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Microstructure and Mechanical Properties of High Strength Two-Phase Titanium Alloys

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1. Introduction

Two-phase titanium alloys constitute very important group of structural materials used in aerospace applications [1-3]. Microstructure of these alloys can be varied significantly in the processes of plastic working and heat treatment allowing for fitting their mechanical properties including fatigue behaviour to the specific requirements [4-6].

The main types of microstructure are (1) lamellar – formed after slow cooling when deformation or heat treatment takes place at a temperature in the single-phase β -field above the so-called beta-transus temperature T_{β} (at which the $\alpha+\beta\rightarrow\beta$ transformation takes place), consisting of colonies of hexagonal close packed (hcp) α -phase lamellae within large body centered cubic (bcc) β -phase grains of several hundred microns in diameter, and (2) equiaxed – formed after deformation in the two-phase $\alpha+\beta$ field (i.e., below T_{β}), consisting of globular α -phase dispersed in β -phase matrix [7-8].

The first type of microstructure is characterized by relatively low tensile ductility, moderate fatigue properties, and good creep and crack growth resistance.

The second microstructure has a better balance of strength and ductility at room temperature and fatigue properties which depend noticeably on the crystallographic texture of the hcp α -phase.

An advantageous balance of properties can be obtained by development of bimodal microstructure consisting of primary α -grains and fine lamellar α colonies within relatively small β -grains (10-20 μm in diameter) [9-10].

In the following sections the relations between microstructure morphology and mechanical properties of selected high strength two-phase titanium alloys were analysed.

Dilatometric tests, microstructure observation and X-ray structural analysis were carried out for cooling rates in the range of $48-0.004^{\circ}\text{C s}^{-1}$ and time-temperature-transformation diagrams were developed for continuous cooling conditions (CCT).

The influence of the quantitative parameters of lamellar microstructure on the tensile properties and fatigue behaviour of selected two-phase titanium alloys was analysed. Rotational bending tests were carried out to determine high cycle fatigue (HCF) strength at 10^7 cycles.

2. High strength two-phase titanium alloys

The materials tested were high strength, two-phase $\alpha+\beta$ titanium alloys: Ti-6Al-4V, Ti-6Al-2Mo-2Cr and Ti-6Al-5Mo-5V-1Cr-1Fe (Table 1).

Alloy	Stability factor of β - phase	Alloying elements content, wt.%							
		K_{β}	Al	Mo	V	Cr	Fe	C	Si
Ti-6Al-4V	0.3	6.1	–	4.3	–	0.16	0.01	–	bal.
Ti-6Al-2Mo-2Cr	0.6	6.3	2.6	–	2.1	0.40	0.05	0.2	bal.
Ti-6Al-5Mo-5V-1Cr-1Fe	1.2	5.8	5.3	5.1	0.9	0.8	0.05	0.15	bal.

Table 1. Chemical composition of the investigated titanium alloys.

Ti-6Al-4V – martensitic $\alpha+\beta$ alloy ($K_{\beta}=0.3$) – is the most widespread titanium alloy (>60% of all titanium alloys produced in USA and EU). Its high applicability results from good balance of mechanical properties and good castability, plastic workability, heat treatability and weldability. Aluminium addition stabilizes and strengthens α phase, increases $\alpha+\beta \leftrightarrow \beta$ transformation temperature and reduces alloy density. Vanadium – β -stabilizer – reduces $\alpha+\beta \leftrightarrow \beta$ transformation temperature and facilitates hot working (higher volume fraction of β -phase). Depending on required mechanical properties following heat treatment can be applied to Ti-6Al-4V alloy: partial annealing ($600\div 650^{\circ}\text{C}$ / 1h), full annealing ($700\div 850^{\circ}\text{C}$ / furnace cooling to 600°C / air cooling) or solutioning ($880\div 950^{\circ}\text{C}$ / water quenching) and ageing ($400\div 600^{\circ}\text{C}$) [1,3].

Ti-6Al-2Mo-2Cr – martensitic $\alpha+\beta$ alloy – known as VT3-1, is one of the first widespread high-temperature titanium alloys used in Russia for aircraft engine elements. Amount of β -stabilizers is similar to Ti-6Al-4V alloy but β -stabilizing factor is higher ($K_{\beta}=0.6$). Mo – stabilises and strengthens β -phase, in the presence of Si increases creep resistance, facilitates plastic working, Cr, Fe – eutectoid elements, stabilise β -phase and strengthen α and β phases in the low and medium temperature range [1].

The alloy is processed by forging, stamping, rolling and pressing. Depending on the application and required properties following heat treatment can be applied to the semiproducts: isothermal annealing (870°C / 1h / furnace cooling to 650°C / holding for 2 h / air cooling),

duplex annealing (880°C / 1h / air cooling and following heating 550°C / 2÷5 h / air cooling) or hardening heat treatment (water quenching and ageing) [1,7].

Ti-6Al-2Mo-2Cr alloy retains its mechanical properties up to 300°C. At the temperature higher than 400°C mechanical properties are reduced due to partitioning of alloying elements proceeding by diffusion.

Ti-6Al-5Mo-5V-1Cr-1Fe transition $\alpha+\beta$ titanium alloy ($K_\beta = 1.2$) is produced in Russia and Ukraine, where is known as VT22. It is characterized by very good mechanical properties thus is mainly used for large, heavy loaded, forged parts for long-term operation at elevated temperature up to 350÷400°C and short-term up to 750÷800°C. Typical applications include disks and blades of low pressure compressors, landing gear elements, engine mount struts and others [1,11].

3. Development of microstructure during continuous cooling

Phase composition of titanium alloys after cooling from β phase range is controlled by cooling rate. Kinetics of phase transformations is related to the value of β -phase stability coefficient K_β resulting from the chemical composition of the alloy [7].

One important characteristic of the alloy is a range of $\alpha+\beta \rightarrow \beta$ phase transformation temperature that determines conditions of thermomechanical processing intended for development of suitable microstructure. Start and finish temperatures of $\alpha+\beta \rightarrow \beta$ phase transformation, vary depending on the contents of β stabilizing elements (Table 2).

Phase transformation	Alloy			
	temperature, °C	Ti-6Al-4V	Ti-6Al-2Mo-2Cr	Ti-6Al-5Mo-5V-1Cr-1Fe
$T_{\alpha+\beta \rightarrow \beta}^{ns}$		890	840	790
$T_{\alpha+\beta \rightarrow \beta}^{ps}$		930	920	830
$T_{\alpha+\beta \rightarrow \beta}^f$		985	980	880
$T_{\beta \rightarrow \alpha+\beta}^s$		950	940	850
$T_{\beta \rightarrow \alpha+\beta}^f$		870	850	810

ns – nucleation start

ps – precipitation start

s – start

f – finish

Table 2. Start and finish temperature of the $\alpha+\beta \rightarrow \beta$ phase transformation for selected titanium alloys ($v_h = v_c = 0.08^\circ\text{C s}^{-1}$)

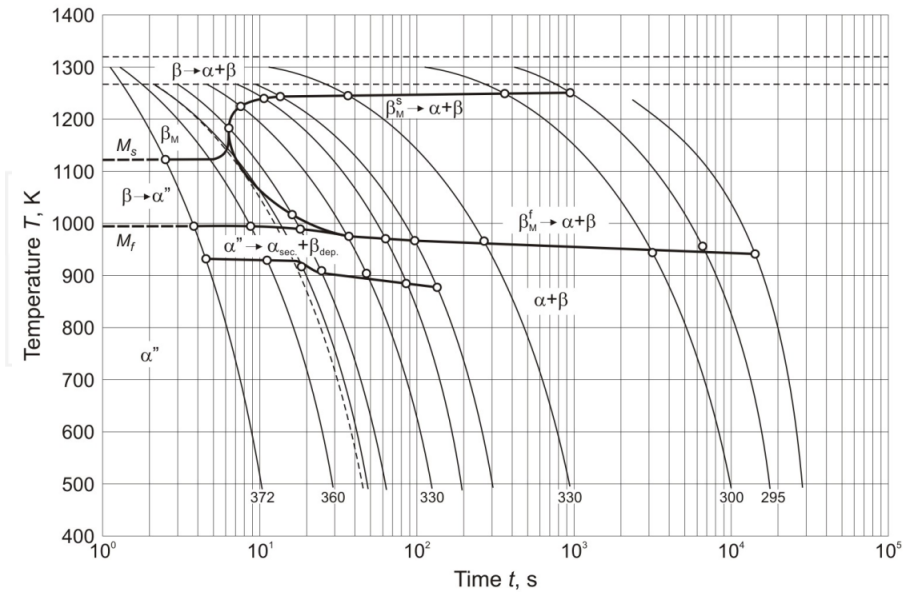


Figure 1. CCT diagram for Ti-6Al-4V alloy.

Cooling of Ti-6Al-2Mo-2Cr and Ti-6Al-4V alloys from above β transus temperature at the rate higher than 18°C s^{-1} leads to development of martensitic microstructure consisting of α' (α'') phases (Fig. 4). Start and finish temperatures of the martensitic transformation $\beta \rightarrow \alpha'$ (α'') or $\beta \rightarrow \alpha''$ do not depend on cooling rate but on β -stabilizing elements content and decrease with increasing K_β value.

For the intermediate cooling rates, down to 3.5°C s^{-1} , martensitic transformation is accompanied by diffusional transformation $\beta \rightarrow \alpha + \beta$ and the volume fraction of martensitic phases decreases to the benefit of stable α and β phases (Figs 1-2). Cooling rates below 2°C s^{-1} lead to a diffusion controlled nucleation and growth of stable α and β phases in the shape of colonies of parallel α -phase lamellae in primary β -phase grains (Fig. 5). For extremely low cooling rates precipitations of TiCr_2 phase were identified in the Ti-6Al-2Mo-2Cr alloy which were formed in eutectoid transformation.

In the transition alloy Ti-6Al-5Mo-5V-1Cr-1Fe martensitic transformation was not observed at any cooling rate. High cooling rate ($>18^\circ\text{C s}^{-1}$) results in metastable β_M microstructure. At lower cooling rates α -phase precipitates as a result of diffusional transformation. At lowest cooling rate, similarly to Ti-6Al-2Mo-2Cr alloy eutectoid transformation occurs and traces of TiCr_2 and TiFe_2 appears [7].

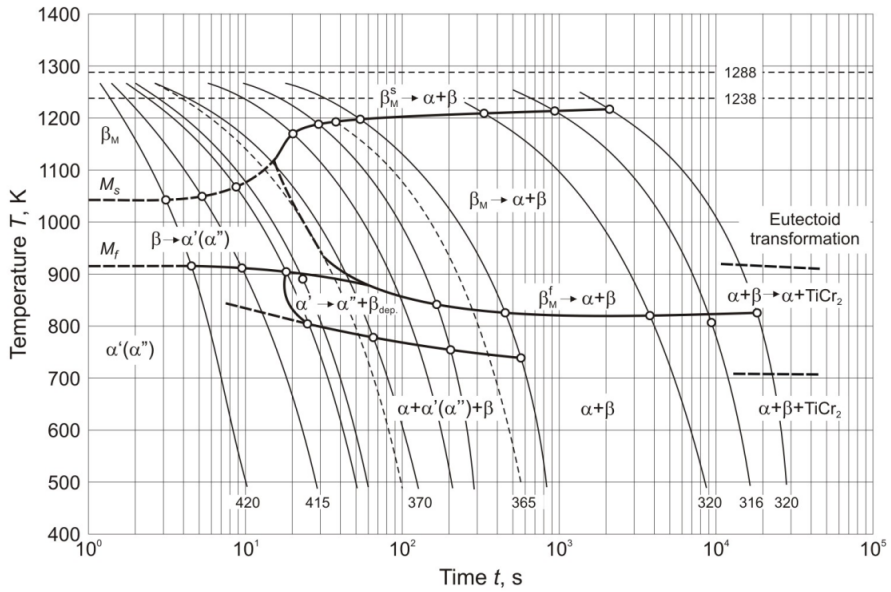


Figure 2. CCT diagram for Ti-6Al-2Mo-2Cr alloy.

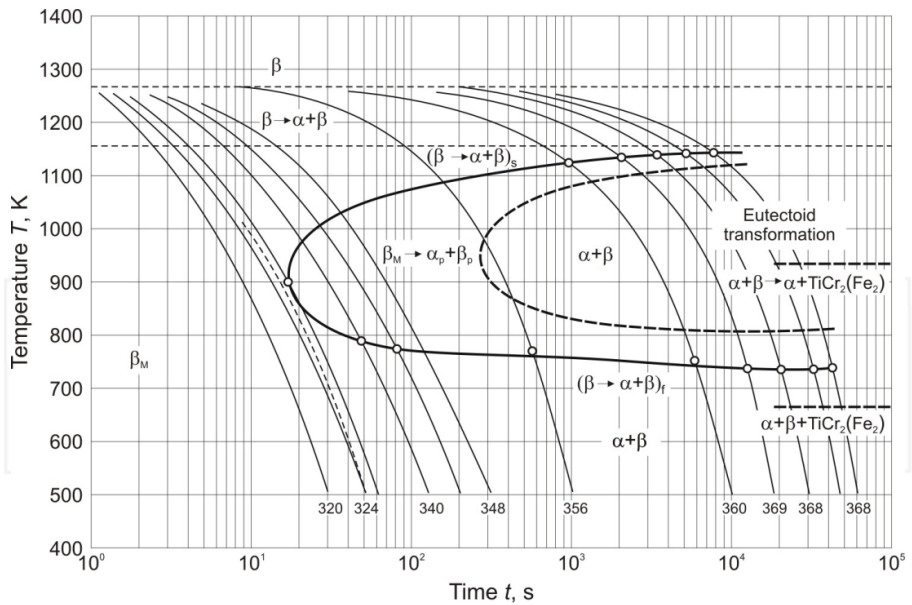


Figure 3. Fig. 3.CCT diagram for Ti-6Al-5Mo-5V-1Cr-1Fe alloy.

Cooling rate, °C·s ⁻¹	Phase composition of the alloy		
	Ti-6Al-4V	Ti-6Al-2Mo-2Cr	Ti-6Al-5Mo-5V-1Cr-1Fe
48-18	$\alpha'(\alpha'')$	$\alpha'(\alpha'')$	β_M
9	$\alpha + \alpha'(\alpha'')$	$\alpha + \alpha'(\alpha'') + \beta$	$\beta_M + \alpha$
7	$\alpha + \alpha'(\alpha'')$	$\alpha + \alpha'(\alpha'')_{trace} + \beta$	$\beta_M + \alpha$
3.5	$\alpha + \alpha'(\alpha'')_{trace} + \beta$	$\alpha + \alpha'(\alpha'')_{trace} + \beta$	$\beta_M + \alpha$
1.2-0.04	$\alpha + \beta$	$\alpha + \beta$	$\alpha + \beta$
0.024-0.004	$\alpha + \beta$	$\alpha + \beta + TiCr_2$	$\alpha + \beta + TiCr_2(Fe_2)$

Table 3. Phase composition of the selected titanium alloys after controlled cooling from the β -phase range [6,9]

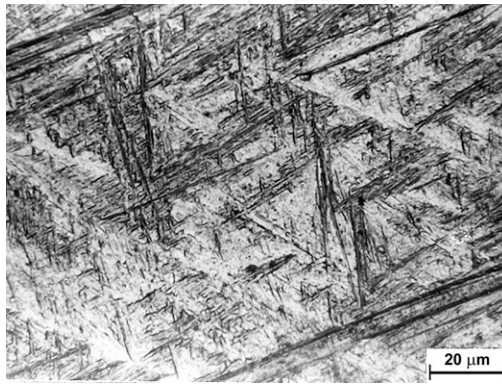


Figure 4. Martensitic microstructure of Ti-6Al-2Mo-2Cr alloy after cooling from 1050°C at a rate of 48°C s⁻¹ (LM-DIC micrograph).

The important parameters for a lamellar microstructure with respect to mechanical properties of the alloy are the β -grain size, size of the colonies of α -phase lamellae, thickness of the α -lamellae and the morphology of the interlamellar interface (β -phase) (Fig. 6) [12-13].

Increase in cooling rate leads to refinement of the microstructure – both α colony size and α -lamellae thickness are reduced. Additionally new colonies tend to nucleate not only on β -phase boundaries but also on boundaries of other colonies, growing perpendicularly to the existing lamellae. This leads to formation of characteristic microstructure called “basket weave” or Widmanstätten microstructure (Fig. 7) [3].

4. Tensile and fatigue properties

Mechanical properties of two phase titanium alloys strongly depend on morphology of particular phases. In the case of the alloys with lamellar microstructure, the thickness of α lamellae and diameter of their colonies have the most significant influence [3,14].

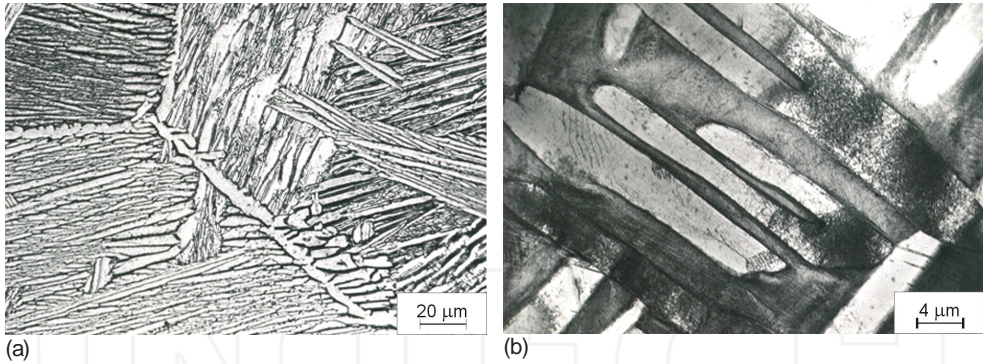


Figure 5. Microstructure of Ti-6Al-2Mo-2Cr alloys after cooling from 1050°C at a rate of 1.2°C s⁻¹: a) LM micrograph, b) TEM micrograph.

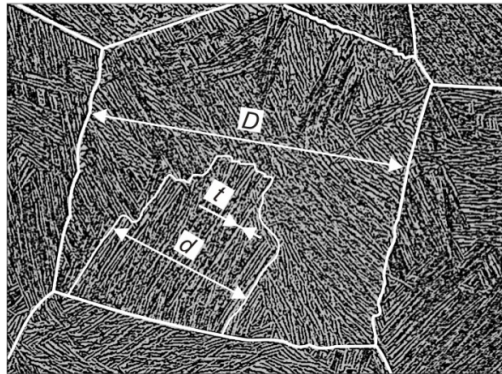


Figure 6. Stereological parameters of lamellar microstructure: D – primary β -phase grain size, d – size of the colony of parallel α -lamellae, t – thickness of α -lamellae.

Refinement of the microstructure results in higher yield stress (Fig. 8a). However the increase of yield stress is moderate unless martensitic phase is present. Tensile elongation increases with increasing cooling rate at first (Fig. 8b). However, after reaching maximum the ductility curve declines. Such behaviour was reported earlier and attributed to the change of fracture mode from ductile transcrystalline for low cooling rates to ductile intercrystalline fracture along continuous α phase layers at primary β grain boundaries [6,8].

The size of the colonies of α lamellae having the same crystallographic orientation have significant influence on the mechanical properties of the alloy as it is a measure of effective slip length [8,15]. However transition to the 'basket weave' type of microstructure makes the determination of colonies size even more difficult. Because of that the thickness of α -lamellae was also taken into account as the quantitative parameter illustrating the effect of microstructure refinement on mechanical properties.

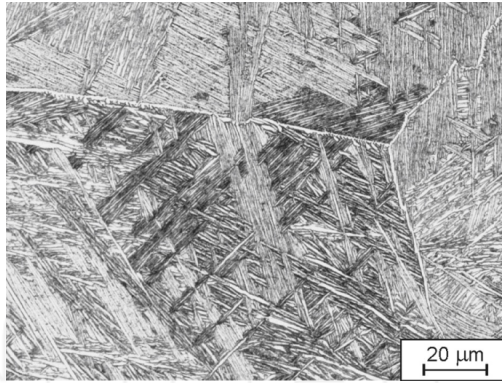


Figure 7. "Basket-weave" or Widmanstätten microstructure of Ti-6Al-4V alloy after cooling from β -phase range at the rate of 9°C s^{-1} .

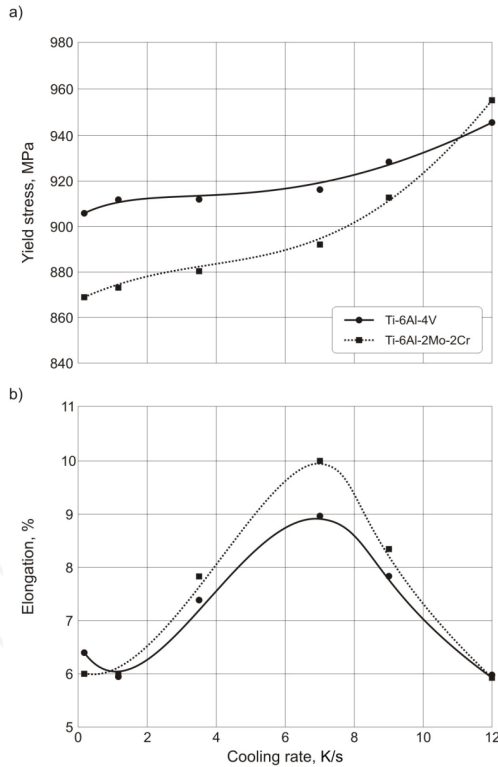


Figure 8. Yield stress and tensile elongation dependence on the cooling rate from β -phase range for selected titanium alloys.

Average thickness of α -phase lamellae, μm	YS, MPa	UTS, MPa	σ_f , MPa	HV
2.4	970	1115	565	326
3.0	928	1068	580	330
5.5	916	1056	570	325
7.6	908	1038	560	336

σ_f – fatigue strength at 10^7 cycles in rotational bending test.

Table 4. Mechanical properties of the Ti-6Al-4V alloy

Average thickness of α -phase lamellae, μm	YS, MPa	UTS, MPa	σ_f , MPa	HV
1.7	980	1136	550	340
2.0	944	1108	575	338
3.6	924	1055	560	336
6.2	922	1024	540	332

Table 5. Mechanical properties of the Ti-6Al-2Mo-2Cr alloy.

Average thickness of α -phase lamellae, μm	YS, MPa	UTS, MPa	σ_f , MPa	HV
0.8	1235	1305	540	348
1.5	1225	1285	555	340
2.6	1186	1262	535	342
4.3	1160	1236	520	336

Table 6. Mechanical properties of the Ti-6Al-5Mo-5V-1Cr-1Fe alloy.

Following values of geometrical parameters of lamellar α -phase, i.e. thickness of the α -lamellae (t) and diameter of the α -phase lamellae colony (d), provided maximum fatigue strength of the investigated alloys:

- Ti-6Al-2Mo-2Cr $t = 2 \mu\text{m}$, $d = 20 \mu\text{m}$,
- Ti-6Al-4V $t = 3 \mu\text{m}$, $d = 30 \mu\text{m}$,
- Ti-6Al-5Mo-5V-1Cr-1Fe $t = 1.5 \mu\text{m}$, $d = 35 \mu\text{m}$.

Fatigue fracture surfaces showed transgranular character with typical ductile surroundings of β -phase around α -phase (Fig. 9). Size of the dimples were closely related to thickness of the α -lamellae and size of the colonies of parallel α -lamellae [16]. No pronounced beach markings

or striations were identified which is an evidence of frequent change of the crack growth direction. This phenomenon along with secondary crack branching are important reasons for advantageous effect of lamellar microstructure on fatigue behaviour.

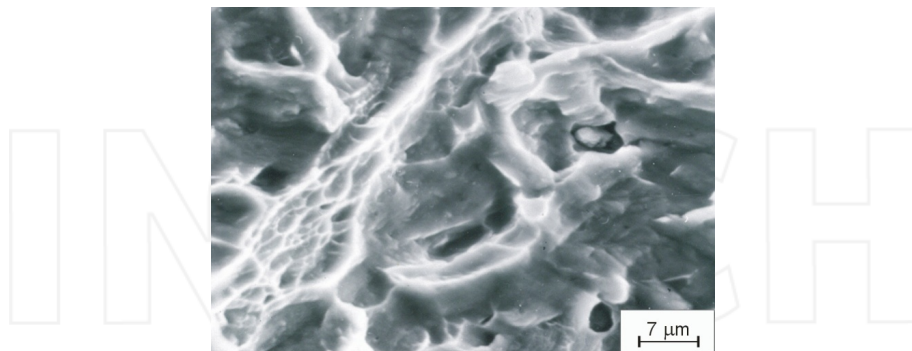


Figure 9. Fatigue fracture surfaces of Ti-6Al-5Mo-5V-1Cr-1Fe alloy cooled from 1020°C at a rate of 0.8°C s⁻¹.

The progress of the crack tip through regions of interfacial β -phase is accompanied by the absorption of large amount of energy due to intensive plastic deformation, contributing to lowering the rate of crack propagation. When thickness of β -phase regions decreases, it cannot absorb sufficient amounts of energy and retard the crack propagation.

5. Summary

Microstructure of two-phase titanium alloys after deformation or heat treatment carried out at a temperature in the range of β -phase stability depends on cooling rate. High cooling rates ($>18^\circ\text{C s}^{-1}$) result in martensitic α' (α'') microstructure for alloys having β stability factor $K_\beta < 1$ and metastable β_M microstructure for alloys with higher contents of β -stabilizers. Low and moderate cooling rates lead to development of lamellar microstructure consisting of colonies of α -phase lamellae within large β -phase grains. Decrease of cooling rate cause increase both in thickness of individual α -phase lamellae and size of the colonies of parallel α -lamellae. This in turn lowers yield stress and tensile strength of the alloys.

Lamellar α -phase microstructure of the alloy heat treated in the β -range has beneficial effect on its fatigue behaviour. This is the result of frequent change in crack direction and secondary crack branching. When α -phase lamellae are too large thin layers of β -phase are not capable to absorb large amounts of energy and retard the crack propagation. In this case the colony of the α -phase lamellae behaves as singular element of the microstructure. This phenomenon is more intensive in the alloy with smaller value of K_β coefficient (Ti-6Al-4V). Sufficient thickness of β -phase surroundings enables absorption of energy in the process of plastic deformation of regions ahead of the crack tip, contributing to slowing the rate of crack propagation and therefore increasing fatigue life.

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