Effects of Annealing/Solution Treating and Aging on Longitudinal and Transverse Tensile Properties of Ti-6Al-4V Forgings

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Approval Page

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Abstract

Forty-eight subsize round-bar tensile samples (ASTM E8) were machined from 3-in-thick forgings of Ti-6Al-4V. Half of these samples were cut so the axis aligned with the longitudinal (LG) direction of the forging, the other half were cut to align with the transverse (LT) direction. Four samples from each direction were subjected to one of six heat treatments so that a total of eight samples received each treatment. Annealed samples were held at 1350°F for two hours, then air cooled. Solution treated and aged (STA) samples were solution treated at one of five temperatures between 1785°F and 1435°F for one hour and then water quenched after a 25 second quench delay in air. Following quenching, all STA samples were aged for 6 hours at 990°F, then air cooled. Prior to tensile testing, the samples were sent back to Weber Metals for chemical milling to remove the brittle alpha-case surface layer formed during heat treatment. A polymer coating was applied to the samples' threaded ends so only the gage length would be milled, and any oxide present was sanded off to avoid preferential etching. Tensile tests showed anisotropy primarily in Tensile strength with the LT direction testing 3-7 ksi higher than the LG direction. A general linear model found a significant difference of Tensile strengths between directions in all heat treatments except for STA 1, which had the highest solution treating temperature. A difference of about 7 ksi was also found between STA 5 samples in Yield strength. No anisotropy was observed in Elongation. It was concluded that higher solution-treating temperatures decreased anisotropy by transforming the textured alpha into beta, effectively erasing any alignment. It was also concluded that the LT direction is stronger than the LG due to non-preferred slip systems being the primary mechanism of deformation.

Keywords: Materials Engineering, Titanium, Ti-6Al-4V, Ti-6-4, Forging, Anisotropy, Tensile testing, Metallography, Alpha-case, Tensile strength, Yield strength, Elongation, Heat treatment, Annealing, Solution-treating

Introduction

Weber Metals (Paramount, CA) is a company that specializes in providing aluminum and titanium forgings to military and commercial aerospace customers. These forgings include landing gear, valve bodies, aircraft wheels, and helicopter rotor hubs. Weber recently expanded their titanium program, and the company is researching titanium, its processing and properties to better understand this alloy system and ultimately provide better products. To that end, this project undertook a study regarding the effects of solution-treating/aging and annealing heat treatments on the longitudinal and transverse properties of Ti-6Al-4V forgings.

Titanium

Titanium is the fourth most abundant structural metal on earth with a concentration of 0.6% in the earth's crust, but it does not naturally occur in metallic form. Instead, titanium is combined as an oxide in the mineral forms rutile, ilmenite, titanomagnetite, anatase, and brookite. Rutile was discovered by Englishman W. Gregor and German M.H. Klaproth in 1790, but commercial pursuit of the extraction and refinement of titanium metal did not begin until the mid-twentieth century.[1]

Extraction of titanium metal from rutile ore begins with the titanium dioxide (TiO_2) in the mineral being converted to titanium tetrachloride ($TiCl_4$). $TiCl_4$ is then reduced to titanium metal and a salt through either the Hunter or Kroll process. In the Hunter process, $TiCl_4$ is reacted with sodium in two steps to produce Ti and NaCl. The more commonly used Kroll process involves a one-step reaction with magnesium to produce Ti and $MgCl_2$. Metal obtained from these two processes is in the form of loosely consolidated sponge which must be further processed into an ingot form to be of any use. The sponge is crushed, washed, compacted, and melted into an ingot through Vacuum arc remelting (VAR). VAR uses the compacted sponge as a consumable anode that is melted by an electric arc into a water-cooled copper crucible under a vacuum. Ingots produced by this method can reach up to 36 inches in diameter.[2] A diagram of this process is shown in Figure 1.

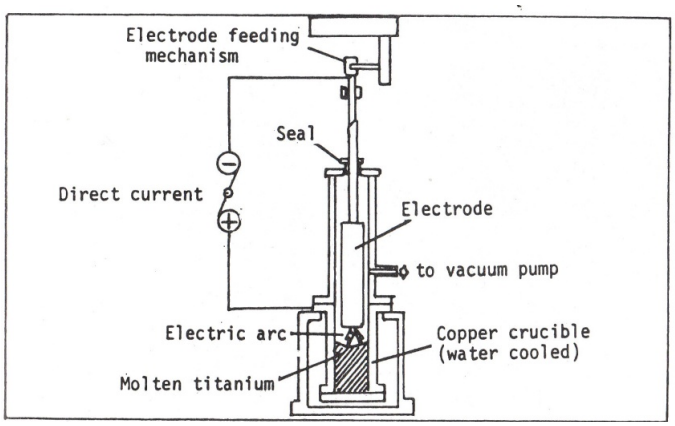

Figure 1: Vacuum arc melting setup for producing titanium ingots.[2]

Titanium is favored in many aerospace and biomedical applications for the following characteristics: High strength-to-weight ratio, good high-temperature performance, good corrosion resistance, and high ductility. The ductility and strength of pure titanium come from its room temperature α phase, which is hexagonal close-packed (HCP) and exists up to 883°C. This phase contains multiple slip systems and twinning planes that contribute to titanium's plastic deformation mechanisms (Figure 2).

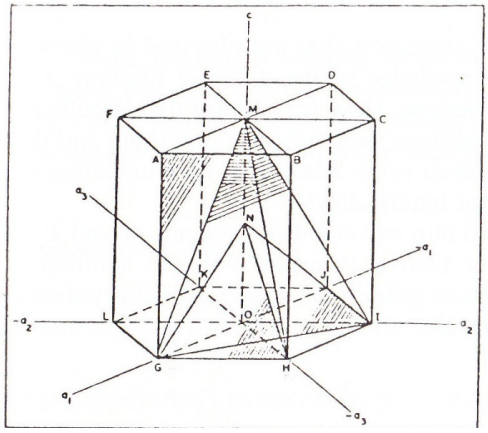

Figure 2: Slip systems and twinning planes in α-titanium. Slip occurs primarily on planes ABHG (prism plane), GHM, and basal planes. The rest of the planes are primarily twinning.[2]

Titanium is allotropic; it exists in different phases in different temperature ranges. In addition to the low-temperature α phase, titanium has a β phase that exists above the 883°C transus. This phase is softer than α and is body-centered cubic (BCC).

To optimize properties for different applications, many different alloying elements are added to titanium. Some of these alloying elements allow the alpha phase to exist at

temperatures higher than the normal transus (alpha stabilizers). Other alloying elements allow the beta phase to exist at temperatures lower than the normal transus (beta stabilizers). Lists of alpha and beta stabilizing elements are shown in Table I.

Table I: Alpha and Beta Stabilizing Elements for Titanium[2].

Alpha Stabilizers	Beta Stabilizers
Aluminum	Vanadium
Gallium	Molybdenum
Germanium	Tantalum
Oxygen	Niobium
Nitrogen	Silicon
Carbon	Copper
	Chromium
	Manganese
	Iron
	Nickel
	Cobalt

The alloys produced by additions of these elements are classified according to how stable the alpha and/or beta phase becomes. Alloys that consist of all alpha to high temperatures are called 'α' or 'near-α' alloys. Likewise, alloys retaining a significant amount of beta in room temperature microstructures are called 'β' alloys. Intuitively, titanium systems containing both stable alpha and beta at ambient temperatures are called 'α-β' alloys, the most prominent of which in the aerospace industry is Ti-6% Al-4% V (Ti-6-4).[2]

Ti-6Al-4V

Ti-6Al-4V, the most commonly used titanium alloy, contains 6% aluminum and 4% vanadium by weight. The percentage of aluminum is enough to raise the beta transus to 1823°F (995°C), giving the alloy good stability and strength up to around 752°F (400°C).[1] This strength can be improved through the precipitation of beta phase during aging, a process made easier by the vanadium's stabilization of beta. The effect of both temperature and solution treating and aging on tensile properties is shown in
Table II.

Table II: Room and Elevated Temperature Tensile Properties of Ti-6-4[2]

	Room Temperature			Elevated Temperature			
Condition	**Yield Strength (ksi)**	**Tensile Strength (ksi)**	**Elong. %**	**Test Temp. (°F)**	**Yield Strength (ksi)**	**Tensile Strength (ksi)**	**Elong. %**
Annealed	134	144	14	600	95	105	14
				800	83	97	18
				1000	62	77	35
Solution Treated + Aged	160	170	10	600	102	125	10
				800	90	116	12
				1000	70	95	22

The room temperature microstructures of Ti-6-4 consist mainly of some form of alpha surrounded by beta, but the arrangements of these phases are affected by heat treatment. When solutionized above the beta transus and furnace cooled, the resulting microstructure contains plate-like grains of alpha separated by intergranular beta. If instead of furnace cooling, the sample is air cooled, a microstructure of acicular alpha (also called 'transformed beta') within prior beta grain boundaries will be observed. Examples of these microstructures are shown in Figure 3.

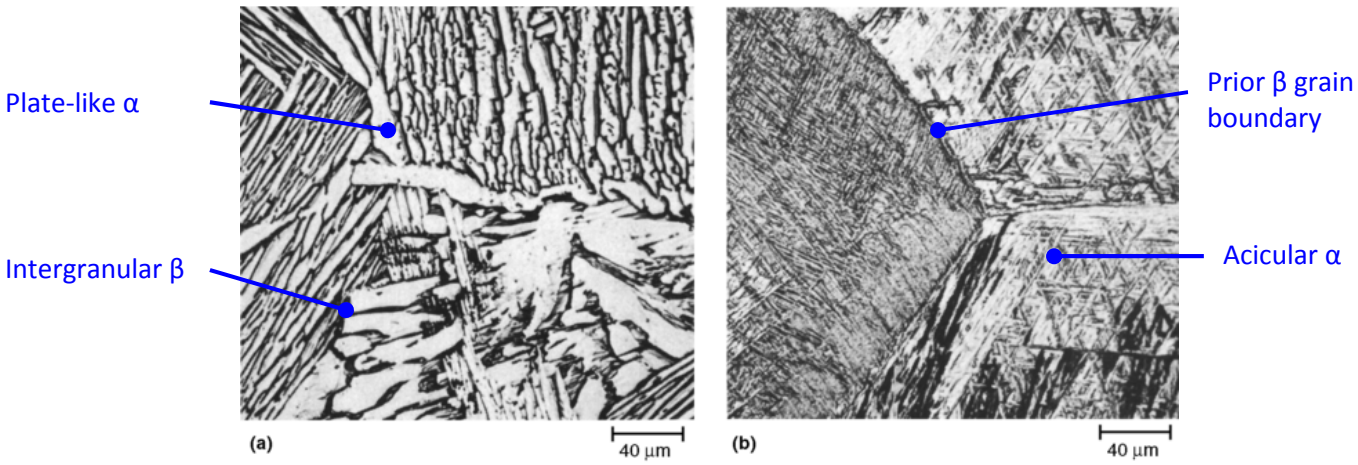

Figure 3: Micrographs of Ti-6Al-4V held for 1 h at 1950 °F (1065 °C), above the beta transus. (a) Furnace cooled. Microstructure of plate-like alpha (light) and intergranular beta (dark). (b) Air cooled. The structure consists of acicular alpha (transformed beta); prior beta grain boundaries are visible. Etchant for both samples: 10 mL HF, 5 mL HNO_3, 85 mL H_2O. 250×. [9]

Much like the iron-carbon system, Ti-6-4 forms martensite when quenched from above or near the transformation temperature. In Ti-6-4 this phase is called alpha-prime (α') and has a heavily twinned HCP crystal structure created by a rapid shear transformation of beta.[2]

Figure 4 shows the alpha-prime structure in a water quenched Ti-6-4 sample. This sample contains some primary alpha grains because it was not heated past the transus, if it was, the structure would be completely martensitic.

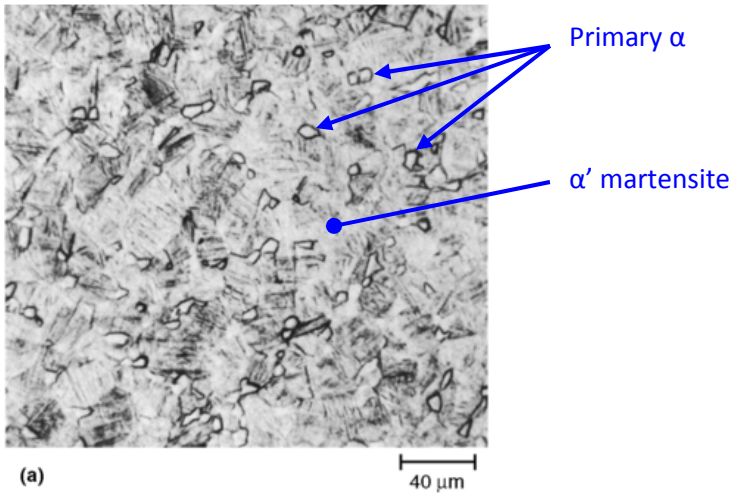

Figure 4: α′ martensite in Ti-6Al-4V. Sample held for 1 h at 1750 °F (955 °C), which is below the beta transus, and water quenched. Equiaxed "primary" alpha grains (light) in α′ matrix. Etched with 10 mL HF, 5 mL HNO3, 85 mL H_2O. 250×.[9]

Alpha-prime martensite is not hard and brittle like steel martensite because interstitials are much more soluble in titanium's HCP structure than iron's BCC, but some strengthening does occur as a result of grain refinement and an increase in dislocation density.[2]

Forging

Ti-6-4 parts are wrought, cast, or produced by powder metallurgy. Of these, wrought parts are the most cost effective and the kind most frequently used in the aerospace industry, making up 95% of the market.[1] One of the more prevalent types of wrought processing is forging: a manufacturing process in which preheated metal is pressed or squeezed into a desired shape (Figure 5).[3]

Figure 5: Forged titanium part from Weber Metals.

The tools used in industrial forging are hammers and presses. Hammers use heavy blows to shape metal while presses use a more controlled, slower application of pressure. Hydraulic presses offer the finest degree of pressure and strain-rate control, and range from 1,200 up to 50,000 tons (Figure 6).[3]

Figure 6: The 33,000 ton forging press at Weber Metals.[4]

Forging Ti-6Al-4V

In the forging of any engineering alloy, the strain rate of deformation and the working temperature at which that deformation takes place are important factors in determining results. While there is a strong interaction between these two factors in forging Ti-6-4, there are serious consequences for using too high a strain rate at any temperature. A study in the Air Force Research Laboratory found that strain rates of 1 s^{-1} or greater caused adiabatic heating of the workpiece resulting in localized flow of the metal and sometimes fracture during deformation.[5]

There are two main options for the mechanical working temperature of Ti-6-4: forging in the beta range, or forging in the alpha + beta range; both result in different as-forged microstructures. When hot-worked in the beta range, deformation occurs while the sample is composed of a single phase, so the sample cools from the single phase region almost as if no deformation occurred. This results in a microstructure of transformed beta quite similar to those previously shown in Figure 3.[6]

When mechanical working takes place in the alpha + beta region, on the other hand, 'texturing' of the microstructure can occur as a result of the HCP alpha structure. Texturing refers to the alignment and elongation of alpha grains in the metal, an example of which is shown in Figure 7. The effects of texturing observed after cold rolling a continuous strip of Ti-6-4 included extreme property directionality as well as an increase in the alloy's susceptibility to aqueous stress-corrosion cracking.[7] It has also been observed that textures developed upon deformation can be further pronounced by a recrystallization anneal following mechanical working. In a study on the effects of hot rolling in Ti-6-4, samples were found to exhibit defined anisotropy of tensile properties after unidirectional rolling and annealing.[8]

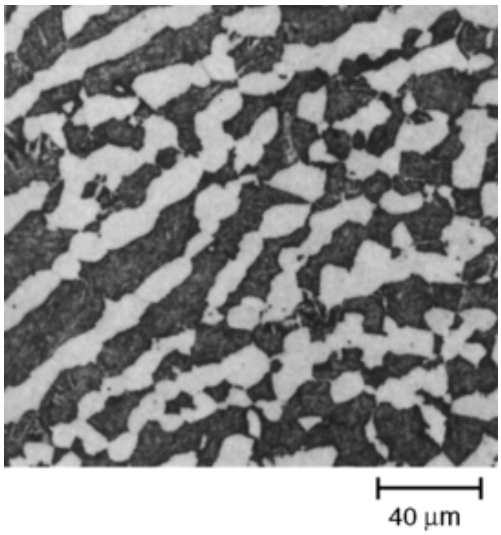

Figure 7: Ti-6Al-4V, forged to 20% reduction at 1750°F (955°C) (below the beta transus). Microstructure is elongated alpha (light) in a matrix of transformed beta. Original microstructure consisted of plate-like alpha. Etchant: Kroll's reagent (ASTM 192), 250x.[9]

Realistic Constraints

The present study is being conducted in partial fulfillment of undergraduate engineering course requirements. Since engineers are expected to understand the importance of their work and its consequences, the purpose of this study has been tied to the realistic constraints of economics and politics.[11]

Economics

Despite its relative abundance in the earth's crust, titanium is an expensive metal to produce and process. This high cost comes from three aspects of the processing. The first is the energy intensive multi-step process required to extract titanium metal from its ore. The second is the fact that large product volumes are necessary for companies to justify running the machines needed for working titanium and its alloys. This is still an issue today since most titanium products have individual requirements and are produced in relatively small batches. The third issue and the one most relevant to the present project is in regard to the unique properties of titanium. Due to its thermal properties, flow properties, and tendency to develop microstructural texture when hot-working, titanium processing must be well thought-out and thoroughly tested.[7] If the results of the project reveal information useful in understanding the effects of texturing mechanisms in forged Ti-6-4, more effective and consistent processing of

forged parts can be designed. This could result in higher-quality components being more economical to produce and affordable for customers in any industry titanium is used.

Politics

More than 50% of all titanium produced is made into Ti-6Al-4V, and over 80% of this is used in the aerospace industry.[1] Since the vast majority of aerospace work is government funded either directly (military) or indirectly (defense contracts), it is safe to say that governments fund nearly 40% of all titanium production. Given the fact that government funds come primarily from tax-paying citizens, it is the responsibility of the government to spend those funds wisely, including on the materials that make up their plane, rocket, and missile components. As engineers we have a duty to make the best product possible for the lowest cost, not only to our company, but to our country. Research in the area of titanium metallurgy such as this project is a step in line with this principle.

Methods and Materials

Heat Treatments

Six heat treatments were explored for this project: an annealing treatment (ANN) and five solution-treating and aging treatments (STA 1-5). The schedules for each treatment are given in Table III. All annealing treatments and solution treatments for STA 1 and STA 2 were done in a CM Rapidtemp furnace (model CM 1710 FL). All other solution and aging treatments were done in one of two Fischer Iso-temp furnaces.

Table III: Heat Treatment Schedules

Treatment Name	Temperature (°F)	Time (hr)	Cooling Method	Aging Temperature (°F)	Aging Time (hr)	Cooling Method
ANN	1350 (±25)	2 (+.25/-0)	Air Cool		N/A	
STA 1	1785 (±15)	1 (+.25/-0)	25 s delay in air, Water Quench	990 (±10)	6 (+.25/-0)	Air Cool
STA 2	1735 (±15)					
STA 3	1635 (±15)					
STA 4	1535 (±15)					
STA 5	1435 (±15)					

Four tensile samples from each direction (LG and LT) were heat treated, resulting in a total of eight samples for each treatment. To reduce the total time spent heat treating, each heat contained either four or eight samples. Annealing, solution, and STA 1 aging treatments were all done in two four-sample heats; aging treatments for STA 2-5 were all done in eight-sample heats. To nullify any effect of different heats, two samples from each direction were included in the four-sample heats, and each eight-sample heat contained all eight samples of one STA designation. To ensure even heating, samples were placed on alumina racking that elevated them about one-quarter inch above the furnace floor. Upon removal from the furnace, all air cooled samples were placed on identical alumina racking resting on a firebrick beside the furnace (Figure 8). This racking also ensured traceability by allowing each sample to be consistently placed in the same position relative to the others in its heat. Sample number and racking position were recorded prior to each heat treatment.

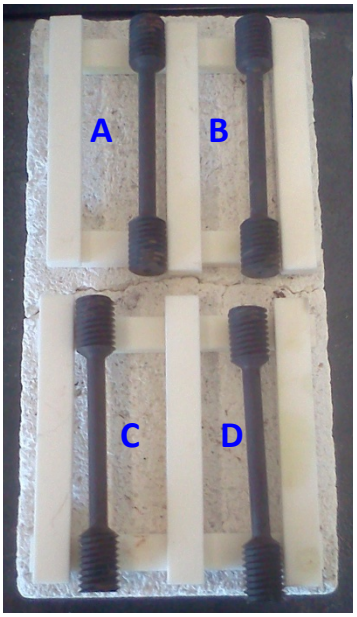

Figure 8: Annealed samples after air-cooling. Each sample was assigned a letter A-D corresponding to its rack position as pictured.

For each treatment, the furnace was allowed to reach temperature and stabilize for 10-15 minutes. Then the door was opened and metal tongs were used to place all four samples on the racking one at a time. The time at which samples were placed in the furnace (time in) was recorded along with the time at which the furnace returned to the proper temperature (time at

temp); treatment time was measured from time at temp. After the appropriate time had elapsed, samples were removed from the furnace. For air cooling samples (after annealing and aging treatments), the furnace door remained open and samples were placed on the cooling rack one at a time. For delay and quenching samples (after solution treatments), one sample at a time was removed, held in air, and then dropped into a water bath with the door being closed between each sample. The time of sample removal (time out) was also recorded. All heat treatments below 1650°F were monitored with a Type K thermocouple connected to an Omega data logger.

Preparation for Chemical Milling

Chemical milling is a surface treatment that consists of dissolving a uniform surface layer off of a part in a chemical bath. This method is commonly used to remove oxides and surface layers from titanium parts. One such brittle alpha-case layer had formed on the samples due to oxygen absorption during heat-treating. This brittle layer caused samples loaded in tension to fracture at elongations far below what was expected, and did not allow accurate tensile properties to be observed. Chemical milling was chosen as a solution to this problem.

The primary area of concern for these samples was the gauge length, since this is the region of the tensile sample that undergoes deformation and fracture. To prepare for chemical milling, the gauge length of each sample was sanded with 220 grit silicon-carbide sandpaper. This removed any loosely adherent oxide scale to avoid preferential material removal in the chemical bath. Then the threaded ends of each sample were treated with two coats of a common polymer dip (Plasti-dip by 'Performix') to preserve the threads and serial numbers so traceability could be maintained following chemical milling. The coats were applied one end at a time, waiting the recommended half-hour between coats. While waiting between coats, samples were arranged so the freshly dipped end extended over the side of a firebrick (Figure 9). Once both coats had been applied to both ends, samples were left to dry overnight.

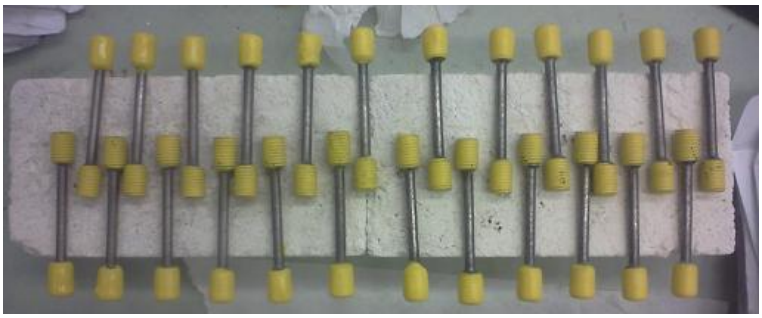

Figure 9: Threaded ends of tensile samples coated so only the gage lengths would be chemical milled.

Tensile Testing

After receiving the samples back from chemical milling, the plastic coatings were removed from the threaded ends and all gauge length diameters were measured to document how much material was removed. As a result of chemical milling, the diameters decreased from 0.252 in to 0.238 in. While this caused our samples to fall outside the diameter tolerance specified in ASTM E8 for 4D subsize samples (1 in gage length, 0.250 ±0.005), testing continued as planned. Tensile testing was conducted on an Instron Tensile Testing System (model 3369) 11,000 lb capacity load cell. The load train for these tests consisted of upper and lower flat-jawed grips and tabbed threaded adapters (Figure 10). The ends of the samples were threaded into the adapters until all threads were engaged before placing the adapter tabs into the grips. To load the sample and adapters into the load frame, the top grip was tightened on one tab and the cross-head was lowered until the lower grip could be tightened on the remaining tab.

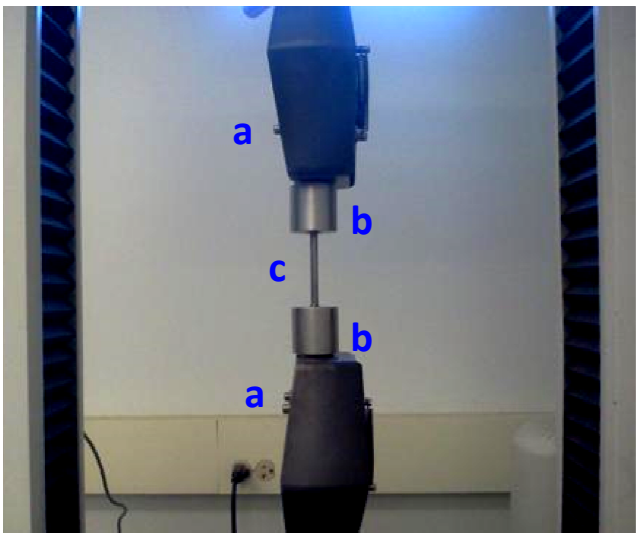

Figure 10: Tensile testing configuration for Ti-6-4 sample tests. (a) upper and lower grips, (b) threaded tab adapters, (c) Ti-6-4 sample.

A 1 in (25.0 mm) gage length extensometer was attached to the sample prior to testing and used to measure the first 1.5% elongation. The extensometer ensured accurate elastic modulus and 0.2% offset yield strength calculations. The crosshead rate of motion was set to 0.1 in/min for the first 1.5% elongation, after which the extensometer was removed and the crosshead rate increased to 0.5 in/min for the remainder of the test. Bluehill software (Instron Corp.) was used to establish the test method and control each test. The test method was configured to take measurements every 50 ms in a format compatible with Microsoft Excel.

Samples were chosen for testing in random order. One bag was filled will all LG samples, another with all LT samples. Before each test a quarter was flipped to determine which bag a sample would be drawn from, heads for LG and tails for LT. After a sample was chosen and the serial number recorded in Bluehill's testing window, the sample was placed in the frame and loaded in tension to fracture.

Metallography

To observe any microstructural trends, metallographic samples were cut from the threaded ends of six tensile bars. This group of six bars consisted of one each LG and LT from the ANN, STA 1, and STA 5 heat treatments to show the annealed microstructure as well as the difference between highest and lowest solution-treating temperatures. When these sections were cut, the orientation of the surface exposed was perpendicular to the direction of the sample, meaning that LG samples produced an LT cross section and vice versa (Figure 3).

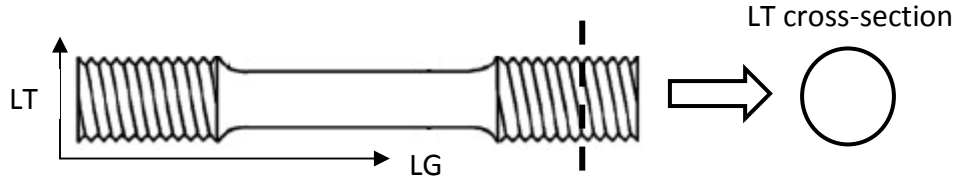

Figure 11: LG direction sample schematic showing LT cross section removed from threaded end for metallography.

The saw used to cut the cross sections was an Allied 'Techcut 4' wafering saw with a 7 in high-concentration diamond blade spinning at 220 RPM. To make polishing easier, the specimens were mounted in blue, mineral-filled diallyl phthalate using a Buehler 'Simplimet 2' 1.25 in mounting press with the cut side of the specimen exposed facing out of the mount.

Upon removing the mounts from the press, the sample serial number was scribed into the back with a vibro-pen to maintain traceability and the edges of the mount rounded on a belt sander.

In preparation for etching, the surface was ground and polished. Grinding was done on wet silicon-carbide sandpaper progressing from 240 grit to 600 grit in four steps. During grinding, light pressure was applied while the sample was pushed across the paper. Samples were occasionally rotated 180° on the same paper to prevent beveling of the surface. Between grits, samples were rotated 90° to more easily differentiate between scratches.

Polishing began with a 9 micron diamond suspension in an ethanol and glycol based lubricant on an eight inch vel-cloth wheel spinning at 200 RPM. Moderate downward force was applied as the sample was moved in a clockwise circle around the pad. After polishing in this manner for approximately 1.5 minutes, the sample was rinsed under hot water, then rinsed with a spray of ethanol, and dried with a hot-air dryer. This procedure was then repeated using pads with 6, 1, and 0.05 micron abrasive suspensions. After the sub-micron final polish, the sample was etched.

Kroll's reagent (10% HF, 5% HNO_3, 85% H_2O) was used for etching. Each sample was held over an empty beaker while enough etchant to cover the surface was dripped onto it. Etchant was left on each sample for 15 seconds, then dumped into the empty beaker. The sample was then dropped into a 4% NaOH solution to neutralize any remaining acid. Samples were then washed using the same method as between each polishing step. Once all samples had been etched, each was individually examined on a Leica DMR optical microscope connected to a PC running Qcapture pro imaging software.

Results

Tensile Testing Without Chemical Milling

The first sample tested after heat-treating (STA 1 LT sample) fractured after only 3% elongation. Since elongations of closer to 10% were expected, testing was stopped until the cause of premature failure could be determined. A cross section of the gauge length was metallographically prepared according to the methods outlined above. This cross section showed the expected microstructure, but also a 200-300 µm thick surface layer that had formed during heat treatment (Figure 12).

Figure 12: Micrograph showing surface layer of brittle alpha-case. 200x, etched with Krolls reagent.

This surface layer was a band of alpha phase that formed due to the high-temperature diffusion of oxygen into the surface of the sample. Since the layer is brittle, it cracked when placed under tension and that crack propagated quickly through the more ductile interior of the sample causing premature fracture. This mechanism was confirmed by examining the fracture surface in a Scanning Electron Microscope (SEM, Figure 13). These findings resulted in the samples being sent out for chemical milling.

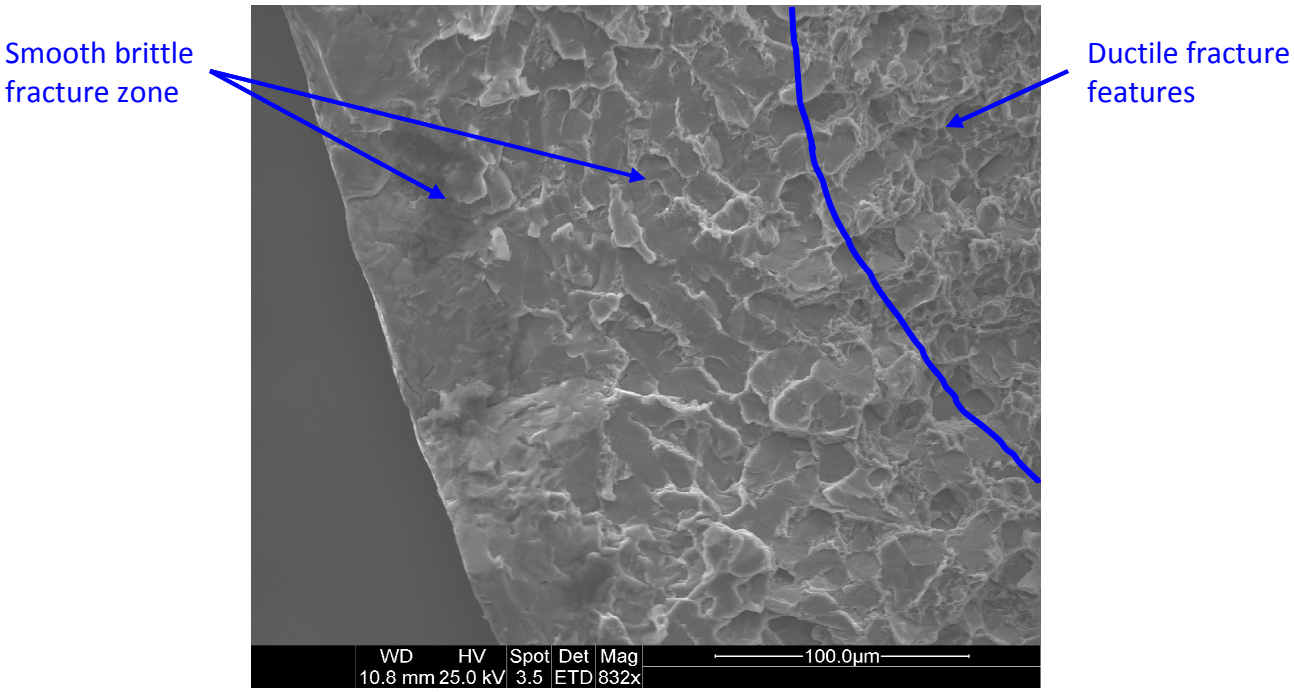

Figure 13: SEM micrograph of premature fracture surface. The surface is smooth at the surface indicating brittle fracture, and dimpled in the interior indicating the more expected ductile fracture mode. The brittle region extends 200-300 µm into the sample, the same distance as the alpha-case.

Tensile Testing After Chemical Milling

After the samples had been chemically milled, they showed properties within the expected ranges. Representative stress-strain curves of LG and LT annealed samples are shown in Figure 14. Average properties of annealed samples are given in Table IV.

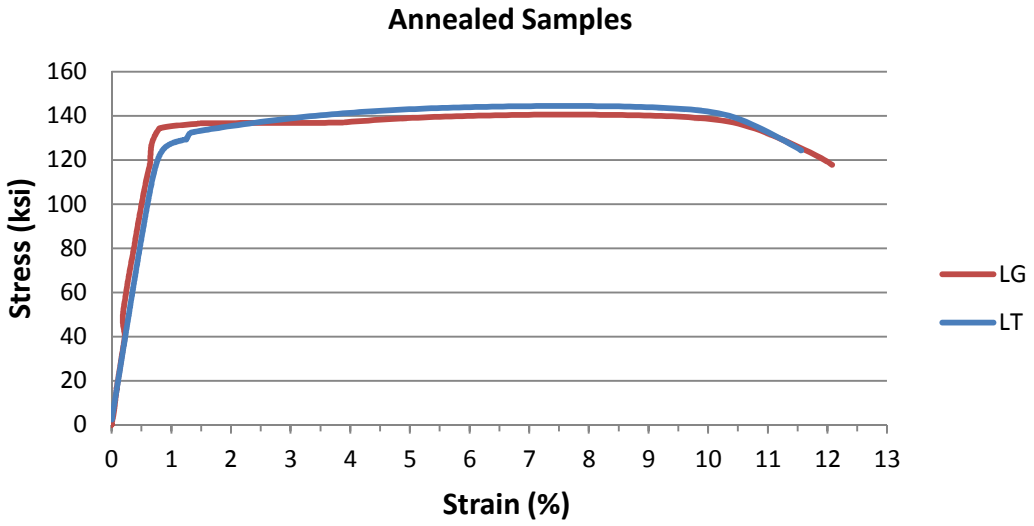

Figure 14: Representative stress-strain curves of annealed Ti-6-4 samples.

Table IV: Average* Tensile Properties of Annealed Samples

Direction	Yield Strength (ksi)	Tensile Strength (ksi)	Elongation (%)
LG	124.95 (± 1.63)	140.73 (± 1.58)	12.19 (± 0.43)
LT	128.49 (± 3.91)	144.42 (± 1.24)	11.30 (± 0.42)

*4 samples tested in each orientation

In the annealed samples, strengths differ by about 4 ksi with the LT direction exhibiting higher values. Figure 15 shows stress-strain curves characteristic of the STA 1 sample group. Average properties of STA 1 samples are given in Table V.

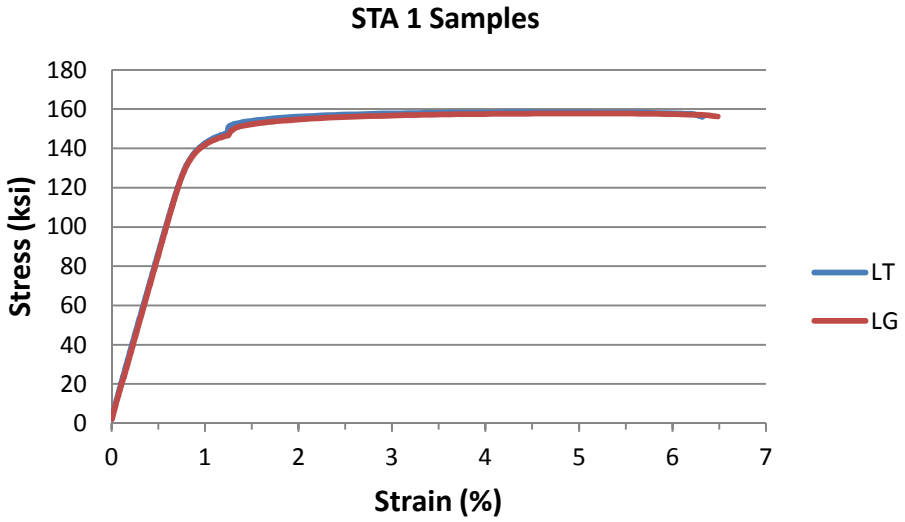

Figure 15: Representative stress-strain curves of STA 1 Ti-6-4 samples.

Table V: Average* Tensile Properties of STA 1 Samples

Direction	Yield Strength (ksi)	Tensile Strength (ksi)	Elongation (%)
LG	144.07 (± 0.96)	156.22 (± 1.12)	6.31 (± 0.53)
LT	143.50 (± 1.17)	158.22 (± 1.08)	6.96 (± 0.57)

*4 samples tested in each orientation

The property values of STA 1 samples fall close together, with average strengths differing by no more than 2 ksi and elongations by less than 1%. Representative stress-strain curves of STA 2 also show similarity in strengths (Figure 16). The curves show a greater disparity in elongation because each curve is taken from the data of one sample's test while the average properties in Table VI are calculated from all four samples of each direction.

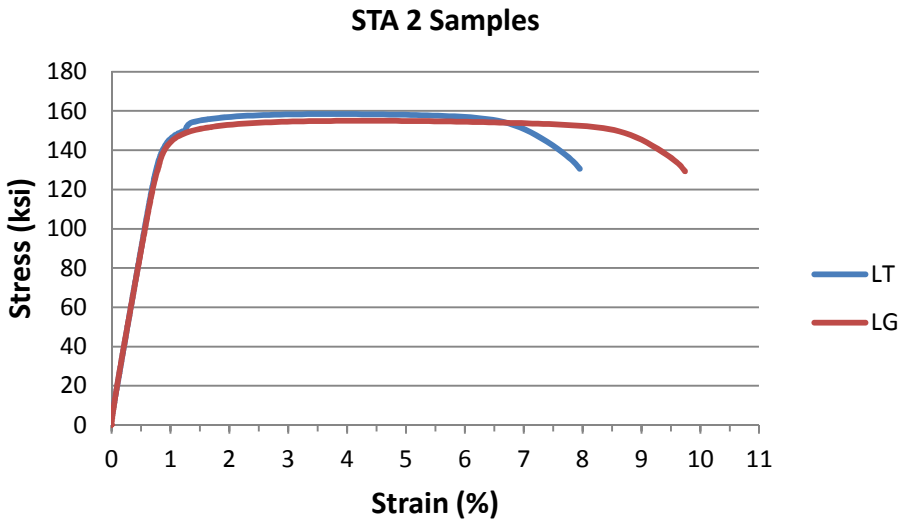

Figure 16: Representative stress-strain curves of STA 2 Ti-6-4 samples.

Table VI: Average* Tensile Properties of STA 2 Samples.

Direction	Yield Strength (ksi)	Tensile Strength (ksi)	Elongation (%)
LG	146.34 (± 3.90)	155.17 (± 0.31)	9.07 (± 1.16)
LT	147.1 (± 0.81)	159.21 (± 1.11)	8.10 (± 0.63)

*4 samples tested in each orientation

The samples that underwent heat treatment STA 3 also showed similar properties on average (Figure 17 and Table VII).

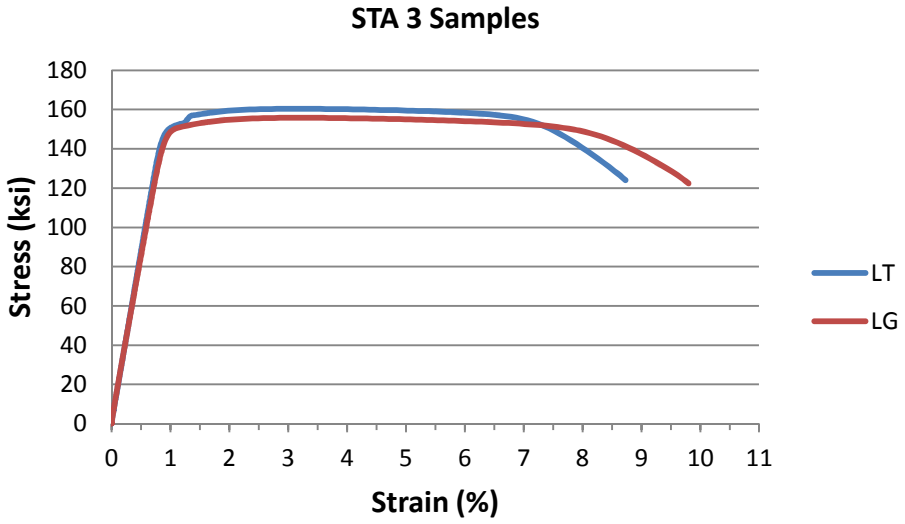

Figure 17: Representative stress-strain curves of STA 3 Ti-6-4 samples.

Table VII: Average* Tensile Properties of STA 3 Samples

Direction	Yield Strength (ksi)	Tensile Strength (ksi)	Elongation (%)
LG	150.25 (± 1.00)	156.20 (± 0.90)	9.37 (± 0.32)
LT	151.89 (± 1.83)	159.59 (± 1.64)	9.16 (± 0.93)

*4 samples tested in each orientation

STA 4 samples continued the trend of seeing approximately 4 ksi higher strengths in the LT direction (Figure 18 and Table VIII).

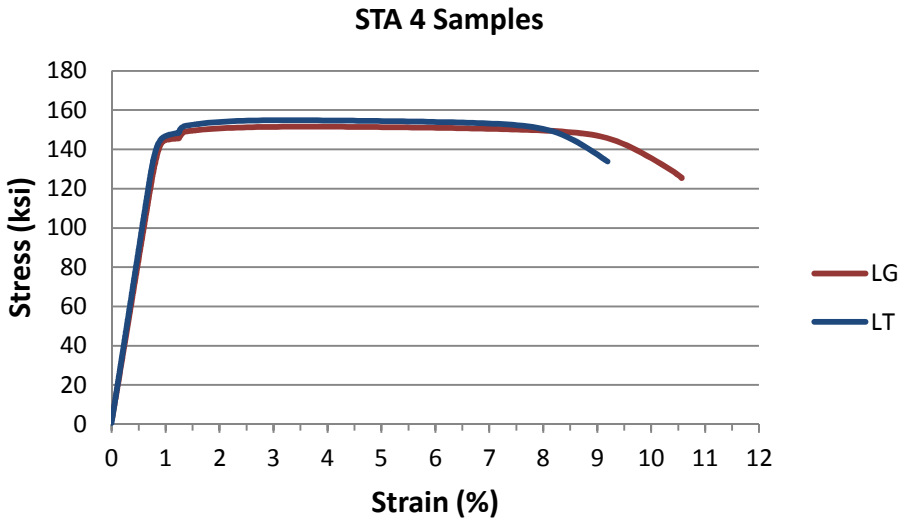

Figure 18: Representative stress-strain curves of STA 4 Ti-6-4 samples.

Table VIII: Average* Tensile Properties of STA 4 Samples

Direction	Yield Strength (ksi)	Tensile Strength (ksi)	Elongation (%)
LG	146.27 (± 2.10)	152.11 (± 1.18)	10.12 (± 0.89)
LT	149.76 (± 2.30)	156.26 (± 0.94)	9.12 (± 0.85)

*4 samples tested in each orientation

The STA 5 treated samples displayed a larger difference between LT and LG directions with the LT properties between 5 and 10% greater than those of the LG direction (Figure 19 and Table IX).

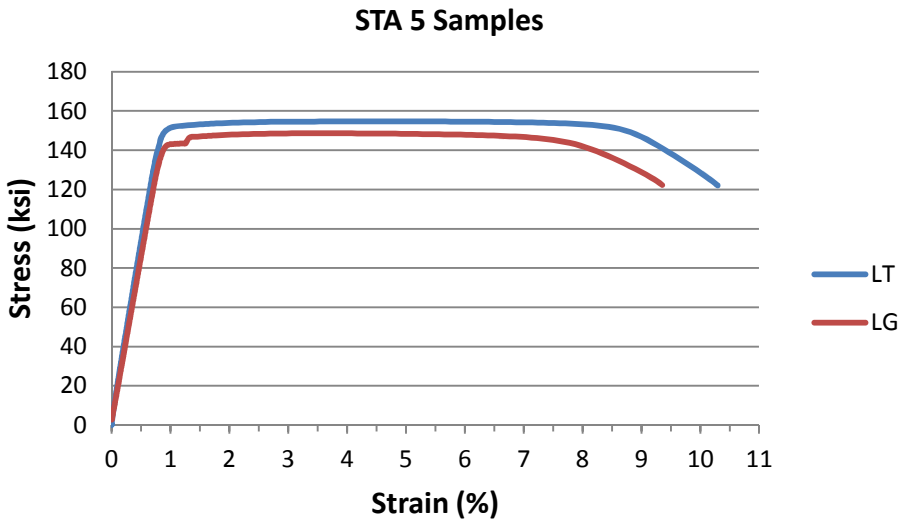

Figure 19: Representative stress-strain curves of STA 5 Ti-6-4 samples.

Table IX: Average* Tensile Properties of STA 5 Samples

Direction	Yield Strength (ksi)	Tensile Strength (ksi)	Elongation (%)
LG	143.82 (± 1.04)	148.77 (± 1.70)	9.32 (± 0.44)
LT	151.57 (± 0.71)	155.98 (± 2.01)	10.28 (± 0.73)

*4 samples tested each direction

Through all of these tests, the largest difference between directions was observed in tensile strength, and this difference was greater in heat treatments of lower temperatures than those of higher temperatures (Table X).

Table X: Average Differences Between Tensile Properties for all Heat Treatments

Heat Treatment	Difference in Yield Strength (ksi)	Difference in Tensile Strength (ksi)	Difference in Elongation (%)
ANN	3.54	3.69	.89
STA 1	.57	2.0	.65
STA 2	.76	4.04	.97
STA 3	1.64	3.39	.21
STA 4	3.49	4.15	1.0
STA 5	7.75	7.21	.96

Metallography

Cross sections of annealed samples showed a clear difference in microstructure between LG and LT directions (Figure 20). The LG direction shows much more elongated alpha phase than the LT direction.

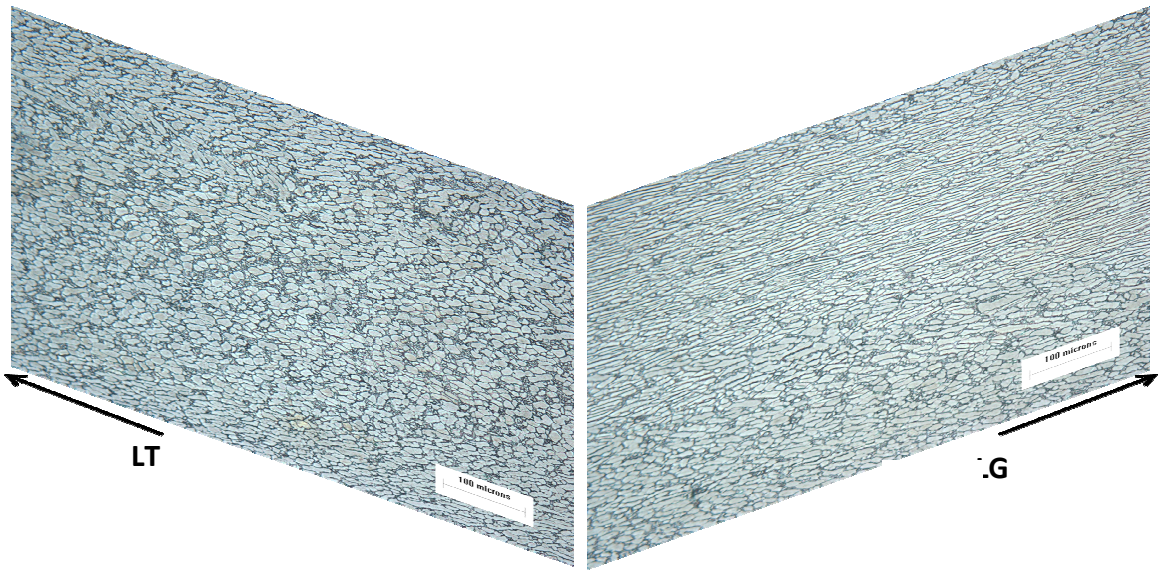

Figure 20: 200x micrographs of annealed samples etched with Kroll's. LG section shows much more elongation of the alpha phase than the LT direction.

STA 1 samples were exposed to the highest solution-treating temperature, and showed no difference in microstructure between directions (Figure 21).

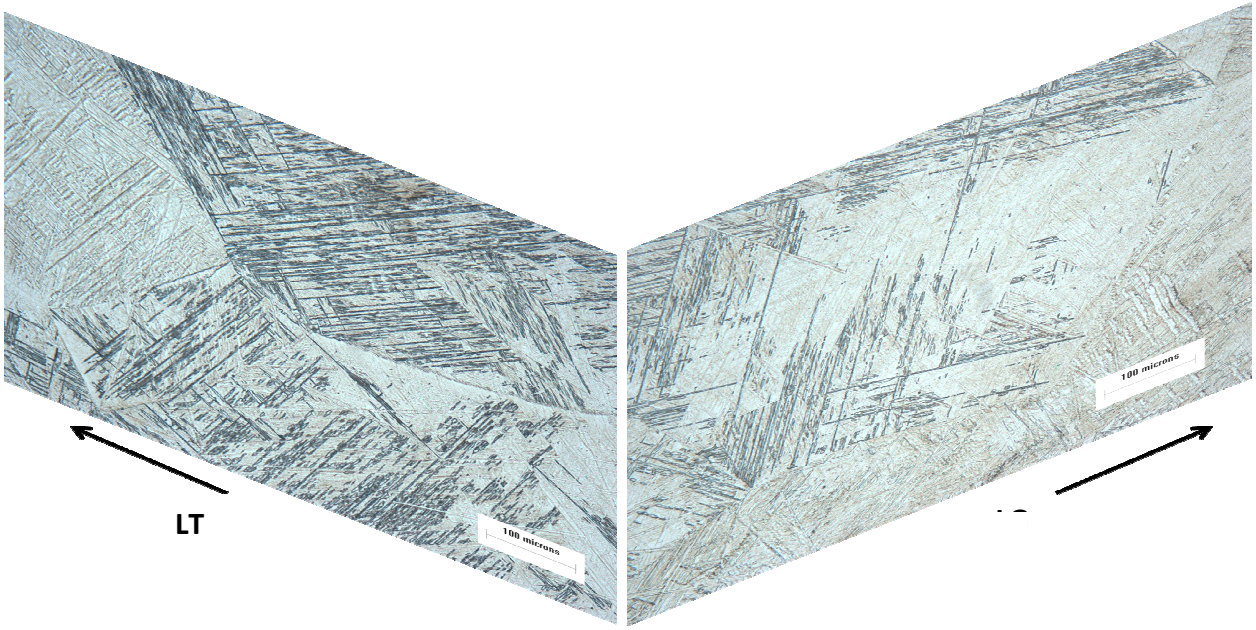

Figure 21: 200x micrographs of STA 1 samples etched with Kroll's. There is no observable difference in the microstructures.

STA 5 samples were exposed to the lowest solution-treating temperature and showed a difference in microstructure similar to that observed in annealed samples (Figure 22).

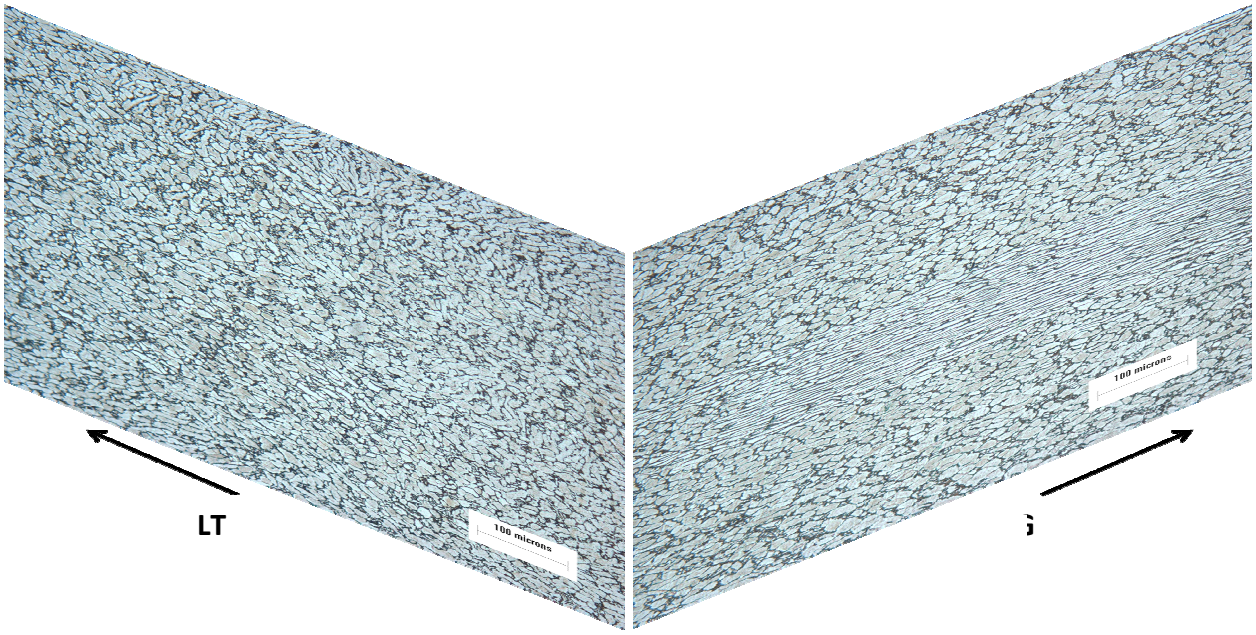

Figure 22: 200x micrographs of STA 5 samples etched with Kroll's. LG section shows more elongation of the alpha phase than the LT direction.

The microstructures of samples STA 2-4 are suspected to fall between those of STA 1 and STA 5. These would move from the completely transformed structure of STA 1 to a finer bimodal structure containing equiaxed alpha grains in a transformed beta matrix. With descending solution-treating temperatures these alpha grains would appear larger and in greater quantity until the structure of STA 5 is reached.

Discussion

Three mechanical properties were examined for directional differences: Tensile strength, Yield strength, and Elongation. To ensure results were statistically significant, Minitab software was used to run a general linear model of the data with an interaction for the factors heat treatment and direction. Boxplots were generated for each property to show the values for each direction and treatment in relation to each other. Confidence intervals for predicted directional differences were also created.

The boxplots of Tensile strength shows a general trend in property values of increasing spread between directions going from STA 1 to STA 5 (Figure 23). Calculated information from the general linear model supports the graph, showing no difference within STA 1, and predicts potentially significant differences (up to 10 ksi) between LG and LT tensile strength depending on heat treatment after forging (Table XI).

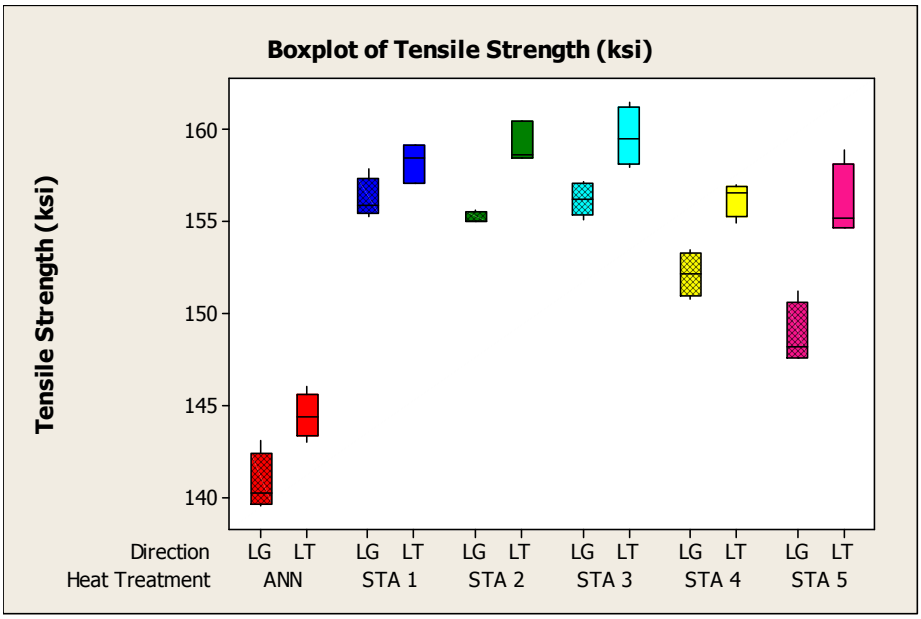

Figure 23: Boxplot of tensile strength data showing a trend of increasing anisotropy with decreasing solution treating temperature.

Table XI: P-values, differences between means of sample data, and 95% confidence intervals predicting differences in the population.

Heat Treatment	P-value	Difference Between Means	95% C.I. for Predicted Difference
ANN	.016	3.69	(.42 – 6.96)
STA 1	.697	-----	-----
STA 2	.014	4.04	(.51 – 7.57)
STA 3	.036	3.4	(.13 – 6.67)
STA 4	.004	4.15	(.88 – 7.42)
STA 5	.000	7.22	(3.95 – 10.48)

Yield strength data does not display as much variation as tensile strength, but STA 5 clearly remains the heat treatment preserving the most anisotropy (Figure 24). The difference between LG and LT in STA 5 is approximately 7.75 ksi with a P-value of .001. A 95% confidence

interval predicts a difference ranging from 2.5 to 13.0 ksi, as significant as that found in tensile strength.

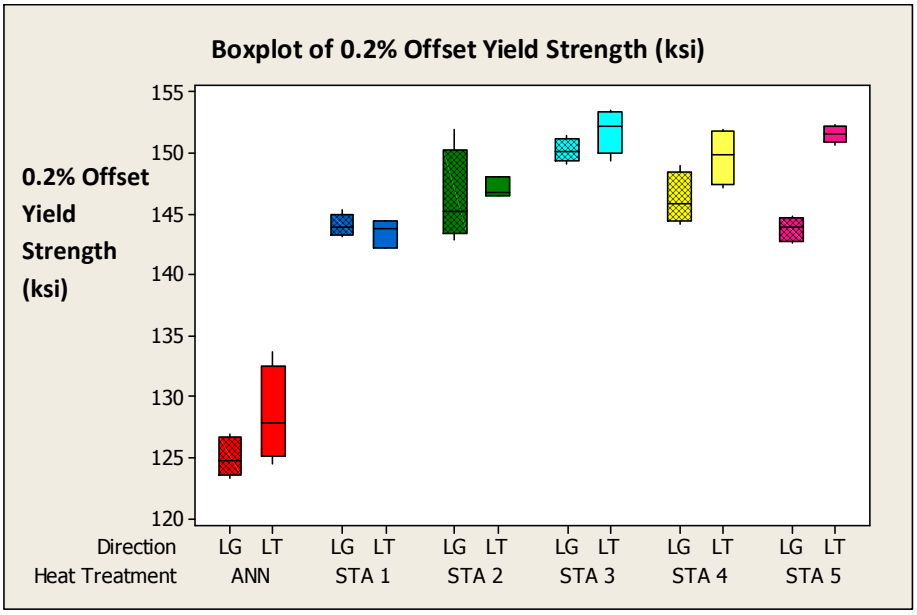

Figure 24: Boxplot of yield strength data showing an appreciable difference between directions only in STA 5 samples.

A plot of elongation showed the expected trend of increasing ductility with lower strength heat treatments, but failed to show any differences between directions (Figure 25). The P-values were greater than 0.69 for all tests.

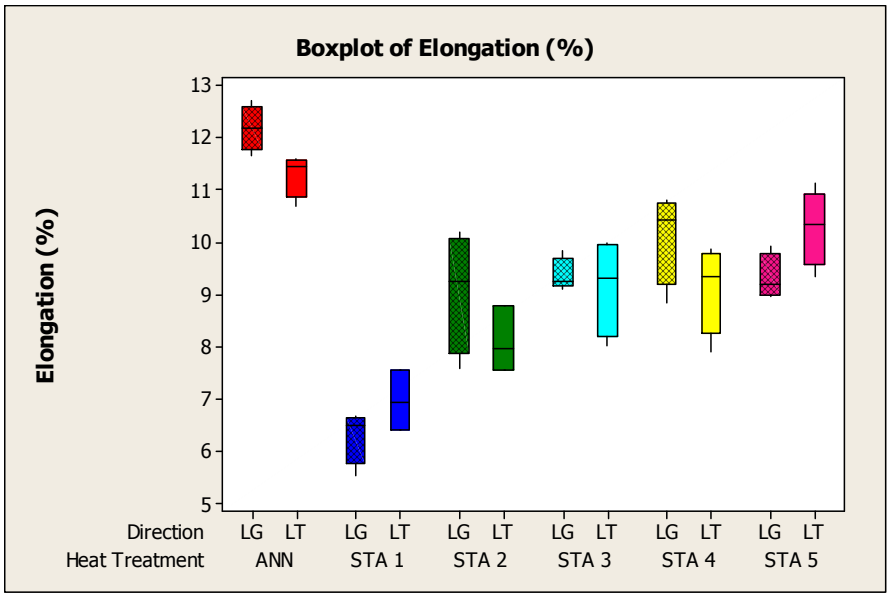

Figure 25: Boxplot of elongation data showing no significant differences between directions.

According to a statistical analysis of the data, tensile strength is the property most affected by forging-induced crystallographic texturing. Since all samples exhibited a plateau region in their stress/strain curves and thus did not experience a great deal of strain hardening, it is reasonable to equate the tensile strength to the stress required for continuous plastic deformation. That is, the tensile strength of these samples is the stress at which dislocations can move through the material until fracture occurs by micro-void coalescence (MVC) or some other ductile mechanism.

Since the favored slip systems of titanium are aligned with the c-axis of HCP crystals, they are also aligned with the LG direction of the forging and thus the LG samples. Textured LG samples contain more alpha grains aligned with the loading direction so the preferred slip systems are readily available for dislocation movement. Conversely, the alpha grains in LT samples are aligned perpendicular to the loading direction, requiring greater stresses to continue moving dislocations through non-favored slip systems. This results in a greater tensile strength observed in the LT direction.

Elongation is a measure of how much plastic deformation occurs (or how much dislocations can move) before necking, or local plastic deformation occurs. This has less to do with orientation, and more to do with grain size and relative volume fractions of phases in which dislocations can move. LT and LG samples from the same heat treatment can be expected to contain the same amounts of phases and grain sizes and thus will have essentially the same elongation.

Another trend that evolved in this data is that anisotropy decreased with increasing solution-treating temperatures. This is explainable given the nature of the anisotropy and the purpose of solution-treatments. The anisotropy arises from a texturing of the alpha phase of during forging. Solution-treating is commonly meant to homogenize a material, effectively erasing the pre-solutionized structure. As the solution-treatment temperature of these samples approaced the beta transus, more and more of the textured alpha phase became ‘erased’, leaving behind a more uniform microstructure and thus less anisotropy.

Conclusions

1. Higher solution-treating temperatures decrease anisotropy by transforming textured alpha into beta, eliminating the aligned structure.
2. Tensile strength is the property most affected by forging-induced texturing resulting in differences of 3-10 ksi between longitudinal and transverse directions.
3. If anisotropy remains after a heat treatment, the transverse direction will be stronger due to non-preferred slip systems being the primary mechanism of deformation.

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