

## 38-6-44 Figures and Tables

- 74 references
- 231 Tables and Figures
- 18 pages Text

### Listing of all the tables and Figures in 38-6-44 Chapter

Table 1.3.1 Specifications for Ti-3Al-8V-6Cr-4Mo-4Zr (Ref. 4)

Table 1.4.1 AMS specified chemical compositions of Ti-3Al-8V-6Cr-4Mo-4Zr alloy (similar to UNS 58640) in percentages by weight in various product forms and conditions (Refs. 5-8)

Table 1.4.2 ASTM specified chemical compositions of Ti-3Al-8V-6Cr-4Mo-4Zr alloy for various product forms (Refs. 9-14)

Table 1.5.1 Common beta-stabilizing isomorphous and eutectoid type elements for titanium alloys and their effect on beta-transus reduction (Ref. 16)

Figure 1.5.2 Pseudo-binary Ti-beta isomorphous phase diagram for multielement titanium alloys, Mo-Eq = Mo + 0.67xV + 0.44xW + 0.28xNb + 0.22xTa + 2.9xFe + 1.6xCr - 1.0xAl (wt. %); the solute-rich Beta-C at Mo-Eq  $\approx$  16 (see 1.5.3) is too stable to decompose isothermally to a  $\beta$  +  $\omega$  phase mixture but instead decomposes into two bcc phases, the solute-rich  $\beta$  and the solute-lean  $\beta'$  phases (In addition, with the eutectoid forming Cr present in Beta-C, the compound TiCr<sub>2</sub> forms after extended aging) (Ref. 18)

Figure 1.5.3 Comparison of the Mo-Eq values for Beta-C alloy (Mo-Eq = 16) and various other commercial beta titanium alloys, where Mo-Eq = Mo + 0.67xV + 0.44xW + 0.28xNb + 0.22xTa + 2.9xFe + 1.6xCr - 1.0xAl (wt. %), and titanium alloys at bottom of the diagram included for comparative purposes (Ref. 18)

Figure 1.5.4 Dependence of beta transus temperature on Mo-Eq of Beta-C (Mo-Eq  $\approx$  16.0) and other titanium alloys (Ref. 18)

Figure 1.5.5 Diagram showing relative positions of Beta-C and other common metastable beta titanium alloys and ranges of solution treatment and age (STA) heat treatments (Ref. 20)

Table 1.5.6 Recommended annealing heat treatment, solution treatment (ST) and age (A) conditions, and stress relief annealing temperature and time for Beta-C (Ref. 23)

Figure 1.5.7 Time-Temperature-Transformation (TTT) diagram for RMI 38644 (Beta-C) alloy in the form of 15.9 mm thick plate that was solution treated at 1200K (927C or 1700F) for 0.5 h, air cooled, and aged for up to 1,000 h at temperatures indicated and air cooled from aging temperatures (Ref. 29)

Figure 1.5.8 Optical micrograph of RMI 38644 (Beta-C) alloy in the form of 15.9 mm thick plate solution treated at 1200K (927C or 1700F) for 0.5 h and air cooled - microstructure consists of coarse  $\beta$ -phase grains (mean linear grain size = 80.5  $\mu$ m) which contain 0.25-1.5  $\mu$ m sized blocky particles, identified by energy dispersive x-ray and TEM diffraction analysis as the hexagonal Ti<sub>x</sub>Zr<sub>y</sub>Si<sub>z</sub> phase (etchant: 15 ml HNO<sub>3</sub>, 5 ml HF, 15 ml lactic acid, and 65 ml H<sub>2</sub>O) (Ref. 29)

Figure 1.5.9 (a) RMI Beta-C (Ti-3Al-8V-6Cr-4Mo-4Zr) sheet solution treated at 1800F for 5 min and water quenched with structure showing metastable  $\beta$ -phase; (b) sample in (a) aged at 975F for 8 h and air cooled resulting in an aged structure consisting of equilibrium  $\alpha$ - and  $\beta$ -phase (both 100X, Etchant: 2% HF + 4% HNO<sub>3</sub>) (Ref. 19)

Figure 1.5.10 (a) RMI Beta-C (Ti-3Al-8V-6Cr-4Mo-4Zr) sheet solution treated at 1500F for 7 min and air cooled with structure showing metastable  $\beta$ -phase and fine  $\alpha$ -phase precipitation (100 X, Etchant: 2% HF + 4% HNO<sub>3</sub>); (b) sample in (a) aged at 950F for 6 h and air cooled resulting in equiaxed prior  $\beta$  grains containing  $\beta$ -phase and fine  $\alpha$ -phase precipitates (100X, Etchant: 30 ml H<sub>2</sub>O<sub>2</sub> + 3 drops HF) (Ref. 19)

Figure 1.5.11 Various silicides (shown in white) that formed with Ti and Zr in the  $\beta$  matrix upon heat treatment of a 1.6 cm bar of Beta-C alloy (0.04 % Si) resulting in a mostly transgranular fracture mode: (a) heat treatment cycle, (b)

SEM photomicrograph (specimen electropolished in an electrolyte of 60 ml perchloric acid, 600 ml methanol, and 360 ml ethylene glycol; unetched) (Ref. 30)

Figure 1.5.12 Absence of silicides in the  $\beta$  matrix of the Beta-C alloy of 1.5.11 upon heat treatment above the silicide solvus (treatment similar to fast cooling of fusion zones of welds) resulting in a transgranular fracture mode: (a) heat treatment cycle, (b) SEM photomicrograph (specimen electropolished in an electrolyte of 60 ml perchloric acid, 600 ml methanol, and 360 ml ethylene glycol; unetched) (Ref. 30)

Figure 1.5.13 Grain boundary silicide precipitation (which led to a loss in ductility) of the Beta-C alloy of 1.5.12 after heat treatment above the silicide solvus followed by reannealing in the  $\beta$  + silicide field at 1144K (1600F) resulting in almost continuous precipitation of grain boundary silicides that reduce ductility and result in an intergranular fracture mode: (a) heat treatment, (b) SEM photomicrograph (specimen electropolished in an electrolyte of 60 ml perchloric acid, 600 ml methanol, and 360 ml ethylene glycol; unetched) (Ref. 30)

Figure 1.5.14 Effect of cyclic heat treatments on dissolution of silicides in Beta-C alloy of 1.5.12 with the resulting microstructure of a  $\beta$  matrix and grain boundary and interior silicides that resulted in a predominantly transgranular fracture mode: (a) heat treatment, (b) SEM photomicrograph (specimen electropolished in an electrolyte of 60 ml perchloric acid, 600 ml methanol, and 360 ml ethylene glycol; unetched) (Ref. 30)

Figure 1.5.15 TTT diagram for solution annealed and water-quenched Beta-C alloy indicating the effects of residual deformation on the kinetics of the  $\beta \rightarrow \alpha$  transformation (direct nucleation of  $\alpha$  from the  $\beta$  phase at higher aging temperatures may lead to inhomogeneous precipitation at the  $\beta$  grain boundaries and the formation of locally weak regions near these grain boundaries, while cold working prior to aging leads to  $\alpha$  phase nucleation through the dislocation cluster sites and more homogeneous precipitation patterns, thus improving mechanical properties) (Ref. 36)

Figure 1.5.16 Microstructure of Beta-C wire (plasma arc melted ingot heated and rolled from 120 to 85 mm diameter on a two high reversing mill, re-heated and hot rolled to 12 mm diameter coil, and finally processed to 7.4-8 mm diameter wire on a standard wire drawing line, finished by solution heat treatment and cold working per AMS 4957): (a) as cold worked exhibiting a single phase  $\beta$  structure a grain size of 20-50  $\mu\text{m}$ , (b) as cold worked and aged at 538C (1000F) for 6 h exhibiting a uniformly aged structure of fine  $\alpha$  phase precipitation within a  $\beta$  matrix (Ref. 32)

Figure 1.5.17 Microstructures of Beta-C alloy cold worked and heat treated: (a) solution treated-age (STA process: 1500F/15 min/AQ + 1050F/24 h/AQ) showing extensive grain boundary  $\alpha$  precipitation and light blotchy zones consisting of unaged  $\beta$  phase at an  $\beta/\alpha$  ratio = 46 vol. %, which can be thermally unstable and subject to continued aging above 350F and (b) preage-solution treated-age (PASTA process: 1150F/8 h/AQ + 1500/15 min/AQ + 1050F/24 h/AQ) showing an improved degree of  $\alpha$  phase precipitation reflected in an  $\alpha/\beta$  ratio = 71 vol. % and significantly improved mechanical properties (100 X, etch not given) (Ref. 34)

Figure 1.5.18 Optical microstructure of Beta-C after solution treatment and a conventional single step aging treatment of 540C/16 h (a) and after a duplex aging treatment of preage at 440C/4 h + 560C/16 h (b) with both heat treatments achieving a similar 0.2% yield strength of 1085 MPa (157.4 ksi); as opposed to the conventional aging, which resulted in precipitate free zones (PFZs) within the  $\beta$  grains that were identified as fatigue crack nucleation sites (see 1.5.19), while the two step aging resulted in a homogeneous precipitation of  $\alpha$  particles within the  $\beta$  grains and superior fatigue properties (see 3.5.1.2) (Ref. 35)

Figure 1.5.19 Fatigue crack initiation (675 MPa and R = -1) in single step aged (solution treated and aged at 540C/16 h) Beta-C (Ti-3Al-8V-4Mo-4Zr) alloy at a PFZ site (see 1.5.18 and 3.5.1.2) (Ref. 35)

Figure 1.5.20 Optical micrographs of Beta-C alloy cylindrical samples machined to 10 mm diameter from 170 mm diameter extruded bar and solution heat treated at 815C for 1 h and water quenched, the directly aged at (a) 480C for 28 h (peak aged), (b) 500C for 28 h, (c) 520C for 28 h, and (d) 480C for 24 h with different volume fractions of precipitate free zones or PFZs (vol-PFZ): a  $\approx$  7%, b  $\approx$  12%, c  $\approx$  35%, and d  $\approx$  12% (ground, polished, and etched in 100 ml H<sub>2</sub>O + 2 ml HNO<sub>3</sub> + 2 ml HF) (Ref. 36)

Figure 1.5.21 Beta-C alloy in 1.5.20 that was solution heat treated to the fully recrystallized condition (920C/30 min) and duplex aged at 440C/12 h + 500C/24 h, which led to improved fatigue resistance (see 3.5.1.2-3.5.1.4): (a) - optical micrograph illustrating complete and homogeneous distribution of the  $\alpha$  phase precipitates, (b) - SEM micrograph showing the arrangement of the  $\alpha$  phase close to the  $\beta$  grain boundaries and inside the  $\beta$  grains, and (c) and (d) - morphology and a difference in size of the  $\alpha$  phase observed in the high resolution SEM micrographs (see 1.5.22) (Ref. 36)

Figure 1.5.22 Solution annealing Beta-C at a hot working temperature of 920C led to complete recrystallization but also to  $\beta$ -grain growth as determined by optical microscopy and electron backscatter diffraction (see 1.5.21) (Ref. 36)

Table 1.5.23 Suggested metallographic techniques for polishing and etching Ti-38-6-44 titanium alloy (Ref. 52)

Table 1.5.24 Effect of solution anneal quench rate and specimen location in 6 in. diameter x 6 in. long samples on full section longitudinal room temperature tensile properties of 6 in. diameter billet of Ti-3Al-8V-6Cr-4Mo-4Zr in the solution annealed and solution annealed and aged condition (Ref. 1)

Figure 1.5.25 (a) Equiaxed microstructure of hot rolled Allvac Ti-38-644 14.3 mm coiled rod with dynamically recrystallized grains as fine as 20  $\mu\text{m}$  and (b) microstructure of the material that was cold drawn 20% showing grains becoming slightly elongated in the longitudinal or drawing direction, depicted by the arrow (Ref. 58)

Figure 1.5.26 Volume percentage of  $\alpha$ -phase as a function of aging time for metastable Beta titanium alloys 38-6-44 (Beta-C), Beta III, and 8-8-2-3 Ti plate products (all solution treated (ST) in air, Beta-C air cooled after ST, while Beta II and 8-8-2-3 Ti were all water quenched after ST) (Ref. 67)

Figure 1.5.27 Tensile strength as a function of the percentage  $\alpha$ -phase for the metastable Beta titanium alloys, including Beta-C, in the solution treated plus aged (STA) condition (compare with 1.5.26) (Ref. 67)

Figure 1.6.1 Relation of tensile strength of solution treated and aged Beta-C alloy in comparison to other  $\beta$  titanium alloys and Ti-6Al-4V ( $\alpha+\beta$  alloy) in relation to size (Ref. 26)

Figure 1.6.2 Hardness (HRC) values for Beta-C alloy samples machined from 19.05 mm rod: A1 - annealed at 800C/0.5 h/AQ + solution treated at 850C/1 h/WQ, AG455 - A1 + aged at 455C/24 h/AQ, AG495 - A1 + aged at 495C/24 h/AQ, and AG540 - A1 + aged at 540C/24 h/AQ (Ref. 38)

Figure 1.6.3 Hardness (HV) response of Beta-C alloy samples, machined to 10 mm diameter from 170 mm diameter extruded bar and solution heat treated at 815C for 1 h and water quenched, then aged between 460 and 560C for 20, 24, and 28 h (see 1.5.20) (Ref. 36)

Figure 1.6.4 Difference in microhardness (HV) between fully aged region and  $\beta$  fleck (PFZ) indicating huge disparity between the microhardness values in regions of inhomogenous  $\alpha$  phase precipitation (see 1.6.3 and 1.5.20) (Refs. 36, 37)

Table 1.6.5 Hardness (HRC) measurements for Beta-C alloy samples made from 41 mm hot rolled bar that was solution treated at 815C/1 h/AQ (ST), machined into two parallel sections 15 mm thick x 41 mm wide at midplane, and aged at 500C (ST/A) for 1-24 h, while some of these machined sections were cold worked (ST/CW) 20-65% and one of the samples that was cold rolled 20% was aged for 24 h at 500C (ST/CW/A) (each HRC value represents an average of 10 measurements) (Ref. 39)

Figure 1.6.6 Hardness (HVN) as a function of cold work ( $\phi = \ln h/h_0$ ,  $h_0$  - initial height and  $h$  - final height) of Beta-C samples machined from 160 mm diameter bar stock, solution treated at 927C/0.5 h/AQ, and cold rolled without aging, showing little work hardening, and after aging 4 h at 400 and 440C, showing a marked increase in hardness with increase in  $\phi$  (neither aging treatment increased hardness without cold working and therefore both aging treatments correspond to the position X in the schematic Beta-C TTT diagram in 1.6.7) (Ref. 40)

Figure 1.6.7 Schematic Beta-C TTT diagram corresponding to results shown in 1.6.6 (Ref. 40)

Figure 1.6.8 Hardness (HVN) profiles in V-notched ( $k_t = 2.5$ ) Beta-C samples deep rolled circumferentially on a hydraulic three-roll device at  $F = 800 \text{ N}$  ( $k_t = 2.5$ ) with samples originally machined from 160 mm diameter bar stock and solution treated at 927C/0.5 h/AQ and tested for hardness in the unaged and aged condition (microstructures shown in 1.6.9) (Ref. 40)

Figure 1.6.9 Notch root area of deep rolled Beta-C samples showing microstructural gradients that corresponded to hardness profiles in 1.6.8 and which led to improved fatigue performance (Ref. 40)

Figure 1.6.10 Hardness (HRC) data (average of five readings per point) used to construct the TTT diagram for Beta-C alloy shown in 1.5.7 with micrographs in 1.5.8 (samples were machined from 15.9 mm thick plate of RMI 38644 alloy that was solution treat at 1200K/0.5 h/AQ and aged at 373 to 973K in protective environments for times up to 1000 h and hardness reading were taken on a plane perpendicular to the rolling direction) (Ref. 29)

Figure 1.6.11 Rockwell hardness (HRC) as a function of aging time for Beta-C alloy after STA and PASTA heat treatments with microstructures shown in 1.5.17 (Ref. 34)

Figure 1.6.12 Dependence of diamond pyramid hardness (HV) on aging time at 570C for solution treated Ti-8823 or Ti-38V-15Si (ST - 900C/15 min/WQ) and Beta-C (ST - 925C/15 min/WQ) alloys (Ref. 65)

Figure 1.6.13 Average Rockwell Hardness (five readings) as a function of aging time for 3-inch (7.6 cm) diameter Beta-C bar for 900F (482C) and 950F (510C) age temperatures with given solution treatments resulting in un-recrystallized (UNRX) and recrystallized (RX) microstructures (Ref. 70)

Figure 1.6.14 Average Rockwell Hardness (five readings) as a function of aging time for 3-inch (7.6 cm) diameter Beta-C bar for 1000F (538C) and 1050F (566C) age temperatures with given solution treatments resulting un-recrystallized (UNRX) and recrystallized (RX) microstructures (Ref. 70)

Figure 1.6.15 Average Rockwell Hardness (five readings) as a function of aging time for 3-inch (7.6 cm) diameter Beta-C bar for 1100F (593C) and 1150F (620C) age temperatures with given solution treatments resulting un-recrystallized (UNRX) and recrystallized (RX) microstructures (Ref. 70)

Table 1.7.1 Product forms available for Beta-C and Beta-C/Pd alloys (Ref. 2)

Figure 1.7.2 Dynamet processes for production of a range of titanium alloy products including Ti-3Al-8V-6Cr-4Mo-4Zr products (ULTRABAR™ PRECISION BAR - a small diameter machining bar that offers Dynamet's best dimensional characteristics and bar-to-bar consistency) (Ref. 41)

Figure 1.7.3 A motorcycle spring (a) and a motorcycle mono-shock assembly (b) manufactured using Allvac Ti-38-644 cost competitive rod material (Ref. 58)

Figure 1.8.1 Comparison of processing routes for Beta-C bar produced using conventional double vacuum arc remelting (VAR) ingot and single melt plasma arc melting (PAM) ingot (Ref. 55)

Figure 1.9.1 Aging response of Ti-3-8-6-4-4 alloy with varying hydrogen content (Ref. 44)

Figure 1.9.2 Total H content measured by hot vacuum extraction as a function of applied cathodic current density ( $I_{app}$ ) to Ti-3Al-8V-6Cr-4Mo-4Zr specimens machined from hot rolled 4.1 cm round bar (precharged 64 h in a solution of 1000 ml H<sub>2</sub>O, 10 ml 36N H<sub>2</sub>SO<sub>4</sub>, and 0.8 g Na<sub>2</sub>P<sub>7</sub>O<sub>4</sub> at 90C): ST - solution treated for 1 h at 815C, STA - solution treated + 30 h at 500C, and CW - ST + cold worked to a 65% reduction in thickness in the longitudinal/transverse (LT) and transverse/longitudinal (TL) direction (lines shown are fits of the equation  $C_H = K I_{app}^{0.5}$ , where K is a constant) (Ref. 45)

Figure 1.9.3 Local fracture initiation stress as a function of total H content (see 1.9.2 for charging conditions) for circumferentially notched tensile specimens of Ti-3Al-8V-6Cr-4Mo-4Zr (0.417 cm diameter with a minimum notch root diameter of 0.241 cm and a notch root radius of 0.054 cm) machined with loading axis parallel to the long axis of the bar: (a) ST (1 h at 815C) for stroke rates from  $1 \times 10^{-6}$  to  $1 \times 10^{-3}$  cm/s, and (b) STA (ST + 30 h at 500C) for stroke rates ranging from  $2 \times 10^{-7}$  to  $1 \times 10^{-4}$  cm/s; beta lattice H content calculated from  $C_{H\beta} = 1.7 C_{H\alpha}$ . Fracture modes are indicated as MV = microvoid processes, F = flutting, IG = intergranular, C = cleavage, and I = process at or near  $\alpha/\beta$  interfaces (Ref. 45)

Figure 1.9.4 Local fracture initiation stress as a function of total H content (see 1.9.2 for charging conditions) for circumferentially notched tensile specimens of Ti-3Al-8V-6Cr-4Mo-4Zr (0.417 cm diameter with a minimum notch root diameter of 0.241 cm and a notch root radius of 0.054 cm) machined with loading axis parallel (L) and transverse (T) to the long axis of the bar with specimens ST + CW to 65% reduction in thickness, tested at a stroke rate of  $2 \times 10^{-5}$  cm/s (see 1.9.3 for comparison) (Ref. 45)

Figure 1.9.5 Local fracture initiation stress as a function of total H content (see 1.9.2 for charging conditions) for circumferentially notched tensile specimens of Ti-3Al-8V-6Cr-4Mo-4Zr (0.417 cm diameter with a minimum notch root diameter of 0.241 cm and a notch root radius of 0.054 cm) machined with loading axis parallel to the long axis of the bar with specimens: RSTU (resolutinized, heat treated and underaged) - ST (1 h at 815C) + 2 h at 830C + 1.5 h at 500C and RSTA (resolutionized, heat treated and peak aged) - ST + 2 h at 830C + 30 h at 500C, tested at a stroke rate of  $2 \times 10^{-5}$  cm/s (see 1.9.3 for ST and STA comparison) (Ref. 46)

Figure 1.9.6 Local fracture initiation stress as a function of total H content (see 1.9.2 for charging conditions) for circumferentially notched tensile specimens of Ti-3Al-8V-6Cr-4Mo-4Zr (0.417 cm diameter with a minimum notch root diameter of 0.241 cm and a notch root radius of 0.054 cm) machined with loading axis parallel to the long axis of the bar with specimens: CW -ST + 65% cold worked, ● - ST + 39% cold work + 0.5 h at 500C, ■ - ST + 12 h at 500C, and ▲ - ST + 100 h at 500C (see 1.9.3 for ST and STA comparison) (Ref. 46)

Figure 1.9.7 Variation in the irreversible hydrogen trapping constant ( $k_a$  - determined from apparent diffusivity of hydrogen in the metal) which correlates with hydrogen embrittlement (HE) for various solution treated and aged beta titanium alloys, including Beta-C in various aging condition (1.593 cm diameter Beta-C rod was hot rolled, solution treated, centerless ground, and aged at 510C for 5, 10, and 20 h and denoted as 5A, 10A, and 20A) with  $k_a$  determined by a potentiostatic pulse technique for hydrogen entry flux for each alloy in 1 mol  $l^{-1}$  acetic acid/1 mol  $l^{-1}$  sodium acetate + 15 ppm  $As_2O_3$  at room temperature (HE susceptibility measured by  $k_a$  was consistent with relative resistance to HE for these alloys) (Ref. 47)

Figure 1.9.8 Relationship of hydrogen charging pressure and hydrogen concentration in aged Beta-C alloy (as-received hot rolled, annealed, and centerless ground rod was aged 28 h at 755K and air cooled to obtain a microstructure of hcp  $\alpha$ -phase in a bcc  $\beta$ -phase matrix) for charging at 625K (a), with x-ray diffraction patterns from regions A, B, and C (b-d): for hydrogen charging, Pd coated samples (to facilitate hydrogen uptake) were exposed to constant hydrogen pressure at 623K for 8 h and cooled quickly, the Pd coating removed and hydrogen content measure by vacuum gas extraction and chromatographic analysis (N.B.: Figure 3(b) shows x-ray diffraction peaks from initial  $\beta$ - and  $\alpha$ -phases only, while 3(c) and 3(d) show the appearance of fcc  $\delta$ -hydride peaks in the concentration range  $15 < \text{total H concentration} < 38$  at. %, with hydride peaks increasing in intensity with increased H concentration and peaks from the  $\alpha$ -phase decreasing while that from the  $\beta$ -phase remains constant.) (see 1.9.9 for tensile results) (Ref. 48)

Figure 1.9.9 Strain to failure in smooth specimens (120 mm long, 6.4 mm diameter, and 8 mm gage length) of aged Beta-C alloy (as-received hot rolled, annealed, and centerless ground rod was aged 28 h at 755K and air quenched) measured by tensile elongation at 295K and  $10^{-5}$  cm/sec displacement rate as a function of hydrogen concentration (see 1.9.8) with open and solid points from two independently prepared series of specimens (see 1.9.8 for H charging conditions), showing a sharp ductile-brittle transition at a H concentration of  $\sim 3.5$  at. % (Ref. 48)

Table 2.1.1.1 Melting range of Beta C alloy (Ti-3.3-Al-8.1V-5.9Cr-4.2Mo-3.9Zr) determined on a 0.375 in. cube of the alloy, resistance heated on a Nb plate in a dry Ar atmosphere, with temperature measured by optical pyrometry by optical pyrometry (Ref. 50)

Table 2.1.2.1 Precipitate phase sequence in RMI 38644 alloy 15.9 mm thick plate, solution treatment at 1200K for 0.5 h, air cooled, and aged at 973K for up to 1,000 h (see 1.5.7) (Ref. 29)

Figure 2.1.2.2 Equilibrium phase fields of Beta-C alloy (see 1.5.11-1.5.14) (Ref. 30)

Figure 2.1.3.1 Thermal conductivity of ATI 38-644 (UNS R58640) titanium alloy in the as rolled and solution treated and aged (STA) condition as a function of elevated temperature (Ref. 3)

Figure 2.1.3.2 Thermal conductivity of Beta-C as a function of elevated temperature compared with that of CP (Commercially Pure) Ti (Ref. 57)

Figure 2.1.4.1 Mean coefficient of linear expansion as a function of temperature of 6 in. square billet of Ti-3Al-8V-6Cr-4Mo-4Zr in the solution annealed and aged condition (Ref. 1)

Table 2.1.4.2 Coefficient of thermal expansion of Beta-C and Beta-C/Pd over the temperature range 32-1,000F (Ref. 2)

Figure 2.2.2.1 Electrical resistivity of Beta-C compared with Grade 2 Ti as a function of temperature (Ref. 57)

Figure 2.3.1.1 Yield strength and corrosion resistance at 315C (600F) of Beta-C compared with other conventional titanium alloys highlighting the excellent corrosion resistance of Beta-C in combination with its excellent elevated temperature strength (note: Beta-C could age in service at temperatures greater than 350C (660F) and therefore is not recommended for use above this temperature) (Ref. 57)

Table 2.3.1.2 General corrosion of Ti-3-8-6-4-4 (Beta-C) alloy in specific media (Ref. 63)

Figure 2.3.1.3 General corrosion in naturally aerated boiling HCl solutions of annealed Ti-3-8-6-4-4 (Beta-C) compared with other titanium alloys (Ref. 63)

Figure 2.3.1.4 General corrosion in naturally aerated boiling HCl solutions of aged Beta-C and Beta-C/Pd compared with other titanium alloys (Ref. 42)

Table 2.3.1.5 Active-to-passive transition acid concentrations for Beta-C, Beta-C/Pd, and other titanium alloys in boiling HCl at which each alloy exhibits a transition from active-to-passive corrosion transition acid concentration assumed to correspond to 0.13 mm/y corrosion rate (e.g., see 2.3.1.3 and 2.3.1.4), below which useful resistance can be expected (acid resistance increases toward the alloys at the bottom) (Ref. 42)

Table 2.3.1.6 Corrosion resistance of Beta-C and Beta-C/Pd compared to other titanium alloys in boiling acid media: mildly reducing aerated HCl, mildly oxidizing FeCl<sub>3</sub> solution, and strongly oxidizing acid HNO<sub>3</sub> solution (Ref. 42)

Table 2.3.1.7 Pitting resistance of annealed Beta-C and Beta-C/Pd compared to other annealed titanium alloys as measured by repassivation potential (lowest potential at which anodic pitting is possible in boiling 5% NaCl solution at pH 3.5 using the galvanostatic method at a constant current density of 200 mA/cm<sup>3</sup> (repassivation potentials above 1 volt imply the alloy can fully resist spontaneous chloride pitting in hot brine service) (Ref. 42)

Figure 2.3.1.8 Approximate temperature limits for crevice corrosion resistance of beta titanium alloys in chloride brines: (1) sweet brines, seawater, pH 3-12; (2) sour brines, pH 3-5; (3) acidic brines, pH 1-2; (4) oxidizing acidic brines, pH 0-2; Group A - Ti Grade 2, Ti-6-4, Ti-15-3-3-3, Ti-13-11-3, Ti-10-2-3; Group B - Beta-C, Ti-8-8-2-3; Group C - Beta-C/Pd, Beta-21S, Ti-15-5, Beta III (Ref. 42)

Table 2.3.1.9 Crevice corrosion of Beta-C compared to CP Ti and Ti Grade 12 measured in 250 g/l NaCl, 15 psi H<sub>2</sub>S, 1 g/l sulfur (pH 3.0) after 720 h at temperatures given (Ref. 57)

Table 2.3.1.10 Ranking of titanium alloy crevice corrosion resistance in NaCl brines at temperatures  $\geq 195\text{F}$  (90C) (Ref. 63)

Figure 2.3.1.11 Comparison of the corrosion of Beta-C and CP Ti in boiling sulfuric acid (a), sulfuric acid inhibited by FeCl<sub>3</sub> (b), hydrochloric acid (c), and nitric acid (d) at concentrations indicated (Ref. 57)

Figure 2.3.1.12 Comparative corrosion rates in boiling HCl solutions of Beta-C, Beta-C/0.045Pd, and Beta-C/0.086Ru sheet (1.8 mm thick) in the ST (816C/30 min/AC at YS of ~827 MPa) and STA (ST + 579C/24 h/AC at YS of ~965 MPa) conditions (Ref. 64)

Figure 2.3.1.13 Effect of acid concentration on corrosion rate of Beta-C and Beta-C/0.045Pd sheet (1.8 mm thick) in the STA condition in 4.5% Na<sub>2</sub>MoO<sub>4</sub>-inhibited HCl solutions at 204C (see 2.3.1.12) (Ref. 64)

Figure 2.3.1.14 Corrosion rate and hydrogen absorption as a function of exposure time in high temperature 20% NaOH solutions of Beta-C, Beta-/0.045Pd, and Beta-C/0.086Ru sheet (1.8 mm thick) in the STA condition STA (816C/30 min/AC + 579C/24 h/AC at YS of ~965 MPa) (Ref. 64)

Figure 2.3.1.15 Absorption of hydrogen as a function of exposure time for Beta-C and Beta-C/0.045Pd alloys in dilute boiling HCl media suggesting that Pd enhancement dramatically improves Beta-C's resistance to hydrogen absorption (Ref. 64)

Table 2.3.1.16 Corrosion potential of standard and enhanced Beta-C alloys in acidic 20% NaCl brine at 90C (volts vs. Ag/AgCl) indicating the more noble (positive) corrosion potentials of the enhanced alloys, especially Beta-C/0.045Pd (these values correlate with alloy resistance shown in 2.3.1.12) (Ref. 64)

Table 2.3.1.17 Full size field tests of downhole tubulars made of Beta-C titanium alloys for the Salton Sea geothermal field, located in the Imperial Valley of California, with geothermal fluids characterized by hypersaline brines containing H<sub>2</sub>S and CO<sub>2</sub> and temperature of 450-600F (Ref. 68)

Figure 2.3.2.1 Effect of test environment on resistance to stress corrosion cracking of

Table 2.3.2.2 Effect of aging temperature on stress corrosion cracking in a 3.5% NaCl distilled water solution of 1.25 in. thick Ti-3Al-8V-6Cr-4Mo-4Zr plate solution annealed at 1,700F for 1 h and air cooled, then aged for 8 h at aging temperatures indicated (Ref. 1)

Figure 2.3.2.3 Yield strength dependence of the lower bound and standardized air fracture toughness compared with the threshold stress intensity factor for side grooved compact tension (CT) specimens\* machined from solution treated (815C/1 h/AC) plus aged (STA) Beta-C bar (plane perpendicular to the longitudinal axis of the bar with crack growth in the radial direction) for experiments conducted at 25C in either moist air or 0.6 M (3.5 wt.%) aqueous NaCl (pH = 5-6) solution at a fixed electrode potential of -150 mV SCE (aqueous NaCl fracture modes for each range of  $\sigma_{YS}$  are indicated) (Ref. 39)

Figure Note on Figure: TG - Transgranular, MVC - Microvoid nucleation, growth and coalescence.

Figure 2.3.2.4 Air fracture toughness and threshold stress as a function of yield strength and cold work reduction for side grooved compact tension (CT) specimens\* machined from solution treated (815C/1 h/AC) plus 20-65% cold rolled (ST/CW) Beta-C bar sections for experiments conducted at 25C in either moist air or 0.6 M (3.5 wt.%) aqueous NaCl (pH = 5-6) solution at a fixed electrode potential of -150 mV SCE (the fracture mode for both air and aqueous NaCl exposure is indicated and the dashed trend lines for the STA data in 2.3.2.3 are shown) (Ref. 39)

Figure 2.3.2.5 Rapid crack growth rate ( $da/dt$ ) as a function of applied stress intensity factor ( $K$ ) for side grooved CT specimens\* machined from solution treated (815C/1 h/AC) plus aged (STA) and ST/CW/A Beta-C bar sections in aqueous NaCl solution with crack growth rates calculated from direct current potential difference based crack length (a) versus time (t) (see also 2.3.2.3 and 2.3.2.4) (Ref. 39)

Figure \*The ST/CW and ST/CW/A specimens for 2.3.2.3, 2.3.2.4, and 2.3.2.5 were in the T-L orientation with the crack plane perpendicular to the rolling plane and crack growth parallel to the cold rolling direction (the original longitudinal axis). Additional ST/CW (65%) Beta-C specimens were in the LT orientation (crack growth perpendicular to the rolling direction). All specimens except ST/CW had a net thickness ( $B_n$ ) of 5.1 mm and width ( $W$ ) of 30 mm, except that the ST/CW (65%) specimens had reduced thickness with  $B_n = 3.8$  mm. The CT specimens were fatigue pre-cracked in moist air, terminating with a  $K_{max} = 15$  MPa $\sqrt{m}$  and a crack length-to-width ratio of 0.53. (Ref. 39)

Figure 2.3.2.6 Temperature thresholds for stress corrosion cracking (SCC) of Beta-C and Beta-C/Pd in different conditions (including the proprietary PASTA heat treatment described in 1.5.17, 3.2.1.4, 3.2.1.5) for a worst case deep gas well sour brine based on the C-ring and slow strain rate test (SSRT) for SCC susceptibility (see 2.3.2.8) (Ref. 42)

Table 2.3.2.7 Fracture toughness of Beta-C plate in the solution treated and aged (STA) condition in air and 3.5% NaCl solution at 25C (Ref. 57)

Figure 2.3.2.8 Temperature thresholds for stress corrosion cracking (SCC) of Beta-C and Beta-C/Pd in sour gas for solution treated and aged (STA) or the proprietary PASTA solution treated and aged conditions based on slow strain rate testing (see 2.3.2.6) (Ref. 57)

Table 2.3.2.9 Results of exposure of C-ring specimens of Beta-C, solution treated and aged (STA) to a yield strength of 175 ksi, to chloride + H<sub>2</sub>S + CO<sub>2</sub> mixtures (Ref. 57)

Figure 2.3.2.10 Effect of HCL and NaNO<sub>3</sub> on  $K_{ISCC}$  of Beta-C in methanol (Note:all titanium alloys including Beta-C are susceptible to SCC in methanol and increasing the halide content decreases time to failure.) (Ref. 57)

Figure 2.3.2.11 Room temperature fatigue crack growth ( $da/dN$ ) as a function of stress intensity ( $\Delta K$ ) measured in compact tensile (CT) specimens oriented in the L-T direction for as-received hot rolled (3.5 mm thick) Ti-3Al-8V-6Cr-4Mo-4Zr plate for as solution heat treated (SHT) material with tensile properties given in 3.2.1.40 (no noticeable acceleration of crack growth rate occurred in going from air to a salt water environment or when frequency was reduced so data are presented as a single scatterband) (Ref. 71)

Figure 2.3.2.12 Room temperature fatigue crack growth ( $da/dN$ ) as a function of stress intensity ( $\Delta K$ ) measured in compact tensile (CT) specimens oriented in the L-T direction for as-received hot rolled (3.5 mm thick) Ti-3Al-8V-6Cr-4Mo-4Zr plate for as-received hot rolled (3.5 mm thick) Ti-3Al-8V-6Cr-4Mo-4Zr plate solution treated (ST) at 800C (LO - 45  $\mu$ m grain size and 20 vol % unrecrystallized  $\beta$  grains) and aged in an Ar atmosphere after LO-ST in a single step (Simplex) and in a two-step (Duplex) process, with tensile properties given in 3.2.1.40 (no noticeable acceleration of crack growth rate occurred in going from air to a salt water environment or when frequency was reduced so data are presented as a single scatterband) (compare with 2.3.2.13) (Ref. 71)

Figure 2.3.2.13 Room temperature fatigue crack growth ( $da/dN$ ) as a function of stress intensity ( $\Delta K$ ) measured in compact tensile (CT) specimens oriented in the L-T direction for as-received hot rolled (3.5 mm thick) Ti-3Al-8V-6Cr-4Mo-4Zr plate solution treated (ST) at 927C (HI - 160  $\mu$ m grain size and fully recrystallized) and aged in an Ar atmosphere after HI ST in a single step (Simplex) and in a two-step (Duplex) process, with tensile properties given in 3.2.1.40 (no noticeable acceleration of crack growth rate occurred in going from air to a salt water environment or when frequency was reduced so data are presented as a single scatterband) (compare with 2.3.1.12) (Ref. 71)

Table 3.1.1 AMS specified minimum (single values) or range of room temperature tensile properties of Grade 19 and 20 alloys in different product forms and conditions (Refs. 5-8)

Table 3.1.2 ASTM specified minimum (single values) or range of room temperature tensile properties of Grade 19 and 20 alloys in different product forms and conditions (Refs. 9-14)

Table 3.2.1.1 Room temperature tensile properties of solution treated and aged (STA) Ti-3Al-8V-6Cr-4Mo-4Zr (Beta-C) in various product forms (Ref. 2)

Table 3.2.1.2 Room temperature tensile properties of Ti-3Al-8V-6Cr-4Mo-4Zr (Beta-C) wire in various conditions (Ref. 2)

Table 3.2.1.3 Effect of solution annealing and aging temperatures on room temperature tensile properties of Beta-C bar (Ref. 20)

Table 3.2.1.4 Tensile properties of solution treated and aged (STA) and preaged and solution treated and aged (PASTA) Beta-C cold pilgered pipe (see 1.5.17) (Ref. 34)

Table 3.2.1.5 Tensile properties of solution treated and aged (STA) and preaged and solution treated and aged (PASTA) Beta-C/Pd cold pilgered pipe (see 3.2.1.4) (Ref. 34)

Figure 3.2.1.6 Room temperature tensile properties of 0.373 in. diameter rod of ATI 38-644 (UNS R58640) titanium alloy after solution treatment, cold drawing, and aging (900F/6 h/air cool) as a function of cold drawing percentage (Ref. 3)

Table 3.2.1.7 Room temperature tensile test results performed on Ti 38-644 (Beta-C) alloy bars extruded at 920C from a 170 mm diameter ingot after solution treatment and single-step aging (A) and duplex aging treatments involving pre-aging (PA) and final aging (FA), and tensile fracture surfaces exhibiting crack propagation as intergranular in the single-step aged material and transgranular in the duplex aged material (see 1.5.20 and 1.5.21 for metallographs, 3.2.1.8 for SEM fractographs, and 3.5.1.4 for fatigue results) (Ref. 36)

Figure 3.2.1.8 SEM fractographs of tensile samples of Ti 38-644 (Beta-C) alloy: (a) simplex aged (ST at 815C/1 h + A at 480C/28 h) exhibiting intergranular fracture and (b) duplex aged (ST at 920C/30 min. + PA at 440C/12 h + FA at 500C/24 h) exhibiting transgranular fracture (see 3.2.1.7 for tensile test results and 3.5.1.4 for fatigue results) (Ref. 36)

Figure 3.2.1.9 Room temperature tensile stress-strain curves for Ti Beta-C tensile specimens machined from 19.05 mm diameter rod, annealed at 800C/0.5 h/AQ + solution treated at 850C/1 h/WQ (BC Annealed), and aged at 455C/24 h/AQ (BC AG 455), 495C/24 h/AQ (BC AG 495), and 540C/24 h/AQ (BC AG 540) (Ref. 38)

Figure 3.2.1.10 Room temperature ultimate tensile strength of Ti Beta-C tensile specimens machined from 19.05 mm diameter rod, annealed at 800C/0.5 h/AQ + solution treated at 850C/1 h/WQ (Annealed), and aged at 455C/24 h/AQ (AG 455), 495C/24 h/AQ (AG 495), and 540C/24 h/AQ (AG 540) (see 3.2.1.8) (Ref. 38)

Table 3.2.1.11 Room temperature tensile properties of Ti Beta-C alloy hot rolled and solution annealed sheet (0.065 in. thick, 1600F/30 min/AQ) in the longitudinal (L) and transverse (T) directions as a function of aging treatment (Ref. 50)

Figure 3.2.1.12 Longitudinal Ti-3Al-8V-6Cr-4Mo-4Zr stress-strain curves at room and elevated temperatures for the solution treated plus aged (STA) condition (Ref. 1)

Figure 3.2.1.13 Transverse Ti-3Al-8V-6Cr-4Mo-4Zr stress-strain curves at room and elevated temperatures for the solution treated plus aged (STA) condition (Ref. 1)

Figure 3.2.1.14 Effect of annealing temperature on room temperature tensile properties of 0.312 in. Ti-3Al-8V-6Cr-4Mo-4Zr round cold drawn fastener stock (Ref. 1)

Figure 3.2.1.15 Effect of cold work on room temperature tensile properties of Ti-3Al-8V-6Cr-4Mo-4Zr fastener stock in two solution annealed conditions (Ref. 1)

Figure 3.2.1.16 Effect of aging on room temperature mechanical properties of Ti-3Al-8V-6Cr-4Mo-4Zr fastener stock cold drawn various percentages (Ref. 1)

Figure 3.2.1.17 Effect of aging temperature on room temperature tensile properties of cold worked Ti-3Al-8V-6Cr-4Mo-4Zr tubing (Ref. 1)

Figure 3.2.1.18 Effect of annealing temperature on room temperature tensile properties of cold worked Ti-3Al-8V-6Cr-4Mo-4Zr tubing (Ref. 1)

Figure 3.2.1.19 Effect of aging temperature on the room temperature tensile properties of annealed Ti-3Al-8V-6Cr-4Mo-4Zr tubing (Ref. 1)

Table 3.2.1.20 Room temperature tensile properties of as-received hot rolled (3.5 mm thick) Ti-3Al-8V-6Cr-4Mo-4Zr plate solution treated (ST) at 800C (LO - 45  $\mu$ m grain size and 20 vol % unrecrystallized  $\beta$  grains) and at 927C (HI - 160  $\mu$ m grain size and fully recrystallized) and aged in an Ar atmosphere after LO and HI ST in a single step (Simplex) and in a two-step (Duplex) process (Ref. 71)



Figure 3.2.1.21 Effect of aging temperature on room temperature longitudinal and transverse tensile properties of 1.25 in. thick Ti-3Al-8V-6Cr-4Mo-4Zr hot rolled plate in the solution annealed condition (Ref. 1)

Figure 3.2.1.22 Effect of annealing temperature on tensile properties of 1 in. thick Ti-3Al-8V-6Cr-4Mo-4Zr hot rolled plate (Ref. 1)

Figure 3.2.1.23 Effect of time at various aging temperatures on the room temperature tensile properties of an 0.625 Ti-3Al-8V-6Cr-4Mo-4Zr pancake forging in the solution annealed condition (Ref. 1)

Figure 3.2.1.24 Effect of various aging treatments on the room temperature mechanical properties of cold drawn (CD) Ti-3Al-8V-6Cr-4Mo-4Zr fastener bar (Ref. 1)

Table 3.2.1.25 Room temperature tensile properties of hot rolled Beta-C alloy sheet (0.065 in. thick) with different processing cycles (Ref. 50)

Table 3.2.1.26 Room temperature tensile properties of induction melted and investment cast Ti-38644 test bars (six 0.250 in. diameter test bars bottom poured from in a cluster in molds coated with pyrolytic graphite), heat treated at 1350F/15 min/rapid cool + 1100F/8 h/rapid cool, and sand blasted before testing (see 3.3.1.7 for tensile properties at 600F) (Ref. 51)

Figure 3.2.1.27 Room temperature tensile properties of induction melted and investment cast test bars of various titanium alloys including Ti-38644 (six 0.250 in. diameter test bars bottom poured from in a cluster in molds coated with pyrolytic graphite) in the solution treated and aged condition after sand blasting before testing (see 3.3.1.8 for tensile properties at 600F) (Ref. 51)

Table 3.2.1.28 Tensile evaluation of 15 mm diameter Beta-C bar hot rolled from a 127 mm diameter plasma arc melted (PAM) ingot (see 1.8.1) and after heat treatment by direct aging and solution treatment + aging (STA) compared with STA properties of vacuum arc remelted (VAR) Beta-C bar (Ref. 55)

Table 3.2.1.29 Effect of various heat treatments on the thermal stability of room temperature tensile and shear properties of 0.312 in. diameter Ti-3Al-8V-6Cr-4Mo-4Zr bar after exposure to 550F for 500 h with no applied stress (Ref. 1)

Table 3.2.1.30 Effect of thermal exposure on room temperature tensile properties of various size Ti-3Al-8V-6Cr-4Mo-4Zr tubing, solution annealed at 1500F/1.5 h/vacuum cooled after exposure to 500F for 100 h with no applied stress (Ref. 1)

Table 3.2.1.31 Effect of aging at 950 and 1,000F for 8 h and air cooling on the tensile properties of 0.003 in. foil of various titanium beta alloys after solution treatment at 1,500F for 15 min. and furnace cooling (Ref. 1)

Figure 3.2.1.32 Effect of thermal exposure treatments prior to aging on the room temperature tensile properties of titanium beta alloys in foil form (Ref. 1)

Table 3.2.1.33 Average tensile test data for direct aging of cold drawn Ti-38-644 rod produced from continuous hot rolled coils (see 3.2.1.34 for comparison of tensile data for rod in STA condition) (Ref. 58)

Table 3.2.1.34 Average tensile test data for Ti-38-644 rod solution treated at 760C and aged at 510C (see 3.2.1.33 for comparison of direct aged cold drawn Ti-38-644 rod) (Ref. 58)

Figure 3.2.1.35 Room temperature yield tensile strength of RMI 38-6-44 sheet (0.5 cm thick) as a function of aging time at 538C for material solution treated at 816C for 1 h in a vacuum furnace, argon quenched, and cold worked 40 and 80% (Ref. 61)

Figure 3.2.1.36 Room temperature ultimate tensile strength of RMI 38-6-44 sheet (0.5 cm thick) as a function of aging time at 538C for material solution treated at 816C for 1 h in a vacuum furnace, argon quenched, and cold worked 40 and 80% (Ref. 61)

Figure 3.2.1.37 Room temperature tensile elongation of RMI 38-6-44 sheet (0.5 cm thick) as a function of aging time at 538C for material solution treated at 816C for 1 h in a vacuum furnace, argon quenched, and cold worked 40 and 80% (Ref. 61)

Figure 3.2.1.38 Room temperature longitudinal (a) and transverse (b) tensile strength and ductility for hot rolled 3-inch (7.6 cm) diameter Beta-C bar solution treated at 1450F (787C), which resulted in an unrecrystallized (UNRX)

grain structure, as a function of aging temperature (see 1.6.14-1.6.15 and 3.2.1.39) (Ref. 70)

Figure 3.2.1.39 Room temperature longitudinal (a) and transverse (b) tensile strength and ductility for hot rolled 3-inch (7.6 cm) diameter Beta-C bar solution treated at 1550F (842C), which resulted in a recrystallized (RX) grain structure, as a function of aging temperature (see 1.6.14-1.6.12.1.38) (Ref. 70)

Figure 3.2.2.1 (see Figure 3.3.2.1) Typical Ti-3Al-8V-6Cr-4Mo-4Zr longitudinal compressive stress-strain curves at room and elevated temperature for the solution treated plus aged (STA) condition (Ref. 1)

Figure 3.2.2.2 (see Figure 3.3.2.2) Typical Ti-3Al-8V-6Cr-4Mo-4Zr transverse compressive stress-strain curves at room and elevated temperature for the solution treated plus aged (STA) condition (Ref. 1)

Table 3.2.2.3 Room temperature compressive properties of Ti Beta-C alloy hot rolled and solution annealed sheet (0.065 in. thick, 1600F/30 min/AQ) in the longitudinal (L) and transverse (T) directions as a function of aging treatment (Ref. 50)

Table 3.2.2.4 Room temperature compressive properties of hot rolled Beta-C alloy sheet (0.065 in. thick) with different processing cycles (Ref. 50)

Table 3.2.3.1 Effect of solution treatment conditions on the dynamic fracture toughness ( $K_{Ic}$ ) and average grain size ( $L_{avg}$ ) of RMI 38644 plate (1.59 cm thick) for conventional unprecracked Charpy V-notch specimens determined in the L-S direction at an initial impact velocity of 130 in./s (Fracture Load -  $P_F$  and General Yield Load -  $P_{GY}$ ) (Ref. 66)

Table 3.2.5.1 Effect of solution treatment temperature on room temperature torsion, modulus, longitudinal and 45° tensile properties of fine grained Ti-38-6-44 bar (Refs. 1, 52)

Table 3.2.5.2 Effect of aging treatment after production annealing at 1,700F/30 min./AC followed by laboratory annealing at 1,500/30 min./AC on torsion and tension properties of 2.625 in. diameter Ti-3Al-8V-6Cr-4Mo-4Zr bar (Ref. 1)

Table 3.2.5.3 Tensile and double shear test results for Beta-C 3/8-24 bolts in the solution treated (1,500F) plus aged (950F) condition (Ref. 52)

Figure 3.2.5.4 Double shear strength of Beta-C bar (0.6 in. diameter) produced from double vacuum arc remelted (VAR) and the more economical single melt plasma arc melted (PAM) ingots (see 1.8.1 and 3.2.1.28) (Ref. 55)

Table 3.2.5.5 Double shear strength of direct aged Beta-C and Ti-185 wire both produced from single melt plasma arc melted (PAM) ingot hot rolled from 120 to 35 mm diameter on a two-high reversing mill, reheated and further processed to 12 mm hot rolled coil, then further processed to by wire drawing to finished size ranging from 7.4 to 8 mm diameter (Beta-C was finished in as cold worked condition per AMS 4957 and Ti-185 was finished in the annealed condition) (Ref. 32)

Table 3.2.5.6 Double shear strength as a function of ultimate tensile strength for Beta-C and candidate titanium alloys for high strength fastener applications (Ref. 60)

Table 3.2.5.7 Room temperature shear strength of Beta-C alloy fastener stock (0.312 in diameter) heat treated as indicated and before exposure to high temperature (see 3.2.5.8 for data after exposure) (Ref. 57)

Table 3.2.5.8 Room temperature shear strength of Beta-C alloy fastener stock (0.312 in diameter) heat treated as indicated and after exposure to to 285C (550F) temperature for 500 h (see 3.2.5.7 for pre-exposure data) (Ref. 57)

Table 3.2.7.1.1 Notch strength ( $K_t = 3.5$ ) of Beta-C and Ti-185 wire both produced from single melt plasma arc melted (PAM) ingot hot rolled from 120 to 35 mm diameter on a two-high reversing mill, reheated and further processed to 12 mm hot rolled coil, then further processed to by wire drawing to finished size ranging from 7.4 to 8 mm diameter (Beta-C was finished in as cold worked condition per AMS 4957 and Ti-185 was finished in the annealed condition) (Ref. 32)

Table 3.2.7.1.2 Comparison of standard room temperature tensile and notched tensile data for Beta-C and candidate titanium alloys for high strength fastener applications (Ref. 60)

Table 3.2.7.2.1 Plane strain fracture toughness properties at room temperature of Ti-3Al-8V-6Cr-4Mo-4Zr for specimens solution annealed at 1500F/15 min/AC plus aged at 1050F/12 h/AC (Ref. 1)

Figure 3.2.7.2.2 Fracture toughness of Ti Beta-C at room temperature as a function of yield strength compared with other titanium alloys (Ref. 2)

Figure 3.2.7.2.3 Effect of solution treatment temperature on room temperature strength-toughness (a) and strength-ductility (b) trends of extruded and heat treated Beta-C alloy showing that the best combination of strength and fracture toughness is achieved using a relatively high temperature beta solution heat treatment at 1700F (strength on horizontal axis is given in ksi) (Ref. 20)

Table 3.2.7.2.4 Room temperature fracture toughness of specimens taken from a 75 mm (3 in.) heat treated Beta-C bar (Ref. 57)

Table 3.2.7.2.5 Room temperature fracture toughness of heat treated Beta-C billet, forging, and plate (Ref. 57)

Table 3.2.7.2.6 Room temperature fracture toughness ( $K_{Ic}$ ) measured on duplicate compact tension specimens machined from hot rolled 3-inch (7.6 cm) diameter Beta-C bar in the solution treated and aged (STA) condition and oriented in C-R and R-L directions (ST at 1450F and 1550F resulted in an unrecrystallized (UNRX) and recrystallized (RX) grain structure respectively) (see 3.3.3.1) (Ref. 70)

Figure 3.2.7.2.7 Values of room temperature J-integral ( $J_{0.2}$  for STA material or upper and lower bounds for ST material) for as-received hot rolled (3.5 mm thick) Ti-3Al-8V-6Cr-4Mo-4Zr plate solution treated (ST) at 800C (LO - 45  $\mu$ m grain size and 20 vol % unrecrystallized  $\beta$  grains) and at 927C (HI - 160  $\mu$ m grain size and fully recrystallized) and aged (STA) in an Ar atmosphere after LO and HI ST in a single step (Simplex) and in a two-step (Duplex) process (tensile properties given in 3.2.1.40) (Ref. 71)

Table 3.2.7.2.8 Room temperature fracture toughness ( $K_I$  calculated from  $J_{0.2}$  per ASTM E813), tearing modulus ( $T = dJ/da$  from 3.2.7.2.7), and plastic zone size ( $r_p$ ) results from compact tensile (CT) specimens oriented in the L-T direction for as-received hot rolled (3.5 mm thick) Ti-3Al-8V-6Cr-4Mo-4Zr plate solution treated (ST) at 800C (LO - 45  $\mu$ m grain size and 20 vol % unrecrystallized  $\beta$  grains) and at 927C (HI - 160  $\mu$ m grain size and fully recrystallized) and aged (STA) in an Ar atmosphere after LO and HI ST in a single step (Simplex) and in a two-step (Duplex) process (tensile properties given in 3.2.1.40 and  $J_{0.2}$  values in 3.2.7.2.7) (Ref. 71)

Figure 3.2.7.2.9 Fracture toughness ( $J_{0.2}$ ) results as a function of yield stress measured in compact tensile (CT) specimens oriented in the L-T direction for as-received hot rolled (3.5 mm thick) Ti-3Al-8V-6Cr-4Mo-4Zr plate solution treated (ST) at 800C (LO - 45  $\mu$ m grain size and 20 vol % unrecrystallized  $\beta$  grains) and at 927C (HI - 160  $\mu$ m grain size and fully recrystallized) and aged (STA) in an Ar atmosphere after LO and HI ST in a single step (Simplex) and in a two-step (Duplex) process (tensile properties given in 3.2.1.40 and  $J_{0.2}$  values in 3.2.7.2.7) (Ref. 71)

Figure 3.3.1.1 Yield strength/density ratio as a function of temperature for Ti Beta-C and competitive metals and alloys (Ref. 2)

Table 3.3.1.2 Typical elevated temperature properties of grade 19 (Ti Beta-C) and 20 (Ti Beta-C/Pd) alloys in the solution treated (ST) and solution treated and aged (STA) conditions (Ref. 2)

Figure 3.3.1.3 Elevated tensile properties of 0.500 in. diameter bar of ATI 38-644 beta Ti alloy (UNS R58640) in the STA condition: solution treated (1,500F) and aged (1,050F/6 h/air cool) (Ref. 3)

Table 3.3.1.4 Short-time tensile test results at 600F of Ti Beta-C alloy hot rolled and solution annealed sheet (0.065 in. thick, 1600F/30 min/AQ) tested in the longitudinal direction as a function of aging treatment (Ref. 50)

Table 3.3.1.5 Tensile test results at -65F of hot rolled Beta-C alloy sheet (0.065 in. thick) with different processing cycles (Ref. 50)

Table 3.3.1.6 Tensile test results at 600F of hot rolled Beta-C alloy sheet (0.065 in. thick) with different processing cycles (Ref. 50)

Table 3.3.1.7 Tensile properties at 600F of induction melted and investment cast Ti-38644 test bars (six 0.250 in. diameter test bars bottom poured from in a cluster in molds coated with pyrolytic graphite), heat treated at 1350F/15 min/rapid cool + 1100F/8 h/rapid cool, and sand blasted before testing (see 3.2.1.26 for room temperature tensile properties) (Ref. 51)

Figure 3.3.1.8 Tensile properties at 600F of induction melted and investment cast test bars of various titanium alloys including Ti-38644 (six 0.250 in. diameter test bars bottom poured from in a cluster in molds coated with pyrolytic graphite) in the solution treated and aged condition after sand blasting before testing (see 3.3.1.27 for room temperature tensile properties) (Ref. 51)

Figure 3.3.1.9 Effect of test temperature on tensile properties of Ti-3Al-8V-6Cr-4Mo-4Zr bar in the solution annealed condition (Ref. 1)

Figure 3.3.1.10 Effect of test temperature on tensile properties of Ti-3Al-8V-6Cr-4Mo-4Zr bar in the solution annealed plus aged condition (Ref. 1)

Figure 3.3.1.11 Effect of test temperature on tensile properties of Ti-3Al-8V-6Cr-4Mo-4Zr billet in the solution annealed plus aged condition (Ref. 1)

Figure 3.3.1.12 Flow stress-strain rate data for Ti-3Al-8V-6Cr-4Mo-4Zr alloy tensile specimens that were initially strained 10-25% at a crosshead strain rate of  $\sim 5 \times 10^3 \text{ s}^{-1}$  to generate a sub-grain microstructure and then tested at increasing Instron crosshead speeds from  $5 \times 10^{-5}$  to  $10^{-3} \text{ s}^{-1}$  at temperatures indicated with Ar gas passed through furnace enclosing specimens during test to assure an inert atmosphere (see 3.3.1.13) (Ref. 69)

Table 3.3.1.13 Elongation to fracture data for Ti-3Al-8V-6Cr-4Mo-4Zr alloy tensile specimens that were initially strained 10-25% at a crosshead strain rate of  $\sim 5 \times 10^3 \text{ s}^{-1}$  to generate a sub-grain microstructure and then tested at increasing Instron crosshead speeds from  $5 \times 10^{-5}$  to  $10^{-3} \text{ s}^{-1}$  at temperatures indicated with Ar gas passed through furnace enclosing specimens during test to assure an inert atmosphere (see 3.3.1.12) (Ref. 69)

Figure 3.3.2.1 Typical Ti-3Al-8V-6Cr-4Mo-4Zr longitudinal compressive stress-strain curves at room and elevated temperature for the solution treated plus aged (STA) condition (Ref. 1)

Figure 3.3.2.2 Typical Ti-3Al-8V-6Cr-4Mo-4Zr transverse compressive stress-strain curves at room and elevated temperature for the solution treated plus aged (STA) condition (Ref. 1)

Figure 3.3.2.3 Effect of test temperature on the compressive yield strength of a Ti-3Al-8V-6Cr-4Mo-4Zr forging in the solution treated plus aged (STA) condition (Ref. 1)

Table 3.3.2.4 Compressive test results at -65F of hot rolled Beta-C alloy sheet (0.065 in. thick) with different processing cycles (Ref. 50)

Figure 3.3.2.5 Compressive test results at 600F of hot rolled Beta-C alloy sheet (0.065 in. thick) with different processing cycles (Ref. 50)

Figure 3.3.3.1 Longitudinal and radial Charpy V-notch impact strength (duplicate transverse specimens with notches oriented longitudinally or radially) of hot rolled 3-inch (7.6 cm) diameter Beta-C bar solution treated at 1450F (787C), which resulted in an unrecrystallized (UNRX) grain structure, and at 1550F (842C), which resulted in a recrystallized (RX) grain structure, and aged to an intermediate strength level (140 ksi UTS or 1102 MPa) at 900 F (482C) as a function of temperature (see 3.2.1.38 and 3.2.1.39) (Ref. 70)

Table 3.3.7.1.1 Sharp notch tensile properties of hot rolled Beta-C alloy sheet (0.065 in. thick) at room, -65F, and 600F temperature (Ref. 50)

Table 3.4.1 Stress rupture and creep properties of grade 19 (Ti Beta-C) and 20 (Beta-C/Pd) alloys in the solution treated and aged (STA) condition (Ref. 2)

Figure 3.4.2 Creep rupture curves for Ti-3Al-8V-6Cr-4Mo-4Zr billet at 500, 700, and 900F in the solution annealed plus aged (STA) condition (Ref. 1)

Table 3.4.3 Thermal and creep stability data for cold drawn, solution treated (ST), and solution treated plus aged (STA) Ti-3Al-8V-6Cr-4Mo-4Zr bar (Ref. 1)

Figure 3.5.1.1 S-N fatigue ( $R = -1$ ) curves for Ti Beta-C specimens machined from 19.05 mm diameter rod (see 3.2.1.8 and 3.2.1.9): (a) annealed at 800C/0.5/AQ + solution treated at 850C/1 h/WQ (UTS = 878 MPa) and (b) annealed and solution treated as in (a) and aged at 455C/24 h/AQ (UTS = 964 MPa) (Ref. 38)

Figure 3.5.1.2 High-cycle fatigue ( $R = -1$ ) behavior of Ti-3Al-8V-6Cr-4Mo-4Zr alloy in the solution heat treated (SHT), Simplex aged at 540C/16 h, and Duplex aged at 440C/4 h + 560C/16 h (see 1.5.18) with 0.2% yield strengths for SHT = 850 MPa, Simplex = 1085 MPa, and Duplex = 1085 MPa (Ref. 35)

Figure 3.5.1.3 S-N fatigue ( $R = -1$ , 50 Hz, air) curves for single aged (simplex) and duplex aged Beta-C plate indicating that long-term aging at a low temperature followed by a short-term age at a higher temperature achieves a more uniform distribution of alpha phase and superior tensile and fatigue strength over simplex aging at a single temperature (see 1.5.19) (Ref. 20)

Figure 3.5.1.4 Statistical determination of the fatigue limit  $\sigma_{50\%}$  and the standard deviation  $s$  by means of the modified staircase method for 15 samples of Ti 38-644 (Beta-C) alloy after duplex aging (ST at 920C/30 min. + PA at 440C/12 h + FA at 500C/24 h) and a limiting number of loading cycles  $N = 2 \times 10^6$  with failure probability  $\sigma_{50\%}$  of duplex aged

Beta-C under tension-compression equal to 700 MPa and under rotating bending equal to 745 MPa (see 3.2.1.7 for tensile test results and 3.2.1.8 for SEM fractographs) (Ref. 36)

Figure 3.5.1.5 Crack propagation curves  $da/dN$  as a function of  $\Delta K$  obtained at  $R = 0.1$  in fully recrystallized (solution annealed at 920C/30 min), direct aged to intermediate strength, and duplex aged (solution annealed 920C/30 min + 440C/12 h + 500C/24 h) Beta-C alloy, with inserts illustrating fatigue crack propagation profiles in relation to microstructural features in (a) duplex aged and (b) solution annealed Beta-C (see 3.2.1.7 for tensile results and 3.5.1.4 for fatigue data) (Ref. 36)

Table 3.5.1.6 Long crack propagation threshold  $\Delta K_{th}$ -values determined by the indirect electric potential method for Beta-C specimens in 3.5.1.6 (Ref. 36)

Table 3.5.1.7 Fatigue limit at  $10^7$  cycles of Beta-C alloy (beta transus,  $\beta_t = 732C$ ) in various conditions with different microstructures (Ref. 49)

Figure 3.5.1.8 S-N curves for Ti-3Al-8V-6Cr-4Mo-4Zr at elevated temperatures for smooth and notched specimens in the solution treated plus aged (STA) condition at a stress ratio of  $R = 0.1$  (Ref. 1)

Table 3.5.1.9 Room temperature smooth tension-tension fatigue tests performed at 60 Hz ( $R = 0.1$ ) on samples machined from coil per ASTM E-466 for Beta-C and candidate titanium alloys for high strength fastener applications (duplicate samples run at 724 MPa or 48-53% of UTS and tests terminated after 1 million cycles) (Ref. 60)

Figure 3.5.1.10 Effects of cold work and tensile strength on the log average fatigue life of Ti-3Al-8V-6Cr-4Mo-4Zr wire of 0.392 in. (10 mm) diameter, shot peened to 0.016-0.018A Almen intensity, with each data point representing the log average of six tests conducted using 3-point bend specimens (Ti-13V-11Cr-3Al data points included for comparison) (Ref. 54)

Figure 3.5.1.11 Effect of surface finish prior to shot peening to 0.016-0.018A Almen intensity on fatigue life at two stress levels for 0.353 in. (9 mm) diameter Ti-3Al-8V-6Cr-4Mo-4Zr wire, with each data point representing the log average of six tests conducted using 3-point bend specimens (Ref. 54)

Figure 3.5.1.12 Fatigue life as a function of tensile strength for recrystallized and unrecrystallized Ti-3Al-8V-6Cr-4Mo-4Zr wire shot peened to 0.016-0.018A Almen intensity, with each data point representing the log average of six tests conducted using 3-point bend specimens (Ref. 54)

Table 3.5.1.13 [Table] Fatigues test results on rigid prototype production springs fabricated from Ti 38-6-44 wire that was cold worked, heat treated, and shot peened (Ref. 54)

Figure 3.5.1.14 Fatigue crack growth as a function of stress intensity range for compact tension specimens of Beta-C alloy solution treated at 900C/30 min/WQ and aged at 350 and 500C with a comparison with data for Ti-6-4 (Ref. 62)

Figure 3.5.2.1 Room temperature fatigue data generated at 1034 MPa (150 ksi) maximum strength,  $R = 0.1$ , and a cyclic rate of 30 Hz for Beta-C alloy bar in the heat treated condition (see 3.2.1.28 for STA conditions) from VAR and single melt PAM ingots (initial PAM data generated at the topmost or low cycle region of the fatigue curve seem to lie within the portion of the curve that would be predicted based on STA bar from VAR ingot) (Ref. 55)

Figure 3.5.2.2 Cyclic stress response behavior of solution treated (ST - 925C/WQ) and aged (see 1.6.12) Beta-C alloy cylindrical specimens (gage = 0.120 in. diameter by 0.32 in. length) tested in completely reversed strain cycling tested under constant plastic strain ( $\epsilon_p$ ) and constant total strain ( $\epsilon_T$ ) on an Instron machine at  $2 \times 10^{-3} \text{ sec}^{-1}$  strain rate (Ref. 65)

Figure 3.5.2.3 Monotonic and cyclic stress-strain curves of solution treated (ST - 925C/WQ) and solution treated and aged Beta C (cyclic flow stress is appreciable lower than the monotonic flow stress due to cyclic softening of the alloy exhibited in 3.5.2.2 at which the cyclic work hardening exponent (defined as  $n' = d \ln \sigma / d \ln c$ , where  $\sigma$  is cyclic stress and  $c$  is cyclic strain) has a value of 0.07 (Ref. 65)

Figure 3.6.2.1 Elastic modulus of ATI 38-644 (UNS R58640) at room temperature as a function of tensile strength (composite modulus line was estimated using results from various degrees of cold drawing + aging process routes) (Ref. 3)

Figure 3.6.2.2 Elastic modulus as a function of tensile yield strength of Beta-C alloy at room temperature (Ref. 57)

Figure 3.6.2.3 Room temperature tensile modulus of elasticity of RMI 38-6-44 sheet (0.5 cm thick) as a function of aging time at 538C for material solution treated at 816C for 1 h in a vacuum furnace, argon quenched, and cold worked 40 and 80% (Ref. 61)

Figure 3.6.2.4 Effect of test temperature on the modulus of elasticity in tension (E) and compression (E<sub>c</sub>) of Ti-3Al-8V-6Cr-4Mo-4Zr billet in the solution treated and aged condition (Ref. 1)

Table 3.6.2.5 Compressive yield strength and modulus at high temperature of Beta-C forgings (6 in. diameter), solution treated at 1,500F (815C), aged at 1,050F (565C) for 4 h, and air cooled (Ref. 57)

Table 3.6.3.1 Shear modulus (G) values for Ti-3Al-8V-6Cr-4Mo-4Zr at high temperatures (Ref. 69)

Figure 3.6.4.1 Longitudinal tangent modulus at room (RT) and elevated temperatures of Ti-3Al-8V-6Cr-4Mo-4Zr billet in the solution treated and aged condition (Ref. 1)

Figure 3.6.4.2 Transverse tangent modulus at room (RT) and elevated temperatures of Ti-3Al-8V-6Cr-4Mo-4Zr billet in the solution treated and aged condition (Ref. 1)

Table 4.1.1 Recommended hot working temperatures for Beta-C and other metastable beta titanium alloys (Ref. 31)

Table 4.1.2 Room temperature press brake bending of 0.065 in. thick sheet of Beta-C alloy in the hot rolled and mill annealed (1600F/30 min./AQ) condition (Ref. 50)

Table 4.1.3 Warm press brake bending of 0.065 in. thick sheet of Beta-C alloy in the hot rolled and mill annealed (1600F/30 min./AQ) condition (Ref. 50)

Table 4.1.4 Effect of forging conditions and heat treatment of Beta-C alloy on mechanical properties (Ref. 57)

Figure 4.1.5 Flow stresses of  $\beta$ -c alloy compared with Ti-6-4, Transage (Ti-2Al-11V-2Sn-11Zr), and Ti-10-2-3 as a function of temperature for the given hot working conditions (Ref. 31)

Figure 4.1.6 Flow stress of Beta-C alloy as a function of strain rate for temperatures of 760-875C with a comparison of Ti-10V-2Fe-3Al at 860C (Ref. 57)

Table 4.2.1 Nominal speeds and feeds for turning Beta-C and other  $\beta$  titanium alloys with high speed steel and carbide tools (Ref. 72)

Table 4.2.2 Nominal speeds and feeds for face milling Beta-C and other  $\beta$  titanium alloys with high speed steel and carbide tools (Ref. 72)

Table 4.2.3 Nominal speeds and feeds for end milling (peripheral) Beta-C and other  $\beta$  titanium alloys with high speed steel tools (Ref. 72)

Table 4.2.4 Nominal speeds and feeds for end milling (peripheral) Beta-C and other  $\beta$  titanium alloys with carbide tools (Ref. 72)

Table 4.2.5 Nominal speeds and feeds for drilling Beta-C and other  $\beta$  titanium alloys with high speed steel drills (Ref. 72)

Table 4.2.6 Nominal speeds and feeds for tapping Beta-C and other  $\beta$  titanium alloys with high speed steel taps (Ref. 72)

Fig Table ure 4.2.7 Nominal speeds and feeds for reaming Beta C and other  $\beta$ titanium alloys with high speed steel and carbide reamers (Ref. 72)

**Figure4.3.1** Relationship of hardness to ultimate strength of weld metal in steel (solid line) and titanium alloys (black dots) (Ref. 73)

Figure 4.3.2 Hardness traverse of TIG butt welds in 0.065 in. Beta-C sheet (Ref. 50)

Table 4.3.3 Mechanical properties of single-pass square TIG butt welds (undressed) made under He gas without filler metal in 0.065 in. thick Beta-C sheet (sheet preparation: milled edge, alkaline clean, HNO<sub>3</sub>-HF acid pickle, draw file and acetone wipe prior to welding) (Ref. 50)

Figure 4.3.4 Comparative strength-to-toughness properties of commercial titanium alloys in a multiple-pass gas thermal arc weld (GTAW) with plate metal and filler metal of the same composition (Beta-C - #16) (Ref. 73)

Table 4.3.5 Chemical composition for AWS classified Beta-C and Beta-C/Pd filler rod or wire (Ref. 73)

Figure 4.3.6 Post-weld heat treated GTAW fusion zone in Beta-C sheet: (a) aged at 900F for 24 h, 275x (b) same heat treatment as (a), 690x, (c) aged at 1100F for 8 h, 275x, and (d) same heat treatment as (c), 690x (heat treatment at 900F

for 24 h resulted in a high ultimate tensile strength (UTS) of 200 ksi and a low elongation of 2.5%, while the increased aging temperature of 1100F for 8 h promoted coarser  $\alpha$ -phase precipitation with a UTS of only 163 ksi with an increased elongation of 8%) (Ref. 74)

Figure 4.4.1 Effectiveness of an aluminum-based oxidation protective coating on titanium (coating was developed for titanium by Nippon Steel and consists of aluminum flakes coated with stearic acid) (Ref. 59)