and theoretically evaluate Ti alloy beams fabricated and tested under both static and fatigue loading.
To develop analysis and design approaches for static and fatigue performance of built-up beams made from both commercially pure titanium and a common titanium alloy.

The study presented in this thesis is part of a larger ongoing research program aimed at establishing the influence of orientation of plate and microstructure on the structural response of welded titanium structures. The thesis is divided into two major sections PART-A and PART-B. The hardness, tensile properties, high cycle fatigue properties and final fracture behavior for Ti-CP (Grade 2) is discussed in PART – A and the titanium alloy Ti-6Al-4V is discussed in PART-B. The literature review related to Ti-CP (Grade 2) is presented in PART-A (Chapter II), while the titanium alloy Ti-6Al-4V is covered in PART-B (Chapter I). The characteristic properties and chemical composition of the as received Ti-CP (Grade 2) is described in PART-A (Chapter III), and for Ti-6Al-4V it is described in PART-B (Chapter II). The test sample preparation and experimental procedures are described in PART-A (Chapters IV and V) for Ti-CP (Grade 2) and PART-B (Chapters III and IV) for Ti-6Al-4V. PART-A (Chapter VI) and PART-B (Chapter V) discuss the two materials under this study, in light of the conjoint and mutually interactive influences of load ratio, maximum stress, intrinsic microstructural features, nature of loading and stress (load)-deformation-microstructural interactions. The conclusions drawn from the analysis of test results for Ti-CP (Grade 2) material are listed in PART – A (Chapter VII) and for the titanium alloy Ti-6Al-4V; they are listed in PART-B (Chapter VI).
PART - A

CHAPTER II

REVIEW OF LITERATURE [Ti – CP (GRADE 2)]

In this chapter, a brief review of the published literature encompassing the domains of processing, microstructure and mechanical behavior of the commercially pure form of titanium is presented. Wiskott et al., [17], studied the mechanical and structural characteristics of commercially pure titanium (Grade 2) welds and solder joints. In recent years, titanium either in the commercially pure form or the alloyed form has been a viable choice for use as an orthopedic and dental implant material. The objective of this research was to determine the applicability of the gathered test data for laser welds and Infrared (IR) brazing using an Au-Pd alloy to the titanium joints. Their study was particularly designed to assess the mechanical resistance, microstructural development, and elemental diffusion of laser welded, electric arc welded and brazed joints using commercially pure titanium as the substrate metal. The native metal was made of grains that were fairly equiaxed and homogeneous. Annealing the machined specimens at 950°C had little to no influence on grain structure. Laser welding of commercially pure titanium produced joint structures that were both identical and continuous with the parent metal. The joints were devoid of visible contamination and no Widmanstatten structures were present [17]. In contrast to the laser joints, electric arc welding was found to produce major alterations in structure of the joint [17]. The heating regimen associated with this technique converted an initially granular parent metal to an acicular structure. The Widmanstatten structures
observed were present up to 15mm from the joints. The results of the mechanical tests revealed the annealed substrate alloy to have a 68% higher UTS value than the ASTM F 67-83 specification for CP Ti grade. However, in a previous report on ASTM Grade 5 and ASTM Grade 2 (Commercially pure titanium whose composition is close to the one used by the researchers in this study), the UTS of standard specimens was found to be in the range of 474 - 479 MPa – a value that was well within 5% of the tensile strength of the specimens. No microstructural alteration was found for the fatigue resistance observed. The arc-welded joints were found to be superior to the native substrate under monotonic tensile load but were inferior when subjected to fatigue loading or cyclical stresses. Judging from the small microstructural differences between the native CP titanium and the annealed CP substrate, the annealing process was actually a stress relief rather than a recrystallization process. This detailed study by the researchers revealed an improvement in cyclic fatigue resistance with annealing [17]. Under fatigue loading conditions, the CP titanium-based brazing as well as the laser welded and electric arc weld performed equally well if not better than an Au-Pd alloy. There was an observable increase in grain size with an increase in heat application. However, these researchers were unable to link a specific features of the microstructure observed or to establish correlation of the elemental analysis with the specimen’s resistance to fatigue stress application [17].

Salishchev et al., [18] examined and documented grain size dependence of certain mechanical properties of commercially pure titanium when deformed at room temperature. A continuous improvement in strength, lower work hardening was observed with a decrease in grain size. The strength was observed to steadily increase with a
decrease in grain size of the annealed titanium specimens. Evaluating the effect of grain size on average strain hardening rate ($\nu_a$) for the annealed state, it was found that the work hardening coefficient decreased as the grain size decreased to 1\(\mu\)m and was found to be relatively independent of grain size. A decrease in the average crystal size resulted in a steady increase in strength, a decrease in hardening, and a shortening of the uniform deformation stage (grain size $d < 20 \\mu$m) [18]. The total plasticity prior to fracture coupled with localized deformation in the neck indicated a relatively weak sensitivity to grain size. The phenomenon associated with grain size effects was attributed to be due to the predominant dislocation density near the grain boundaries [18].

The microstructure and mechanical properties of commercial purity titanium severely deformed by accumulative roll bonding (ARB) process was studied by Terada et al., [19]. Commercial purity titanium was deformed by accumulative roll bonding (ARB) process up to 8 cycles (equivalent to strain of 6.4) at ambient temperature. The microstructure of the ARB-processed specimens revealed two kinds of characteristic ultrafine microstructures. The first was a lamellar boundary structure that was elongated along the rolling direction (RD), which has also been reported for the accumulative roll bonding processed cubic metals [19]. The lamellar boundary spacing decreased with an increase in the accumulative roll bonding strain and reached about 80nm following 5 cycles of ARB. The second microstructure was the equiaxed grains having a mean grain size of 80-100nm. Ultra fine grain (UFG) microstructures were formed in the CP-Ti highly deformed by the accumulative roll bonding process [19]. Deformation twins were observed to form at the lower strain below 50\% rolling reduction but disappeared at a higher reduction above 75\%. Therefore, deformation twining was found to have little
influence on evolution of the ultrafine microstructure [19]. It was suggested that fine equiaxed grains formed during recovery and the inhomogeneously deformed regions, such as, the macroscopic shear bands and micro shear bands, where very large shear strain tends to localize. Local adiabatic heating arising from low thermal conductivity of titanium does assist in the recovery [19].

I.P.Semenova et al., [20] documented the enhanced fatigue strength of commercially pure titanium processed by severe plastic deformation. In their study, they evaluated the high-cycle fatigue behavior of smooth and notched samples of ultrafine-grained titanium prepared by severe plastic deformations and compared it with the corresponding properties of conventional titanium. They found that a combination of high strength and enhanced ductility of the ultrafine-grained titanium led to an increase in fatigue endurance limit [20]. Commercially pure (CP) titanium (Grade4) [Ti: balance, C: 0.052%, O: 0.34%, Fe: 0%, N: 0.015 % (wt %)] was the material used to conduct the research study. The microstructure of the titanium rods subsequent to processing was characterized to be a combination of (i) Equiaxed ultrafine grains and sub grains having an average grain size of 200-nm and (ii) A high dislocation density in the cross-section. The typical structure in the longitudinal section, i.e., examined along the rod length, consisted of the α- grains elongated along the rod axis. The interior of the elongated grains was fragmented by sub grains about 200nm in size and having low-angle boundaries. They found the fatigue endurance limit of the ultra-fine grain samples following $10^7$ cycles to increase from 350 MPa to 590 MPa due to a significant increase in the strength of the material that possessed sufficient ductility. These researchers showed that an additional enhancement of the fatigue limit, up to 610Mpa, was possible
by means of annealing at 350° C for 6h. The annealing enabled increasing ductility without appreciable loss of strength. The fatigue endurance limit of the UFG titanium is strongly influenced by notch size and shape [20]. The notch sensitivity of the ultrafine grain titanium was found to be greater than that of the coarse grained CP Ti counterpart, but lower than the one for the Ti-6Al-4V alloy.

Takao et al., [21] studied the low cycle fatigue behavior of commercially pure titanium. Owing to its excellent strength-weight ratio, commercially pure titanium is generally chosen for use in heat exchangers both in power stations and in chemical plants. In an aggressive environment, the material undergoes cyclic deformation and their study did shed some light on the low-cycle fatigue behavior of this material. The high cycle fatigue of the titanium material chosen is characterized by microscopic cracking initiating within the slip bands between grains and joining together to form macroscopic cracks that grow through the microstructure culminating in final fracture. A similar behavior can be observed for low cycle fatigue criteria. Thus, crack initiation and crack propagation behavior for the commercially pure titanium was studied by a careful examination of the fracture surface of the test specimen that was subject to cyclic strain conditions in low cycle fatigue. The 0.2% proof stress of the material was determined to be 243MPa. A tensile strength of 372 MPa, elongation of 35.4%, a reduction in area of 67.0%; true fracture stress of 793MPa, elastic modulus of 110GPa and average grain size of 76μm were determined from mechanical testing of the specimens. Further, it was observed that, the fatigue hardening and softening behavior was not very remarkable in commercial purity titanium. Irreversible localized deformation was observed along the grain boundaries, which gradually accumulates during subsequent loading cycles.
culminating in the formation of cracks. Since the number of slip systems for commercial purity titanium is less than other metals having a face-centered cubic or body-centered cubic crystal structure, maintaining the continuity of plastic deformation is difficult [21]. Therefore slip band growth is obstructed by the surrounding grains resulting in a concentration of deformations along the grain boundaries [21].
PART- A

CHAPTER III

MATERIAL [Ti – CP (GRADE 2)]

The first material chosen for this study was commercially pure (CP) titanium (Grade 2) referred to henceforth as Ti - CP (Grade 2). This material was provided by TICO Titanium based in Wixom, Michigan, USA. The material was provided by the manufacturer in a fully annealed condition. The Ti-CP (Grade 2) is receptive to heat treatment and can be solution heat treated and annealed to achieve the desired strength and properties. The quantities supplied by TICO are shown in Figure 1, and were as follows:

i. Plate 0.375” TK X 3.00” wide X 36.00” long six pieces

ii. Sheet 0.125” TK X 4.00” wide X 36.00” long three pieces

The edges of the plates and sheets were sheared. The chemical composition of the plates was provided by the supplier and is as shown in Table 1 along with other details.
Table 1: Material specifications, nominal chemical composition (in weight percent) and material properties given by TICO for the Ti-CP (Grade 2).

<table>
<thead>
<tr>
<th>Description</th>
<th>0.375” Thick Plate</th>
<th>0.125” Thick Sheet</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Chemical Composition</strong></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Fe Iron</td>
<td>0.13</td>
<td>0.08</td>
</tr>
<tr>
<td>Oxygen</td>
<td>0.16</td>
<td>0.1</td>
</tr>
<tr>
<td>N Nitrogen</td>
<td>0.002</td>
<td>0.0105</td>
</tr>
<tr>
<td>C Carbon</td>
<td>0.012</td>
<td>0.014</td>
</tr>
<tr>
<td>• Residual Element (Each)</td>
<td>less than 0.10</td>
<td>• Residual Element (Each) less than 0.10</td>
</tr>
<tr>
<td>• Residual Element (Total)</td>
<td>less than 0.40</td>
<td>• Residual Element (Total) less than 0.4</td>
</tr>
<tr>
<td>• Final Product Hydrogen</td>
<td>0.0032</td>
<td>• Final Product Hydrogen 11/10 PPM</td>
</tr>
<tr>
<td>• Titanium Remainder</td>
<td></td>
<td>• Titanium Remainder</td>
</tr>
<tr>
<td><strong>Average Tensile Strength, psi</strong></td>
<td>73,500</td>
<td>70,250</td>
</tr>
<tr>
<td>Yield 0.2% Offset</td>
<td>52,100</td>
<td>49,500</td>
</tr>
<tr>
<td>Elongation %</td>
<td>33</td>
<td>25.75</td>
</tr>
<tr>
<td>Reduction in Area %</td>
<td>59</td>
<td></td>
</tr>
<tr>
<td>Anneal Condition</td>
<td>Yes</td>
<td>1370F, HGA 3 to 11 Min</td>
</tr>
<tr>
<td>Guided Bend Test</td>
<td></td>
<td>Pass</td>
</tr>
</tbody>
</table>
Figure 1: The as received plate stock for the Ti–CP (Grade 2) material.
PART-A

CHAPTER IV

TEST SAMPLE PREPARATION [Ti – CP (Grade 2)]

The test samples used in this study were prepared as per the specifications prescribed in the standard ASTM E-8. The samples were precision machined from the as-provided annealed plate stock both in transverse (T) and longitudinal (L) orientations. In case of the longitudinal orientation, the specimens were machined such that the major stress axis was parallel to the rolling direction, whereas, in case of the transverse orientation, the specimens were machined such that the major stress axis is perpendicular to the rolling direction of the alloy plates. The specimens measured 0.250 inch in diameter at the grips and 0.125 inch in diameter at the gage section and 0.5 inch in length. A schematic of the test specimen is as shown in Figure 2. In order to minimize the effects of surface irregularities and finish, the gage sections of the specimens were mechanically polished using progressively finer grades of silicon carbide impregnated emery paper (320 grit, 400 grit and 600 grit) to remove any circumferential scratches or residual marks.
Figure 2: A schematic of the longitudinal and transverse flat test specimen used for Mechanical testing (Tensile and Cyclic fatigue).
PART- A

CHAPTER V

EXPERIMENTAL PROCEDURES [Ti-CP (GRADE 2)]

5.1 Initial microstructure characterization

An initial characterization of the microstructure of the as-provided material was done using a low magnification optical microscope. Samples of desired size were cut from the as-received stock of the titanium material Ti-CP (Grade 2) and mounted in bakelite. The mounted samples were then wet ground on progressively finer grades of silicon carbide impregnated emery paper using copious amounts of water both as a lubricant and as a coolant. Subsequently, the ground samples were mechanically polished using five-micron alumina solution. Fine polishing to a perfect mirror-like finish of the surface of the titanium material, i.e., Ti-CP (Grade 2), was achieved using five-micron diamond solution as the lubricant. The polished samples were subsequently etched using a reagent that is a solution mixture of 5-ml of nitric acid (HNO₃), 10 ml of hydrofluoric acid (HF) and 85 ml of water (H₂O). The polished and etched surface of the sample of Ti-CP (Grade 2) was observed under an optical microscope and photographed using bright field illumination technique.
5.2 Hardness testing

Hardness of a material is the mechanical property defined as the resistance offered by the material to indentation i.e. permanent deformation and cracking [22]. The hardness was measured using a Vickers (HV) micro hardness testing machine. The test itself is quite simple and nondestructive that is easy to perform and is widely used for the purpose of determining the mechanical properties of monolithic metals and their alloy counterparts.

The macro-hardness measurements (RC) were made on a Rockwell hardness machine using an indentation load of 140 Kgf, a minor load of 10 Kgf, 120 degree diamond cone, a dwell time of 10 seconds and the value read on the ‘C’ scale. The macro-hardness tests were also done on the polished surface of each Ti-CP (Grade 2) test specimen. Two samples of, Ti-CP (Grade 2) were examined for micro-hardness and macro-hardness measurements. The results of the micro-hardness and macro-hardness tests are summarized in Table 2 and Table 3.
Table 2: A compilation of microhardness measurements made on Ti-CP (Grade 2).

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Trial 1</th>
<th>Trial 2</th>
<th>Trial 3</th>
<th>Trial 4</th>
<th>Trial 5</th>
<th>Average</th>
</tr>
</thead>
<tbody>
<tr>
<td>Sample 1 A</td>
<td>213.59</td>
<td>193.56</td>
<td>193.65</td>
<td>193.67</td>
<td>200.58</td>
<td>199.37</td>
</tr>
<tr>
<td>Ti-CP (Grade 2)</td>
<td>R_c</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Sample 1 B</td>
<td>175.59</td>
<td>178.58</td>
<td>167.26</td>
<td>169.26</td>
<td>193.67</td>
<td>176.82</td>
</tr>
<tr>
<td>Ti-CP (Grade 2)</td>
<td>R_c</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
</tbody>
</table>

Table 3: A compilation of macrohardness measurements made on Ti-CP (Grade 2).

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Trial 1</th>
<th>Trial 2</th>
<th>Trial 3</th>
<th>Trial 4</th>
<th>Trial 5</th>
<th>Average</th>
</tr>
</thead>
<tbody>
<tr>
<td>Sample 1A</td>
<td>260</td>
<td>260</td>
<td>272</td>
<td>254</td>
<td>260</td>
<td>261.2</td>
</tr>
<tr>
<td>(Ti-CP (Grade 2))</td>
<td>R_c</td>
<td>24</td>
<td>24</td>
<td>26</td>
<td>23</td>
<td>24.2</td>
</tr>
<tr>
<td>Sample 1B</td>
<td>260</td>
<td>266</td>
<td>279</td>
<td>254</td>
<td>260</td>
<td>263.8</td>
</tr>
<tr>
<td>(Ti-CP (Grade 2))</td>
<td>R_c</td>
<td>24</td>
<td>25</td>
<td>27</td>
<td>23</td>
<td>24.6</td>
</tr>
</tbody>
</table>
5.3 Mechanical testing

Tensile and high cycle fatigue tests were conducted on the specimens to systematically study the tensile deformation, cyclic fatigue and final fracture behavior for the commercially pure titanium. The details of the tests are given in the following sections.

5.3.1 Tensile tests

Uniaxial tensile tests were performed on a fully-automated, closed-loop servo-hydraulic mechanical test machine [INSTRON-8500 Plus] using a 100 kN load cell. The tests were conducted at room temperature (300 K) and in the laboratory air (Relative Humidity of 55 pct) environment. The test specimens were deformed at a constant strain rate of 0.0001/sec. An axial 12.5-mm gage length clip-on type extensometer was attached to the test specimen at the gage section using rubber bands. The stress and strain measurements, parallel to the load line, and the resultant mechanical properties, such as, stiffness, strength (yield strength and ultimate tensile strength), failure stress and ductility (strain-to-failure) was provided as a computer output by the control unit of the test machine. The average values of yield strength obtained from the tensile tests were used for calculating the following information required for the high cycle fatigue tests.

(a) Yield load based on the yield stress obtained from the tensile tests.

(b) Maximum stresses, as percentages of the yield stress of the alloy for purpose of testing the specimens under conditions of stress amplitude-controlled fatigue.
5.3.2 High cycle fatigue tests

The stress-amplitude controlled high cycle fatigue tests were performed using a sinusoidal waveform at a stress ratio \([R = \frac{\sigma_{\text{min}}}{\sigma_{\text{max}}}]\) of 0.1. The cyclic loading of the test specimen was conducted at a frequency of 5Hz. At a given stress ratio \([R = \frac{\sigma_{\text{minimum}}}{\sigma_{\text{maximum}}}]\), the fatigue tests were conducted over a range of stress amplitudes to establish the variation of maximum stress \((\sigma_{\text{maximum}})\) with cyclic fatigue life \((N_f)\).

Three specimens in the transverse orientation had been subjected to a stress percentage marginally higher than the yield stress condition (approximately 6%). The data collected from the high cycle fatigue test is used to establish the following relationships.

1. Variation of maximum stress \((\sigma_{\text{max}})\) with fatigue life \((N_f)\).
2. Variation of the ratio of maximum stress/yield stress \((\sigma_{\text{max}}/\sigma_y)\) with fatigue life \((N_f)\).
3. Variation of maximum elastic strain \((\sigma_{\text{max}}/E)\) with fatigue life \((N_f)\).
4. Variation of the ratio of maximum stress/ultimate tensile stress \((\sigma_{\text{max}}/\sigma_{\text{UTS}})\) with fatigue life \((N_f)\).

The fatigue test data for the cylindrical samples of the Ti–CP (Grade 2) (at \(R = 0.1\)) are summarized in Appendix A.

5.4 Failure-damage analysis

The fracture surfaces of the deformed and failed tensile specimens of i.e., Ti-CP (Grade 2), were comprehensively examined in a scanning electron microscope (SEM) to determine the macroscopic fracture mode and to concurrently characterize the fine scale
topography of the tensile fracture surface for the purpose of establishing microscopic mechanisms governing fracture. The distinction between the macroscopic mode and microscopic fracture mechanisms is based entirely on the magnification level at which the observations are made. The macroscopic mode refers to the overall nature of failure while the microscopic mechanisms relate to the local failure processes, such as: (i) microscopic void formation, (ii) microscopic void growth and coalescence, and (iii) nature, intensity and severity of the fine microscopic and macroscopic cracks dispersed through the fracture surface. The samples for observation in the SEM were obtained from the failed tensile specimens by sectioning parallel to the fracture surface.
PART-A
CHAPTER VI
RESULTS AND DISCUSSION [Ti-CP (GRADE 2)]

6.1 Initial Microstructure

The microstructure of an alloy is an important factor that determines its mechanical properties to include tensile properties, fracture toughness, fatigue resistance and resultant fracture behavior. The optical microstructure of Ti-CP (Grade 2) is shown in Figure 3 at three different magnifications. At a higher magnification (500 x), the intrinsic microstructural constituents of Ti-CP (Grade 2), is shown in Figure 3c bringing out clearly the size, morphology, volume fraction, and distribution of the alpha (α) and beta (β) phases.

Low magnification observation of the microstructure of the Ti-CP (Grade 2) essentially revealed the primary alpha (α) grains to be intermingled with small yet noticeable pockets of the beta (β) grains. High magnification observation revealed very fine alpha (α) phase lamellae located within the beta (β) grain (Figure 3b).
Figure 3: Optical micrographs showing the key micro-constituents for the titanium material Ti–CP (Grade 2) at three different magnifications.
6.2 Hardness

Hardness is defined as the resistance offered by a material to permanent indentation or deformation. A hardness test is the most economical, reliable and efficient method for determining the mechanical properties of a metal. The presence of intrinsic variations in the microstructural features i.e., the alpha and beta phases, does cause a marked variation of the properties measured across a cross-section of the Ti-CP (Grade 2) sample.

6.2.1 Microhardness measurement

The Vickers microhardness measurements were made from edge-to-edge across the center of a sample mounted in bakelite. Multiple measurements were made in order to compensate for the variation of hardness values across the sample cross-section. The cumulative effect of strengthening observed in the material can be attributed to the following:

(a) The presence of the intrinsic microstructural features like the alpha and beta phases

(b) A net weakening effect, that can be ascribed to the presence of processing-related features like (i) fine microscopic voids and pores, and (ii) fine microscopic cracks that is intercepted by the pyramidal indenter resulting in a net decreased value for microhardness of the specimens. [22,23].

It can be further observed that, the hardness values when plotted reveal marginal spatial variability with an average value of 200kg/mm² for the Ti-CP (Grade 2). The variation of microhardness values across the length of fully annealed Ti-CP (Grade 2) material is shown in Figure 4.
6.2.2 Macrohardness measurement

The macrohardness values (Table 2) across the length of the specimen were measured based on the Rockwell C scale, with an average value of 260 kg/mm² for the Ti-CP (Grade 2) material. The variation of macrohardness values across the length of fully annealed Ti-CP (Grade 2) is shown in Figure 5. The bar graph depicting the average microhardness and macrohardness values of the material Ti-CP (Grade 2) is shown in Figure 6.
Figure 5: A profile showing the variation of macrohardness values across the length of fully annealed Ti-CP (Grade 2).

Figure 6: Bar graph depicting the average microhardness and macrohardness values of Ti-CP (Grade 2).
6.3 Tensile Properties

The room temperature tensile properties of Ti-CP (Grade 2) are summarized in **Table 4**. The results reported are the mean values based on duplicate tests. The elastic modulus, yield strength, ultimate tensile strength, elongation-to-failure and strength at failure (fracture) were provided as an output of the PC based data acquisition system. The yield strength was determined by identifying the stress at a point on the engineering stress versus engineering strain curve where a straight line drawn parallel to the elastic portion of the stress versus strain curve at 0.2% offset intersects the curve. The ductility is reported as elongation-to-failure over a gage length of 12.7 mm. This elongation was measured using a clip-on extensometer that was attached to the gage section of the test specimen [6].

**Table 4**: Room temperature tensile properties for Ti-CP (Grade 2) [Results are mean values based on duplicate tests]

<table>
<thead>
<tr>
<th>Material</th>
<th>Orientation</th>
<th>Elastic Modulus</th>
<th>Yield Strength</th>
<th>UTS</th>
<th>Elongation GL=0.5&quot; (%)</th>
<th>Reduction in Area (%)</th>
<th>Tensile Ductility ln(Ao/Af) (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ti-CP (Grade-2)</td>
<td>Longitudinal</td>
<td>16</td>
<td>115</td>
<td>63</td>
<td>432</td>
<td>14.7</td>
<td>43.4</td>
</tr>
<tr>
<td></td>
<td>Transverse</td>
<td>15</td>
<td>108</td>
<td>55</td>
<td>382</td>
<td>18.9</td>
<td>47</td>
</tr>
<tr>
<td>ASM Handbook</td>
<td></td>
<td>15.2</td>
<td>105</td>
<td>40-60</td>
<td>275-410</td>
<td>50</td>
<td>344</td>
</tr>
</tbody>
</table>
The representative stress versus strain curves for Ti-CP (Grade 2) for both the longitudinal (L) and transverse (T) orientations are shown in Figure 7 and Figure 8. The modulus of elasticity for Ti-CP (Grade 2) was determined to be 115GPa in the longitudinal orientation and 108GPa in the transverse orientation. The yield strength for Ti-CP (Grade 2) in the longitudinal orientation was determined to be 432 Mpa (63 ksi), which is 15 % higher than the value for the transverse orientation [382 Mpa (55 ksi)]. The ultimate tensile strength is 561 MPa (81 Ksi) in the longitudinal orientation and 465 MPa (67 ksi) in the transverse orientation, a noticeable difference of 20 pct. The ultimate tensile strength is higher than the yield strength indicating a noticeable work hardening beyond yield. The elongation of Ti-CP (Grade 2) alloy was 14.7% in the longitudinal orientation and 18.9% in the transverse orientation. The reduction-in-area was 43.4% in the longitudinal orientation and 47% in the transverse orientation. The engineering stress versus engineering strain curves for the material in both the longitudinal and transverse orientations are shown in Figure 7. Observation of the variation of stress with plastic strain, is shown in Figure 8, and reveals that Ti-CP (Grade 2) to be harder in the longitudinal orientation compared to the transverse orientation. The strain hardening exponent (n) of this material is 0.097 for the longitudinal (L) orientation and 0.094 for the transverse (T) orientation.
Figure 7: Influence of test specimen orientation on engineering stress versus engineering strain curve for Ti – CP (Grade 2).

Figure 8: The monotonic stress versus strain curve for Ti – CP (Grade 2)
6.4 Tensile Fracture Behavior

An exhaustive examination of the tensile fracture surfaces of both the longitudinal and transverse specimens in a scanning electron microscope (SEM) revealed the specific role played by intrinsic microstructural features and microstructural effects on strength and ductility properties of the chosen titanium material. Representative fractographs of the tensile fracture surface of Ti-CP (Grade 2) for the longitudinal orientation are shown in Figure 9 and for the transverse orientation in Figure 10.

Due to the occurrence and presence of fairly high localized stress intensities at selected points through the microstructure, the corresponding strain is conducive for: (i) the nucleation and early growth of the microscopic voids, and (ii) the eventual coalescence of the microscopic and macroscopic voids to occur at low to moderate stress levels [25, 26]. This results in the early initiation and presence of fine microscopic cracks distributed through the fracture surface [16]. During far-field loading, the fine microscopic voids tend to grow, or increase in size, and eventually coalesce with each other. The halves of these voids are the shallow dimples visible on the fracture surface. Coalescence of the fine microscopic cracks results in a macroscopic crack that is favored to propagate in the direction of the major stress axis. The irreversible plastic deformation that is occurring at the “local” level is responsible for aiding the growth of the cracks through the microstructure. The fine microscopic and macroscopic cracks tend to propagate in the direction of the major stress axis, suggesting the important role played by normal stress in promoting tensile deformation. Since a gradual extension of both the microscopic and macroscopic cracks under quasi-static loading essentially occurs at the high local stress intensities comparable with the fracture toughness of the material, the
presence of a population of fine microscopic and macroscopic voids does exert a detrimental influence on strain-to-failure ($E_t$) associated with ductile fracture. A gradual destruction of microstructure of the titanium material culminates in conditions at the “local” level that are conducive to crack extension to occur at locations of the prevailing local high stress intensities. The microscopic cracks tend to propagate in the direction of the tensile stress axis, suggesting the important role of normal stress in enhancing tensile deformation. Since crack extension under quasi-static loading occurs at the high local stress intensities comparable with the fracture toughness of the material, the presence of a healthy population of voids of varying size has a detrimental influence on strain-to-failure associated with ductile fracture.

On a macroscopic scale, tensile fracture of this material in the longitudinal orientation was at an inclination to the far-field stress axis (Figure 9a). Careful observation of the different regions of the fracture surface was made at high magnification to delineate the intrinsic fracture features. Isolated and randomly distributed microscopic voids and a healthy population of ductile dimples were found covering the transgranular fracture regions (Figure 9b). High magnification observations in the transgranular region revealed the dimples to be shallow and of varying size coupled with fine microscopic cracks (Figure 9c). The overload region revealed a combination of fine microscopic cracks, microscopic voids of varying size and randomly distributed through the surface, and a large population of shallow dimples, features reminiscent of locally brittle and ductile failure mechanisms (Figure 9d).
For the transverse specimen, overall morphology of the tensile fracture surface was also at an inclination to the tensile stress axis (Figure 10a). High magnification observation in the transgranular region revealed features similar to the longitudinal counterpart, i.e., a combination of microscopic voids of varying size and a large population of ductile dimples (Figure 10b). The dimples were shallow in nature and of varying size (Figure 10c). Observation in the tensile overload region revealed features that were quite similar to those observed in the transgranular region, i.e., microscopic voids of varying size and shallow dimples of non-uniform size and shape. The features observed through the fracture surface were reminiscent of locally ductile and brittle failure mechanisms governing the tensile response of the material under study, Ti – CP (Grade 2).
Figure 9: Scanning electron micrographs showing the tensile fracture surface of Ti-CP (Grade 2) specimen (orientation: Longitudinal) deformed in tension:

(a) Overall morphology of tensile fracture surface at an inclination to stress axis.
(b) High magnification of (a) showing population of voids and shallow dimples of varying size covering the transgranular region.
(c) High magnification of (b) showing the morphology and distribution of dimples and fine microscopic cracks features reminiscent of locally ductile and brittle failure mechanisms.
(d) The overload region showing microscopic voids, cracks and shallow dimples.
Figure 10: Scanning electron micrographs showing the tensile fracture surface of Ti-CP (Grade 2) specimen (orientation: Transverse) deformed in tension:

(a) Overall morphology of tensile fracture surface at an inclination to stress axis.
(b) High magnification of (a) showing isolated voids and shallow dimples of varying size covering the transgranular region.
(c) High magnification of (b) showing the morphology and distribution of dimples and fine microscopic voids features reminiscent of locally ductile failure mechanisms.
(d) The overload region showing randomly distributed microscopic voids, microscopic cracks and pockets of shallow dimples.
6.5 High Cycle Fatigue Resistance

The cyclic degradation of the strength of a material through the accumulation of damage is known as fatigue. The fatigue life ($N_f$) of a structure or a component is determined as the sum of the number of cycles to crack initiation ($N_i$) plus the number of cycles of crack propagation ($N_p$) eventually culminating in failure by fracture. A detailed study of the cyclic fatigue response of the CP (Grade 2) titanium is important for its selection and eventual use in a spectrum of performance-critical applications. A comprehensive understanding of the fatigue behavior of commercially pure titanium metal is rendered difficult by the presence and role of intrinsic microstructural features and their specific role in governing the deformation and fracture processes during cyclic loading. Studies have found and documented that the fatigue properties of the titanium alloys to be governed by the conjoint influence of (a) surface defects, (b) crystallographic texture, (c) intrinsic microstructural effects spanning the size and morphology of the beta grains, colonies of alpha platelets, and (d) the presence and role of other intrinsic microstructural features [10,27,28]. A few independent studies have shown that during stress-controlled high-cycle fatigue, the structure-property relationship in commercially pure titanium and titanium alloys is influenced by the presence, role and contribution of intrinsic microstructural features. Alloy microstructures that produce a low resistance to crack initiation, even at low values of applied stress, will usually yield the best endurance during high-cycle fatigue testing and even tensile resistance [29,30].

The crack initiation characteristics of commercially pure titanium subjected to high-cycle fatigue have been reported in literature as follows:
(a). Initiation of microscopic cracks occurs along the grain boundaries in the form of multiple slip bands within the grains which eventually coalesce to increase in size comparable to the grain dimensions.

(b). The formation of multiple stress concentrations, resulting from the progressive build-up of dislocations in the form of micro-cracks, along the grain boundaries followed by formation of voids.

The formation of voids and the subsequent phenomenon of void coalescence results in a change in the macroscopic stress response behavior of the Ti-CP (Grade 2) material characterized by gradual growth of the crack along the surface of the specimen due to strain localization. With continued cyclic deformation, this eventually leads to failure of the specimen.

The prime objective of conducting the stress-amplitude controlled fatigue tests was to establish the existence of a fatigue limit or endurance limit. The results of the axial stress-amplitude controlled tests are shown in Figure 11, in which the maximum stress ($\sigma_{\text{maximum}}$) is plotted as a function of cycles to fracture or failure ($N_f$). At the load ratio studied ($R = 0.1$), the variation of maximum stress ($\sigma_{\text{maximum}}$) with cyclic fatigue life ($N_f$) reveals increasing fatigue life with a decrease in stress amplitude. This trend is typical of most non-ferrous metals and their composite counterparts [6, 31]. However, commercially pure titanium does not appear to exhibit a well-defined endurance limit. At an equivalent maximum stress, the fatigue life of the transverse (T) specimen of CP (Grade 2) titanium is noticeably greater than the longitudinal (L) counterpart. This is shown in Figure 12. In an attempt to understand and rationalize the influence of ductility of the Ti-CP (Grade 2) material on high cycle fatigue response, the test data is
re-plotted as the variation of maximum elastic strain with cycles-to-failure ($N_f$). Herein, the maximum elastic strain is taken to be the ratio of maximum stress ($\sigma_{\text{maximum}}$) to the elastic modulus (E) of the Ti-CP (Grade 2). This is shown in Figure 13. At equivalent values of maximum elastic strain, the transverse specimen reveals noticeably improved fatigue life as compared to the longitudinal (L) counterpart.

To understand and rationalize the specific role of microstructure of the Ti-CP (Grade 2) on high cycle fatigue life at a load ratio of 0.1, the test data is re-plotted to take into account the strength of the Ti-CP (Grade 2) alloy in the two orientations, namely longitudinal (L) and transverse (T). When the maximum stress ($\sigma_{\text{maximum}}$) is normalized by the material yield strength ($\sigma_{\text{YS}}$), the transverse (T) specimens have noticeably improved fatigue life when compared to the longitudinal (L) specimens at equivalent values of the ratio (Figure 12). A near similar trend is observed when the maximum stress ($\sigma_{\text{maximum}}$) is normalized by the ultimate tensile strength ($\sigma_{\text{UTS}}$) of the material along the specific orientation. At equivalent values of the ratio [$\sigma_{\text{maximum}}/\sigma_{\text{UTS}}$], the transverse (T) orientation reveals improved fatigue life when compared to the longitudinal (L) counterpart (Figure 14).
Figure 11: Variation of maximum stress ($\sigma_{\text{max}}$) with fatigue life ($N_f$) for Ti-CP (Grade 2) at a stress ratio of ($R = 0.1$)

Figure 12: Variation of the ratio of maximum stress/yield stress ($\sigma_{\text{max}}/\sigma_y$) with fatigue life ($N_f$) for Ti-CP (Grade 2) at a stress ratio of ($R = 0.1$)
Figure 13: Variation of maximum elastic strain ($\sigma_{\text{max}}/E$) with fatigue life ($N_f$) for Ti-CP (Grade 2) at a stress ratio of ($R = 0.1$)

Figure 14: Variation of the ratio of maximum stress/ultimate tensile stress ($\sigma_{\text{max}}/\sigma_{\text{UTS}}$) with fatigue life ($N_f$) for Ti-CP (Grade 2) at a stress ratio of ($R = 0.1$)
6.6 Cyclic Fatigue Fracture

A careful examination of the cyclic fatigue fracture surfaces of the deformed and failed specimens was conducted in a JEOL scanning electron microscope (SEM) at the following levels of magnifications:

(A) At low magnification for the purpose of identifying (i) the region of microscopic crack initiation, and the nature and characteristics of early crack growth, (ii) the domain of stable crack growth, and (iii) final fracture, i.e., the region of overload.

(B) At gradually increasing magnification in both the region of stable crack growth and overload with the purpose of identifying or establishing (i) the nature of damage initiation, (ii) nature of both microscopic and macroscopic crack growth through the alloy microstructure, and (iii) other fine scale features on the fracture surface in the region of overload.

For the Ti-CP (Grade 2), the fracture surfaces revealed marginal to no difference in topographies at the different values of maximum stress ($\sigma_{\text{maximum}}$) and resultant fatigue life ($N_f$). On a microscopic scale, the nature, morphology, and volume fraction of the intrinsic features on the fatigue fracture surfaces were found not to vary with maximum stress and resultant fatigue life. Only representative fractographs of the fatigue fracture surfaces of the Ti-CP (Grade 2), at different values of maximum stress and resultant cyclic fatigue life, are shown in Figure 15 to Figure 22.
6.6.1 Cyclic Fatigue Fracture: Longitudinal Orientation

At a maximum stress of 405 MPa, the longitudinal specimen had a fatigue life of 45,532 cycles. Scanning electron microscopy observations at low magnifications revealed the overall morphology to comprise a small yet distinct region of crack initiation and early microscopic crack growth, which appeared like the radiance of the sun (Figure 15a). Careful high magnification observations revealed the region to be flat and near featureless, populated with extremely fine microscopic cracks that were randomly distributed (Figure 15b). High magnification observation of the region of slow and stable crack growth region revealed pockets of shallow striations, reminiscent of localized micro-plastic deformation (Figure 15c). In the region of overload, observation revealed a healthy population of dimples intermingled with voids of varying size and shape, features reminiscent of locally ductile failure mechanisms (Figure 15d). In the terminal stages of stable crack growth and approaching overload, the fatigue fracture surface revealed fine striations intermingled with a healthy population of fine microscopic cracks. The fine microscopic cracks grow in increments and eventually coalesce to form macroscopic cracks (Figure 16).

For a test specimen that was cyclically deformed at a lower maximum stress of 366 MPa with a resultant fatigue life of 125,559 cycles, the fatigue fracture surface revealed distinct regions of crack initiation and overload (Figure 17a). The transition from crack initiation and early crack growth to stable crack growth to overload was distinct as seen in (Figure 17b). High magnification observation in the region of slow and stable crack growth revealed pockets of well-defined striations reminiscent of localized microplastic deformation (Figure 17c). The overload region was rough and
comprised essentially of microscopic voids of varying size and shape intermingled with shallow dimples, features reminiscent of locally ductile failure processes (Figure 17d). High magnification observation in the region immediately prior to the onset of a short region of unstable crack growth and overload revealed pockets of striations within the grains, microscopic cracking along the grain boundaries, and microscopic cracking at and along grain boundary triple junctions (Figure 18).

Thus, for the longitudinal (L) orientation the microscopic fracture features observed for the specimens deformed at high maximum stress, resultant short fatigue life (N_f), and low maximum stress, resultant long fatigue life (N_f), were essentially similar.
Figure 15: Scanning electron micrographs of the fatigue fracture surface of Ti-CP (Grade 2) specimen (orientation: Longitudinal) deformed in cyclic fatigue under a maximum stress of 405MPa with a fatigue life ($N_f$) of 45,532 cycles:

(a) Overall morphology showing the different regions.
(b) Fine microscopic cracks in the region of early crack growth.
(c) Pockets of shallow striations in the region of stable crack growth.
(d) Healthy population of dimples and voids of varying size and shape in the region of overload.
Figure 16: Scanning electron micrographs of the fatigue fracture surface of Ti-CP (Grade 2) specimen (orientation: Longitudinal) deformed in cyclic fatigue under a maximum stress of 405 MPa with a fatigue life ($N_f$) of 45,532 cycles, showing growth and eventual coalescence of the fine microscopic cracks to form a macroscopic crack.
Figure 17: Scanning electron micrographs of the fatigue fracture surface of Ti-CP (Grade 2) specimen (orientation: Longitudinal) deformed in cyclic fatigue under a maximum stress of 366MPa with a fatigue life ($N_f$) of 125,559 cycles:

(a) Overall morphology showing distinct regions of crack initiation and overload.

(b) The transition region from crack growth to overload.

(c) Pockets of well defined striations in the region of stable crack growth.

(d) Microscopic voids of varying size and shape intermingled with dimples on the overload surface.
Figure 18: Scanning electron micrographs of the fatigue fracture surface of Ti-CP (Grade 2) specimen (orientation: Longitudinal) deformed in cyclic fatigue under a maximum stress of 366MPa with a fatigue life ($N_f$) of 125,559 cycles, showing microscopic cracking along grain boundaries and along grain boundary triple junctions.
6.6.2 Cyclic Fatigue Fracture: Transverse Orientation

Compared to samples oriented in the longitudinal direction, the transverse (T) oriented specimens of the Ti-CP (Grade 2) provided evidence of enhanced cyclic fatigue resistance at the same value of maximum stress at a load ratio of 0.1. At a maximum stress of 399 MPa and resultant fatigue life of 72,773 cycles, the intrinsic features observed on the fracture surface are as shown in (Figure 19). The overall fracture surface revealed distinct regions of crack initiation, stable crack growth followed by overload (Figure 19a). High magnification observation in the region between stable crack growth and overload revealed a healthy population of voids of varying size and shape intermingled with dimples; features reminiscent of locally ductile failure mechanisms (Figure 19b). The overload region revealed a population of voids of varying size and dimples. During repeated cyclic deformation, the fine microscopic voids grow and eventually coalesce to form a microscopic crack (Figure 19c).

The test specimen cyclically deformed at a lower maximum stress of 387 MPa had a cyclic fatigue life of 234,351 cycles. The overall morphology of the fatigue fracture surface is as shown in (Figure 20a). Scanning electron microscopy observations of the region of stable crack growth (Figure 20b) at higher magnifications revealed a gradual progress of fatigue damage through the microstructure in the radial direction (Figure 20b). High magnification observation of the transgranular regions of the fatigue fracture surface revealed pockets of well-defined striation like features (Figure 20c). Microscopic voids of varying size and shape coupled with shallow dimples were found covering the region of unstable crack growth (Figure 22d). The transgranular regions in the domain of stable crack growth were populated with fine microscopic cracks (Figure
which tend to grow progressively during repeated cyclic deformation and eventually coalesce to form macroscopic cracks having a random orientation and distribution (Figure 22b). At the higher allowable magnifications of the electron microscope fine and shallow striation-like features, reminiscent of localized micro-plastic deformation, were found covering the transgranular region of the fatigue fracture surface (Figure 22c).
Figure 19: Scanning electron micrographs of the fatigue fracture surface of Ti-CP (Grade 2) specimen (orientation: Transverse) deformed in cyclic fatigue under a maximum stress of 399MPa with a fatigue life ($N_f$) of 72,773 cycles:

(a) Overall morphology showing the regions of crack initiation and early crack growth and overload.

(b) Voids of varying size and shape and a healthy population of dimples.

(c) Growth and coalescence of the fine microscopic voids to form a microscopic crack.
Figure 20: Scanning electron micrographs of the fatigue fracture surface of Ti-CP (Grade 2) specimen (orientation: Transverse) deformed in cyclic fatigue under a maximum stress of 387MPa with a fatigue life ($N_f$) of 234,351 cycles:

(a) Overall morphology showing the two distinct regions of crack initiation and early crack growth and overload.
(b) Progress of fatigue damage in the region of stable crack growth.
(c) High magnification of (b) showing well-defined striations in the region of stable crack-growth.
(d) Microscopic voids of varying size and shape along with shallow dimples in the region of unstable crack growth.
Figure 21: Scanning electron micrographs of the fatigue fracture surface of Ti-CP (Grade 2) specimen (orientation: Transverse) deformed in cyclic fatigue under a maximum stress of 387MPa with a fatigue life ($N_f$) of 234,351 cycles, showing the transgranular region in the region of stable crack growth with a population of fine microscopic cracks and well defined striations.
Figure 2: Scanning electron micrographs of the fatigue fracture surface of Ti-CP (Grade 2) specimen (orientation: Transverse) deformed in cyclic fatigue under a maximum stress of 356MPa with a fatigue life ($N_f$) of 957,815 cycles:

(a) Overall morphology showing the two distinct regions of crack initiation and early crack growth and overload.

(b) Progress of fatigue damage in the region of stable crack growth.

(c) High magnification of (b) showing well defined striations in the region of stable growth.

(d) Microscopic voids of varying size and shape along with shallow dimples in the region of unstable crack growth.
Under the influence of a far-field applied stress, the fine microscopic voids present on the fatigue fracture surface did appear to have undergone limited growth thereby providing evidence for contributions from the presence of constraint-induced triaxiality on failure of the Ti-CP (Grade 2) matrix. The limited growth of the fine microscopic voids under the influence of an externally applied far-field stress coupled with lack of their coalescence as the dominant fracture mode inhibits the fracture from being all-ductile. This suggests that the plastic deformation properties of Ti-CP (Grade 2) in the as-received fully annealed condition are visibly altered by the presence of “local” constraints. Under the influence of repeated cyclic stressing, only a few of the fine microscopic voids appeared to have undergone limited growth to facilitate their linkage by coalescence. The halves of these voids are the shallow dimples that were observed covering the transgranular and overload regions of the fatigue fracture surface. The lack of formation of ductile dimples, as a dominant fracture mode, is attributed to the presence of “local” constraints in plastic flow on the commercially pure titanium metal matrix and not to the limited ductility of the metal.

The occurrence of microscopic cracking at and along the grain boundaries and grain boundary triple junctions is a direct result of the strength of the titanium material in this particular orientation. This is attributed to the presence of a large area fraction of weakened high angle grain boundaries that are parallel to the longitudinal direction. The preference for cracking along the grain boundaries results in a gradual loss of the through-thickness constraint causing an essentially plane strain fracture process to be divided into several plane stress fractures. The resultant beneficial influence of crack tip plasticity and tensile ductility is offset or limited by the occurrence of constraints in
deformation. It is the intrinsic microstructural features coupled with constraints in matrix deformation that promote the occurrence of failure through the grain boundaries. Continued propagation of the microscopic cracks along with the macroscopic cracks, through the commercially pure titanium microstructure, causes a progressive drop in load-carrying capability of the test specimen and eventually culminating in failure.

Based on an observation of the fracture surfaces of the test samples that were cyclically deformed over the range of stress amplitudes, the fracture plane of the deformed and failed specimen was essentially perpendicular to the loading axis, suggesting the importance of tensile stress in inducing and/or promoting fracture. Few of the voids that were created by the presence of intrinsic microstructural features did not grow extensively in the direction of the applied stress, which is generally the case for ductile failure of metallic materials. The observed lack of extensive void growth suggests that the overall strain that induces fracture is controlled by the conjoint influence of the void nucleation strain, crack propagation strain, and linkage strain.
PART- A

CHAPTER VII

CONCLUSIONS [Ti-CP (GRADE 2)]

A study of the high-cycle fatigue and final fracture behavior of test specimens machined from a commercially pure titanium (Grade 2) fully annealed plate stock for the two different orientations, i.e. longitudinal and transverse, provides the following key observations:

1. The as-received material in the fully annealed condition revealed the primary alpha (α) grains to be intermingled with small pockets of the beta (β) grains. Observation at the higher allowable magnifications of the optical microscope revealed very fine alpha (α) phase lamellae located within the beta (β) grains.

2. A critical comparison of the elastic modulus of this material for the two orientations reveals the modulus to be 8% higher in the longitudinal direction as compared to the transverse orientation.

3. The yield strength and tensile strength of this CP (Grade 2) titanium is higher for the longitudinal orientation when compared to the transverse orientation. The ultimate tensile strength of the CP titanium plate stock is marginally higher than the yield strength indicating the tendency for strain hardening of the material beyond yield to be low.
4. The Ti-CP (Grade 2) plate stock in both the longitudinal (L) and transverse (T) orientations had adequate ductility, quantified in terms of elongation-to-failure ($\epsilon_f$), which makes it readily receptive and easily conducive for mechanical working operations. Both elongation-to-failure and reduction-in-area were marginally higher in the longitudinal orientation when compared to the transverse orientation.

5. The maximum stress ($\sigma_{\text{maximum}}$) versus fatigue life ($N_f$) curve shown by this material is quite different from those non-ferrous metals that exhibit a well-defined endurance limit. When compared at equal values of maximum stress at a load ratio of 0.1, the fatigue life of the transverse specimen is noticeably greater than the longitudinal counterpart. At equivalent values of maximum elastic strain, the transverse specimens revealed noticeably improved fatigue life when compared one-on-one to the longitudinal counterparts.

6. At a chosen load ratio of 0.1, the cyclic fatigue fracture surfaces revealed minimal differences in the nature and volume fraction of the intrinsic features on the fracture surface as a function of maximum stress and resultant cyclic fatigue life. The regions of crack initiation and early microscopic crack growth and stable crack growth were essentially flat and transgranular. The region of stable crack growth did reveal a population of fine microscopic and macroscopic cracks coupled with “pockets” of well-defined striations reminiscent of the occurrence of micro-plastic deformation at the local level. The overload region was covered with a healthy population of dimples of varying shape and fine microscopic voids of varying size reminiscent of the locally operating failure mechanisms.
PART-B
CHAPTER I
LITERATURE REVIEW [Ti-6Al-4V]

In this chapter, a brief review of the published literature encompassing the domains of processing, microstructure and mechanical behavior of the titanium alloy Ti-6Al-4V is presented. Among the earliest titanium alloys developed in the United States during the early 1940s, Ti-6Al-4V emerged to be the most popular alloy, which was widely chosen and used for aerospace and other performance-critical applications spanning both the defense and civilian sectors. Owing to its excellent mechanical properties like high specific strength (σ/ρ), stiffness, corrosion resistance, good erosion resistance in environments spanning a range of aggressiveness, acceptable mechanical properties at elevated temperature coupled with the ability to withstand and function at elevated temperatures, this alloy of titanium has gradually found varied use in aerospace and other space-related applications. The main factors that facilitate the selection and use of titanium alloys are the following:

(a) Weight reduction (substitute to steel or nickel-base alloys)

(b) Application temperature (substitute to aluminum, steel or nickel based alloys)

(c) Corrosion resistance (substitute to aluminum alloys or low alloyed steels)

(d) Galvanic compatibility with polymer matrix (substitute to aluminum alloys)

(e) Space limitation (substitute to Aluminum alloys).
The need to reduce costs while concurrently not compromising on the growing demand for maximization of component performance has motivated manufacturers of end components to adopt a life cycle approach to material selection for purposes of design. Besides, the need for a healthy combination of good mechanical properties, other factors like cost, machinability, castability and weldability are also important and must be considered in the selection and use of a titanium alloy for a performance specific application. The other areas where titanium alloys have been chosen for use include the following [32]

1. To prevent the propagation of fatigue cracks in aircraft fuselages by placing thin narrow rings around the aluminum fuselage.
2. Use of Ti-3Al-2.5V for hydraulic tubing in aircrafts has resulted in a weight reduction of up to 40% compared to tubes that are made of steel.
3. Commercial purity titanium is used where high corrosion resistance is required coupled with moderate strengths. Aircraft floors surrounding onboard kitchens and toilets are the areas where emphasis and importance of the corrosive environment comes into picture.
4. The piping system for de-icing equipment is manufactured from unalloyed titanium.
5. For other important parts, such as, aircraft landing gear and cockpit windows have been constructed from forged titanium despite its huge cost.

Compared with the aircrafts manufactured for use in the civil commercial sector, the use of alloys of titanium in military aircraft is noticeably higher [32]. The alloys of titanium account for 30-50% of the weight of a modern fighter jet. Most of this weight can be
found in the engine bay of an aircraft where temperatures are so high that aluminum and its alloys cannot be used as a viable alternative. The key difference in the selection and use of titanium alloys in commercial aircraft market compared to the military aircraft market is that overall cost effectiveness is important for commercial aircraft, while high performance requirements govern their selection and use for the military aircraft. The noticeably high machining cost that is typical of titanium forgings has created a need for an effective optimization of the forging processes.

Major improvements in gas turbine efficiency have been made possible due to the introduction of titanium alloys for the key rotating parts. The need for more efficient use of titanium alloys is restricted by the inefficiency of the extremely conservative nature of fatigue life prediction for the alloys of titanium. Bache et al [33] have tried to examine the early crack growth phase and the factors involved in the formation of quasi-cleavage facets. The parameters like stress redistribution, microstructure and grain-refinement were studied. It has been established through earlier research that manipulation of the microstructural conditions in near α (alpha) and α/β (alpha/beta) alloys can be used to achieve the desired mechanical properties. While a fine grain size gives good ultimate and yield strength and resistance to fatigue crack initiation, a coarse lamellar structure provides improved toughness, fatigue crack propagation resistance and creep compatibility. Bimodal structures can be produced, which give an optimized combination of both fatigue and creep resistance. Traditional design procedures, which rely on ‘life to first crack’ methodology have been in use and they rely primarily on the stress-fatigue life data obtained from S-N curves combined with statistical data to establish a safe cyclic life for a given set of operating stress conditions. Though these techniques have been
proven to be safe, they are conservative and thus inefficient. In spite of improved processing technologies and reliable non-destructive evaluation methodologies, the assumption of inherent flaws in the material for calculating the safe life was considered to be conservative and inefficient. Studies are being conducted in order to create a ‘total life’ philosophy, which combines the best of these two design philosophies. But a thorough understanding of fatigue crack initiation and crack growth behavior is essential for such studies.

Crack initiation in the α / β (alpha/beta) titanium alloys was found to be associated with formation of quasi-cleavage facets, which are formed under both static and cyclic loading conditions [33]. These facets are considered to be the result of separation at the intense slip bands under the action of a tensile stress normal to the plane of slip [33]. Stress-redistribution is assumed to play a major role in this process. The effects of grain size and mechanical anisotropy have clearly shown that processing techniques controlling these features have a direct effect on the fatigue performance through control of the formation of quasi-cleavage facets [33]. Also, under certain circumstances, the deliberate introduction of a strong global texture was found to significantly improve the fatigue response of the materials chosen and studied in comparison with critical in-service stresses or strains. The mechanical performance is governed by the basal texture and strong textures can be achieved by rolling [33]. In these studies, a uniform strain rate (\( \dot{\varepsilon} \)) of 0.0016 % / min was applied to cylindrical specimens of 6mm diameter and 10mm gage length. An increase in elastic modulus, yield stress and ultimate tensile stress was often reported for the transverse orientation [33]. However, strain control fatigue data at 20°C indicates a stronger fatigue performance for specimens
oriented in the longitudinal direction. Interestingly, under load control conditions, where the maximum and minimum stress levels are fixed, the fatigue performance characteristics are reversed with transverse orientations providing longer fatigue lives [33].

Tokaji et al [34] have studied the fatigue properties of near -alpha titanium alloys, which are considered as high temperature resistance alloys. Fatigue tests were performed on Ti-6Al-4V at the elevated temperatures of 623K and 723K and the fatigue behavior of the material was carefully examined and evaluated. Fatigue specimens having 6mm width, 4mm thickness and 20 mm gauge length were machined in the longitudinal direction. A shallow notch of depth of 0.5mm was introduced for the purpose of observation of crack initiation and its subsequent growth through the alloy microstructure. The mechanical properties of the material at room temperature were determined to be:

(i). Proof stress = $\sigma_{0.2} = 880$Mpa,
(ii). Tensile strength = $\sigma_B = 933$MPa,
(iii). Elongation ($\epsilon_f$) of 14%, and
(iv). Reduction in area (RA) of 26%.

Cyclic fatigue test were performed using an electro-hydraulic fatigue testing apparatus at a frequency of 10Hz and a sinusoidal wave form at: (a) room temperature, (b) 623K, and (c) 723K. The temperature of the specimen was controlled using an induction heating system and temperature control was found to be well within 1% of the desired temperature [34].
The researchers noted that the tensile strengths obtained using fatigue specimens at 623K was 602 MPa and at 723K, it was measured to be 402MPa. Their results revealed a significant reduction in strength with an increase in test temperature. The failure and fracture surfaces were significantly different between samples deformed at room temperature and the sample deformed at an elevated temperature. Based on the results obtained from their exhaustive experiments, the researchers have concluded that an increase in temperature led to a significant loss of fatigue strength [34]. A similar trend was observed for fatigue strength in terms of fatigue ratio at the higher temperatures and the fracture surfaces were found to be predominantly brittle for the sample cyclically deformed at the higher temperatures than the sample cyclically deformed at ambient temperature. Early crack initiations and rapid growth of the small fatigue cracks were observed at the higher temperatures indicating that fatigue life is governed by the formation, growth and eventual coalescence of small fatigue cracks [34].

Morrissey et al., [27] collected data on the high-cycle fatigue behavior of titanium alloys at low stress ratios and frequencies. However, for applications like turbine engines, higher frequencies and stress values will be experienced by the chosen material. Due to the inherent difficulties in achieving test conditions that can simulate high frequencies and loads, limited data was collected and documented. Since titanium is a strain-rate sensitive material their independent investigation aimed at studying the fatigue behavior of Ti-6Al-4V at higher frequencies of the order of 1 kHz, which was closer to the actual service conditions. The fatigue tests were performed at 70 Hz, 400 Hz and 1800 Hz at three different stress ratios [R] of R =0.1, R= 0.5 and R =0.8. The Ti-6Al-4V alloy was subjected to heat treatment. The material properties in the longitudinal direction were
determined to be $\sigma_y = 980 \text{ MPa}$ and $\sigma_{UTS} = 1030 \text{ MPa}$. Two different testing systems were utilized in their study with the 70 Hz and 400 Hz tests performed on a custom built high frequency testing apparatus consisting of a pneumatic actuator for applying the static load and an electromagnetic shaker for the superimposed high frequency load. The tests at 1800Hz were performed using a prototype design for applying loads at very high frequencies. For the tests at 70 Hz and 400 Hz, the researchers found the scatter in test data to be little (less than 5%) [27]. However, there was an appreciable amount of scatter for the tests at the highest frequency, i.e., 1800Hz. Further, they observed a significant frequency effect for the Ti-6Al-4V at the lower stress ratios. As the frequency increased, the stress required for failure of the sample following $10^7$ cycles was considerably higher [27]. Many materials to include Ti-6Al-4V show a strain rate dependence of plastic flow during a constant strain rate test. Morrissey et al [27] determined the effect to be smaller for the stress ratios used in their testing. However, at the higher rates, the effect was magnified. A relatively small increase in the maximum stress was observed when the frequency was increased from 70 Hz to 400 Hz and a larger increase in the maximum stress when the frequency was increased to 1800Hz. As the frequency of loading was increased, the mobile dislocations had little time to overcome the obstacles through thermal activation. Consequently there is an apparent reduction in the accumulated plastic strain. They observed that this to be possible only at the higher stress ratios where plastic strain can occur. Since fatigue is a phenomenon induced by plasticity, they were able to convincingly rationalize the observed increase in fatigue strength at the higher frequencies [27].
Filip et al., [35] attempted to study the effect of heat treatment on microstructural properties of two titanium alloys and its intrinsic influence on strength. The alloys of titanium studied were (i) Ti-6Al-4V, and (ii) Ti-6Al-2Mo-2Cr. They observed the heat treatment to result in a wide variation in properties of the two titanium alloys. The properties like yield stress, tensile stress and ductility were noticeably improved through a bimodal microstructure, whereas the alloy having a lamellar microstructure was characterized to have a combination of high fracture toughness and high fatigue crack propagation resistance. For example, with an increase in the cooling rate and content of β-stabilizing elements, the thickness and length of the α-phase decreased. For a cooling rate in the range of 1.2 K/s to 40 K/s, the microstructure comprised of: (i) stable α and β phases, and (ii) martensitic phases α’ (α”) [35]. However, at a low cooling rate of θc <1.2 Ks⁻¹ a stable lamellar microstructure was observed.

The yield stress of the specimen increased at a constant cooling rate and a probable cause for the observed behavior was proposed to be due to refinement of the lamellar microstructure. A reduction in slip length and a decrease in ductility arising from an inter-granular fracture mode resulted in intermediate rates of cooling causing the largest tensile elongation. Regardless of the observed difference in fracture toughness values, it was observed that all of the specimens revealed an identical fracture surface, which indicated a fracture mechanism characterized by rapid and unstable crack propagation [35].

Very few studies have documented the role of foreign-object damage (FOD) in the initiation and early growth of small surface fatigue cracks under conditions of high
cycle fatigue (HCF). Peters et al., [36] studied the effect of microstructural characteristics on fatigue crack initiation and fatigue crack growth by comparing a fine grain bimodal microstructure with a coarse grain beta (β) annealed lamellar microstructure. They subjected the Ti-6Al-4V alloy to carefully controlled thermo-mechanical treatment to obtain the required microstructures. In their study, they determined the uniaxial tensile properties at an initial stress rate of $8 \times 10^{-4} \text{s}^{-1}$. The lamellar microstructure revealed higher yield strength (975MPa) compared to the bimodal counterpart (915MPa). In their research study, the foreign-object-damage (FOD) was simulated by high velocity impact (200-300 m/s) of steel spheres onto the specimen surface. Subsequent to the impact, the specimens were cycled using a sinusoidal waveform at load ratios of $R=0.1$ and $R=0.5$. A significant reduction in fatigue strength was observed for both microstructures: (i) lamellar, and (ii) bimodal. The primary cause was determined to be the premature initiation of fatigue cracks due to stress concentration at the zones of foreign object damage (FOD) [36]. Also, they observed and recorded an increase in impact velocity to result in an increase in damage for both microstructures with the FOD initiated microscopic cracks providing preferred nucleation sites for growth of the fatigue crack. However, low velocity impacts (< 250 m/s) revealed no microscopic cracking indicating that low velocity indentations do not provide a realistic simulation of foreign object damage. A noticeable reduction in high cycle fatigue strength was attributed to the creation of preferred sites for the premature initiation of fatigue cracks in the zones of FOD indentation.
PART-B

CHAPTER II

MATERIAL [Ti-6Al-4V]

The second material chosen for this study, was provided by Allegheny Technologies ATI Wah Chang (Albany, OR) in the form of six pieces of Ti-6Al-4V to the following dimensions:

• Plate 0.25” TK X 12.00” wide X 36.00” Long six pieces

The Ti-6Al-4V alloy is receptive to heat treatment and can be solution heat treated and annealed to achieve the desired strength and properties. The Ti-6Al-4V alloy was provided by the manufacturer in the fully-annealed condition. The material supplied by ATI Wah Chang is a commercial titanium alloy and therefore the chemical composition is proprietary to the supplier. From published literature for similar titanium alloys, the chemical composition is expected to be as shown in Table 5.

Table 5: Nominal chemical composition (in weight percent) of the titanium alloy Ti-6Al-4V

<table>
<thead>
<tr>
<th>Material</th>
<th>Ti</th>
<th>Al</th>
<th>N</th>
<th>V</th>
<th>C</th>
<th>Fe</th>
<th>H</th>
<th>O</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ti-6Al-4V</td>
<td>90.0</td>
<td>6.0</td>
<td>0.05</td>
<td>4.0</td>
<td>0.1</td>
<td>0.4</td>
<td>0.02</td>
<td>0.20</td>
</tr>
</tbody>
</table>
The test samples used in this study were prepared as per the specifications prescribed in the standard ASTM E-8. The samples were precision machined from the as-provided annealed plate stock both in transverse (T) and longitudinal (L) orientations. In case of the longitudinal orientation, the specimens were machined such that the major stress axis was parallel to the rolling direction, whereas, in case of the transverse orientation, the specimens were machined such that the major stress axis is perpendicular to the rolling direction of the alloy plates. The test samples used in this study were identical to the Ti-CP (Grade 2) specimens. A schematic of the test specimen used for tensile and high-cycle fatigue tests is given in the Figure 2.
PART-B
CHAPTER IV
EXPERIMENTAL PROCEDURES [Ti-6Al-4V]

4.1 Initial microstructure characterization

An initial characterization of the microstructure of the as-provided material was done using a low magnification optical microscope. Samples of desired size were cut from the as-received stock of the Ti-6Al-4V and mounted in bakelite. The mounted samples were then wet ground on progressively finer grades of silicon carbide impregnated emery paper using copious amounts of water both as a lubricant and as a coolant. Subsequently, the ground samples were mechanically polished using five-micron diamond solution. Fine polishing to a perfect mirror-like finish of the surface of the Ti-6Al-4V was achieved using one-micron diamond solution as the lubricant. The polished samples were subsequently etched using a reagent that is a solution mixture of 5-ml of nitric acid (HNO₃), 10 ml of hydrofluoric acid (HF) and 85 ml of water (H₂O). The polished and etched surface of the samples of Ti-6Al-4V was observed under an optical microscope and photographed using bright field illumination technique.

4.2 Hardness testing

Hardness of a material is the mechanical property defined as the resistance offered by the material to indentation i.e. permanent deformation and cracking [22].
The hardness was measured using a Vickers (HV) micro hardness test machine. The test itself is simple and a nondestructive method that is easy to perform and widely used for the purpose of determining the mechanical properties of monolithic metals and their alloy counterparts.

The macro-hardness measurements (RC) were made on a Rockwell hardness machine using an indentation load of 140 Kgf, a minor load of 10 Kgf, 120 degree diamond cone, a dwell time of 10 seconds and the value read on the ‘C’ scale. The macro-hardness tests were also done on the polished surface of the titanium alloy test specimen. Two samples of Ti-6Al-4V were examined for micro-hardness and macro-hardness measurements. The results of the micro-hardness and macro-hardness tests are summarized in Table 6 and Table 7.
Table 6: A compilation of macrohardness measurements made on the Ti-6Al-4V

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Trial 1</th>
<th>Trial 2</th>
<th>Trial 3</th>
<th>Trial 4</th>
<th>Trial 5</th>
<th>Average</th>
</tr>
</thead>
<tbody>
<tr>
<td>Sample 2A</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Ti-6Al-4V</td>
<td>Vickers hardness</td>
<td>340.41</td>
<td>325.20</td>
<td>332.72</td>
<td>334.61</td>
<td>340.46</td>
</tr>
<tr>
<td><strong>Rc</strong></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Sample 2B</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Ti-6Al-4V</td>
<td>Vickers hardness</td>
<td>330.88</td>
<td>342.44</td>
<td>332.74</td>
<td>327.11</td>
<td>338.51</td>
</tr>
<tr>
<td><strong>Rc</strong></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

Table 7: A compilation of macrohardness measurements made on the Ti-6Al-4V

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Trial 1</th>
<th>Trial 2</th>
<th>Trial 3</th>
<th>Trial 4</th>
<th>Trial 5</th>
<th>Average</th>
</tr>
</thead>
<tbody>
<tr>
<td>Sample 2A</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Ti-6Al-4V</td>
<td>Vickers hardness</td>
<td>286</td>
<td>286</td>
<td>302</td>
<td>279</td>
<td>286</td>
</tr>
<tr>
<td><strong>Rc</strong></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Sample 2B</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Ti-6Al-4V</td>
<td>Vickers hardness</td>
<td>286</td>
<td>294</td>
<td>310</td>
<td>279</td>
<td>286</td>
</tr>
<tr>
<td><strong>Rc</strong></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>
4.3 Mechanical testing

Tensile and high cycle fatigue tests were conducted on the specimens to systematically study the tensile deformation, cyclic fatigue and final fracture behavior for the Ti-6Al-4V alloy. The details of the tests are given in the following sections.

4.3.1 Tensile tests

Uniaxial tensile tests were performed on a fully-automated, closed-loop servo-hydraulic mechanical test machine [INSTRON-8500 Plus] using a 100 kN load cell. The tests were conducted at room temperature (300 K) and in the laboratory air (Relative Humidity of 55 pct) environment. The test specimens were deformed at a constant strain rate of 0.0001/sec. An axial 12.5-mm gage length clip-on type extensometer was attached to the test specimen at the gage section using rubber bands. The stress and strain measurements, parallel to the load line, and the resultant mechanical properties, such as, stiffness, strength (yield strength and ultimate tensile strength), failure stress and ductility (strain-to-failure) was provided as a computer output by the control unit of the test machine. The average values of yield strength obtained from the tensile tests were used for calculating the following data required for the high-cycle fatigue tests.

(a) Yield load based on the yield stress obtained from the tensile tests.

(b) Maximum stresses, as percentages of the yield stress of the alloy for purpose of testing the specimens under conditions of stress amplitude-controlled fatigue.
4.3.2 High cycle fatigue tests

The stress-amplitude controlled high cycle fatigue tests were performed using a sinusoidal waveform at a stress ratio \([R = \frac{\sigma_{\text{min}}}{\sigma_{\text{max}}}]\) of 0.1. The cyclic loading of the test specimen was conducted at a frequency of 5Hz. At a given stress ratio \([R = \frac{\sigma_{\text{minimum}}}{\sigma_{\text{maximum}}}]\), the fatigue tests were conducted over a range of stress amplitudes to establish the variation of maximum stress \((\sigma_{\text{maximum}})\) with cyclic fatigue life \((N_t)\).

The data collected from the high-cycle fatigue test is used to establish the following relationships.

1. Variation of maximum stress \((\sigma_{\text{max}})\) with fatigue life \((N_t)\).
2. Variation of the ratio of maximum stress/yield stress \((\sigma_{\text{max}}/\sigma_y)\) with fatigue life \((N_t)\).
3. Variation of maximum elastic strain \((\sigma_{\text{max}}/E)\) with fatigue life \((N_t)\).
4. Variation of the ratio of maximum stress/ultimate tensile stress \((\sigma_{\text{max}}/\sigma_{\text{UTS}})\) with fatigue life.

The fatigue test data for cylindrical samples of the Ti – 6Al-4V (at \(R = 0.1\)) are summarized in Appendix B.

4.4 Failure-damage analysis

The fracture surfaces of the deformed and failed tensile specimens of Ti-6Al-4V were comprehensively examined in a scanning electron microscope (SEM) to determine the macroscopic fracture mode and to concurrently characterize the fine scale topography of the tensile fracture surface for the purpose of establishing microscopic mechanisms governing fracture. The distinction between the macroscopic mode and microscopic
fracture mechanisms is based entirely on the magnification level at which the observations are made. The macroscopic mode refers to the overall nature of failure while the microscopic mechanisms relate to the local failure processes, such as: (i) microscopic void formation, (ii) microscopic void growth and coalescence, and (iii) nature, intensity and severity of the fine microscopic and macroscopic cracks dispersed through the fracture surface. The samples for observation in the SEM were obtained from the failed tensile specimens by sectioning parallel to the fracture surface.
5.1 Initial Microstructure

The microstructure of an alloy is an important factor that determines its mechanical properties to include tensile properties, fracture toughness, fatigue resistance and resultant fracture behavior. The optical microstructure of the Ti-6Al-4V alloy is shown in Figure 23 at three different magnifications. These figures reveal that the as-received, undeformed microstructure of the Ti-6Al-4V alloy to be slightly different when compared to Ti-CP (Grade2). The difference is primarily in the volume fraction, morphology, size, and distribution of the intrinsic micro-constituents in the microstructure.

At equivalent high magnification (500 x), the intrinsic microstructural constituents of the Ti-6Al-4V, is shown in Figure 23b bringing out clearly the size, morphology, volume fraction and distribution of the alpha (α) and beta (β) phases. Observations of the Ti-6Al-4V alloy over a range of magnifications spanning very low to high magnification revealed a duplex microstructure consisting of the near equiaxed alpha (α) and transformed beta (β) phases.
The primary near equiaxed shaped alpha (α) grains (light in color) was well distributed in a lamellar matrix with transformed beta (dark in color) (Figure 23c). The size of both the alpha grains and the transformed β grains range from 10 to 15 microns. The presence of trace amounts of aluminum and oxygen in the Ti-6Al-4V alloy contributes to stabilizing and strengthening the alpha (α) phase, which is overall beneficial for enhancing hardenability, increasing strength, and improving the response kinetics of the alloy to heat treatment [26].

5.2 Hardness

Hardness is defined as the resistance offered by a material to permanent indentation. A hardness test is the most economical, reliable and efficient method for determining the mechanical properties of a metal. The presence of intrinsic variations in the microstructural features i.e., the alpha and beta phases, does cause a variation in the properties measured across a cross-section of the alloy sample.

5.2.1 Microhardness measurement

The Vickers microhardness measurements were made from edge-to-edge across the center of a sample mounted in Bakelite. Multiple measurements were made in order to compensate for the variation of hardness values across the sample cross-section. The cumulative effect of strengthening observed in the material can be attributed to the presence of the intrinsic microstructural features like the alpha and beta phases and also a net weakening effect, due to the presence of processing-related defects like (i) fine microscopic voids and pores, and (ii) fine microscopic cracks that are intercepted by the pyramidal indenter resulting in a decreased value for microhardness of the specimens [23,
It is further observed that the hardness values when plotted reveal marginal spatial variability with an average value of 330 kg/mm$^2$ for the Ti-6Al-4V alloy. We conclude from a comparative study of the hardness values of Ti-6Al-4V and Ti-CP (Grade 2), that the commercially produced and annealed alloy is noticeably harder than commercially pure (Grade 2) titanium.
Figure 23: Optical micrographs showing the key micro-constituents, their morphology and distribution, in the Ti-6Al-4V alloy at three different magnifications.
5.2.2 Macrohardness measurement

The macrohardness values (Table 6) across the length of the specimen were measured on the Rockwell C scale, with an average value of 290 kg/mm$^2$ for the Ti-6Al-4V alloy. However, the micro-hardness value of 330 kg/mm$^2$ is significantly higher than the macro-hardness value of 290 kg/mm$^2$ as shown in Figure 26. The observed lower macro-hardness for the Ti-6Al-4V alloy when compared to its micro-hardness is attributed to the presence of the processing-related defects, such as (i) fine microscopic voids and pores, and (ii) fine microscopic cracks.
Figure 25: A profile showing the variation of macrohardness values across the length of fully annealed Ti-6Al-4V.

Figure 26: Bar graph depicting the average microhardness and macrohardness values of Ti-6Al-4V alloy.
5.3 Tensile Properties

The room temperature tensile properties of the Ti-6Al-4V alloy are summarized in Table 8. The results reported are the mean values based on duplicate tests. The elastic modulus, yield strength, ultimate tensile strength, elongation to failure and strength at failure (fracture) were provided as an output of the PC-based data acquisition system. The yield strength was determined by identifying the stress at a point on the engineering stress versus engineering strain curve where a straight line drawn parallel to the elastic portion of the stress versus strain curve at 0.2% offset intersects the curve. The ductility is reported as elongation-to-failure over a gage length of 12.7 mm. This elongation was measured using a clip-on extensometer that was attached to the gage section of the test specimen [24].

Table 8: Summary of the room temperature tensile properties of Ti-6Al-4V
[Results are mean values based on duplicate tests]

<table>
<thead>
<tr>
<th>Material</th>
<th>Orientation</th>
<th>Elastic Modulus</th>
<th>Yield Strength</th>
<th>UTS</th>
<th>Elongation GL=0.5” (%)</th>
<th>Reduction in Area (%)</th>
<th>Tensile Ductility ln(Ao/Af) (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ti-6Al-4V</td>
<td>Longitudinal</td>
<td>18</td>
<td>126</td>
<td>137</td>
<td>948</td>
<td>154</td>
<td>1060</td>
</tr>
<tr>
<td></td>
<td>Transverse</td>
<td>20</td>
<td>137</td>
<td>152</td>
<td>1047</td>
<td>171</td>
<td>1181</td>
</tr>
<tr>
<td>ASM Handbook</td>
<td></td>
<td>-</td>
<td>114</td>
<td>-</td>
<td>115-160</td>
<td>800-1100</td>
<td>130-180</td>
</tr>
</tbody>
</table>

81
The representative stress versus strain curves for the Ti-6Al-4V alloy for both the longitudinal (L) and transverse (T) orientations are shown in Figure 27 and Figure 28. The modulus of elasticity of the Ti-6Al-4V alloy was determined to be 126 GPa in the longitudinal orientation and 137 GPa in the transverse orientation. The yield strength of the alloy in the annealed condition is determined to be 948 MPa (137 ksi) in the longitudinal orientation and 1047 MPa (152 ksi) in the transverse orientation. The ultimate tensile strength of the alloy is 1060 MPa (154 ksi) in the longitudinal orientation and 1181 MPa (171 ksi) in the transverse orientation. The ultimate tensile strength of the Ti-6Al-4V alloy is marginally higher than the yield strength, i.e., about 10 percent, indicating the work hardening rate beyond yield to be low. The elongation-to-failure (εf) of the alloy is 7.8% in the longitudinal orientation and 11.5% in the transverse orientation. The reduction-in-area of the alloy was 24% in the longitudinal orientation and 22% in the transverse orientation and accords well with the annealed condition of the alloy microstructure. The yield strength and tensile strength values of the alloy conform well with the values obtained and reported by the manufacturer (ATI Wah Chang) and recorded in ASM Metals Handbook [37]. Variation of stress with plastic strain, as shown in Figure 27 reveals the Ti-6Al-4V to be marginally harder in the transverse orientation when compared to the longitudinal orientation. The strain hardening exponent (n) of the alloy is 0.095 for the transverse orientation and 0.085 for the longitudinal orientation as shown in Figure 28.
Figure 27: Influence of test specimen orientation on engineering stress versus engineering strain curve of the Ti-6Al-4V alloy.

Figure 28: The monotonic stress versus strain curve for the Ti-6Al-4V alloy.
5.4 Tensile Fracture Behavior

An exhaustive examination of the tensile fracture surfaces of both the longitudinal and transverse specimens in a scanning electron microscope (SEM) revealed the specific role played by intrinsic microstructural features and microstructural effects on strength and ductility properties of the specific titanium material. Representative fractographs of the tensile fracture surface of the Ti-6Al-4V alloy for the longitudinal specimen are shown in Figure 29 and for the transverse orientation in Figure 30.

Due to the occurrence and presence of fairly high localized stress intensities at selected points through the microstructure, the corresponding strain is conducive for: (i) the nucleation and early growth of the microscopic voids, and (ii) the eventual coalescence of the microscopic and macroscopic voids to occur at low to moderate stress levels [25, 26]. This result in the early initiation and presence of very fine microscopic cracks distributed through the fracture surface [16]. During far-field loading, the fine microscopic voids tend to grow, or increase in size, and eventually coalesce with each other. The halves of these voids are the shallow dimples visible on the fracture surface. Coalescence of the fine microscopic cracks results in a macroscopic crack that is favored to propagate in the direction of the major stress axis. The irreversible plastic deformation that is occurring at the “local” level is responsible for aiding the growth of the cracks through the microstructure. The fine microscopic and macroscopic cracks tend to propagate in the direction of the major stress axis, suggesting the important role played by normal stress in promoting tensile deformation. Since a gradual extension of both the microscopic and macroscopic cracks under quasi-static loading essentially occurs at the high local stress intensities comparable with the fracture toughness of the material, the
presence of a population of fine microscopic and macroscopic voids does exert a
detrimental influence on strain-to-failure (ε_f) associated with ductile fracture. A gradual
destruction of microstructure of the titanium material culminates in conditions at the
“local” level that are conducive to crack extension to occur at locations of the prevailing
local high stress intensities. The microscopic cracks tend to propagate in the direction of
the tensile stress axis, suggesting the important role of normal stress in enhancing tensile
deforation. Since crack extension under quasi-static loading occurs at the high local
stress intensities comparable with the fracture toughness of the material, the presence of a
healthy population of voids of varying size has a detrimental influence on strain-to-failure
associated with ductile fracture.

At the macroscopic level, tensile fracture of the test specimen machined from the
plate stock was essentially normal to the far-field stress axis. Overall morphology of the
tensile fracture surface appeared to be rough and layered (Figure 29a). Observation of
the fracture surface at higher magnifications revealed macroscopic and fine microscopic
cracks (Figure 29b) and a population of shallow dimples of varying size and shape
covering the transgranular fracture surface (Figure 29c). At the higher allowable
magnification of the SEM, a population of fine microscopic voids and shallow dimples
were found covering the region of tensile overload (Figure 29d). An observation of the
fracture surface of the same test specimen at a different location revealed the
transgranular regions to be covered with pockets of shallow dimples and randomly
distributed fine microscopic voids. The morphology of the fine microscopic voids on the
transgranular fracture surface can be seen in Figure 29c.
The fractographs of the test specimen machined from the transverse orientation of the as-provided plate stock are shown in Figure 30. The failure was typical cup-and-cone type. The overall morphology of the cone surface was rough (Figure 30a). An array of fine microscopic cracks were surrounded by a population of dimples and voids of varying size, features reminiscent of both locally brittle and locally ductile failure mechanisms (Figure 30b). At regular intervals, deep-seated macroscopic cracks were evident on the fracture surface (Figure 30c). The macroscopic cracks result as a consequence of the growth and eventual coalescence of the fine microscopic voids. The overload region revealed a population of dimples of varying size and shape, and fine microscopic voids (Figure 30d). Overall, the presence of a sizeable population of fine microscopic voids and macroscopic voids coupled with isolated microscopic cracks contributes to lowering the actual strain-to-failure associated with ductile fracture. Isolated cracks in deep pockets, surrounded by a population of shallow dimples of varying size and fine microscopic voids was observed at higher magnification (Figure 30b). Observation of the transgranular surface region revealed the fine microscopic cracks to be surrounded by a population of microscopic voids and dimples reminiscent of locally brittle and ductile failure mechanisms. The dimples observed were of varying size but shallow in depth.
Figure 29: Scanning electron micrographs of the tensile fracture surface of Ti-6Al-4V specimen (orientation: Longitudinal) deformed in tension, showing:

(a) Overall morphology of tensile fracture surface
(b) High magnification of (a) showing macroscopic crack in the transgranular region
(c) A microscopic crack surrounded by population of shallow dimples
(d) The overload region showing coalescence of microscopic voids to form a microscopic crack and surrounded by shallow dimples
Figure 30: Scanning electron micrographs of the tensile fracture surface of Ti-6Al-4V specimen (orientation: Transverse) deformed in tension, showing:

(a) Overall cup and cone morphology of tensile fracture surface
(b) High magnification of (a) showing microscopic crack, population of voids and dimples in the transgranular region
(c) High magnification of (b) showing the morphology and distribution of microscopic cracks, voids and dimples on the transgranular fracture surface
(d) The overload region showing coalescence of microscopic voids to form a microscopic crack and surrounded by shallow dimples.
5.5. High cycle fatigue resistance

The cyclic degradation of the strength of a material through the accumulation of damage is known as fatigue. The fatigue life \( (N_f) \) of a structure or a component is defined as the sum of the number of cycles to crack initiation \( (N_i) \) plus the number of cycles of crack propagation \( (N_p) \), which eventually leads to failure by fracture. A detailed study of the cyclic fatigue response of the Ti-6Al-4V is important for its selection and subsequent use in a wide array of performance-critical applications. An understanding of failure and damage mechanisms governing failure is made difficult by the underlying intrinsic microstructural features. It has been established by previous research that the fatigue properties of the titanium alloys are governed by the cumulative influence of: (a) surface defects, (b) crystallographic texture, and (c) microstructural characteristics to include the size and morphology of the beta grains, colonies of alpha platelets [38]. Alloy microstructures that produce a low resistance to crack initiation, even at low values of applied stress, will usually yield the best endurance during high cycle fatigue testing and even tensile testing [21]. The cyclic fatigue test is the most widely used technique to establish the endurance limit of a chosen metal by essentially determining the variation of maximum stress \( (\sigma_{\text{maximum}}) \) or stress amplitude \( (\Delta \sigma/2) \) with fatigue life as quantified by the number of cycles to failure \( (N_f) \). Metals, alloys and other materials based on metal matrices, which are used at stress levels below the endurance limit may be cycled indefinitely.

The high-cycle fatigue test is the most widely used technique to establish the endurance limit of a metal by essentially determining the variation of maximum stress \( (\sigma_{\text{maximum}}) \), or stress amplitude \( (\Delta \sigma/2) \), with fatigue life \( (N_f) \). For small, highly stressed
components, fatigue life is often controlled by cyclic plasticity and relatively rapid growth of the fine microscopic cracks through the microstructure of the titanium alloy. For larger components that operate at low stress levels, the fatigue life is controlled by microscopic crack formation and their subsequent growth through the microstructure. The kinetics of crack nucleation in this Ti-6Al-4V alloy is governed by mutually interactive influences of the following: (i) the rate of cooling, (ii) volume fraction and size of the alpha phase, and (iii) local stress concentration effects or surface residual stress [9-12]

At the load ratio studied (R = 0.1), the variation of maximum stress ($\sigma_{\text{maximum}}$) with cyclic fatigue life ($N_f$) reveals increasing fatigue life with a decrease in stress amplitude. This trend is typical of most non-ferrous metals and their composite counterparts [2-3]. At the higher levels of equivalent maximum stress, the results indicate the fatigue life to be more or less identical in both the transverse and longitudinal orientations.
Figure 31: Variation of maximum stress ($\sigma_{\text{max}}$) with fatigue life ($N_f$) for Ti-6Al-4V alloy at a stress ratio of ($R=0.1$).

Figure 32: Variation of the ratio of maximum stress/yield stress ($\sigma_{\text{max}}/\sigma_y$) with fatigue life ($N_f$) for Ti-6Al-4V alloy at a stress ratio of ($R=0.1$).
Figure 33: Variation of maximum elastic strain ($\sigma_{\text{max}}/E$) with fatigue life ($N_f$) for Ti-6Al-4V alloy at a stress ratio of ($R = 0.1$).

Figure 34: Variation of the ratio of maximum stress/ultimate tensile stress ($\sigma_{\text{max}}/\sigma_{\text{UTS}}$) with fatigue life ($N_f$) for Ti-6Al-4V alloy at a stress ratio of ($R = 0.1$).
5.6 Cyclic Fatigue Fracture

The fracture surfaces of the high-cycle fatigue test specimens were exhaustively studied under a JEOL scanning electron microscope (SEM) at the following levels of magnification:

(a) In order to identify the site of microscopic crack initiation, the region of stable crack growth and to study the kinetics of early crack growth, a low magnification was used.

(b) Gradually increasing levels of magnification were used for studying the nature of damage initiation, crack propagation, cyclic fracture behavior and overload region morphology.

For the Ti-6Al-4V alloy, marginal difference was observed in the topographies at various levels of maximum stress and corresponding fatigue life for both the longitudinal and transverse specimens. On a microscopic scale, the nature and morphology of the fracture surfaces did not vary extensively with maximum stress and fatigue life. The representative fractographs of the fatigue fracture surfaces of Ti-6Al-4V at different values of maximum stress and corresponding fatigue life are shown in Figure 35 to Figure 43.

It is well known that high-cycle fatigue life, in general is dependent on the initiation and propagation of the fatigue cracks. Much of the fatigue life is spent in initiation of these fatigue cracks [10]. Microstructural features whether formed during initial heat treatments or during subsequent fabrication of the welded structural members, can change the high-cycle fatigue characteristics of the material. Different welding
methods can cause a significant variations in the microstructural features like the formation and presence of fine microscopic pores that act as crack initiation sites [10]. The microstructure of the as-provided Ti-6Al-4V alloy used in this study is characterized as a duplex microstructure consisting of the near equiaxed alpha (α) and transformed beta (β) phases. Previous researchers have established fatigue cracking in the Ti-6Al-4V alloy to initiate along grain boundaries. During continued cyclic deformation of the specimens, a gradual accumulation of fracture features across the transgranular surface leads to the formation of localized failure sites with resultant accumulation of strain at the microscopic levels. When the strain at these localized stress concentration exceeds a critical value, microscopic crack initiation occurs, which is fuelled by a continued application of stress leading to a progressive accumulation of micro-plastic deformations at the ‘local’ level [23]. Continuous cyclic deformation coupled with the gradual crack propagation resulting from coalescence of the fine microscopic voids leads to a macroscopic crack and a resultant drop in the load-carrying capability of the test specimen eventually culminating in failure.

5.6.1 Cyclic Fatigue Fracture: Longitudinal orientation

At a maximum stress of 829MPa, the longitudinal specimen had a fatigue life of 255,334 cycles. Low magnification images of scanning electron microscopy revealed the existence of a confined region of crack initiation and propagation. Early microscopic crack growth can be observed to be radially extending from a point of crack initiation (Figure 35b). On carefully studying these regions at high magnifications, we observe this region to be essentially featureless, containing very fine microscopic cracks, which exhibit a random distribution across this region (Figure 35c). For almost all the samples
at varying maximum stress levels, the overload region revealed the presence of shallow
dimples indicative of a locally ductile failure mechanism (Figure 35d and Figure 36d).

For a test specimen cyclically deformed at a maximum stress of 811MPa with a
resultant fatigue life of 674,128 cycles, distinctive crack initiation and overload regions
were observed. (Figure 36a). In the terminal stages of stable crack propagation,
formation of a transition zone between stable and unstable crack growth occurs (Figure
36b). High magnification observations of the stable and unstable crack growth regions
reveal a healthy population of macroscopic and fine microscopic voids (Figure 36c),
which, during a repetitive cyclic loading can lead to the phenomenon of void coalescence
eventually resulting in the formation of fine microscopic cracks and shallow dimples
(Figure 36d). The transgranular region (Figure 37b), when observed at high
magnifications showed a healthy distribution of pockets of fine shallow striations (Figure
37c) in the region of early crack initiation and propagation and well defined striations
(Figure 38) in the region of stable crack growth indicative of localized micro-plastic
deformation mechanism. Thus, for the longitudinal orientation, the microscopic fracture
characteristics observed across a range of maximum stress values are essentially similar.
Figure 35: Scanning electron micrographs of the fatigue fracture surface of the Ti-6Al-4V specimen (Orientation: Longitudinal) deformed in cyclic fatigue at a maximum stress of 870 MPa at a stress ratio of $R = 0.1$ and resultant fatigue life ($N_f$) of 35,357 cycles, showing:

(a). Overall morphology of failure.
(b). The region between stable and unstable crack growth showing a healthy population of fine microscopic voids and dimples.
(c). High magnification of (b) showing a population of macroscopic and fine microscopic voids.
(d). Void coalescence to form a fine microscopic crack and shallow dimples indicative of locally ductile failure.
Figure 36: Scanning electron micrographs of the fatigue fracture surface of the Ti-6Al-4V specimen (Orientation: Longitudinal) deformed in cyclic fatigue at a maximum stress of 829 MPa at a stress ratio of $R = 0.1$ and resultant fatigue life ($N_f$) of 255,334 cycles, showing:

(a). Overall morphology of failure.
(b). The region of crack initiation and radial outward progression of fatigue damage.
(c). The transgranular surface: flat and near featureless in the region of early crack growth.
(d). Shallow dimples covering the region of overload reminiscent of locally ductile failure.
Figure 37: Scanning electron micrographs of the fatigue fracture surface of the Ti-6Al-4V specimen (Orientation: Longitudinal) deformed in cyclic fatigue at a maximum stress of 811 MPa at a stress ratio of $R = 0.1$ and resultant fatigue life ($N_f$) of 674,128 cycles, showing:

(a). Overall morphology of failure.
(b). The transgranular surface in the region of early crack growth.
(c). High magnification of (b) showing pockets of fine shallow striations indicative of localized micro-plastic deformation.
(d). Shallow dimples in the region of overload.
Figure 38: Scanning electron micrographs of the fatigue fracture surface of the Ti-6Al-4V specimen (Orientation: Longitudinal) deformed in cyclic fatigue at a maximum stress of 811 MPa at a stress ratio of R = 0.1 and resultant fatigue life ($N_f$) of 674,128 cycles, showing well defined striations in the region of stable crack growth indicative of micro-plastic deformation.
5.6.2 Cyclic Fatigue Fracture: Transverse orientation

Following the same trend as the specimens oriented in longitudinal direction, the transverse oriented specimens of Ti-6Al-4V alloy exhibit almost identical fatigue lives at higher stress levels at a load ration of 0.1. At a maximum stress of 962 MPa and resultant fatigue life of 40,664 cycles, the intrinsic fracture surface characteristics are shown in Figure 39. For a specimen subjected a maximum stress of 900MPa and resultant fatigue life of 425,984 cycles, the overall fracture surface morphology revealed the presence of distinctive regions of crack initiation and radial propagation across the transgranular surface (Figure 40a). High magnification of the region of early crack growth (Figure 40b) showed a random distribution of fine microscopic cracks (Figure 40c).

The characteristic features on the overload surface like fine microscopic voids and shallow dimples can be observed on the fracture surface for specimens across a range of maximum stress values (Figure 40d and Figure 41b). Fine and shallow striations were found in random locations in the region of stable crack growth indicative of a localized ductile failure mechanism (Figure 41a). The test specimen cyclically deformed at a lower maximum stress of 823 MPa had a fatigue life of 581,951 cycles. The overall morphology of the fracture surface showing the radial propagation of fatigue damage is shown in Figure 42a. Scanning electron microscopy of the transgranular fracture surface region revealed a population of fine microscopic cracks evenly distributed across the region (Figure 42b).
High magnification observations of the fine microscopic cracks revealed the non linear nature of crack propagation as seen in Figure 42c. Irrespective of the maximum stress applied to cyclically deform the specimen, the region of unstable crack growth is characterized by the presence of a healthy population of shallow dimples (Figure 42d and Figure 43).
Figure 39: Scanning electron micrographs of the fatigue fracture surface of the Ti-6Al-4V specimen (Orientation: Transverse) deformed in cyclic fatigue at a maximum stress of 962 MPa at a stress ratio of $R = 0.1$ and resultant fatigue life ($N_f$) of 40,664 cycles, showing:

(a). Overall morphology of failure.

(b). The region of early crack growth: flat and near featureless.

(c). Pockets of shallow striations in the region of stable crack growth.

(d). A healthy population of fine microscopic voids in the region of overload.
Figure 40: Scanning electron micrographs of the fatigue fracture surface of the Ti-6Al-4V specimen (Orientation: Transverse) deformed in cyclic fatigue at a maximum stress of 900 MPa at a stress ratio of $R = 0.1$ and resultant fatigue life ($N_f$) of 425,984 cycles, showing:

(a). Overall morphology of failure showing region of initiation and radial progression of fatigue damage.

(b). High magnification of region of early crack growth.

(c). High magnification observation of the region of early crack growth showing a random distribution of fine microscopic cracks.

(d). A population of voids and dimples in the region of overload.
Figure 41: Scanning electron micrographs of the fatigue fracture surface of the Ti-6Al-4V specimen (Orientation: Transverse) deformed in cyclic fatigue at a maximum stress of 900 MPa at a stress ratio of $R = 0.1$ and resultant fatigue life ($N_f$) of 425,984 cycles, showing:

(a). Fine and shallow striations found in random pockets in the region of stable crack growth.

(b). Healthy population of fine microscopic voids of varying shape intermingled with dimples.
Figure 42: Scanning electron micrographs of the fatigue fracture surface of the Ti-6Al-4V specimen (Orientation: Transverse) deformed in cyclic fatigue at a maximum stress of 823 MPa at a stress ratio of $R = 0.1$ and resultant fatigue life ($N_f$) of 581,951 cycles, showing:

(a). Overall morphology of showing radial propagation of fatigue damage.
(b). A healthy population of fine microscopic cracks covering the transgranular fracture surface region.
(c). High magnification of (b) showing non-linear nature of fine microscopic cracks.
(d). A healthy population of shallow dimples in the region of unstable crack growth.
Figure 43: Scanning electron micrographs of the fatigue fracture surface of the Ti-6Al-4V specimen (Orientation: Transverse) deformed in cyclic fatigue at a maximum stress of 823 MPa at a stress ratio of $R = 0.1$ and resultant fatigue life ($N_f$) of 581,951 cycles, showing pockets of shallow striations reminiscent of local microplastic deformation.
PART-B
CHAPTER VI
CONCLUSIONS [Ti-6Al-4V]

A study of the high cycle fatigue and final fracture behavior of test specimens machined from a Ti-6Al-4V fully annealed plate stock for the two different orientations, i.e. longitudinal and transverse provides the following key observations:

1. The Ti-6Al-4V alloy exhibited a duplex microstructure consisting of the near equiaxed alpha (α) and the transformed beta (β) phases. The primary near equiaxed shaped alpha (α) grains was well distributed in a lamellar matrix with transformed beta grains.

2. The yield strength and ultimate tensile strength of Ti-6Al-4V is nearly equal both in the longitudinal orientation and in the transverse orientation indication a more or less isotropic nature of the material. In both orientations, the tensile strength is higher than the yield strength indicating the occurrence of noticeable work hardening beyond yield.

3. There is a marginal to no influence of microstructure on high-cycle fatigue life of both orientations of the alloy and conforms well with the observed results, within the limits of experimental scatter, to no difference in cyclic fatigue life at a given load ratio (R=0.1).
4. At the chosen load ratio of 0.1 the fatigue fracture surfaces revealed minimal to no difference in the nature, distribution and volume fraction of the intrinsic features on the fracture surface as a function of maximum stress and resultant fatigue life. The region of crack initiation and early microscopic crack growth and a short, yet distinct, region of stable crack growth were essentially flat and transgranular. A population of fine microscopic and macroscopic cracks was evident in the region of stable crack growth approaching overload. Also evident in this region were pockets of well defined striations indicative of the occurrence of micro-plastic deformation at the local level. The region of overload was covered with a population of shallow dimples, fine microscopic voids and isolated macroscopic void and cracks, features reminiscent of the locally governing failure mechanisms. During repeated cyclic loading, the fine microscopic cracks coalesce to form macroscopic crack.


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APPENDICES
APPENDIX A

TABLES SHOWING HIGH CYCLE FATIGUE TEST DATA, AND STRESS AND LOAD CALCULATIONS FOR Ti-CP (GRADE 2)

Table 7: A summary of the high-cycle fatigue tests conducted on Ti-CP (Grade 2) (Orientation: Longitudinal) specimens at a load ratio of R=0.1

<table>
<thead>
<tr>
<th>SPECIMEN ID</th>
<th>Nf</th>
<th>σYS %</th>
<th>σmax</th>
<th>σmax/σYS</th>
<th>σmax/E</th>
<th>σmax/σUTS</th>
</tr>
</thead>
<tbody>
<tr>
<td>P4-L7</td>
<td>45532</td>
<td>94</td>
<td>406.08</td>
<td>0.940</td>
<td>0.003531</td>
<td>0.724</td>
</tr>
<tr>
<td>P4-L12</td>
<td>49463</td>
<td>93</td>
<td>401.76</td>
<td>0.930</td>
<td>0.003494</td>
<td>0.716</td>
</tr>
<tr>
<td>P4-L10</td>
<td>72565</td>
<td>92</td>
<td>397.44</td>
<td>0.920</td>
<td>0.003456</td>
<td>0.708</td>
</tr>
<tr>
<td>P4-L11</td>
<td>125564</td>
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<td>367.2</td>
<td>0.850</td>
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<td>0.655</td>
</tr>
<tr>
<td>P4-L6</td>
<td>409467</td>
<td>80</td>
<td>345.6</td>
<td>0.800</td>
<td>0.003005</td>
<td>0.616</td>
</tr>
</tbody>
</table>

Table 8: A summary of the high-cycle fatigue tests conducted on Ti-CP (Grade 2) (Orientation: Transverse) specimens at a load ratio of R=0.1

<table>
<thead>
<tr>
<th>SPECIMEN ID</th>
<th>Nf</th>
<th>σYS %</th>
<th>σmax</th>
<th>σmax/σYS</th>
<th>σmax/E</th>
<th>σmax/σUTS</th>
</tr>
</thead>
<tbody>
<tr>
<td>P4-T14</td>
<td>47639</td>
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<td>1.060</td>
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<td>P4-T17</td>
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<td>1.010</td>
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<td>P4-T15</td>
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<td>P4-T18</td>
<td>957816</td>
<td>87.5</td>
<td>334.25</td>
<td>0.875</td>
<td>0.003095</td>
<td>0.719</td>
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</table>
Table 9: A summary of the input data and fatigue life for the high-cycle fatigue tests conducted on Ti-CP (Grade 2) (Orientation: Longitudinal) specimens at a load ratio of R=0.1.

<table>
<thead>
<tr>
<th>Specimen Id</th>
<th>Percent of yield Stress (%)</th>
<th>Maximum Stress (σ&lt;sub&gt;max&lt;/sub&gt;)</th>
<th>Maximum Load (P&lt;sub&gt;max&lt;/sub&gt;)</th>
<th>Minimum Stress (σ&lt;sub&gt;min&lt;/sub&gt;)</th>
<th>Minimum Load (P&lt;sub&gt;min&lt;/sub&gt;)</th>
<th>Mean value of Load</th>
<th>Amplitude of Load</th>
<th>HCF Test Completed?</th>
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<tr>
<td></td>
<td>Ksi</td>
<td>Mpa</td>
<td>Kips</td>
<td>k-N</td>
<td>Ksi</td>
<td>Mpa</td>
<td>Kips</td>
<td>k-N</td>
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<tr>
<td>L7, L5</td>
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<td>59.21</td>
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<td>0.722</td>
<td>3.22</td>
<td>0.92</td>
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<tr>
<td>L12</td>
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<td>58.27</td>
<td>401.76</td>
<td>0.715</td>
<td>0.711</td>
<td>3.18</td>
<td>0.93</td>
<td>40.18</td>
</tr>
<tr>
<td>L10</td>
<td>92.5</td>
<td>57.96</td>
<td>399.60</td>
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<td>3.16</td>
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<td>0.92</td>
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<tr>
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<td>Mpa</td>
<td>Kips</td>
<td>k-N</td>
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<td>Mpa</td>
<td>Kips</td>
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<td>0.661</td>
<td>2.94</td>
<td>0.86</td>
<td>37.15</td>
</tr>
</tbody>
</table>

Test 1 | Repeat | Test 1 | Repeat
L5 | failed | L11 | failed
L6 | error | L12 | failed
L7
L8 | failed | L9 | failed
L10 | DNF

Specimen σ<sub>0</sub> (Mpa) | Yield Stress | Initial dia. of Specimen (mm) | Area of Specimen (mm<sup>2</sup>) | Stress Ratio | Area of Specimen (in<sup>2</sup>) | Yield Load (k-N) | Test 1 | Repeat | Test 1 | Repeat
1.9 | 420.0 | 8.175 | 0.125 | 7.9178 | 0.01927 | 8.4928 | 0.928 |
1.9 | 420.0 | 8.175 | 0.125 | 7.9178 | 0.01927 | 8.4928 | 0.928 |
Average | 420.0 | 8.175 | 0.125 | 7.9178 | 0.01927 | 8.4928 | 0.928 |
Table 10: A summary of the input data and fatigue life for the high-cycle fatigue tests conducted on Ti-CP (Grade 2) (Orientation: Transverse) specimens at a load ratio of R=0.1.

<table>
<thead>
<tr>
<th>Specimen Id</th>
<th>Percent of yield Stress (%)</th>
<th>Maximum Stress ($\sigma_{max}$) (Ksi)</th>
<th>Maximum Load ($P_{max}$) (kN)</th>
<th>Minimum Stress ($\sigma_{min}$) (Ksi)</th>
<th>Minimum Load ($P_{min}$) (kN)</th>
<th>Mean value of Load</th>
<th>Amplitude of Load</th>
<th>HCF Test Completed</th>
</tr>
</thead>
<tbody>
<tr>
<td>T6</td>
<td>93.5</td>
<td>51.84</td>
<td>357.44</td>
<td>0.636</td>
<td>0.280</td>
<td>0.352</td>
<td>0.500</td>
<td>Yes</td>
</tr>
<tr>
<td>T5</td>
<td>90.0</td>
<td>49.90</td>
<td>344.06</td>
<td>0.612</td>
<td>0.272</td>
<td>0.337</td>
<td>0.498</td>
<td>Yes</td>
</tr>
<tr>
<td>T4</td>
<td>87.5</td>
<td>48.82</td>
<td>334.50</td>
<td>0.595</td>
<td>0.261</td>
<td>0.327</td>
<td>0.457</td>
<td>Yes</td>
</tr>
<tr>
<td>T7</td>
<td>80.0</td>
<td>46.58</td>
<td>321.52</td>
<td>0.572</td>
<td>0.242</td>
<td>0.318</td>
<td>0.415</td>
<td>Yes</td>
</tr>
<tr>
<td>T8</td>
<td>75.0</td>
<td>41.58</td>
<td>286.72</td>
<td>0.510</td>
<td>0.220</td>
<td>0.281</td>
<td>0.329</td>
<td>Yes</td>
</tr>
<tr>
<td>T1</td>
<td>70.0</td>
<td>38.81</td>
<td>267.60</td>
<td>0.476</td>
<td>0.212</td>
<td>0.262</td>
<td>0.315</td>
<td>Yes</td>
</tr>
<tr>
<td>T2</td>
<td>65.0</td>
<td>36.94</td>
<td>248.49</td>
<td>0.442</td>
<td>0.197</td>
<td>0.243</td>
<td>0.307</td>
<td>Yes</td>
</tr>
<tr>
<td>T3</td>
<td>60.0</td>
<td>32.27</td>
<td>229.37</td>
<td>0.408</td>
<td>0.182</td>
<td>0.225</td>
<td>0.299</td>
<td>Yes</td>
</tr>
</tbody>
</table>

Yield Stress: 882.8 Mpa (55,400 ksi)  
Initial dia. of Specimen: 8.175 mm (0.325 in)  
Area of Specimen: 7.9178 mm² (0.01097 in²)  
Test 1 Repeat  
T1 failed  
T7 failed  
T4 Stopped  
T5 failed  
T9 Buckled  
T6 failed  
T10  
T11  
T19
### APPENDIX B

**TABLES SHOWING HIGH CYCLE FATIGUE TEST DATA.**

**STRESS AND LOAD CALCULATIONS FOR Ti-6Al-4V**

**Table 11:** A summary of the high-cycle fatigue tests conducted on Ti-6Al-4V (Orientation: Longitudinal) specimens at a load ratio of R=0.1.

<table>
<thead>
<tr>
<th>SPECIMEN ID</th>
<th>( N_f )</th>
<th>( \sigma_{YS} ) %</th>
<th>( \sigma_{max} )</th>
<th>( \sigma_{max}/\sigma_{YS} )</th>
<th>( \sigma_{max}/E )</th>
<th>( \sigma_{max}/\sigma_{UTS} )</th>
</tr>
</thead>
<tbody>
<tr>
<td>L7</td>
<td>35737</td>
<td>94.5</td>
<td>870.35</td>
<td>0.95</td>
<td>0.006908</td>
<td>0.85</td>
</tr>
<tr>
<td>L5</td>
<td>38740</td>
<td>92</td>
<td>847.32</td>
<td>0.92</td>
<td>0.006725</td>
<td>0.83</td>
</tr>
<tr>
<td>L12</td>
<td>255334</td>
<td>90</td>
<td>828.9</td>
<td>0.90</td>
<td>0.006579</td>
<td>0.81</td>
</tr>
<tr>
<td>*L8</td>
<td>34724</td>
<td>89.5</td>
<td>824.3</td>
<td>0.90</td>
<td>0.006542</td>
<td>0.81</td>
</tr>
<tr>
<td>*L11</td>
<td>1e6+57365</td>
<td>89</td>
<td>819.69</td>
<td>0.89</td>
<td>0.006505</td>
<td>0.80</td>
</tr>
<tr>
<td>L3</td>
<td>674128</td>
<td>88</td>
<td>810.48</td>
<td>0.88</td>
<td>0.006432</td>
<td>0.79</td>
</tr>
<tr>
<td>*L10</td>
<td>1.00E+06</td>
<td>87.5</td>
<td>805.88</td>
<td>0.88</td>
<td>0.006396</td>
<td>0.79</td>
</tr>
<tr>
<td>L9</td>
<td>(*7.3)x 10^5</td>
<td>85/85.5</td>
<td>787.46</td>
<td>0.86</td>
<td>0.006250</td>
<td>0.77</td>
</tr>
<tr>
<td>L6</td>
<td>1.00E+06</td>
<td>82.5</td>
<td>759.83</td>
<td>0.83</td>
<td>0.006030</td>
<td>0.74</td>
</tr>
</tbody>
</table>

**Table 8:** A summary of the high-cycle fatigue tests conducted on Ti-6Al-4V (Orientation: Transverse) specimens at a load ratio of R=0.1.

<table>
<thead>
<tr>
<th>SPECIMEN ID</th>
<th>( N_f )</th>
<th>( \sigma_{YS} ) %</th>
<th>( \sigma_{max} )</th>
<th>( \sigma_{max}/\sigma_{YS} )</th>
<th>( \sigma_{max}/E )</th>
<th>( \sigma_{max}/\sigma_{UTS} )</th>
</tr>
</thead>
<tbody>
<tr>
<td>T6</td>
<td>40664</td>
<td>93.5</td>
<td>962.12</td>
<td>0.94</td>
<td>0.007023</td>
<td>0.83</td>
</tr>
<tr>
<td>T5</td>
<td>202376</td>
<td>90</td>
<td>926.1</td>
<td>0.90</td>
<td>0.006760</td>
<td>0.80</td>
</tr>
<tr>
<td>T4</td>
<td>425984</td>
<td>87/87.5</td>
<td>900.38</td>
<td>0.88</td>
<td>0.006572</td>
<td>0.78</td>
</tr>
<tr>
<td>T7</td>
<td>445229</td>
<td>84</td>
<td>864.36</td>
<td>0.84</td>
<td>0.006309</td>
<td>0.75</td>
</tr>
<tr>
<td>*T10</td>
<td>3824</td>
<td>80</td>
<td>823.2</td>
<td>0.80</td>
<td>0.006309</td>
<td>0.75</td>
</tr>
<tr>
<td>T11</td>
<td>581951</td>
<td>80</td>
<td>823.2</td>
<td>0.80</td>
<td>0.006090</td>
<td>0.71</td>
</tr>
<tr>
<td>T8</td>
<td>34427</td>
<td>79.5</td>
<td>818.06</td>
<td>0.80</td>
<td>0.005971</td>
<td>0.71</td>
</tr>
<tr>
<td>T12</td>
<td>1000000</td>
<td>75</td>
<td>771.75</td>
<td>0.75</td>
<td>0.005633</td>
<td>0.67</td>
</tr>
<tr>
<td>T9</td>
<td>Buckled</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
</tbody>
</table>
Table 5: A summary of the input data and fatigue life for the high-cycle fatigue tests conducted on Ti-6Al-4V (Orientation: Longitudinal) specimens at a load ratio of $R=0.1$

<table>
<thead>
<tr>
<th>Specimen Id</th>
<th>Percent of yield Stress (%)</th>
<th>Maximum Stress ($\sigma_{\text{max}}$) (Mpa)</th>
<th>Minimum Stress ($\sigma_{\text{min}}$) (Mpa)</th>
<th>Maximum Load ($P_{\text{max}}$) (kips)</th>
<th>Minimum Load ($P_{\text{min}}$) (kips)</th>
<th>Mean value of Load (kips)</th>
<th>Amplitude of Load (kips)</th>
<th>HCF Test Completed ?</th>
<th>Cycles (N)&lt;sub&gt;f&lt;/sub&gt;</th>
</tr>
</thead>
<tbody>
<tr>
<td>L7</td>
<td>94.5</td>
<td>126.23</td>
<td>12.42</td>
<td>85.65</td>
<td>0.152</td>
<td>0.838</td>
<td>3.70</td>
<td>Yes</td>
<td>85787</td>
</tr>
<tr>
<td>L5</td>
<td>122.89</td>
<td>847.31</td>
<td>12.29</td>
<td>84.73</td>
<td>0.151</td>
<td>0.829</td>
<td>3.690</td>
<td>Yes</td>
<td>88740</td>
</tr>
<tr>
<td>L6</td>
<td>121.56</td>
<td>838.11</td>
<td>12.16</td>
<td>83.81</td>
<td>0.149</td>
<td>0.820</td>
<td>3.650</td>
<td>Yes</td>
<td>8910</td>
</tr>
<tr>
<td>L12</td>
<td>120.22</td>
<td>825.90</td>
<td>12.02</td>
<td>82.89</td>
<td>0.148</td>
<td>0.811</td>
<td>3.609</td>
<td>Yes</td>
<td>255884</td>
</tr>
<tr>
<td>L8</td>
<td>119.55</td>
<td>824.30</td>
<td>11.96</td>
<td>82.43</td>
<td>0.147</td>
<td>0.807</td>
<td>3.589</td>
<td>Yes</td>
<td>84724</td>
</tr>
<tr>
<td>L11</td>
<td>118.89</td>
<td>819.69</td>
<td>11.89</td>
<td>81.97</td>
<td>0.146</td>
<td>0.802</td>
<td>3.569</td>
<td>Yes</td>
<td>10&lt;sup&gt;6&lt;/sup&gt; - 87965</td>
</tr>
<tr>
<td>L9</td>
<td>117.55</td>
<td>810.48</td>
<td>11.76</td>
<td>81.05</td>
<td>0.144</td>
<td>0.793</td>
<td>3.529</td>
<td>Yes</td>
<td>674128</td>
</tr>
<tr>
<td>L10</td>
<td>116.88</td>
<td>805.88</td>
<td>11.69</td>
<td>80.59</td>
<td>0.143</td>
<td>0.789</td>
<td>3.509</td>
<td>Yes</td>
<td>1000000</td>
</tr>
<tr>
<td>L9 (Rep)</td>
<td>114.88</td>
<td>792.06</td>
<td>11.49</td>
<td>79.21</td>
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<td>0.775</td>
<td>3.449</td>
<td>Yes9</td>
<td>83000</td>
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<tr>
<td>L9</td>
<td>113.54</td>
<td>782.85</td>
<td>11.35</td>
<td>78.29</td>
<td>0.139</td>
<td>0.766</td>
<td>3.409</td>
<td>Yes</td>
<td>780000</td>
</tr>
<tr>
<td>L6</td>
<td>110.87</td>
<td>764.43</td>
<td>11.09</td>
<td>76.44</td>
<td>0.136</td>
<td>0.748</td>
<td>3.329</td>
<td>Yes</td>
<td>6120</td>
</tr>
<tr>
<td>L6</td>
<td>110.20</td>
<td>759.83</td>
<td>11.02</td>
<td>75.98</td>
<td>0.135</td>
<td>0.724</td>
<td>3.309</td>
<td>Yes</td>
<td>6090</td>
</tr>
<tr>
<td>L6</td>
<td>109.54</td>
<td>755.22</td>
<td>10.95</td>
<td>75.52</td>
<td>0.134</td>
<td>0.708</td>
<td>3.289</td>
<td>Yes9</td>
<td>6050</td>
</tr>
<tr>
<td>L6</td>
<td>108.20</td>
<td>746.01</td>
<td>10.82</td>
<td>74.60</td>
<td>0.131</td>
<td>0.700</td>
<td>3.269</td>
<td>Yes9</td>
<td>6080</td>
</tr>
<tr>
<td>L6</td>
<td>106.86</td>
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<td>10.69</td>
<td>73.68</td>
<td>0.131</td>
<td>0.692</td>
<td>3.228</td>
<td>Yes9</td>
<td>6080</td>
</tr>
<tr>
<td>L6</td>
<td>100.18</td>
<td>690.75</td>
<td>10.02</td>
<td>69.08</td>
<td>0.123</td>
<td>0.676</td>
<td>3.008</td>
<td>Yes9</td>
<td>5550</td>
</tr>
<tr>
<td>L6</td>
<td>93.51</td>
<td>644.70</td>
<td>9.35</td>
<td>64.47</td>
<td>0.115</td>
<td>0.631</td>
<td>2.807</td>
<td>Yes9</td>
<td>5162</td>
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<tr>
<td>L6</td>
<td>86.83</td>
<td>598.65</td>
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<td>59.87</td>
<td>0.107</td>
<td>0.586</td>
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<tr>
<td>L6</td>
<td>80.15</td>
<td>552.60</td>
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<td>55.26</td>
<td>0.098</td>
<td>0.541</td>
<td>2.406</td>
<td>Yes9</td>
<td>4432</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Specimen</th>
<th>$\sigma_{\text{min}}$ (Mpa)</th>
<th>Yield Stress</th>
<th>Initial dia. of Specimen</th>
<th>Area of Specimen (mm$^2$)</th>
<th>Stress Ratio</th>
<th>Average Yield Load (kips)</th>
</tr>
</thead>
<tbody>
<tr>
<td>L1</td>
<td>948.0</td>
<td>291.0</td>
<td>8.175</td>
<td>7.9178</td>
<td>0.1</td>
<td>7.2918</td>
</tr>
<tr>
<td>L4</td>
<td>894.0</td>
<td>291.0</td>
<td>0.155</td>
<td>0.10292</td>
<td></td>
<td>1.6893</td>
</tr>
<tr>
<td>Average</td>
<td>-</td>
<td>291.0</td>
<td>0.155</td>
<td>0.10292</td>
<td></td>
<td>1.6893</td>
</tr>
</tbody>
</table>

Orientation: Longitudinal
<table>
<thead>
<tr>
<th>Specimen</th>
<th>Yield Stress (MPa)</th>
<th>Initial dia. of Specimen</th>
<th>Area of Specimen (mm²)</th>
<th>Stress Ratio</th>
<th>T1</th>
<th>T2</th>
<th>T3</th>
<th>T4</th>
<th>T5</th>
<th>Test 1</th>
<th>Test 1</th>
<th>Test 1</th>
<th>Test 1</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>710</td>
<td>7.9178</td>
<td>0.0127</td>
<td>-</td>
<td>1</td>
<td>1</td>
<td>1</td>
<td>1</td>
<td>1</td>
<td>Failed</td>
<td>Failed</td>
<td>Failed</td>
<td></td>
</tr>
<tr>
<td>1</td>
<td>1928</td>
<td>7.9178</td>
<td>0.0127</td>
<td>-</td>
<td>1</td>
<td>1</td>
<td>1</td>
<td>1</td>
<td>1</td>
<td>Failed</td>
<td>Failed</td>
<td>Failed</td>
<td></td>
</tr>
<tr>
<td>2</td>
<td>1650</td>
<td>7.9178</td>
<td>0.0127</td>
<td>-</td>
<td>1</td>
<td>1</td>
<td>1</td>
<td>1</td>
<td>1</td>
<td>Failed</td>
<td>Failed</td>
<td>Failed</td>
<td></td>
</tr>
<tr>
<td>3</td>
<td>304</td>
<td>7.9178</td>
<td>0.0127</td>
<td>-</td>
<td>1</td>
<td>1</td>
<td>1</td>
<td>1</td>
<td>1</td>
<td>Failed</td>
<td>Failed</td>
<td>Failed</td>
<td></td>
</tr>
</tbody>
</table>

A summary of the input data and fatigue life for the high-cycle fatigue tests conducted on Ti-6Al-4V (Orientation: Transverse) specimens at a load ratio of R=0.1.
APPENDIX C

PROCEDURE FOR PERFORMING THE TENSION TEST ON THE INSTRON-8500 SERVO HYDRAULIC TESTING MACHINE

Step 1: Switch on the POWER.

1. Switch on the power switch on the power surge bar.

2. Turn on the water flow for the hydraulic plant.

3. Switch on the computer and the CPU.

4. Switch <ON> the actuator.

5. Press <LOW> button for 5 seconds.

6. Then press <HIGH>

7. Check for the status in the message console on the monitor to make sure that it does not show any errors or warnings.

Step 2: Calibrate the load.

1. Right click on the load module on the console, and click on “Calibrate”.

2. Click “Next” four times until the screen shows “Start” and click “Start” for the calibration to complete.
3. The calibrate signal in the load area will blink continuously and stop, upon completion of CALIBRATION.

4. Uncheck the ticked boxes for overwriting the calibration settings and click “Finish”

**Step 3: Specimen Loading.**

1. Press <Load Protect> <ON>

2. Fix the specimen into the grips, this can be done by fixing the lower end of the specimen first and then moving the actuator <UP> so that the specimen is in place and the then the upper end of the specimen can be fixed firmly into the grips.

**Step 4: Recalibrate the load with specimen attached.**

1. This is done because, while fixing the specimen some load was exerted on the actuators. To negate that load-effect, we need to re-calibrate the load.

2. Turn load protector <OFF>

3. Right click on the load module on the console, and click on “Calibrate”.

4. Click “Next” four times until the screen shows “Start” and click “Start” for the calibration to complete.

5. The calibrate signal in the load area will blink continuously and stop, upon completion of CALIBRATION.
6. Uncheck the ticked boxes for overwriting the calibration settings and click “Finish”.

**Step 5: Calibration of the strain gage.**

1. Attach the strain gage plug to the strain socket.

2. Right click on the load module on the console, and click on “Calibrate”.

3. Click “Next” four times until the screen shows “Start” and click “Start” for the calibration to complete.

4. The calibrate signal in the load area will blink continuously and stop, upon completion of CALIBRATION.

5. The calibration of the strain gage is done WITHOUT fixing the strain gage to the specimen.

**Step 6: Fixing the strain gage to the specimen.**

1. Once the specimen is firmly in the grips and the strain gage has been calibrated, the strain gage must be fixed onto the specimen at the center of the gage length of the specimen (the region where failure is predicted).

2. Ensure that the PIN remains in the strain gage until both the arms of the strain gage are fixed onto the specimen with the aid of rubber bands.

3. Right Click on the load module to turn the <LOAD PROTECT> <ON>
4. Once the arms of the strain gage are fixed onto the specimen with the help of rubber bands, remove the pin.

5. It is advisable that the pin be removed gently, so as not to disturb the strain calibration.

6. The calibration can be done again to ensure the reading of strain on the console is close to zero.

**Step 7: Choosing the program to run the tension test.**

1. Choose Bluehill 2.0 on the desktop.

2. Create a Test method: choose a tension test method present in the list as per the sample and its geometry.

3. File → Open → Test methods: (UDAY/Ti-6Al-4V Tension test).

4. Click OK.

5. Check for the parameters of the tensile test method: Check all the specimen dimensions, test parameters and data storage parameters.

6. Set Strain control rate = 0.6%/min (This will depend on the material being tested)

7. Once the settings are set up, return to the HOME SCREEN.

8. Select the TEST icon.

9. Enter a file name for the test. (Example: Ti-6Al-4V (L-3))

10. Click OK and click next until the “Start Test” button is enabled.
11. Click on “Start Test” button. The test will begin and the graph on the computer screen will show the variation of stress with strain (if chosen).

12. The test will stop when the sample fails.

13. Remove one of the two rubber bands; insert the PIN back into the strain gage and remove the other rubber band, and place the strain gage safely in the box.

14. Right Click on the load module to turn the <LOAD PROTECT> <ON>

15. Remove the specimen from the grips.

16. The computer is programmed to give the print out of the data and the graph for reference.

17. Click on “Next” until “Finish Test” button is enabled. Click “Finish Test”.

18. Ensure that you see a green progress bar indicating that the raw data collected during the test is saved to the specified location.

19. Press < low> on the actuator controller for 5 seconds and then switch <OFF>.

20. Turn off the water supply and switch <OFF> the computer CPU and the monitor.
APPENDIX D

PROCEDURE FOR PERFORMING A HIGH CYCLE FATIGUE TEST ON THE INSTRON-8500 SERVO HYDRAULIC TESTING MACHINE

Step 1: Switch on the POWER.

1. Switch on the power switch on the power surge bar.

2. Turn on the water flow for the hydraulic plant.

3. Switch on the computer and the CPU.

4. Switch <ON> the actuator.

5. Press <LOW> button for 5 seconds.

6. Then press <HIGH>

7. Check for the status in the message console on the monitor to make sure that it does not show any errors or warnings.

Step 2: Calibrate the load.

1. Right click on the load module on the console, and click on “Calibrate”.

2. Click “Next” four times until the screen shows “Start” and click “Start” for the calibration to complete.
3. The calibrate signal in the load area will blink continuously and stop, upon completion of CALIBRATION.

4. Uncheck the ticked boxes for overwriting the calibration settings and click “Finish”.

**Step 3: Specimen Loading.**

1. Press <Load Protect> <ON>

2. Fix the specimen into the grips, this can be done by fixing the lower end of the specimen first and then moving the actuator <UP> so that the specimen is in place and the then the upper end of the specimen can be fixed firmly into the grips.

**Step 4: Recalibrate load with specimen attached.**

1. This is done because, while fixing the specimen some load was exerted on the actuators. To negate that load-effect, we need to re-calibrate the load.

2. Turn load protector <OFF>

3. Right click on the load module on the console, and click on “Calibrate”.

4. Click “Next” four times until the screen shows “Start” and click “Start” for the calibration to complete.

5. The calibrate signal in the load area will blink continuously and stop, upon completion of CALIBRATION.
6. Uncheck the ticked boxes for overwriting the calibration settings and click “Finish”.

**Step 6: Choosing the program for the FATIGUE test.**

1. Choose the SAX program from the desktop.

2. Create a Test method: choose a fatigue test method present in the computer as per the sample and its geometry using File→ Open→ Test methods: (UDAY/Ti-6Al-4V Fatigue test).k OK.

3. Click on the waveform generator, ensure that all the test requirements reflect on the screen, example: the waveform, this can be sinusoidal, triangular etc., start ramp: 0.1 second, the frequency: 5Hz

4. Check for the parameters of the fatigue test method: Enter the values for the amplitude and mean load values based on the calculations.

5. The end of the test was set at 1,000,000 cycles.

6. Save the test at a desired location with a proper name.

7. Ensure that the cycle counter is reset to ZERO before the test begins.

8. Go to Test → Start Test. Click OK to start the test.
Step 7: Removal of specimen after failure.

1. Once the test sample has failed, lower the actuator to ensure sufficient space for the removal of the failed specimen.

2. Ensure that the raw data is saved at the specified location.

3. Press <low> on the actuator controller for 5 seconds and then switch <OFF>.

4. Turn off the water supply and switch <OFF> the computer CPU and the monitor.
APPENDIX E
PROCEDURE FOR PERFORMING LOOP SHAPING FOR A GIVEN STRESS RATIO
AND MATERIAL ON THE INSTRON-8500 SERVO HYDRAULIC TESTING
MACHINE

Step 1: Switch on the POWER.

1. Switch on the power switch on the power surge bar.

2. Turn on the water flow for the hydraulic plant.

3. Switch on the computer and the CPU.

4. Switch <ON> the actuator.

5. Press <LOW> button for 5 seconds.

6. Then press <HIGH>

7. Check for the status in the message console on the monitor to make sure that it does not show any errors or warnings.

Step 2: Calibrate the load.

1. Right click on the load module on the console, and click on “Calibrate”.

2. Click “Next” four times until the screen shows “Start” and click “Start” for the calibration to complete.
3. The calibrate signal in the load area will blink continuously and stop, upon completion of CALIBRATION.

4. Uncheck the ticked boxes for overwriting the calibration settings and click “Finish”.

**Step 3: Specimen Loading.**

1. Press <Load Protect> <ON>

2. Fix the specimen into the grips, this can be done by fixing the lower end of the specimen first and then moving the actuator <UP> so that the specimen is in place and the then the upper end of the specimen can be fixed firmly into the grips.

**Step 4: Recalibrate load with specimen attached.**

1. This is done because, while fixing the specimen some load was exerted on the actuators. To negate that load-effect, we need to re-calibrate the load.

2. Turn load protector <OFF>

3. Right click on the load module on the console, and click on “Calibrate”.

4. Click “Next” four times until the screen shows “Start” and click “Start” for the calibration to complete.

5. The calibrate signal in the load area will blink continuously and stop, upon completion of CALIBRATION.
6. Uncheck the ticked boxes for overwriting the calibration settings and click “Finish”.

**Step 5: Loop shaping with a test specimen.**

1. Right click on the load module on the console, and click on “General” Option.

2. Click on the “PID” tab and note down the values for the fields proportional, integral, derivative and Lag.

3. Click on the “Properties” tab and modify the value of the field, “Tuning disturbance amplitude” to a small value say 0.5 kN.

4. Click on “Auto-tune now” and wait for the process to complete.

5. Now, Click on the “PID” tab and note down the values for the fields’ proportional, integral, derivative and Lag. Repeat steps 3 and 4 with gradually incrementing the Tuning disturbance amplitude value until the values of the above field remains more or less constant.

**Step 7: Removal of specimen after Loop tuning.**

1. Once the loop tuning is complete, enable specimen protect and carefully remove the specimen.

2. Press < low> on the actuator controller for 5 seconds and then switch <OFF>.

3. Turn off the water supply and switch <OFF> the computer CPU and the monitor.