J. K. Gregory[1] and L. Wagner[2]

1 GKSS Research Center, W-2054 Geesthacht, Germany
2 Technical University Hamburg-Harburg, W-2100 Hamburg 90, Germany

HEAT TREATMENT AND MECHANICAL BEHAVIOR IN Beta-C™

INTRODUCTION

Owing to the precipitation of the stable α-phase at temperatures between 350-600 °C, the solute-rich meta-stable β-titanium alloy Beta-C™ (Ti-3Al-8V-6Cr-4Zr-4Mo) can be heat treated to strength levels ranging from 850 MPa (in the as-solution heat treated condition) to over 1500 MPa [1]. Normally, precipitation is heterogeneous, resulting in an uneven distribution of α-precipitates such that some areas are precipitate-free. One way to obtain a uniform α-phase distribution is to carry out "duplex aging,"[2] where a pre-age at 425-480 °C produces fine, uniformly distributed precipitates, and a final age at 480-590 °C completes the aging treatment. Evaluation of fatigue and fracture toughness behavior for Beta-C heat treated both by direct aging and by duplex aging to a yield stress of 1080 MPa shows that duplex aging results in superior smooth specimen fatigue limits with no sacrifice in toughness or ductility. The corre- lation between strength, fatigue behavior, and toughness is discussed for several heat treated conditions.

HEAT TREATMENT

Material was received from Robert Zapp Werkstofftechnik in Dusseldorf in the form unidirectionally rolled plate. A solution heat treatment of 1/2 h at 927 °C on 13.5 mm thick samples followed by air cooling was found to be suitable to achieve complete recrystallization. The following six aging treatments were performed after this solution heat treatment (SHT):

16 h 500 °C	4 h 455 °C + 16 h 555 °C
16 h 530 °C	4 h 440 °C + 16 h 560 °C
16 h 540 °C	1000 h 250 °C

MICROSTRUCTURE

Figs. 1 - 6 show the microstructures obtained after the heat treatments performed. The as-SHT condition in Fig 1 is characterized by equiaxed β-grains with a diameter of 160 mm. After direct aging at 530 °C as well as 540 °C (Figs. 2 and 3), precipitation of the a-phase (the dark areas) is inhomogeneous, leaving regions roughly 25 by 100 mm which are precipitate free (the light areas). Duplex aging with a pre-age at 455 °C results in a somewhat more homogeneous distribution of the a-precipitates (Fig. 4), while an even distribution is obtained after a pre-age at 440 °C (Fig. 5). The long-term age of 1000 h at 250 °C does not produce any changes which can be seen by optical microscopy (Fig. 6).

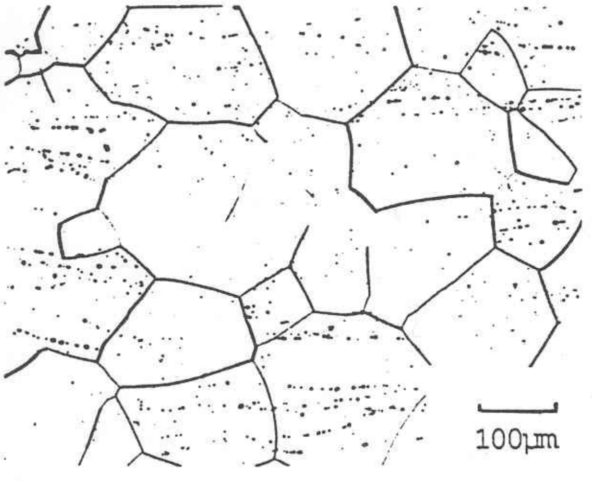


Fig.1: Microstructure of as-SHT condition

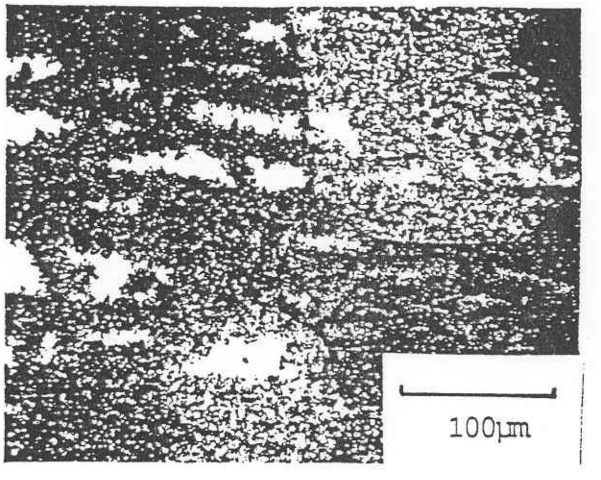


Fig.2: Microstructure after aging 16 h at 530 °C

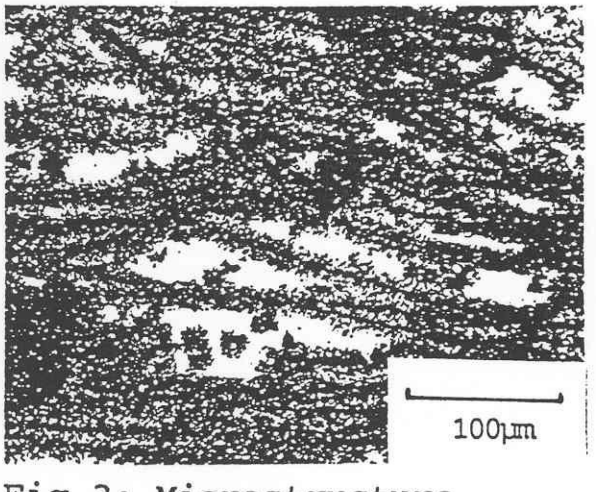


Fig.3: Microstructure after aging 16 h at 540 °C

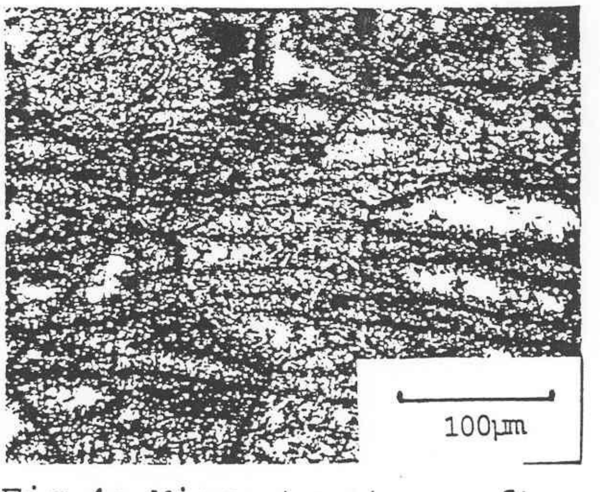


Fig.4: Microstructure after 4 h 455 °C + 16 h at 555 °C

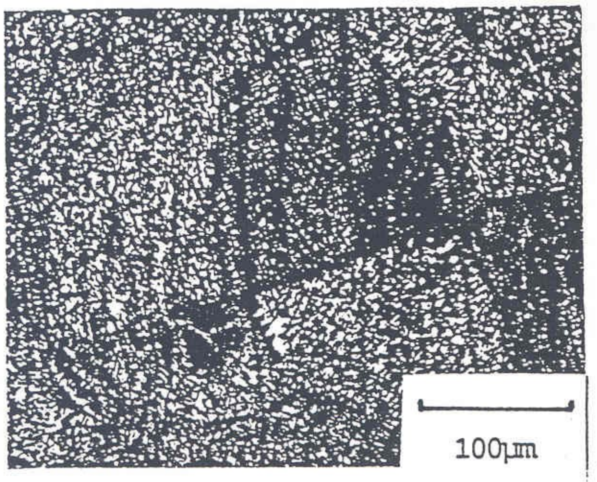


Fig.5: Microstructure after 4 h 440 °C + 16 h at 560 °C

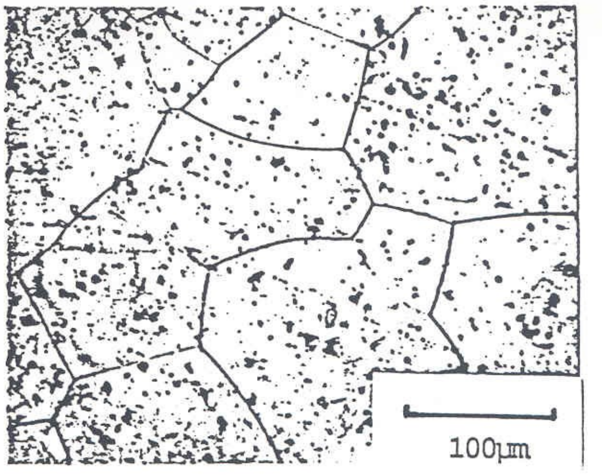


Fig.6: Microstructure after aging 1000 h at 250 °C

TENSILE PROPERTIES

The tensile test results far the heat treated conditions evaluated are shown in Table 1. All tests are for the tensile direction parallel to the rolling direction. For aging times of 16 h in the temperature range chosen, heat treatments which cause the α-phase to precipitate result in 0.2 % offset stresses of 1000-1140 MPa. The 0.02 % offset values for α-aged material vary between 960 MPa (after 16 h at 500 °C) and 1035 MPa (after aging 16 h at 530 °C or 4 h 440 °C + 16 h 560 °C). The loss of ductility after heat treating 1000 h at 250 °C is likely to be due to the formation of the ω-phase, which can only be seen with transmission electron microscopy.

Table 1: Tensile Properties of Beta-C after various heat treatments.

Heat Treatment	E (GPa)	$\sigma_{0.02}$ (MPa)	$\sigma_{0.2}$ (MPa)	UTS (MPa)	elong (%)	RA (%)
as-SHT	88	820	850	850	25	62
16 h 500 °C	94	960	1065	1130	9	21
16 h 530 °C	96	1035	1140	1220	12	21
16 h 540 °C	106	1005	1085	1165	13	23
4 h 455 °C + 16 h 555 °C	99	1000	1065	1215	14	27
4 h 440 °C + 16 h 560 °C	103	1035	1085	1140	12	24
1000 h 250 °C	114	--	--	1180	0	0

FATIGUE BEHAVIOR

Fatigue behavior was evaluated on smooth, electro-polished specimens under rotating beam loading (fully reversed) in air at 50 Hz. The S-N curves in Fig. 7 show the influence of a 16 hour direct aging treatment at 500 °C, 530 °C, and 540 °C. In general, the endurance limit scales with the strength level $\sigma_{0.02}$. Accordingly, the lowest endurance limit is exhibited by the as-SHT condition and the highest by the condition aged 16 h at 530 °C.

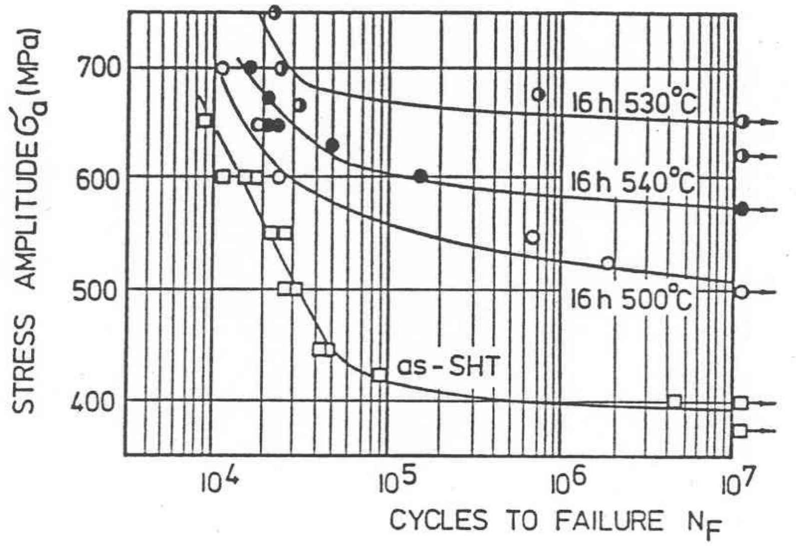


Fig. 7: S-N curves at R=-1 comparing as-SHT condition with those aged 16 h at 500 °C, 530 °C and 540 °C.

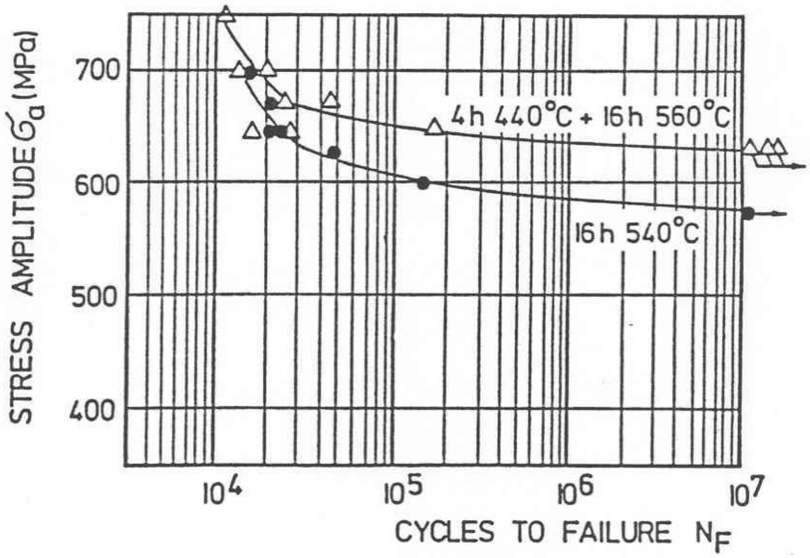
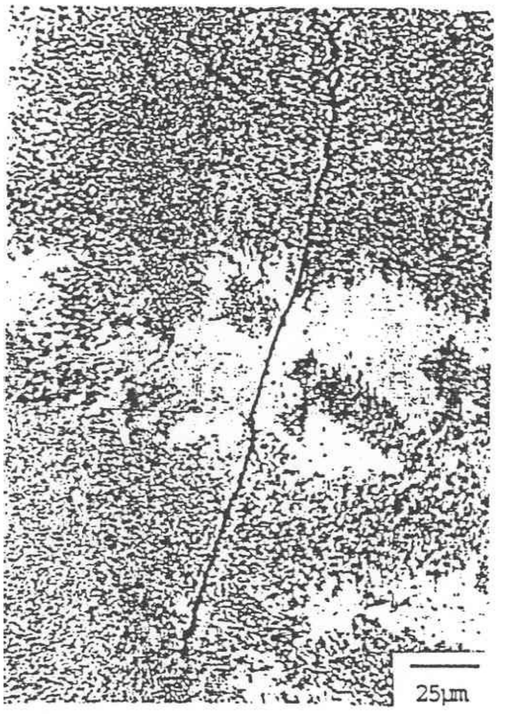


Fig. 8: S-N curves at R=-1 comparing the condition 16 h at 540 °C with 4 h 440 °C + 16 h 560 °C.

The influence of the α-phase distribution on fatigue life can be seen in the S-N curves shown in Fig. 8. The superior behavior of the material with the pre-age of 4h at 440 °C can be explained by examining the crack nucleation mechanisms. Fig. 9 shows a crack nucleation site in a fatigue sample which had been aged 16 h 540 °C. Evidently, cracks can nucleate more easily in the precipitate-free regions.



<u>Fig. 9:</u> Crack nucleation at precipitate-free regions for material direct-aged 16 h at 540 °C.

Fatigue crack growth tests were performed on CT-type specimens in air at 10 Hz at a load ratio of R=0.l. The resulting curves are shown in Fig. 10. Compared to the as-SHT condition, α-aged Beta-C exhibits crack growth rates which are slightly higher (almost a factor of two), while ω-aged Beta-C exhibits crack growth rates roughly a factor af two lower. The lower fatigue crack growth rates in the ω-aged material are surprising in view of the extremely brittle nature of this condition. Since the high elastic modulus (Table 1) can only partly explain this comparatively high resistance to crack growth, crack closure effects are thought to be responsible for these lower crack growth rates. Salt water has been shown to have no effect on the da/dN-ΔK curves for as-SHT and α-aged Beta-C at loading frequencies of 10, 1 and 0.1 Hz [3]. The possibilities far influencing the da/dN-ΔK behavior in Beta-C are quite limited when compared with those which can be achieved in (α+β) alloys such as Ti-6Al-4V, where significant variations in phase morphology can be induced [4].

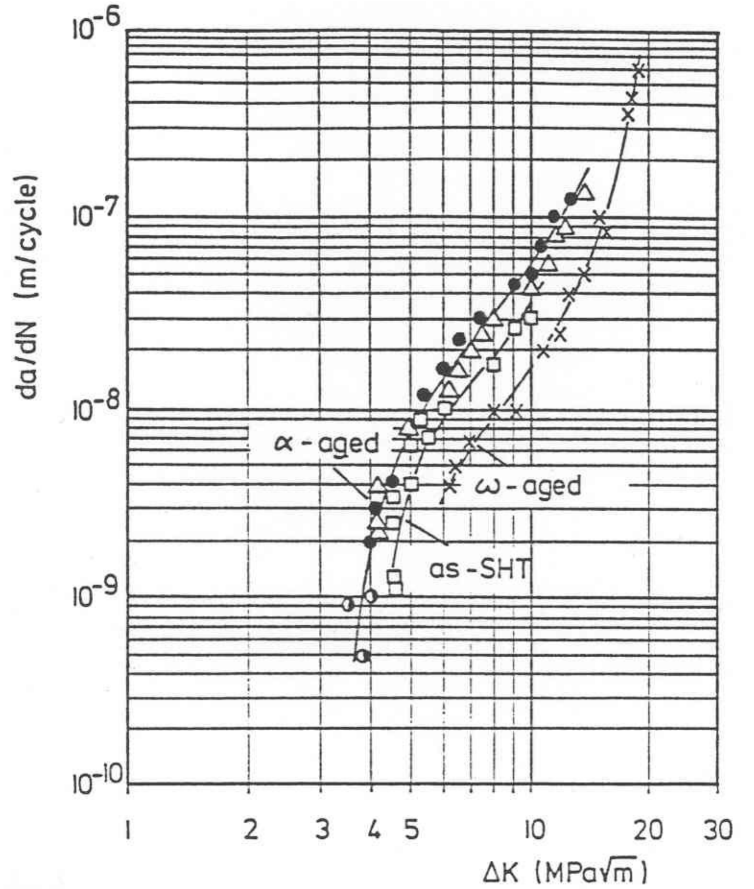


Fig. 10: da/dN-ΔK-curves far as-SHT, α-aged, and ω-aged Beta-C tested in air at 10 Hz, load ratio = 0.1.

FRACTURE TOUGHNESS

Values for the fracture toughness K_{Ic} were measured according to the ASTM method E647 (with the exception that a/W - 0.6). The highest fracture toughness was found for the as-SHT condition and the lowest for the ω-aged (1000 h 250 °C) condition. As is shown in Fig. 11, fracture toughness increases with increasing ductility. Hardness, fracture toughness, and fatigue limits for the various microstructures are summarized in Table 2. For microstructures with inhomogeneous α-phase distributions (as in Fig.3) two microhardness values are obtained, reflecting the difference in local properties of the weaker precipitate-free regions (305 HV) and the stronger precipitate-containing regions (408 HV).

Table 2: Hardness, fracture toughness, and fatigue limit values for the microstructures investigated

Heat Treatment	Micro-hardness HV 0.025	Macro-hardness HB 2.5/187.5	K_{Ic} $MPa\text{-}m^{1/2}$	σ_{a10^7} MPa
as-SHT	320	272	96	390
16 h 500 °C		342	70	500
16 h 530 °C		387	78	650
16 h 540 °C	305,408	365	84	575
4 h 455 °C + 16 h 555 °C		341	82	625
4 h 440 °C + 16 h 560 °C	385	349	86	625
1000 h 250 °C		396	22	--

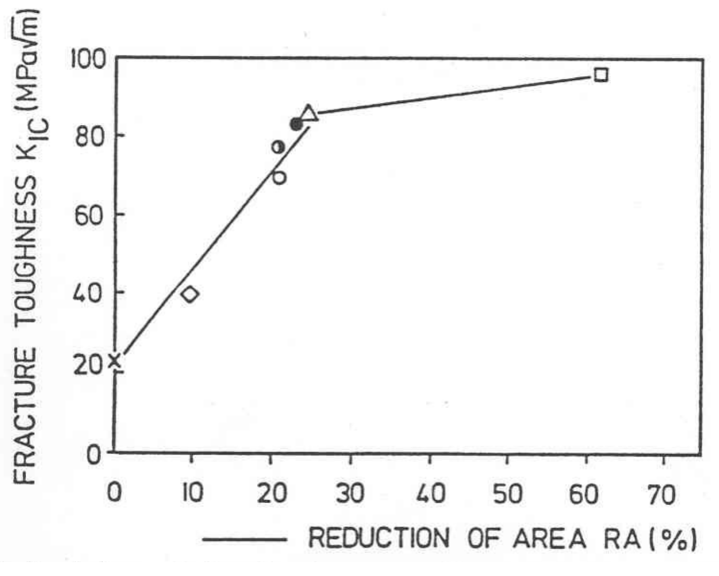


Fig. 11: Relationship between fracture toughness K_{Ic} and ductility as assessed by reduction of area (RA).

For a given strength level, both the fracture toughness and the fatigue limit are higher in duplex annealed than in direct aged microstructures. While the higher fracture toughness of the duplex aged conditions is likely to be a result of the somewhat higher ductilities

(Table 1), the lower fatigue limit in the direct aged microstructures is thought to be a result of the early crack nucleation which occurs in the weaker precipitate-free regions (Fig. 9).

SUMMARY

For Beta-C, the highest ductility and fracture toughness are found in the as-SHT condition, which also exhibits the lowest yield stress and fatigue limit. Long term exposure (1000 h) of this as-SHT condition to temperatures of about 250 °C causes severe embrittlement, owing to the precipitation of the metastable ω-phase.

After direct aging at temperatures between 500 -540 °C, precipitation of the α-phase is inhomogeneous, leaving precipitate-free regions in a hardened matrix. These weak regions adversely affect fatigue strength through early crack nucleation in these areas.

Duplex aging results in a homogeneously precipitation-hardened microstructure. The absence of weak regions leads to fatigue limits superior to those of direct-aged material having the same yield stress, at no sacrifice in tensile ductility or fracture toughness.

Compared to (α+β) titanium alloys, fatigue crack growth rates cannot be markedly varied by altering the microstructure, presumably because no significant change in phase morphology can be brought about by heat treatment.

REFERENCES

1. H.J. Rack, in: Titanium '80 Science and Technology, ed. H. Kimura and O. Izumi, TMS-AIME, Warrendale, PA (1980) 1627-1636.
2. D.E. Thomas, in: Sixth World Conference on Titanium, ed. P. Lacombe, R. Tricot, and G. Béranger (les édition de physique) (1989) 147-152.
3. H.-E. Krugmann and J.K. Gregory, in: Microstructure / Property Relationships in Titanium Alloys and Titanium Aluminides, eds. R.R. Boyer and Y. Kim, TMS-AIME, Warrendale, PA (1991) in press.
4. G.W. Kuhlmann and A.K. Chakrabaty, in: Microstructure Fracture Toughness and Fatigue Crack Growth

Rate in Titanium Alloys, eds. A.K. Chakrabaty and J.C. Chesnutt, TMS-AIME, Warrendale, PA (1987) 3-15.