Evaluation of the Mechanical Behaviour of a Beta-C Titanium Alloy in Aged and Overaged Condition (*)

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Abstract

The fatigue property of a Beta-C titanium alloy was studled by compact tension specimens from a rolled plate. Preliminary investigation of the alloy's response to aging at temperatures in the range 450°-600° was followed by hardness test and then fatigue tests in the LT and TL directions on solution treated specimens and on those aged for 20 hrs at 450°, 500° and 575°C.

The crack propagation rate curves were characterised by higher threshold values for the solution treated specimens, by comparison with the aged ones whereas there was a trend to lower values of crack velocity for higher range of ΔK in specimens aged at the lower temperatures. It is suggested that this behaviour is dependent on the delay effect. Support for this view was also provided by the SEM study of the fracture surfaces.

Micrographic examination of the structure seemed to point to a certain insensitivity of the crack propagation rate to the structure of specimens aged in the whole temperature range examined.

The overall results suggest that a narrower aging temperature range from 475° to 500°C provides the best combination of tensile strength and fatigue property.

Riassunto

È stato studiato il comportamento a fatica di un laminato in lega di titanio Beta-C mediante provini CT. In particolare, dopo un primo esame della risposta all'invecchiamento della lega nel campo di temperatura 450°-600°C, le prove di fatica sono state condotte nelle direzioni LT e TL per i campioni solo solubilizzati e per i campioni invecchiati a 450°, 500° o 575°C per 20 ore.

Le curve di velocità di propagazione della cricca di fatica sono caratterizzate da più elevati valori di soglia per i campioni solo solubilizzati, mentre si ha una tendenza a minori incrementi di velocità di propagazione della cricca per i più alti valori di ΔK nei campioni invecchiati alle temperature minori. Si può ipotizzare che questo comportamento sia legato ad effetti di ritardo, come evidenzia anche l'esame SEM delle superfici di frattura.

L'esame micrografico della struttura sembra confermare una certa insensibilità della velocità di propagazione della cricca con lo stato strutturale degli invecchiati nel campo di temperature esaminato.

Il complesso delle prove svolte ha permesso di individuare nell'intervallo di temperatura di invecchiamento 475°-500°C la miglior combinazione di proprietà tensili e resistenza a fatica.

Introduction

The commercial production of new types of β titanium alloys was an important feature of R & D in this field during the Seventies and Eighties. A representative example is the metastable beta alloy known as Beta-C alloy produced by RMI company (USA) [1-5]. Its trade designation "38-6-44" stems from its composition: Ti-3Al-8V-6Cr-4Mo-4Zr.

In addition to the general properties of the beta alloys, such as excellent forgeability, high strength and equally sound fatigue behaviour, Beta-C displays good hot and cold machinability, coupled with excellent corrosion resistance. This alloy is also suitable for the less technologically sophisticated sectors, since its specifications are not particularly restrictive, and thus has a very broad market potential [6]. In Italy, it has not yet been applied on a wide scale; but its direct production, could be the source of interesting prospects in the near future.

This paper examines the fatigue behaviour of Beta-C alloy. Particular attention is directed to the combination of strength and fatigue properties, in relation (inter alia) to possible working textures and as function of various aging treatments, a subject on which there is little in the specific literature.

Fatigue behaviour, or more precisely the subcritical crack propagation rate, can be represented by semi-empirical models allowing the fatigue crack growth rate to be analytically expressed with a very small number of characteristic experimental parameters.

The relations proposed by Paris, by Forman and by Collipriest are primarily used for this purpose. The last of these [7] is employed in this paper. The equation is:

$$da/dN = C(K_{Ic}\Delta K_{th})^{n/2} \exp \{\ln[K_{Ic}/\Delta K_{th}]^{n/2} \ \mathrm{tgh}^{-1} \ Z\}$$

where: $Z = \{\ln(\Delta K^2/[(1-R)K_{Ic}\Delta K_{th}])\} \ / \ \{\ln[(1-R)K_{Ic}/\Delta K_{th}]\}$.

The pattern of the fatigue curve is determined by the inverse hyperbolic tangent function [7].

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Experimental part

Alloy Beta-C was supplied by the manufacturer (RMI, USA) as solution treated (815°C: ½ hr; A.C.) 20 mm thick rolled plate. Its chemical composition was: Al 3.3% - V8.1% - Cr 5.8% - Mo 3,9% - Zr 3.8% - Fe 0.08% - Nb 0.08% - C 0.02% - N 0.015% - H 80 ppm.

Specimens for aging and for tensile and fatigue tests were cut with a milling machine.

The initial aging tests were conducted on 20*25*4 (mm) specimens in an argon furnace across the range 450-600°C for periods of up to 130 hours for the lowest temperature.

Vickers hardness measurements were taken with a Wolpert machine (50 kg load-operating time 15 s.)

Longitudinal and transverse tensile stress specimens (length 50 mm; gauge length 20 mm; cross-section $\sim 16\ mm^2$) were aged at 475°, 525°, 550° and 575°C for 5, 12 and 18 hrs respectively. YS, UTS and El% were then determined with an Instron 1193 machine.

Two-stage aging tests were carried out at 425°C (8 hrs) 575°C (10 hrs); 475°C (3 hrs) 525°C (10 hrs); 475°C (3 hrs) 575°C (5 hrs) and 475°C (5 hrs) 575°C (5 hrs) to activate initially the formation of β' zones ($\beta \rightarrow \beta + \beta'$) and then the $\beta + \beta' \rightarrow \beta + \alpha$ transformation. Since the second-stage temperature is higher and hence characterised by a faster α_s growth kinetics according to the TTT diagram [9], it should quickly lead to high hardness values. Distribution of the precipitates is fully dependent on the first stage, which should be suitable to supply abundant and uniform β' nucleation.

Fatigue tests were run according to ASTM E 647 in a lab environment on machines whose load was generated by an electromagnet and amplified mechanically with a frequency of 9.5 Hz. Compact test specimens 12.5 mm thick and 40 mm wide (Fig. 1) in longitudinal-transverse (LT) and transverse-longitudinal (TL) configurations were initially examined in the solution treated state and after aging at 450, 500 and 575°C for 20 hrs. These temperatures and this time were chosen to give the highest strength over a period compatible with treatment times usually employed in industrial practice.

Lower temperatures provide higher strength, but require much longer times, as it can be seen in the part of the paper showing the aging experiments.

Sinusoidal loads were applied with a ratio $R = 0.5$ ($R = Pmin/Pmax$). Crack lengths were measured at x 40 light microscope and checked after failure of the specimen. The polynomial equation was then employed to determine the crack growth rate from these data. The results were processed with a program specially prepared for da/dN-ΔK curves.

Collipriest's semi-empirical relation [7] was used to interpolate the experimental data. This model, together with those of Paris and Forman, provides the basis for prediction of the fatigue behaviour of a real component.

Metallographic examinations (etching reagent: 1 ml HF (40%), 5 ml HNO_3, 94 ml distilled water) and fracture surface examination were carried with a light microscospe and a Hitachi S-2500 scanning electron microscope (SEM) equipped with a Kevex 8000 EDS.

Results and discussion

Aging treatments

As it can be seen in Fig. 2, all the aging curves display a similar qualitative pattern with regard to increases in hardness as a function of increasing time. Except at 600°C, where no substantial variations occur, hardness progressively increased in specimens aged at lower temperatures. The optimum treatment range for this mechanical property is thus between 450°C and 500°C. This, of course, necessitates aging times to reach the maximum hardness of a lot of hours or even of several days at $T \approx 450$°C.

Fig. 3 compares the 500°C aging curves of 3.5% cold-worked and unstretched specimens. Plastic deformation causes dislocations and structure defects in general. By thus increasing the concentration of active sites for α_s, nucleation and precipitation, it should result in faster hardening. It will be observed, however, the curves are rather close to each other and the maximum hardness values are practically the same. Thermomechanical treatment seems therefore superfluous for practical purposes.

The hardness values obtained after duplex aging are shown in Table 1. It is clear that not even this procedure furnished a satisfactory response.

TABLE 1 - Hardness values after duplex aging

Aging treatment	HV (average values)
425°C - 8 h/525°C - 10 h	365
425°C - 8 h/575°C - 10 h	297
475°C - 3 h/525°C - 10 h	371
475°C - 3 h/575°C - 5 h	325
475°C - 5 h/575°C - 5 h	334

Fig. 4 illustrates as an example UTS, YS (0,2%) and El% values for specimens parallel to the main rolling direction and aged at temperatures indicated. Since no great differences (1-2% for strenght; 5% for elongation) are observed for T specimens, it is clear that, in this case too, higher strength is obtained at lower temperature, though with a loss of plasticity. If consideration is given to the necessity of not decreasing strongly ductility, as well as the avoidance of excessively long treatment times, aging temperatures of nearly 500°C are better, except in particular cases.

From metallography, Fig. 5 shows the structure of the solution treated alloy in three directions; longitudinal (L), transverse (T) and short transverse (S). As for mechanical properties, there is not appreciable difference between L and T, not surprisingly in view of the type of product (relatively thick plate). A rather coarse and poorly uniform grain size is also evident.

Micrographs of specimens aged to the maximum hardness at the four temperatures are shown in Fig. 6. At 475°C (Fig. 6a); very fine precipitation of α_s (dark phase) is perfectly distributed within the β matrix (light phase): the precipitates are very small (0.1 - 0.5 μm). At 500°C (Fig. 6b), precipitation is still fairly uniform, but particle sizes have grown to 1 - 1.5 μm. Their size increases still further at 525°C (Fig. 6c), though their distribution is still sufficiently uniform for high hardening. At 550°C (Fig. 6d), however, this is no longer the case. Large (3-6 μm) α_s islands are elongated in bands and very randomly precipitated. In Fig. 7, indeed, taken at a low magnification (x 300) to obtain an overall view of a specimen aged at 600°C, it can be seen that only a few grains have been involved in the precipitation phenomena. Back scattering shows the dark lower atomic weight α_s zones, EDS analysis confirmed depletion of β-genic elements (V, Cr, Mo) in the precipitates, as illustrated in the corresponding Rx maps (Fig. 8).

Fatigue tests

The results of the fatigue tests carried out on solution treated alloy (Fig. 9a) and after aging at 450°, 500° and 575°C (Figs. 9b, c, d) show that LT ant TL specimens are very similar in behaviour.

Fig. 10 gives the Collipresst curves for the TL results only, using $K_{IC}=90$ MPa $\sqrt{m}$ [5].

Crack propagation rates are faster in aged specimens at low ΔK values and in solution treated specimens when ΔK exceeds 10 MPa $\sqrt{m}$; here, indeed, propagation is slower at the lower aging temperature. This is also true for LT specimens.

Examination of the fracture surface of fatigue failured specimens revealed a substantially similar initial morphology in both the solution treated and the aged specimens. In the early stages when crack propagation in slower (lower ΔK values), all specimens display transgranular propagation with the cleavage-like steps, as illustrated in Fig. 11.

The solution treated specimens, were rougher. This feature was less evident after aging, especially at 450° and 500°C, when extensive areas of cleavage crack propagation appeared (Fig. 12). In specimens aged at 450° and 500°C, this characteristic developed over the whole fracture surface from the early to the final stages. In the solution treated specimens and in those aged at 570°C, the typical fatigue streaks (Fig. 13) could be seen about 20 mm from the notch ($\Delta K \approx 12$ MPa $\sqrt{m}$). The distance between the striations (slightly less than 1 μm) was in reasonable agreement with the measured growth rates. These, as already stated, were lower for the solution treated or aged at 575°C specimens than for those aged at 450°C and 500°C.

This result is apparently in conflict with the few literature data concerning fatigue tests on the alloy Beta-C [3] and β alloys in general [8]. Its explanation may lie in the higher delay behaviour presented at high ΔK values by specimens aged at 450 and 500°C, as is equally apparent from the greater roughness of their fracture surfaces. Similar delay behaviour in fatigue crack propagation have already been described for another β alloy (Ti - 10 V - 2 Fe - 3 Al) [11]. This phenomenon will be the subject of further investigation when aging at temperatures in the 350°-400°C range will be investigated.

As it is well known [2, 3, 9, 10], aging of Beta-C alloy at temperatures up to 400°C primarily results in the precipitation of β′ and/or α phases (*), depending on the temperature and duration of the treatment, initially in coherent form and subsequently as very fine precipitates uniformly distributed within the β matrix, showing very high hardness and yield point, but also rather low plasticity, less dynamic fracture toughness, and lower fatigue limits [3]; e.g. hardness >450 Vickers and YS >1600 MPa, but an ultimate elongation of only 1-2%.

Aging at 450-500°C, on the other hand, appears to offer the best combination of tensile properties and fatigue resistance (as shown by compact tension specimens) UTS of 1250 MPa and a yielding point of 1200 MPa, can be easily reached togheter with an ultimate elongation of the order of 5%, plus very small differences in this case between LT and TL specimens thanks to persistence of a very uniform and fine α precipitate structure (Fig. 6). At higher temperatures (e.g. 550°C), rather coarse α crystals begin to appear (Fig. 6d), while at 600°C precipitation is totally inhomogeneous with precipitate impoverishment of some grains (Fig. 7). An increase in aging temperature, therefore, results in a progressive loss of strength with a parallel gain in ductility.

The fatigue behaviour during the second step ($\Delta K \approx 10$ MPa $\sqrt{m}$) of specimens aged at 450°C is apparently better than those treated at higher temperatures and thus in possession of higher plasticity, though with slightly lower resistance to tensile stress. It should be noted, however, that these differences are always rather small.

Conclusions

The aging phenomena underlying the structural transformation of the metastable beta phase in Beta-C titanium alloy enable a wide range of mechanical property values to be obtained, especially with regard to yielding, ductility, fracture toughness, and fatigue resistance.

(*) The Ω-phase, which is typical in β-alloys aged at low temperatures, does not precipitate in the Beta-C alloy but in very small amounts and by aging at temperatures lower or equal to 300°C [2].

The series of tests described in this paper showed that an aging temperature range of 475-500°C gives the best combination of tensile properties and fatigue resistance. The lowest of these temperatures is preferable if a high yield limit has to be reached together with acceptable plasticity and good fatigue behaviour. This, of course, necessitates longer treatment times.

This type of aging leads to a structure composed of substantial quantities of α precipitates that are incoherents, but fine and uniformly distributed in the β matrix. Intermediate stretching, itself less advantageous economically, and duplex aging led to no greater improvement in the results.

Acknowledgements

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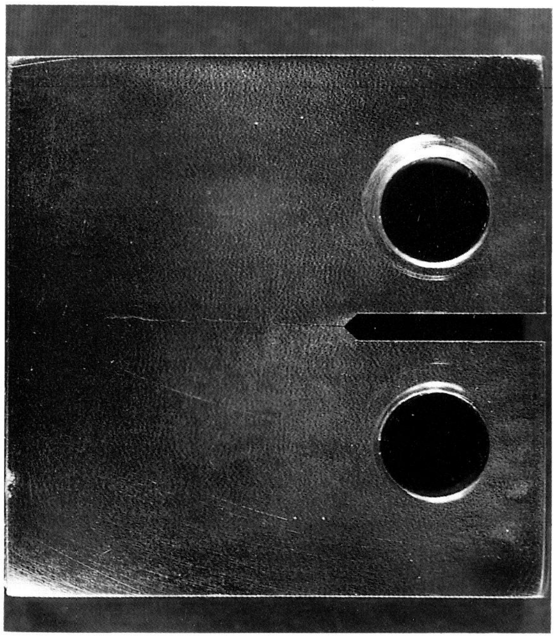

Fig. 1:
Compact tension specimen after the fatigue test.

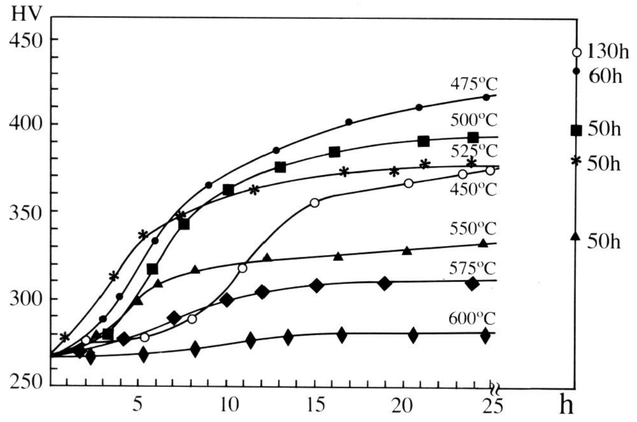

Fig. 2:
Aging curves at the temperatures indicated after solution treating (815°C-½ h, A.C.)

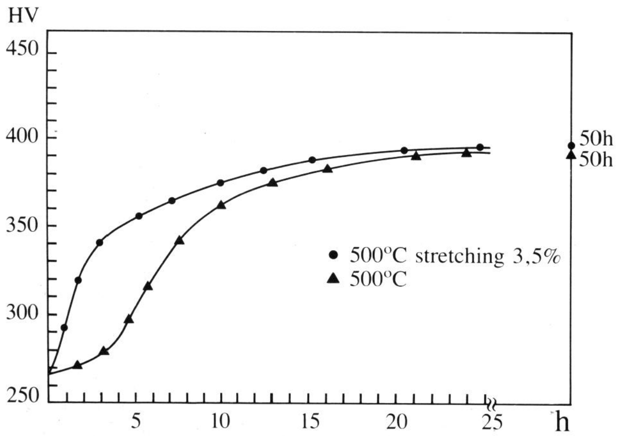

Fig. 3:
Comparison of 500°C aging curves for a specimen 3.5% cold worked and an unstretched specimen.

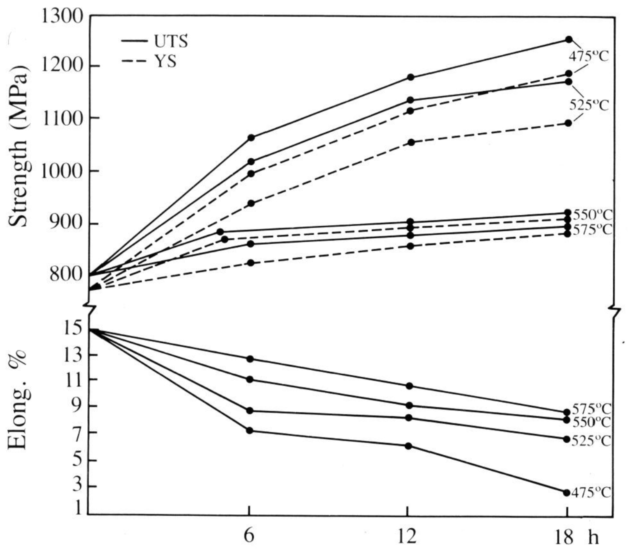

Fig. 4:
Mechanical properties (UTS, YS, El%) versus aging times at the temperatures indicated.

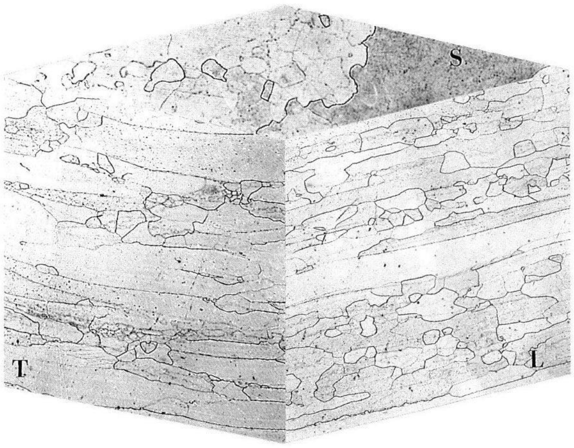

Fig. 5:
Optical micrograph for L, T and S sections of the solution treated alloy (x 40).

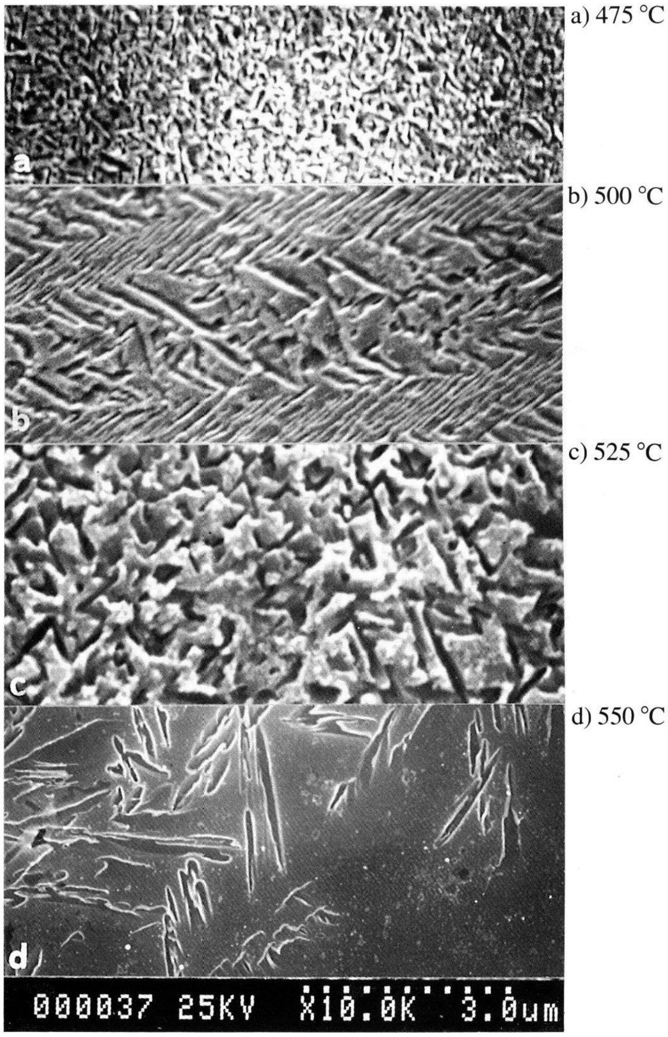

Fig. 6:
SEM micrographs of specimens aged at:
(a) 475°C (b) 500°C (c) 525°C (d) 550°C.

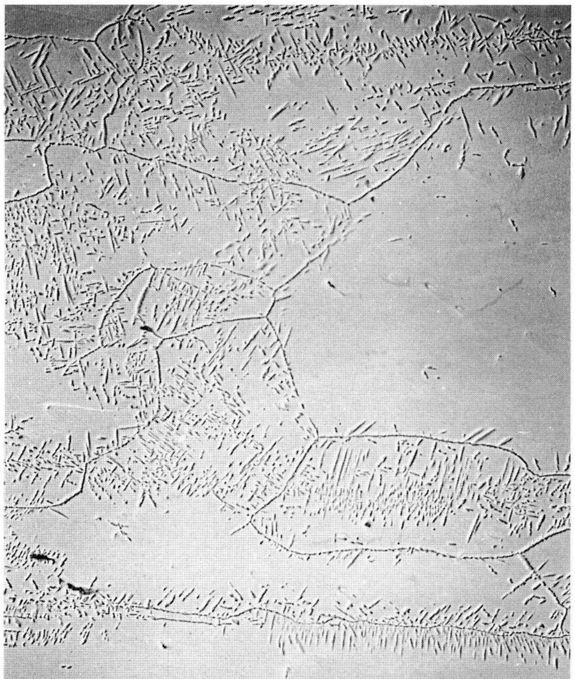

Fig. 7:
SEM micrograph of a specimen aged at 600°C.

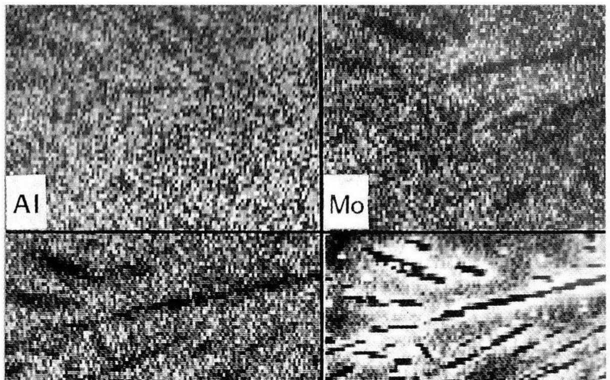

Fig. 8:
X-rays maps for Al, Mo and Cr and SEM micrograph of a specimen aged at 600°C.

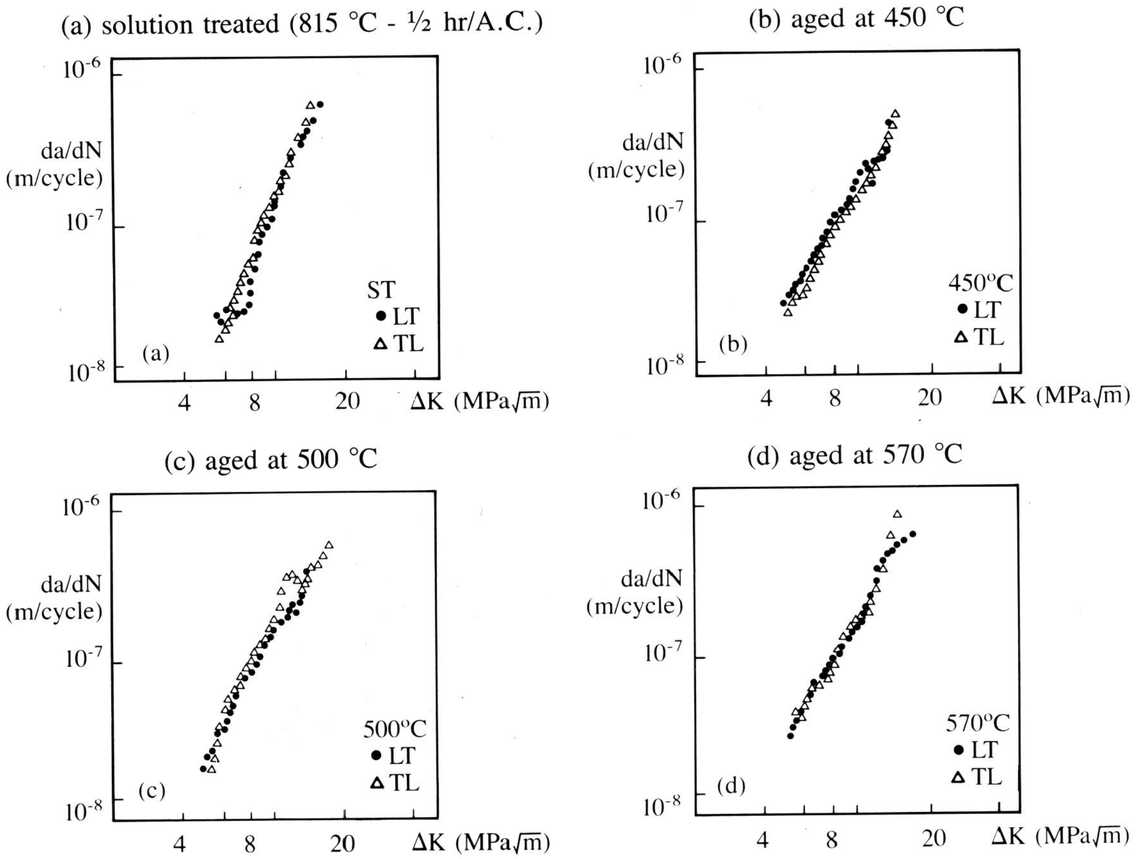

Fig. 9:
Experimentals fatigue test results for LT and TL specimens.

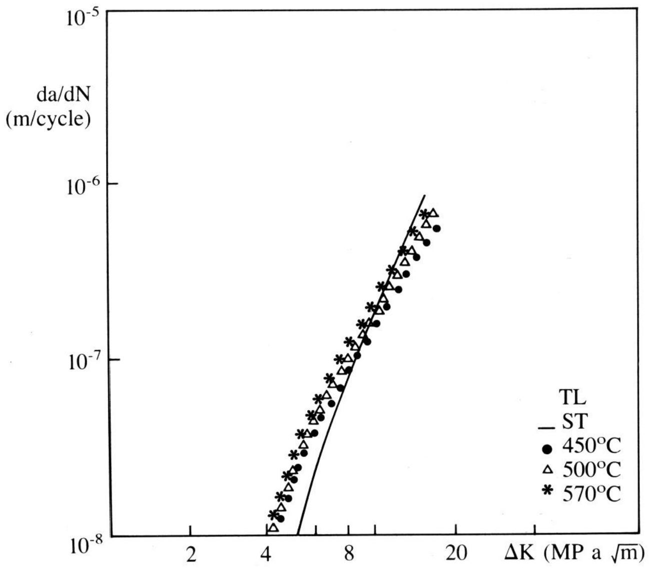

Fig. 10:
Fatigue crack growth rate of Beta-C alloy in the TL direction after aging, at temperatures indicated. Data interpolated with Collipriest's model:

COLLIPRIEST COEFFICIENTS	SOLUTION TREATED	AGED AT 450°C	500°C	570°C
C (10^{-9})	0.097	1.09	1.26	1.27
n	2.85	1.88	1.9	1.95

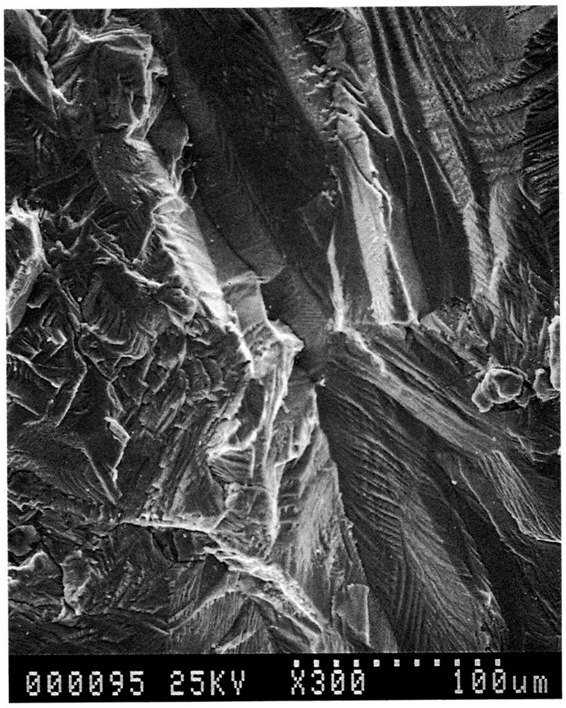

Fig. 11:
SEM micrograph of the fatigue fracture surface of the solution treated specimen: about 6 mm from the notch.

Fig. 12:
SEM micrograph of the fatigue fracture surface of 450°C aged specimen:
about 4 mm from the notch.

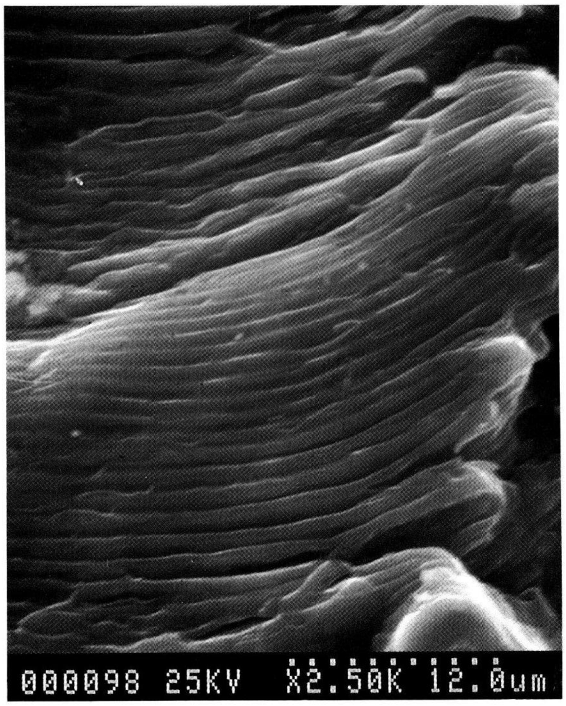

Fig. 13:
SEM micrograph of the fracture surface in solution treated specimen:
about 20 mm from the notch ($\Delta K \approx 12$ Mpa $\sqrt{m}$).
The presence of progressive fatigue striations in clearly evident.