Acta Materialia 103 (2016) 341–351

Contents lists available at ScienceDirect

Acta Materialia

journal homepage: www.elsevier.com/locate/actamat

Full length article

The effect of aluminium on twinning in binary alpha-titanium

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ARTICLE INFO

Article history:
Received 17 July 2015
Received in revised form 23 September 2015
Accepted 23 September 2015
Available online 6 November 2015

Keywords:
Deformation twinning
Ti–Al binary alloys
Neutron diffraction
EBSD
Short range ordering

ABSTRACT

The deformation mechanisms of binary Ti–Al model alloys (0–13.1 at.% Aluminium) have been investigated with respect to the twinning activity using in-situ loading in combination with neutron diffraction as well as detailed post mortem electron backscatter diffraction analysis. A consistent starting grain size and texture was generated for all alloys promoting tensile twinning during compression testing. Long-wavelength neutron diffraction and selected area diffraction transmission electron microscopy analysis were carried out to detect evidence of Aluminium ordering and Ti_3Al formation.

It was found that raising the Aluminium content in Titanium does first slightly enhance twinning, with $\{10\,\bar{1}2\}\langle 10\,\bar{1}1\rangle$ tensile twinning being by far the dominant type, while the critical residual intergranular strains for twin initiation decreases. This suggests that either the lowering of stacking fault energy by Aluminium or its solute solution strengthening effect are important factors. At around 7 at.% Aluminium a turning point in twinning activity was noticed and a further increase in Aluminium did result in a dramatic loss of twinning activity particularly when the material had been exposed to an additional low temperature age. The dramatic decrease of twinning activity is strongly correlated with increasing evidence of short range ordering and also early signs of Ti_3Al-formation in case of the highest Aluminium content. In addition, electron backscatter diffraction analysis revealed that the formation of Aluminium ordered zones do severely hinder growth of twin boundaries.

1. Introduction

Despite the importance of aluminium as an alloying element in commercial titanium alloys, its effect on deformation mechanism activity is still poorly understood. This is particularly true regarding its effect on twinning. In metals with a hexagonal close packed (hcp) crystal structure, twinning is often considered an important deformation mechanism as it offers shear with a <c> component. The only other mechanism that includes <c> component shear is pyramidal <c+a> slip, which is known to have a critical resolved shear stress (CRSS) about 3–4 times that of prismatic <a> slip in typical titanium alloys [1,2]. In principle, twinning improves formability and impact resistance because it increases work hardening rate during deformation [3]. Adding Al to Ti is known to reduce elongation to failure [4], cause unstable shear [1] and it has been argued that 6 wt.% Al (10 at.% Al) completely switches off twinning [2]. However, it is still unclear why Al affects twinning in Ti alloys.

Aluminium and oxygen are potent solid solution strengthening elements in α-Ti [5]. Aluminium is a substitutional element whereas oxygen is interstitial [2]. The effect of Al on the slip and twin activity has been studied in some detail in large single-crystal compression samples [1]. This work showed that the CRSS for basal and prismatic <a>-slip increases with increasing Al content. It was also shown that at 12 at.% Al, CRSS for basal <a> slip is slightly higher than prismatic <a> slip. In principle, this trend can be explained by Al decreasing the stacking fault energy on the basal plane [6,7], which is also known to have a significant effect on the high-temperature creep performance in α-Ti alloys [8,9]. The work of Williams et al. [1] also showed that CRSS for <c+a> slip is consistently more difficult than <a> slip.

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http://dx.doi.org/10.1016/j.actamat.2015.09.048

Regarding twin activity and twin morphology, significant differences have been observed when comparing commercially pure (CP) Ti [10,11] and for instance Ti–6Al–4 V [12,13]. While in CP-Ti twin activity tends to be very high during the early stage of deformation and twins are readily observable, twin activity has often been reported to be less frequent, or even absent, in Ti–6Al–4 V [2]. The reason for this is not entirely clear but it has been suggested it is related to the smaller grain size, high Al content and the presence of Ti_3Al precipitates in Ti–6Al–4 V [2]. However, more recently, it has been shown that compression testing of Ti–6Al–4 V at room temperature leads to the development of a normal basal texture, which is generally associated with activation of $\{10\bar{1}2\}\langle 10\bar{1}1\rangle$ tensile twinning. Moreover, the same study suggests that despite the difficulty of twin nucleation in Ti–6Al–4 V, once twins have nucleated their growth is very rapid resulting in the consumption of entire parent grains, making the detection of twins by conventional means difficult [12].

For metals with an fcc crystal structure it has long been established that the propensity for twin nucleation is related to the stacking fault energy (SFE) as the stacking fault can act as a nucleation site for twin formation [14–16]. Interestingly, in the absence of ordering, Al does lower the SFE on the basal plane in Ti very substantially [9,17]. Although twins in Ti do not usually involve stacking faults on the basal plane, according to one of the models of heterogeneous twin nucleation, twins initiate and grow from stacking faults on twinning planes, forming zonal dislocations with a $\langle 10\bar{1}\bar{1}\rangle$ Burgers vector in the case of $\{10\bar{1}2\}\langle 10\bar{1}1\rangle$ tensile twins [18]. Although the effect of increasing Al content on this process cannot be easily estimated, its effect on basal SFE suggests it should make twin nucleation easier.

One way in which Al could affect twin activity is through solute solution strengthening [4,5], however it is not immediately clear whether this would promote or hinder twinning. On one hand, if solid solution strengthening affects the CRSS for slip but not for twinning, a higher flow stress should lead to higher twinning activity, simply because plastic deformation occurs at higher stress levels. However, adding Al not only strengthens the material but also changes the character of slip. This is a consequence of the tendency to create short range ordering (SRO) followed by long range ordering (LRO), i.e. Ti_3Al (α_2 – DO_{19} structure), depending on the level of Al addition and the heat treatment procedure. Studying ordering in Ti–Al alloy systems is notoriously difficult. A TEM study of Ti 15at.%Al (500 ppm O) after ageing for 80 h at 550 °C identified spherical α2 precipitates [19], which coarsened to ellipsoidal precipitates along the <c>-axis during continued ageing [19,20]. Signs of ordering have also been commonly reported at Al concentration as low as 8.5 at.% [2,4,21–23]. Namboodhiri et al. identified signs of SRO down to 7 at.% Al by resistivity measurements [24]. An adapted version of the phase diagram in Ref. [25] is presented in Fig. 1. Reported Ti_3Al domain sizes and ageing times are given for several concentrations and temperatures as marked in the phase diagram [1,4,20]. For comparison, the heat treatments used in the current study are also indicated.

These ordered domains or phases are expected to strengthen the material as shearing will either result in the formation of a diffuse anti-phase boundary (DAPB) in the case of SRO or the more familiar anti-phase boundary (APB) when LRO is fully developed [26]. First principle calculations suggest that the energy for the formation of a DAPB is about 4–8 times smaller than the APB energy for Ti_3Al. It has long been suggested that ordering promotes slip localisation [27] and more recent TEM investigations have indeed confirmed the formation of coupled dislocations in deformed Ti-6 wt.% Al [28]. The change in slip character to more localized slip with increased Al content could make the accommodation of the twinning strain in both the parent grain and the twinned volume more difficult, hindering twin growth. In aged material, the presence of Ti_3Al particles will also have an effect. Research on the TiAl–Ti_3Al system has consistently demonstrated that although TiAl forms deformation-twins, Ti_3Al does not [29–34], suggesting that ordering hinders twinning. In addition, some interesting recent studies of twin boundary mobility in a Mg–Gd alloy have also shown that ordering on twin boundaries exerts a very strong pinning force [35].

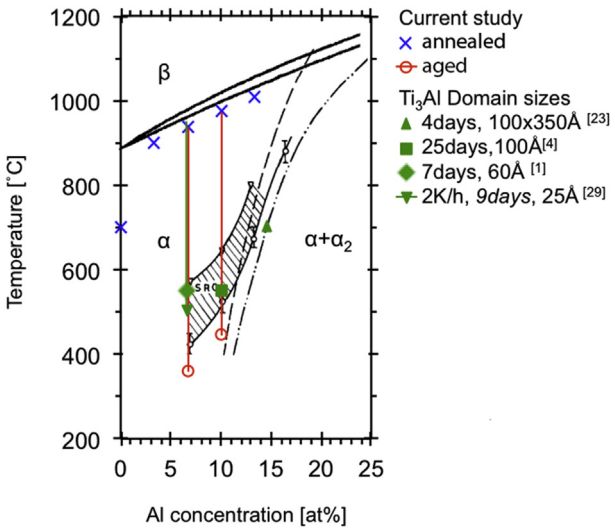

Fig. 1. Ti-rich side of the Ti–Al phase diagram adapted from Refs. [1,23,29,45,52] showing positions of annealing and ageing temperatures.

In summary, there is no convincing explanation for how adding Al to Ti affects twinning. Increasing the Al content increases the flow stress and decreases the SFE, all of which should promote twinning. On the other hand, it leads to slip localization and ordering, both of which should suppress twinning. In light of this lack of a clear understanding, and the absence of a systematic studies of the effect of Al on polycrystalline materials, a series of deformation studies were carried out on binary Ti–Al alloys with Al contents ranging from 0 to 13 at.%. To facilitate a meaningful comparison, great care was taken to generate starting microstructures with very similar recrystallized grain size and macroscopic texture. The microstructure was first characterised by SEM/EBSD, TEM and long-wavelength neutron diffraction to determine starting texture, grain size, lattice spacing and provide evidence of α_2 and SRO. In-situ neutron diffraction compression experiments were carried out to quantify twin activity and measure elastic strains along the c-axis in the grain family associated with the parent grains of twins. The methodology used is similar to previous work carried out on commercially pure Ti, Ti–6Al–4 V, Zr alloys and Mg alloys [36–41] to monitor the $\{10\bar{1}2\}\langle 10\bar{1}1\rangle$ tensile twin activity during compression testing. In the present work, the twin analysis was further complimented by detailed post-mortem EBSD analysis.

2. Materials and methods

2.1. Material preparation

For the purpose of this research, 200 g binary Ti–Al alloy buttons with different Al content up to 13 at.% Al (8 wt.% Al), were double melted in a tungsten arc furnace under an inert gas atmosphere. This was followed by β forging at 1100 °C at the TIMET research facility in Witton, UK. Subsequently, the buttons were cross-rolled into a bar (14 × 14 × 260 mm) on a “2 high Robertson mill”