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Effect of β grain growth on variant selection and texture memory effect during $\alpha \rightarrow \beta \rightarrow \alpha$ phase transformation in Ti–6 Al–4 V

G.C. Obasi^{*}, S. Biroasca, J. Quinta da Fonseca, M. Preuss

Manchester Materials Science Centre, The University of Manchester, Grosvenor Street, Manchester M1 7HS, UK

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Abstract

In the present study, the texture evolution and the role of β grain growth on variant selection during $\beta \rightarrow \alpha$ phase transformation have been investigated in Ti–6 Al–4 V with and without 0.4 wt.% yttrium addition. The aim of adding yttrium was to control β grain growth above the β transus by pinning grain boundaries with yttria. Both materials were first thermomechanically processed to generate similar starting microstructures and crystallographic textures. Subsequently, both materials were solution-heat-treated above the β transus followed by slow cooling to promote growth of the α lath structure from grain boundary α . Additional interrupted slow cooling experiments were carried out to identify the α lamellae that nucleate first from β grain boundaries. Detailed electron backscatter diffraction analysis was carried out and it was found that the β heat treatment did not generate new texture components although the intensities of the individual components changed dramatically depending on the alloy/ β grain size. Variant selection was assessed by comparing measured α texture components with predicted α texture components based on the high-temperature β texture assuming equal variant selection. It was found that with increasing β grain size variant selection intensified favouring the $\{\phi_1, \Phi, \phi_2\} \{90^\circ, 30^\circ, 0^\circ\}$ texture component. Interrupted cooling experiments revealed that α nucleates first on β grain boundaries that are formed by two β grains having a common (110) normal and that these α lamellae display almost exclusively a $\{\phi_1, \Phi, \phi_2\} \{90^\circ, 30^\circ, 0^\circ\}$ orientation. Consequently, the dominance of this variant with increasing β grain size can be related to the relative free growth of this particular α texture component into an “empty” β grain.

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1. Introduction

Microstructures in most engineering alloys are controlled by thermomechanical processing. Titanium alloys are no exception and due to their low temperature hexagonal close packed (hcp) crystal structure, they are also particularly prone to develop strong crystallographic deformation textures that can significantly affect the mechanical properties of the final product [1]. Titanium, as well as zirconium, exhibits an α hcp phase at low temperature, which transforms to a β body-centred-cubic (bcc)

phase at high temperature. By adding β stabilizing alloying elements it is possible to design two-phase titanium alloys, which generally display superior mechanical properties compared to a single-phase alloy. Titanium alloys are usually processed from an ingot by using vacuum melting and subsequently subjected to a series of thermomechanical processing stages in the single β phase and the $\alpha + \beta$ phase field [2]. β heat treatment is part of the processing route in order to chemically homogenize the material. Due to the cleanliness of typical titanium alloys, and hence the absence of carbides or other high-temperature non-metallic precipitates, heat treatment in the single-phase β region leads to very rapid grain growth. The phase transformation during heating ($\alpha \rightarrow \beta$) and cooling ($\beta \rightarrow \alpha$) is governed by the so-called Burgers orientation relationship $\{0002\}_\alpha \parallel \{110\}_\beta$

^{*} Corresponding author. Tel.: +44 (0)7941 726825; fax: +44 (0)161 3068840.

E-mail address: Gideon.obasi@postgrad.manchester.ac.uk (G.C. Obasi).

and $\langle 11-20 \rangle_{\alpha} \parallel \langle 111 \rangle_{\beta}$ with six possible β orientation variants during the $\alpha \rightarrow \beta$ phase transformation and 12 possible α orientations that can transform from a single parent β grain during $\beta \rightarrow \alpha$ phase transformation [3]. The two parallel planes are both closest packed planes while the two parallel directions correspond to the nearest neighbour direction [4]. The crystallographic orientation relationship is represented schematically in Fig. 1. If all 72 variants occur with equal probability during an $\alpha \rightarrow \beta \rightarrow \alpha$ phase transformation the final α texture should be essentially randomized regardless of the initial α texture. However, it has been demonstrated by a number of researchers [5–8] that during both displacive and diffusional phase transformation of titanium and zirconium alloys, the initial α texture is to a certain extent retained after a full heat treatment cycle. This “texture memory” effect has been reported widely for single-crystal titanium [7] and polycrystalline titanium and zirconium alloys [8,9] and has often been attributed to variant selection. For example, for zirconium and titanium, Romero et al. [5] and Lonardelli et al. [6] demonstrated that perfect texture memory is a result of incomplete α -to- β phase transformation. Although it is clear that any retained α during heat treatment can act as a nucleation site during cooling, it was also suggested that the absence of β grain growth due to the pinning of retained α might contribute to the perfect memory effect [5].

Variant selection mechanisms in both bcc and hexagonal materials during phase transformation have often been linked to transformation strain and elastic anisotropy [10,11]. In the case of β grains transforming to α under unconstrained conditions, contraction in the range of 10% is expected along a $\langle 100 \rangle_{\beta}$ direction and contraction of 2% and expansion of 10% along the two perpendicular $\langle 110 \rangle_{\beta}$ directions [12]. Bhattacharyya et al. [13] reported that in the case of neighbouring β grains with a common

$(110)_{\beta}$ pole, $(0002)_{\alpha}$ variants are selected on both sides of the prior β grain boundary with almost the same crystallographic orientation. The two $(0002)_{\alpha}$ variants are closely aligned with the common $\langle 110 \rangle_{\beta}$ pole in preference to the other five $\langle 110 \rangle_{\beta}$ poles. Even though numerous studies on the crystallography of these “special β grain pairs” [14–16] confirmed the suggestion made by Bhattacharyya, the fundamental concept responsible for this variant selection mechanism is poorly understood. This observation is of fundamental importance from a practical point of view since these large regions of practically identical crystallographic orientation are suspected to be the root source for the formation of macrozones [17]. More recently, Crayon [18] suggested that the β grains with the common $(110)_{\beta}$ planes are β variants inherited from the α parents that transform to β during the α - β phase transformation.

The aim of the present work has been to shed further light on the α variant selection mechanism described by Bhattacharyya et al. [13], to investigate the effect of β grain growth and also to study its practical impact on texture formation. Since β grain growth is very difficult to control in an alloy such as Ti–6 Al–4 V alloy, a second variant was produced that includes 0.4 wt.% yttrium in order to pin β grain boundaries in the high-temperature single-phase region [19]. In-house thermomechanical processing was carried out to achieve the best comparable microstructure and texture of the two alloys before undertaking β heat treatments to study variant selection.

2. Experimental methods

2.1. Material

The material used in the present study was Ti–6 Al–4 V produced by conventional ingot metallurgy and uniaxially rolling and provided by TIMET, UK. Some of this material was remelted at the University of Birmingham to add 0.4 wt.% of yttrium. The nominal chemical compositions of both alloys are shown in Table 1. Differential scanning calorimetry (DSC) measurements of both alloys determined a β transus of 990 °C, confirming indirectly that the remelting procedure for the yttrium-containing alloy had not changed the alloy chemistry apart from the yttrium addition (no significant oxygen pickup). Blocks of 40 × 40 × 20 mm were cut from both alloy variants and heat-treated at 1050 °C followed by water quenching. These blocks were subsequently unidirectional-rolled using a calibrated furnace set to 950 °C. The soaking time of the samples was 1 h before starting the rolling process. A total reduction of 75% was achieved by having a 15% reduction per pass with 2 min reheating between each pass. The rolling and heat treatment processes are shown schematically in Fig. 2.

The rolled material was subsequently heat-treated for 8 h at 950 °C followed by 1 °C min^{−1} cooling in a tube furnace under argon atmosphere. Samples with dimensions of 10 × 10 × 5 mm³ were cut from the centre of the small

Fig. 1. Schematic representation of the orientation relationship between the α phase (hcp) and β phase (bcc) showing the Burgers relationship (BR).

Table 1
Elemental compositions of the two variants of Ti–6 Al–4 V alloy.

Elements	Fe	V	Al	C	O	N	B	Y	Ti
wt. %	–	4.0	6.0	0.1	0.17	0.3	.005	.001	Bal.
wt. %	–	4.0	6.0	0.1	0.17	.03	.005	0.4	Bal.

Fig. 2. The schematic representation of β quenching (H), unidirectional rolling (UD), recrystallization (RX) and solution heat treatment (β -HT) of Ti–6 Al–4 V and Ti–6 Al–4 V–0.4 Y processes.

slabs and subjected to β heat treatments in an argon atmosphere at 1050 °C and 1150 °C for 30 min and again subsequently cooled at 1 °C min^{−1} to room temperature.

Mid-sections of the β heat-treated samples were prepared following standard grinding/polishing procedures and etched in 2% HF, 8% HNO₃ and 90% H₂O (Kroll's reagent) for 10 s for optical inspection. Electron backscatter diffraction (EBSD) was used to characterize the texture inherited from the $\alpha \rightarrow \beta \rightarrow \alpha$ transformation. To improve Kikuchi pattern quality, the samples were OPS polished (1 h) and etched repeatedly with the last polishing step taking 2 h.

A CamScan field emission gun scanning electron microscope (FEGSEM), equipped with HKL Channel 5 software, was used for EBSD data acquisition and analysis. Due to the coarse β grain size and large α colony structure, a step size of 10 μ m was used, which allowed covering an area of 7 × 7 mm² during each scan in order to measure a reasonable number of β grains for texture analysis. The EBSD orientation data were used to plot pole figures and the orientation distribution functions (ODFs). One of the experimental difficulties when studying variant selection is the requirement to determine the high-temperature β texture. The most reliable in situ techniques, which are not compromised by surface effects, are neutron and high-energy synchrotron X-ray diffraction. Although they can provide excellent information on the global texture, they do not provide information on the microstructural scale, i.e. information on neighbouring β grains. In order to determine the high-temperature β texture, the α orientation maps, recorded by EBSD, were used to reconstruct the β grain structure based on the Burgers orientation relationship between the parent and daughter phases. The software used for this β reconstruction has been developed at the University of Sheffield within IMPETUS (Institute for Microstructural and Mechanical Process Engineering). The software checks each α lath and identifies the neighbouring

laths that are most likely to have the same parent β orientation. In order to obtain a successful reconstruction a minimum allowable misorientation angle between α variants of a common β grain is defined, which in the present case was 3°. Further details on the methodology are described in Refs. [20–23] and details regarding the software can be found in Ref. [20]. The information from the reconstruction was subsequently reloaded into Channel 5 to plot the β orientation maps. These maps were also used to measure β grain size distributions based on the linear intercept method (ASTM E112) [24] and identify neighbouring β grains with common $\langle 110 \rangle$ poles. The total number of measured β grains was ~ 150 for each temperature condition (1050 °C and 1150 °C). In order to assess the level of variant selection the expected α texture, assuming no occurrence of variant selection during the $\beta \rightarrow \alpha$ transformation, was calculated from the reconstructed β texture. The transformations and the representations of texture (ODFs and pole figures) were carried out using TEXTAN-III, an in-house software developed by Bate [25]. In TEXTAN-III, the Burgers orientation relationship is applied as a rotation of Euler angles, taking into account the crystal symmetries of the parent and the daughter crystals. It applies the transformation to experimental data by using the spherical harmonic method proposed by Bunge [26].

3. Results

3.1. Microstructure and texture evolution

Optical micrographs of Ti–6 Al–4 V with and without yttrium are shown in Fig. 3a and b after recrystallizing the material at 950 °C for 8 h. The figures reveal in both cases a fully equiaxed primary α microstructure with a mean grain size of 18 μ m and retained β at the triple junctions of primary α (α_p). The α textures of the two alloys at this stage are presented in the form of ODF sections and representative quarter pole figures, in Fig. 4. The ODF sections presented here are the $\varphi_2 = 0$ for α phase texture and $\varphi_2 = 45$ for β phase texture [27].

The ODF sections at $\varphi_2 = 0^\circ$ of each alloy have a strong transverse texture component at Euler angles $\{\varphi_1, \Phi, \varphi_2\}$ $\{0^\circ, 90^\circ, 0^\circ\}$ with maximum intensity of 16 times random intensity. This orientation corresponds to the basal pole components rotated towards TD (transverse direction during rolling) in the (0002) pole figure as shown in Fig. 4c and d. This is the so-called transverse texture often found in Ti–6 Al–4 V deformed at temperatures above 900 °C. The second component is a weak orientation component

Fig. 3. Optical micrographs of samples recrystallized (RX) at 950 °C for 8 h: (a) Ti-6 Al-4 V and (b) Ti-6 Al-4 V-0.4 Y.

Fig. 4. EBSD measured ODF ($\phi_2 = 0$) section and pole figure of α RX texture: (a) Ti-6 Al-4 V, (b) Ti-6 Al-4 V-0.4 Y, (c) Ti-6 Al-4 V, and (d) Ti-6 Al-4 V-0.4 Y.

at Euler angles $\{\phi_1, \Phi, \phi_2\}$ $\{90^\circ, 30^\circ, 0^\circ\}$ with maximum intensity of six times random intensity. This orientation corresponds to basal poles rotated about 30° away from ND (=normal direction to the rolling plane) towards RD (rolling direction) in the $(0002)\alpha$ pole figure. The similarity of the grain morphology and the α textures in both alloys demonstrates that yttrium additions did not have any significant effect on the microstructure produced during thermomechanical processing. Hence, an almost identical starting microstructure was achieved in the two alloys before undertaking any β heat treatment for variant selection studies.

Optical micrographs of the fully transformed lamellar microstructures are presented in Fig. 5 for Ti-6 Al-4 V and Ti-6 Al-4 V-0.4 Y heat-treated at 1050 °C and 1150 °C. The microstructures are populated by large lamellar colonies of α . In the case of conventional Ti-6 Al-4 V the prior β grains are easily identified by continuous film

of grain boundary α (α_{GB}) in Fig. 5a and b, whereas in the yttrium-containing alloy, Fig. 5c and d, grain boundary α was less distinct. However, in both cases it was possible to confirm that α colonies had mainly grown from grain boundary α . There was no indication that α had predominantly nucleated from yttria particles. In order to reduce uncertainty in determining the β grain size in Ti-6 Al-4 V-0.4 Y, all β grain size measurements were based on reconstructed EBSD maps of the β phase. The grain size distribution and mean grain size was determined by defining high-angle boundaries to have a misorientation of at least 5° . Fig. 6 compares the grain size distribution of the two alloys when heat-treated at 1050 °C (Fig. 6a) and 1150 °C (Fig. 6b). It can be seen that conventional Ti-6 Al-4 V displays a wider grain size distribution than the yttrium variant for both heat-treatment temperatures with a significant tail towards grain sizes exceeding 1 mm. Table 2 summarizes the mean grain size values for the four conditions. It shows

Fig. 5. Optical micrographs of β heat-treated samples: (a) Ti-6 Al-4 V at 1050 °C, (b) Ti-6 Al-4 V at 1150 °C, (c) Ti-6 Al-4 V-0.4 Y at 1050 °C, and (d) Ti-6 Al-4 V-0.4 Y at 1150 °C.

Fig. 6. The β grain size distributions of samples after β heat treatment: (a) Ti-6 Al-4 V and Ti-6 Al-4 V-0.4 Y at 1050 °C and (b) Ti-6 Al-4 V and Ti-6 Al-4 V-0.4 Y at 1150 °C.

Table 2

Mean grain size of Ti-6 Al-4 V and Ti-6 Al-4 V-0.4 Y showing β grain size variation at 1050 °C and 1150 °C.

Temperature (°C)	Ti-6 Al-4 V (μm)	Ti-6 Al-4 V-0.4 Y (μm)	Bin width (μm)
1050	450	210	100
1150	480	250	100

that the mean β grain size of conventional Ti-6 Al-4 V is about twice as large as for Ti-6 Al-4 V-0.4 Y when heat-treated at the same temperature. It can also be seen that increasing the temperature to 1150 °C further coarsens the β grains in both alloys although the temperature effect is small compared to the effect of yttria-pinning β grain boundaries.

The α transformation texture of both alloys after β heat treatment at 1050 °C and 1150 °C are shown in the ODF sections at $\varphi_2 = 0^\circ$ in Fig. 7. The two alloys show similar texture components but significant differences in the texture intensity. The α textures of the conventional Ti–6 Al–4 V presented in Fig. 7a and b are composed of a comparatively weak $\langle 0001 \rangle$ fibre texture component with maximum intensity of six times random while the second component with Euler angles $\{\varphi_1, \Phi, \varphi_2\} \{90^\circ, 30^\circ, 0^\circ\}$ is very strong with maximum intensities of 12 ($T = 1050^\circ\text{C}$) and 14 ($T = 1150^\circ\text{C}$) times random. In the case of the yttrium-containing alloy it is obvious from Fig. 7c and d that the same texture components are found but that the intensities of the $\langle 0001 \rangle$ fibre texture and the second component at the Euler angles $\{\varphi_1, \Phi, \varphi_2\} \{90^\circ, 30^\circ, 0^\circ\}$ are very similar. If one compares the strength of the $\{90^\circ, 30^\circ, 0^\circ\}$ texture component between the four samples it can be seen that the conventional alloy heat-treated at 1150 °C shows the strongest intensity, followed by the conventional alloys heat-treated at 1050 °C and the yttrium variant heat-treated at 1150 °C and at 1050 °C. Therefore, there seems to be a clear link between texture intensity and β grain size. When comparing the ODF sections at $\varphi_2 = 0^\circ$ in Figs. 4 and 7 it also becomes clear that the main texture components are the same even though their intensities are different. This suggests that the transformation texture is affected by texture memory as previously reported for titanium alloys by a number of researchers (for example, Refs. [7,8]).

3.2. Variant selection

As mentioned before, in order to judge the level of variant selection, one has to compare the measured α texture after β -to- α phase transformation with the predicted α texture based on the β texture assuming no variant selection.

Fig. 7. EBSD measured ODF ($\varphi_2 = 0$) section of α transformation texture: (a) Ti–6 Al–4 V β HT at 1050 °C, (b) Ti–6 Al–4 V β HT at 1150 °C, (c) Ti–6 Al–4 V–0.4 Y β HT at 1050 °C, and (d) Ti–6 Al–4 V–0.4 Y β HT at 1150 °C.

This was carried out for the four different conditions in the present study, Ti–6 Al–4 V with and without yttrium and heat-treated at 1050 and 1150 °C. The texture results are presented in Fig. 8 for Ti–6 Al–4 V and Fig. 9 for Ti–6 Al–4 V–0.4 Y. In all cases, the β texture displayed comparatively strong α -fibre and weak γ -fibre texture components, which is a characteristic of hot-rolled bcc materials. The strong α texture at the Euler angles $\{90^\circ, 30^\circ, 0^\circ\}$ was not the product of the strong α -fibre component, as would be expected, but was the product of one of the weak γ -fibre texture components at the Euler angles $\{\varphi_1, \Phi, \varphi_2\} (\{90^\circ, 66^\circ, 0^\circ\} \pm 5^\circ)$ as shown with the dotted line in Fig. 8a. The discrepancy between the modelled and measured α texture for Ti–6 Al–4 V heat-treated at 1050 and 1150 °C is clearly visible when comparing Fig. 8c and d with Fig. 8e and f. The intensities of the two main texture components from the modelled α texture are about six times random intensity while the measured α texture showed 12 times random intensity when heat-treated at 1050 °C and 14 times random intensity when heat-treated at 1150 °C. In the case of Ti–6 Al–4 V–0.4 Y a very different picture is seen. Here, the modelled (Fig. 9c and d) and measured (Fig. 9e and f) α textures look far more similar although the intensities of the measured texture components are still slightly higher than what would be expected from the predictions assuming no variant selection. The intensities of the modelled α texture components is about six times random intensity whereas the measured α texture components show intensities of seven times random intensity when heat-treated at 1050 °C and eight times random intensity when heat-treated at 1150 °C. Both alloys also indicate that the discrepancy between predicted and measured α texture increases with increasing heat treatment temperature.

In summary, the results of the texture measurements clearly indicate that there is a strong correlation between β grain growth and level of variant selection when heat treating Ti–6 Al–4 V and the yttrium-containing variant above the β transus. In the following section, the possible root causes are discussed for such correlation.

4. Discussion

It is clear from the Burgers relationship that the transformation from α to β and back to α , assuming no variant selection, will result in multiplying the α orientation components 72-fold since a single α grain can theoretically give rise to six β orientations and a single β grain should transform to 12 α orientations [27]. Consequently, even an exceptionally strong and sharp α texture should be changed dramatically when such the material is heat-treated above the β transus. In the present investigation, it was observed in Fig. 7 that the β heat treatment did not produce new α texture components compared to the starting recrystallization texture shown in Fig. 4. Instead, only the intensities of the already present components changed with the intensity of the weak $\{\varphi_1, \Phi, \varphi_2\} \{90^\circ, 30^\circ, 0^\circ\}$ texture components

Fig. 8. Ti-6 Al-4 V EBSD reconstructed β texture heat-treated at 1050 °C (a) and at 1150 °C (b). EBSD calculated α texture from β texture according to BR (1050 °C) and (d) at 1150 °C. EBSD measured α texture at 1050 °C (e) and 1150 °C (f).

Fig. 9. Ti-6 Al-4 V-0.4 Y EBSD reconstructed β texture heat-treated at 1050 °C (a) and at 1150 °C (b). EBSD calculated α texture from β texture according to BR (1050 °C) and (d) at 1150 °C. EBSD measured α texture at 1050 °C (e) and 1150 °C (f).

being strengthened whereas the intensity of the $\langle 0001 \rangle$ fibre texture was reduced. The retained texture components are an indication of some texture memory in Ti-6 Al-4 V, which is not related to any retained α grains, as previously reported in Ref. [6], since the material was heat-treated well above the β transus (+50 °C and +150 °C for Ti-6 Al-4 V).

In addition, the present study has demonstrated that with increasing β grain size the $\{90^\circ, 30^\circ, 0^\circ\}$ texture component is strengthened significantly. An important factor of the present work is that the samples were cooled down slowly from above the β transus promoting α lath growth from grain boundary α (which sits on the prior β grain boundaries). Fig. 10a displays the orientation of the α lath structure using IPF colouring in combination with the β grain boundaries (black lines, at least 5° misorientation criterion) that were determined from the β reconstruction.

The analysis of these α colonies in combination with β grain boundaries showed evidence of α variants with similar orientations (same colours) on both sides of the prior β grain as previously reported in Refs. [13,14]. The β grain boundaries of the 150 grains studied here can be divided into two groups: Group 1 represents boundaries with α variants of similar orientations nucleating on both sides of the prior β grain boundaries and Group 2 represents grain boundaries with α variants of different orientations nucleating on both sides of the grain boundaries, see Fig. 10b. As expected, a common $\langle 110 \rangle \beta$ pole of the two neighbouring grains was identified if the grain boundary belonged to Group 1, as demonstrated in Fig. 10c. It was also found that in this case the misorientation between the two common $\langle 110 \rangle \beta$ poles did not exceed 10°. For better visualization, the three-dimensional (3-D) crystal

Fig. 10. (a) α IPF map superimposed on reconstructed β high-angle grain boundary map showing Group 1 (close crystallographic relationship) and Group 2 (without close crystallographic relationship), (b) pole figures of Group 2 without close alignment of (0002) pole with the (110) pole in adjacent β grains, (c) pole figures of Group 1 showing close alignment of (0002) pole with the (110) pole of adjacent β grains, (d) the three-dimensional crystal orientation of group 2 GB α variants and adjacent β grains, and (e) the three-dimensional grain orientation of Group 1 GB α variants and the adjacent β grains.

orientation of the α products and β grains are shown Fig. 10d and e. Although the two β grain orientations in Fig. 10e are not completely similar, they both have one (110) plane with similar poles and consequently α products nucleated on both sides of the β grain boundary with the same (0002) crystallographic orientation. Importantly, these α products correspond in most cases to the texture component with Euler angles $\{\phi_1, \Phi, \phi_2\}$ $\{90^\circ, 30^\circ, 0^\circ\}$, which is the texture component that has become dominant after transformation.

Fig. 11a and b highlights the regions in the orientation map that represent the $\{90^\circ, 30^\circ, 0^\circ\}$ texture component. Since these maps also include the β grain boundaries, it can be seen clearly that most of the $\{90^\circ, 30^\circ, 0^\circ\}$ texture component regions are across β grain boundaries providing evidence that the development of the strong $\{90^\circ, 30^\circ, 0^\circ\}$

texture component is indeed related to the variant selection mechanism of common $\langle 110 \rangle \beta$ poles. Fig. 11 also demonstrates that the $\{90^\circ, 30^\circ, 0^\circ\}$ texture component regions have grown substantially larger in coarse β grains (Fig. 11a) compared to small β grains (Fig. 11b), providing the link between β grain size and dominance of the $\{90^\circ, 30^\circ, 0^\circ\}$ texture component.

Due to the Burgers relationship, α variants within the same β grain can only produce five different misorientation angles and axes, which are $60^\circ/\langle 11-20 \rangle$, $60.83^\circ/\langle 1.377, -1, 2.377, 0.359 \rangle$, $63.26^\circ/\langle 10, 5, 5, -3 \rangle$, $90^\circ/\langle 1, -2.38, 1.38, 0 \rangle$, and $10^\circ/\langle 0001 \rangle$. The ratio between the five is 2:3:2:2:1 in the absence of any variant selection [28]. Since in the present case different levels of variant selections have been observed, the relative misorientation distributions have been plotted and compared to the case of no variant selection in Fig. 12.

Fig. 11. Regions in the IPF orientation map corresponding to the texture component $\{90^\circ, 30^\circ, 0^\circ\}$: (a) Ti-6 Al-4 V and (b) Ti-6 Al-4 V-0.4 Y.

Fig. 12. Relative misorientation angle distribution of inherited α phase showing the misorientations of reference alloy without variant selection and Ti-6 Al-4 V and Ti-6 Al-4 V-0.4 Y HT at 1050 °C and 1150 °C.

Since the 60°, 60.83°, 63.26° misorientation angles were indistinguishable, they were merged in Fig. 12 to a single misorientation angle labelled 60°. Fig. 12 shows that the measured misorientation angle distributions differ from the expected one. The difference is relatively small for Ti-6 Al-4 V-0.4 Y heat-treated at 1050 °C, which displays the smallest β grain growth, and largest in the case of conventional Ti-6 Al-4 V, which also has the largest β grains. This further confirms the increasing level of variant selection with increasing β grain size. It can also be seen that the frequency of the 10° misorientation angle goes up with increasing β

grain size. This can be interpreted as an indication of the increasing length scale of Group 1 β grain boundaries resulting in low misorientation α colonies across the former β grain boundary, which is also evident when comparing Fig. 12a and b.

In order to further understand the significant growth of the (90°, 30°, 0°) texture component with increasing β grain size, early growth of the α lamellae from β grain boundaries was studied by applying interrupted cooling experiments. A Ti-6 Al-4 V sample was heat-treated at 1150 °C for 30 min and cooled at 1 °C min⁻¹ to 975 °C followed by water quenching. Even though fast quenching does not enable one to retain the β phase, the different lath morphology, generated during quenching compared to slow cooling, made it easy to identify the α lamellae that had formed during the early stage of cooling. Fig. 13a and b shows SEM micrographs that were recorded from such samples. A large number of β grain boundaries were studied in this way and it was found that in the early stage of diffusional transformation the nucleation of α products on both sides of β grain boundaries were largely favoured (Fig. 13a) compared to α lamellae nucleating on only one side of the β grain boundaries (Fig. 13b). It has been suggested in Ref. [13] that the interfacial energy of the grain boundary is minimized if the α variants on both sides of the boundaries select one specific (110) β plane common to both β grains. The similar d-spacing values of the (0002) α and (110) β planes would then decrease the critical

Fig. 13. SEM micrographs of Ti-6 Al-4 V β heat-treated at 1150 °C for 30 min and cooled at 1 °C min⁻¹ to 975 °C followed by water quench, showing (a) α lath growing on both sides of prior β grain boundary, (b) α lath on one side of prior β grain boundary, (c) EBSD map of α lath on both sides of prior β grain boundary, (d) EBSD measured ODF and pole figure corresponding to α lath on both sides of prior β grain boundary, and (e) EBSD measured ODF and pole figure corresponding to α lath on one side of prior β grain boundary.

nucleus size for α precipitation and consequently reduce the activation energy for α nucleation. The crystallographic orientation of early grown α variants was investigated by EBSD. Fig. 13c shows α lamellae grown on both sides of a β grain boundary during the early stage of cooling (between the dotted lines). The ODF section of $\varphi_2 = 0$ and quarter pole figure calculated from this region are shown in Fig. 13d and e. It can be seen that these early α lamellae again correspond to the $\{\varphi_1, \Phi, \varphi_2\} \{90^\circ, 30^\circ, 0^\circ\}$ texture component, which is the dominant texture component after $\beta \rightarrow \alpha$ phase transformation shown in Figs. 8 and 9. In contrast, Fig. 13e shows the texture component of an early grown α lamellae region only grown on one side of the prior β grain boundary (rare situation). This α product provides a different texture component contributing to the weak $\langle 0001 \rangle$ fibre texture component seen in Figs. 8 and 9. The nucleation advantage shown here will lead to early growth of α laths on both sides of the grain boundaries until the colony encounters other lamellae growing from different direction. Since at this stage, other α variants are hardly present in the β grain, the growth of this particular α variant is mainly limited by the β grain size. In other words, if the β grain is very large, the α variant nucleating early can grow very far and become a far more dominant texture component (in case this variant is indeed of the same texture component, see Fig. 12a) compared to early α variants growing in a smaller β grain (Fig. 12b). It should be noted that this work did not find any indication that yttria particles themselves acted as early nucleation sites. However, they might have played a minor role and further investigations are continued in this field.

5. Conclusions

Two variants of Ti–6 Al–4 V were studied and heat-treated at two different temperatures in order to shed light on the role of β grain size on α variant selection and texture memory. Grain growth was controlled in one variant by having yttria pinning the β grain boundaries during high-temperature heat treatment. Care was taken to generate similar starting microstructures and textures before β heat treatment to single out β grain size as a variable and study its effect. The ability to reconstruct β grains of EBSD maps enabled a detailed study, which led to the following conclusions:

- The grain boundary pinning effect due to precipitation of Y_2O_3 at prior β grain boundaries was evident at 1050 °C and 1150 °C. This led to half the β grain size observed in Ti–6 Al–4 V–0.4 Y compared to conventional Ti–6 Al–4 V when β heat-treated at the same temperature. A similar but less pronounced effect of grain growth was observed when comparing heat treatment temperature for the same alloy.
- The β heat treatment did not change the main texture components, although the individual texture component intensities had changed dramatically after β heat treatment. When comparing the two alloys, significant differences were observed in terms of texture intensities although the starting textures were almost identical. Consequently, these texture variations are an indication of texture memory and variant selection.
- It was shown that β heat treatment of both Ti–6 Al–4 V variants, with a recrystallization texture dominated by Euler angles $\{\varphi_1, \Phi, \varphi_2\} \{0^\circ, 90^\circ, 0^\circ\}$, results in variant selection based on neighbouring β grains with nearly parallel $\langle 110 \rangle$, which leads to strengthening of the $\{90^\circ, 30^\circ, 0^\circ\}$ texture component.
- Variant selection during the β -to- α phase transformation is significantly stronger in conventional Ti–6 Al–4 V, which develops a larger β grain size, compared to Ti–6 Al–4 V–0.4 Y. It was found that nucleation of the α lath on both sides of prior β boundaries is preferred to nucleation on only one side and that such early nucleation is associated with β grain pairs having a common ($<10^\circ$ deviation) $\langle 110 \rangle$ pole. Such early α nucleation promotes large regions of α texture component with Euler angles $\{\varphi_1, \Phi, \varphi_2\} \{90^\circ, 30^\circ, 0^\circ\}$, which corresponds to the strong texture component that was observed after transformation.
- It is proposed that the growth of the $\{90^\circ, 30^\circ, 0^\circ\}$ texture component is mainly limited by the β grain size and consequently, this texture component grows in dominance with increasing β grain size.

Finally, it is important to note that the observed effect of β grain size on variant selection and texture strengthening is highly relevant from an application point of view since the β grain size does increase from the outer surface to the core of a billet. As demonstrated here, large regions with a particular texture component might therefore be expected deep inside a billet. Such regions might be difficult to break up from a crystallographic point of view with the potential of resulting in macrozone formation during further thermomechanical processing.

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