THE PHASE AT THE α/β INTERFACE IN THE Ti 685 ALLOY

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Introduction

The TA 6 Zr 5 D titanium alloy (Ti 685), the weight composition of which is given in Table 1, is very used by the Air Service Engineering because of its convenient mechanical properties of creeping in the 500-550°C temperature range. In fact it is particularly available to make the compressor discs of the turbo-motors of the firms ROLLS-ROYCE, TURBOMECA, SNECMA ...

From 1976 [1] we have investigated this alloy in order :

- firstly to define the different fields of transformation which may occur during various types of continuous or interrupted coolings, started from 1050°C, after an homogeneization of the alloy, in the β_0 field or from 1000°C in the $(\alpha+\beta)$ field. The cooling rates (as referred to V_R in the following sections) were ranging from 50°C/ s (quenching) to 0.0005°C/s (slow cooling);
- secondly, the different types of microstructures formed during or at the end of these coolings were studied.

During the drawing of the TRC (Transformations occuring during continuous cooling) and the TRI (Transformations occuring during quenching after continuous cooling) diagrams and for the slow cooling rate field, a phase as referred to L was found at the α/β interface. The present paper reports a further investigation of this L phase.

Table 1 . Chemical compositions

Materials	compositions, wt%							
	Al	Zr	Mo	Si	C	Fe	H	Sn
Ti 685	6.05	5.1	0.6	0.28	0.01	0.02	0.0028	-
Ti 6242	6.2	3.8	2	-	0.021	-	0.0016	1.95
Ti 651	6.23	4.22	0.95	0.262	0.035	0.023	0.008	2.2
T 40	-	-	-	-	0.01	0.04	0.002	-

Experimental procedure

The samples were solutionized in the β_0 field or the $(\alpha+\beta)$ fields for one hour. The L phase at the α/β interface was mainly studied by the TEM (Transmission Electron Microscopy and electron microdiffraction) with a JEOL 100 C at a tension of 100 kV.

The thin foils were obtained by conventional electropolishing technics indicated by BLACKBURN [2].

The chemical composition of the phases present in the alloy were determined :

- either in bulk samples by quantitative analysis made with a CAMEBAX apparatus which is both a scanning microscope and an electron microprobe (for the Ti 685 alloy, the irradiated volume being about 2 μm^3);

- or in thin foils with the STEM (Scanning Transmission Electron Microscope).
In this device, the size of the microprobe may reach 150 Å and the analysis was made using the technics of compound standard samples [3].

Results

The TRC diagram pointing out the L phase field

For the Ti 685 alloy, the TRC diagram obtained from samples solutionized at 1050°C in the β_0 field is shown in Figure 1. In order to draw this diagram, more numerous methods of investigation than those indicated in the preceding section were necessary, which are in particular, the dilatometric and crystallographic analysis, the optical microscopy carried out either at room temperature or with a hot-stage equipment and the hardness measurements .

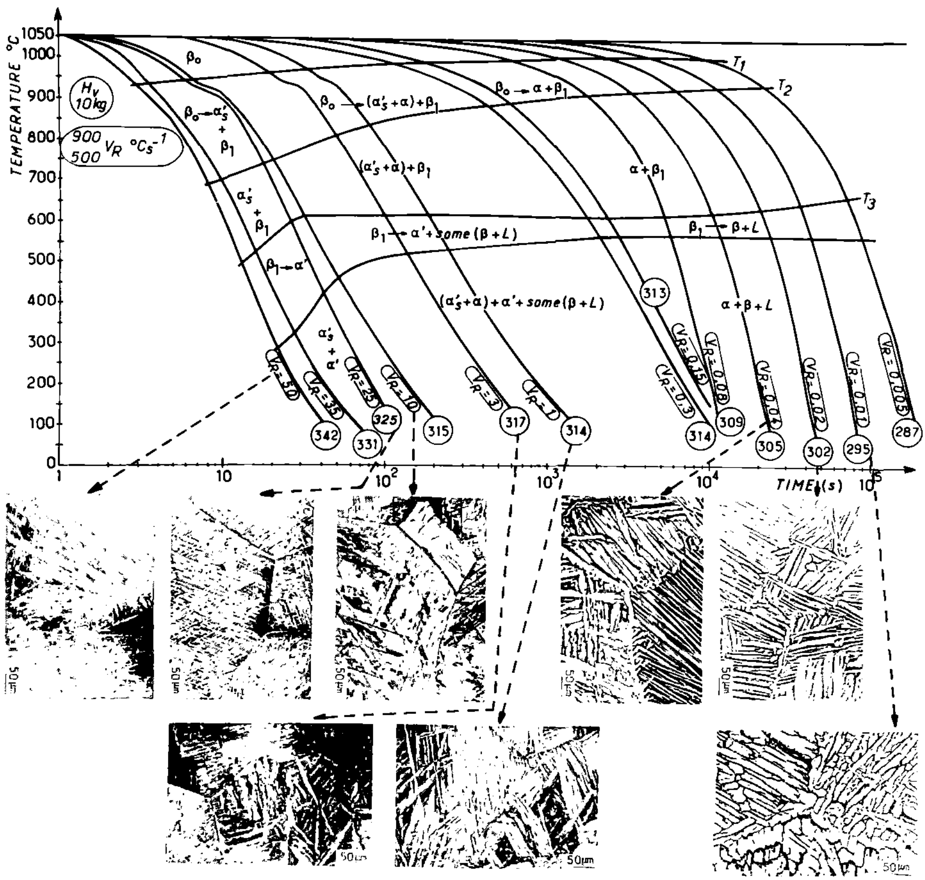

Fig.1 T.R.C. diagram of Ti 685

At 20°C, three different types of microstructures are observed depending upon the rate V_R used to cool the sample from the temperature of homogeneization T_H :

- the martensite structure field for $V_R \geqslant 10°C/s$,
- the Widmanstätten structure field for $V_R \leqslant 0.1°C/s$,
- and the mixed structure field which are used by the Air Service Engineering ($0.1 \leqslant V_R \leqslant 10°C/s$).

On the optical micrographs of the Widmanstätten structure, only the two α and β phases are observed. But with the scale of the electron microscopy this structure appears more complicated. In fact at the α/β interface, the L phase was found.

Morphology of the L phase

It mainly appears to be formed of two layers which can or cannot exist together:
- either a monolithic layer, called ML (Fig. 2-a) (Bright Field),
- or a striated layer, called SL (Fig. 2-b) (Bright Field).

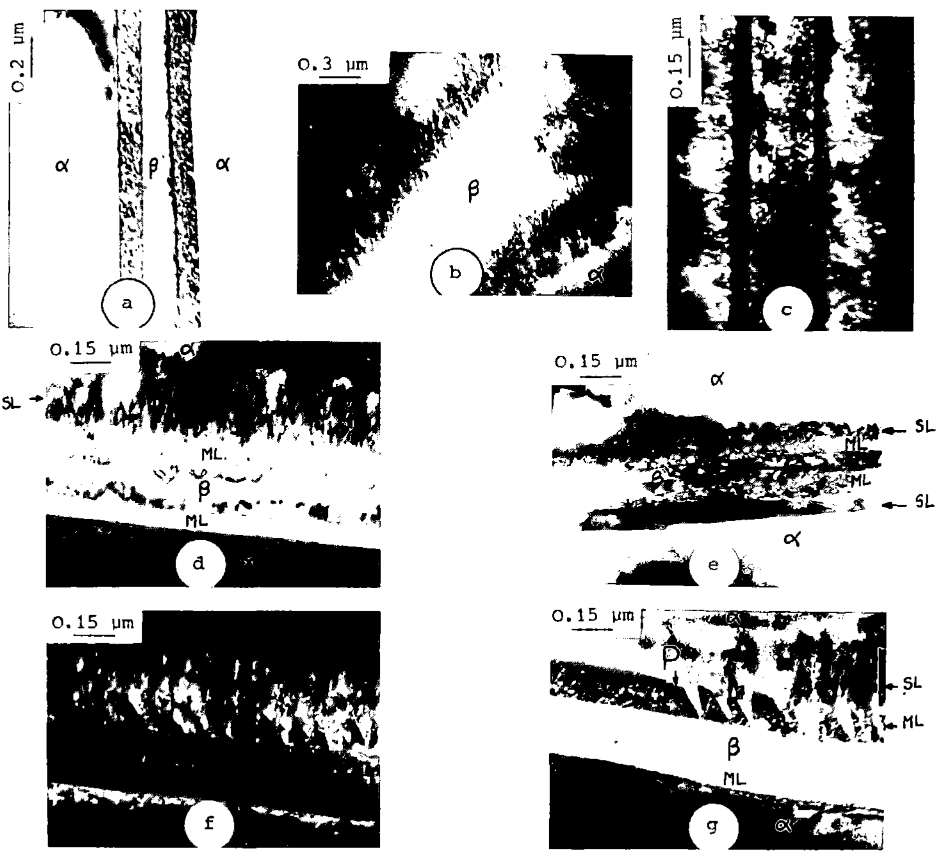

Fig.2 T.E.M. of Ti 685 cooled 0.01°C/s from 1050°C to 20°C illustrating monolithic (ML) and striated (SL) layer.

In figure 2-c (Dark field) the ML and Sl layers are observed simultaneously and are situated symmetrically on both sides of the β phase. Furthermore, these two forms can exist dissymetrically on both of the β phase as illustrated by the figure 2-d (Bright field). It must be indicated that the striations of the SL layer are more or less well defined as shown in figure 2-e (Bright field). In this particular case the SL/α interface is nearly a straight line.

When the striations of the SL layer are well defined, it can be noticed that they initiate at the ML/β interface in the ML layer and extend into the α phase. So the SL/α interface is very serrated as shown in figure 2-f (Dark field).

The striations of SL layer are not always present along the ML layer as illustrated by figure 2-g (Bright field). In fact, from the point P, it can be noted that these striations disappear. Sometimes in the ML layer, one can observe a tangle of lines which may reflect dislocations lines (see Fig.2-a).

At the SL/α interface, some dislocations lines are sometimes observed.

For a given VR, the width of the L phase is largely varying from one α/β interface to another. However, a mean value of the L phase width has been calculated and plotted as a function of VR in figure 3.

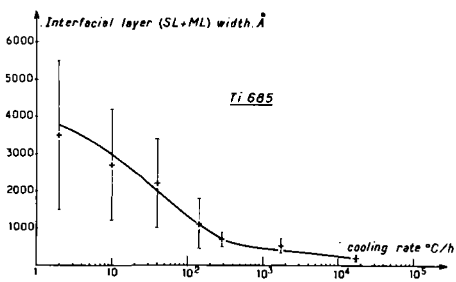

Fig. 3

Interfacial layer width as a function of cooling rate from 1050°C to 20°C.

This mean value represents the sum of the widths of the ML and SL layers. The important error plotted for this width value represents the extreme value of the size of the interface phase found in the thin foil and shows the great heterogeneity of this interface.

Crystallography of the L phase

- The diffraction patterns, the bright and dark fields micrographs show that the ML phase has a f.c.c. crystallographic structure with a parameter $\underline{a} \simeq 4.37 \pm 0.10$ Å. Between the ML, β and α phases, two different orientation relationships were found in the same sample.

Relation (1) $\{110\}_\beta \ // \ (00.1)_\alpha \ // \ \{111\}_{ML}$
$\langle 111\rangle_\beta \ // \ \langle 1\bar{2}10\rangle_\alpha \ // \ \langle 110\rangle_{ML}$

This relation represents the well known orientation relationship of Burgers between the α and β phases and that of Kurdjumov and Sachs between the β and ML phases.

Relation (2) $\{110\}_\beta \ // \ (00.1)_\alpha \ // \ \{001\}_{ML}$
$\langle 111\rangle_\beta \ // \ \langle 1\bar{2}10\rangle_\alpha \ // \ \langle 110\rangle_{ML}$

The relation (1) which exhibits the parallelism between the closed packed crystallographic planes and directions of the different phases which exist is mostly observed.

- The SL phase has also a c.f.c. crystallographic structure with the same parameter $\underline{a}$ as that of the ML phase. When the striations of the SL phase are well defined, they give arced reflections on the diffraction patterns.

This well illustrates the slight desorientation which exists between the striations.

In the case where the α/β interface is composed of both the ML and SL phases, the two following orientation relationships were determined between the different phases .

Relation (3)	$\{110\}_\beta$	//	$(00.1)_\alpha$	//	$\{111\}_{ML}$	//	$\{111\}_{SL}$
	$<111>_\beta$	//	$<1\bar{2}10>_\alpha$	//	$<110>_{ML}$	//	$<110>_{SL}$
Relation (4)	$\{110\}_\beta$	//	$(00.1)_\alpha$	//	$\{00.1\}_{ML}$	//	$\{111\}_{SL}$
	$<111>_\beta$	//	$<1\bar{2}10>_\alpha$	//	$<110>_{ML}$		$<110>_{SL}$

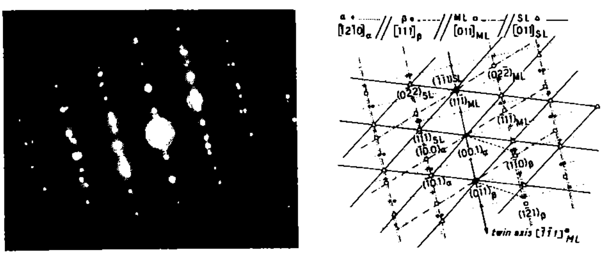

Fig.4 S.A.D. pattern illustrating Burgers relation between α and β and twin relation between ML and SL.

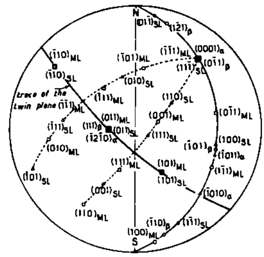

Fig.5 Stereographic projection representing twin relation.

Figure 4 illustrates the relation (3) which was mostly found and shows that the diffraction patterns given by both the ML and SL phases are symmetrical from the $<111>^*_{ML,SL}$ axis. Therefore a twin relationship exists between these two ML and SL phases. The twin plane is $\{111\}_{ML,SL}$. The stereographic projection represented in figure 5 shows this twin relationship. Furthermore this relationship is confirmed by the diffraction pattern shown in figure 6, which exhibits another orientation between the α and β phases which results from that of Burgers (see the stereographic projection shown in figure 7).

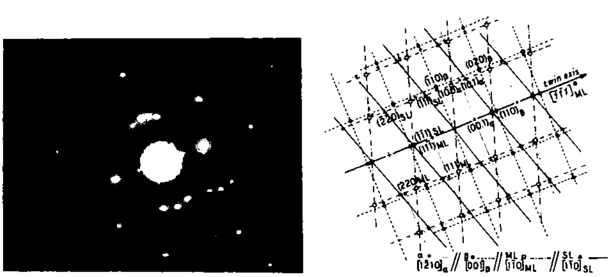

Fig.6 S.A.D. pattern confirming twin relation between ML and SL.

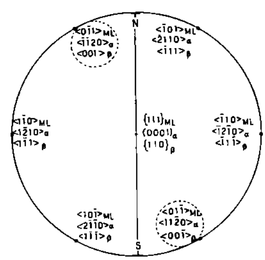

Fig.7 Stereographic projection illustrating Burgers relation and its consequences.

Chemical composition of the L phase

The chemical analysis was made with a Ti 685 sample which has been previously solutionized at 1050°C and then cooled to 20°C with $V_R \simeq 0.01$°C/s [4]. In figure 8, the bright field micrograph of an interface zone is shown. In this zone, around the electron beam,dark spot appears due to the contamination of the thin foil which occurs during the analysis. However it must be noted that the size of this spot is greater than that of the microprobe itself which appears like a light spot. The results of the quantitative analysis expressed as (wt %) are indicated in Table 2. The error which was assigned to each value depends upon the number of measurements carried out in different zones of the thin foil.

From this Table, it can be noted that the ML and Sl phases have the same chemical composition. The X-ray emission spectrum relative to the ML and SL phases is shown in figure 9.

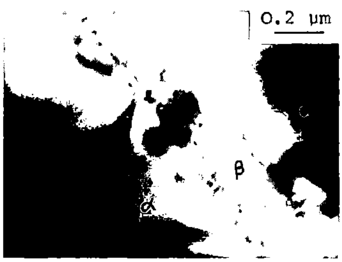

Fig.8 T.E.M. illustrating the analysed zone.

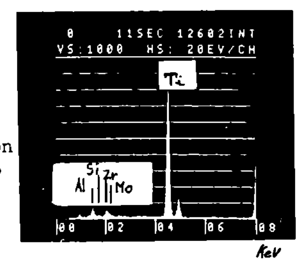

Fig.9 X-Ray emission spectrum relative to ML and SL.

Table 2 Chemical compositions (STEM analysis)

Phase	compositions, wt%				
	Ti	Al	Zr	Mo	Si
α	88.4 ± 2.5	6.8 ± 1.0	4.6 ± 1.0	-	0.22 ± 0.10
β	75.5 ± 2.5	2.0 ± 0.6	6.3 ± 2.0	15.7 ± 2.5	0.48 ± 0.20
SL ML	88.8 ± 2.5	5.9 ± 1.0	4.5 ± 1.0	0.5 ± 0.3	0.26 ± 0.10

Formation conditions of the L phase

- In the Ti 685 alloy

In order to determine the conditions for formation of the L phase, the homogeneization of the samples was made in the β_0 field at 1050°C and in the ($\alpha+\beta$) field at 1000°C. For this latter temperature the content in betagene element (Mo) of the β phase is greater than that of the β_0 phase. This fact is evident by considering the pseudo-binary sections published by [5] for the Ti-Al-Mo-Zr diagram. The comparison of the TRC diagram obtained from 1050°C and the TRC diagram draws from 1000°C shows a shift of transformation fields from the slower cooling rates towards the faster cooling rates. In particular at the end of a cooling carried out with $V_R \simeq 5$°C/s, the presence of the L and β phases is clearly detected by the T.E.M.

To determine the temperature from which the formation of the L phase begins, TRI type thermal treatments were carried out.

The following results were obtained from the alloy solutionized at 1000°C, cooled slowly at the rate of 0.04°C/s, then quenched to 20°C:

- the samples quenched from 980–830°C do not exhibit the L phase. In fact, the interface between the α grains consists of acicular martensite, α'_a (see Fig.10). The composition of α'_a is indicated in Table 3 ,
- the samples quenched from temperatures lower than 830°C, present the three phases α , β and L at 20°C. The width of the β phase is decreasing with the quenching temperature T. Moreower, its composition varies continuously with T, in particular its betagene element (Mo) content is increasing when T is decreasing (Table 3).

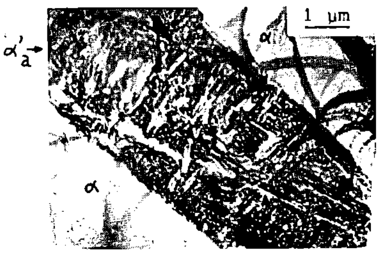

Fig. 10

T.E.M. of Ti 685 cooled 0.04°C/s from 1000°C to 900°C and water quenched illustrating acicular martensite.

Table 3 Chemical compositions of α'_a and β phases

Quenching temperature T	900°C	780°C	450°C
Phases analysed at 20°C	α'_a	β	β
wt% Ti	86	82	79
wt% Al	4	2.7	1.7
wt% Zr	7.4	8	8
wt% Mo	2.3	7	11
wt% Si	0.3	0.3	0.3

- In different titanium alloys

The L phase was found in different titanium alloys : such as the Ti 6242 and Ti 651 [6] , the compositions of which are given in Table 1. In particular, these alloys are more enriched in Mo than the Ti 685. For the same type of thermal treatment (T_H and V_R identical), the Ti 6242 alloy exhibits a L phase, the mean value of the width of which is greater than that observed in the Ti 685 alloy. For $V_R \simeq 0.01$°C/s, its value is 0.4 µm instead of 0.25 µm in the Ti 685. In the titanium of commercial purity T 40, the composition of which is given in Table 1, the L phase does not exist.

Discussion

The existence of an α/β interface phase, with a more or less complicated morphology has been shown by other authors [7, 8, 9] , especially in the TA 6 V alloy. For this alloy the crystallographic structure of the monolithic layer was described as c.f.c. with a parameter $\underline{a} \simeq 4.2$ Å by [7] , and as f.c.t. with a parameter $\underline{a} \simeq 4.34 \pm 0.03$ Å and $c/a \simeq 1.13$ by [9] .The crystallographic structure of the striated layer has been previously described by RHODES [10] as hcp identical to that of the type 2 α. This fact was confirmed by HALL [9].

Recently, RHODES and PATON [11] have shown that the structure of the striated layer may be considered as c.f.c. , with a twin relationship between the ML and SL layers.

The experimental results obtained by the TEM on the Ti 685 alloy and reported in the present paper are in good agreement with those obtained by [11] on the TA 6 V alloy.

The structural transformations occuring during slow continuous cooling and in particular the formation of the L phase in the Ti 685 can be explained as follows :

- from the T_1 temperature until the T_2 temperature (see Fig. 1) by diffusional mechanism, there is a continuous decomposition of the β_0 phase with the c_o Mo content following the reaction

$$\underset{c_0}{\beta_0} \longrightarrow \alpha + \underset{c_1}{\beta_1}$$

The α phase nucleates from the β grain boundaries. These α nuclei and the β_0 phase are related by the orientation relationship of BURGERS. These nuclei grow in the form of α plates. As a function of the decreasing temperature, the growing of these plates is shown in figure 11. With the scale of the hot-stage optical microscopy, it can be noted that there is no formation of new α plates at the β interface between two α plates.

Fig. 11 Hot stage optical micrographs showing the growing of α plates from β boundaries (a,b,c) after polishing and etching (d).

During a TRI treatment, carried out between T_1 and T_2 , there is the formation of the α'_a martensite occuring during quenching according to the following reaction :

$$\alpha + \underset{c_1}{\beta_1} \longrightarrow \alpha + \underset{c_1}{\alpha'_a}$$

The observation by the TEM of the acicular martensite and not of the lath martensite allows us to think that the c_1 Mo content of the β_1 phase is higher than the c_0 Mo content of the β_0 phase. In fact, for a Ti-X binary system, where X is a betagene element, an evolution of the nature of the phase formed during quenching from the β field is observed when the X element of the alloy is increasing. Firstly, lath martensite is formed, secondly acicular martensite and finally equilibrium β phase [12] . Furthermore the analysis of α'_a confirms that $c_1 > c_0$.

- from the T_3 temperature, the ML phase forms. The mechanism which play a role in this formation cannot be indicated in this paper because the in-situ formation of the L phase in the hot-stage TEM has not yet been observed.

Versus the decreasing temperature and to account for the different dilation coefficients of the various phases α, ML, β , defects nucleate in the ML phase and accomodate the thermal stresses. The figure 2-a shows that a great density of dislocations is sometimes observed in the ML phase;

- from the T_4 temperature ($T_4 < T_3$) the stress level is so important that the ML phase shears by mechanical twinning, giving rises to the SL phase. The SL and ML twin phases have of course the same composition confirmed by the STEM analysis.

The striations which composed the SL layer which nucleate from the ML/β interface are in general perpendicular to this interface. These striations can extend into the β phase. The striations generally grow with the long direction parallel to $[00.2]_\alpha$. But for a well-defined orientation between the α, β ,ML and SL phases and taking into account that the α/β interface is firstly a straight line then becomes curvated, the trace analysis of the striations is not very significant.

Conclusions

- In Ti 685, the development of the α/β interface is depending upon both the homogeneization temperature in the β_0 or ($\alpha+\beta$) fields and the cooling rate V_R. Its final thickness is a function of V_R .
- The interface presents two forms : a monolithic (ML) and a striated (SL) layer. These layers have a f.c.c. crystallographic structure and are in general in twin relationship with the $<111>^*_{ML}$ twin axis.
- The formation of the interface and its width is influenced by the Mo content of the β phase.
- The ML and SL layers have identical composition. This latter is very slightly enriched in Mo compared to that of the α phase, so the interface seems to be a transition phase in the $\beta \longrightarrow \alpha$ transformation.

Acknowledgments

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