Creep Mechanisms and Interface-Enhanced Deformation Twinning in a Two-Phase Lamellar TiAl Alloy

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Deformation mechanisms and the role of interfaces in deformation twinning of a two-phase [TiAl (γ)/Ti3Al (α2)] lamellar alloy creep deformed at elevated temperatures have been investigated. Since the multiplication of lattice dislocations within both γ and α2 lamellae is very limited at a low stress level due to a refined lamellar microstructure, the glide of interfacial dislocations on both γ/α2 and γ/γ interfaces (i.e., interface sliding) becomes an important deformation mode. Obstacles such as impinged lattice dislocations can impede the movement of interfacial dislocations, which glide in a cooperative fashion along the lamellar interfaces. The impediment of dislocation motion subsequently causes a dislocation pile-up in front of obstacles as creep strain accumulates. When the crystals deform at high stress level, deformation twinning becomes a predominant deformation mode. Deformation twins are found to nucleate from the interfaces as a result of a local stress concentration generated from dislocation pile-ups. It is suggested that the deformation twinning in lamellar TiAl/Ti3Al crystals can be viewed as a stress relaxation process for the concentration of stress at the head of each dislocation pile-up. An interface-assisted twinning mechanism is accordingly proposed and discussed.
Introduction

Two-phase TiAl alloys with a lamellar microstructure where TiAl (γ) and Ti3Al (α2) phases are present as alternating sheets are appealing for applications in advanced turbine engine components due to their high specific strength, stiffness, creep resistance and environmental resistance at elevated temperatures. However, a major drawback of the alloys for the structural applications is their inadequate fracture resistance (ductility and toughness) at low temperatures. In order to overcome this obstacle, a recent development of the alloys has focused on refining the γ/α2 lamellar microstructure through advanced processing routes such as a powder metallurgy (P/M) process. The mechanical properties of P/M lamellar alloys with a refined microstructure have been demonstrated to be superior to those of the lamellar alloys fabricated by conventional ingot metallurgy (I/M) [1]. Deformation twinning provides not only an additional deformation mechanism to facilitate the room temperature ductility [2] but also effective dislocation obstacles for enhancing creep resistance of the lamellar alloys. It has been demonstrated that the alloys become more creep-resistant at a low stress level (138 MPa) after mechanically twinned at a high stress level (518 MPa) [3,4]. Since the lamellar interfaces can provide additional nucleation sites for deformation twins (DT hereafter), the interfaces essentially promote twinning processes in the alloys. Although the formation of DT can be attributed to the homogeneous glide of 1/6<112> twinning dislocations on every {111} plane [5], the underlying twinning mechanism still remains unclear. To better design the lamellar alloys for high temperature applications, it is of importance to understand and gain insights for the role of lamellar interfaces in the twinning process as well as the mechanical behavior of the alloys. Accordingly, this investigation has been conducted in order to elucidate the mechanisms of deformation twinning in two-phase lamellar TiAl alloys.

Experimental

A two-phase lamellar TiAl alloy [a nominal composition of Ti-47Al-2Cr-2Nb (at. %)] was fabricated at Oak Ridge National Laboratory by a powder metallurgy process involving a hot-extrusion of gas-atomized titanium aluminate powder at 1400°C. After extrusion, the alloy was stress-relieved at 900°C in a vacuum (~10^-4 Pa) for 2 h. Creep tests were conducted at 760°C in a dead-load creep machine with a lever arm ratio of 16:1. Detailed information regarding creep experiment and the experimental data were reported elsewhere [1]. For current study, the deformation substructures of two tested specimens [one crept at a stress of 138 MPa (creep strain: ~0.25%) and the other crept at a stress of 518 MPa (creep strain: -3.6%)] were investigated. TEM foils were prepared by twin-jet electropolishing in a solution of 60 vol % methanol, 35 vol % butyl alcohol and 5 vol. % perchloric acid at ~15 V and ~30°C. The microstructures of the crept alloys were examined using a JEOL-200CX transmission electron microscope equipped with a double-tilt goniometer stage. Images of dislocations were recorded using a weak-beam dark field (WBDF) imaging technique under g (2g) two-beam diffraction conditions.

Results and Discussion

Microstructure

Figure 1 shows a typical edge-on lamellar microstructure of the two-phase TiAl alloy. In general, there exist two types of lamellar interfaces within the alloy [6] (1) The γ/α2 interphase interface which has a usual orientation relationship (0001)α2 || (111), and <1120>α2 || <110>γ. (2) The γ/γ twin-related interfaces which includes true-twin (180° rotational) and pseudo-twin (60° and/or 120° rotational) interfaces. The width of α2 lamella ranges from 10 to 50 nm, and that of γ lamella ranges from 150 to 300 nm. Figure 2 shows a typical dislocation substructure within a lamellar grain. Both lattice dislocations (LD hereafter) within γ lamellae and a high density of interfacial dislocations (ID hereafter) on inclined lamellar interfaces can be clearly seen. The density of ID is much greater than that of LD, and the LD are mainly threading dislocation lines which terminate their two ends at lamellar interfaces. While those ID on semi-coherent γ/α2, and γ/γ pseudo-twin interfaces are 1/6<112> or 1/3<112> type [] misfit dislocations,
those on γ/γ true-twin interface are 1/6<112> type geometry necessary dislocations for accommodating the departure of the twin interface from the exact twin plane.

Glide of interfacial dislocations

Since the volume fraction of the α₂ lamella is small (~10%), the deformation strain is mainly carried by γ lamellae through the multiplication of LD (mainly threading dislocations). The critical shear stress (τ_b ≈ μhl/l, where l is the distance between two pinning points) required to bow the threading dislocations however is considerably high because of thin γ lamellae. Taking μ ~ 59 GPa (at 760 °C), h ~ 0.56 nm, l ~ 100 – 300 nm (interface spacing), and τ_b ~ 165 – 330 MPa, the critical stresses (τ_b^* ≈ 2τ_b) are estimated to be 330 – 660 MPa, which are greater than the applied stress of 138 MPa.

Since the multiplication of lattice dislocations within γ lamellae is very limited due to a refined lamellar microstructure, the glide of interfacial dislocations (i.e., interface sliding) becomes a predominant deformation mode. This has been demonstrated in the result of a room temperature in situ straining study reported elsewhere [7] Figures 4 and 5 show the deformation substructures of a soft lamellar grain (oriented nearly ~45° with respect to the stress axis) within a specimen tested at a low stress level (138 MPa). The ID on both the γ/γ and γ/α₂ interfaces become mobile at elevated temperatures even with a low shear stress (~69 MPa) resolved on the interfaces. As a result, interface sliding occurs due to the movement of ID and resulting in the formation of grain boundary steps (Fig. 4). The wavy appearance of interfacial dislocation lines (site A in Fig. 4) suggests that solute or impurity atoms may have dragged the movement of ID. The mobility of ID can also be impeded by dislocation obstacles formed by the impingement of LD on the lamellar interface as shown in Fig. 5. Here in Fig. 5(a), several ID are piled up as a result of the impingement of LD on a γ/γ interface. Furthermore, a more effective obstacle can be formed if a number of impinged LD react with ID to form dislocation nodes as shown in Fig. 5(b).
Consequently, the nodded ID becomes less mobile and resulting in a dislocation pile-up.

Deformation twinning

When the alloy is creep deformed at a high stress level, deformation twinning starts to engage as a predominant deformation mode. Figure 6 shows the formation of (T 11) [211]-type deformation twins (DT hereafter) within a specimen tested at 518 MPa. Notice that one of the
(111) twin lamellae (marked by an arrow) was emitted from the upper interface, and eventually will be blocked by the lower interface if it continues to grow. This observation suggests that the DT is originally nucleated from the lamellar interfaces as a result of a local stress concentration caused by the pile-up of interfacial dislocations. Accordingly, it is proposed that deformation twinning in lamellar TiAl/Ti3Al can be viewed as a stress relaxation process to relieve the local stress concentration caused by the pile-up of interfacial dislocations during deformation. The effective stress (τ) at the tip of the pile-up of n dislocations is τ = nτi[8], where τi is the resolved shear stress acting on the interface. To relieve the stress concentration, DT in γ lamellae is therefore formed by a dislocation reaction based upon a stair-rod cross-slip mechanism [9,10]. As for an example of the (111)-type DT formed in an alloy deformed at a high stress level, the corresponding dislocation reaction (dissociation) is proposed to be 1/6[121] (111) → 1/6[011] (100) + 1/6[112] (111). The (111) type DT is accordingly formed by a successive cross-slip of the twinning dislocations 1/6[112] on the (111) plane and leaving the stair-rod dislocations 1/6[011] on the (100) plane. Twin (stacking) faults are subsequently formed on the interfaces between the γ lamellae and DT. This is schematically illustrated in Fig. 5(b).

Fig. 3 A WBDF image showing a typical microstructure of a nearly 45°-oriented lamellar grain within a sample creep deformed at 138 MPa. Grain boundary ledges (indicated by arrows) were formed due to interface sliding. The wavy appearance of interfacial dislocation lines is noted at site A.

Fig. 4 (a) A WBDF image showing the pile-up of several ID (indicated by an arrow) on a γ/γ related interface. (b) A WBDF image showing that the motion of interfacial dislocations is impeded as a result of the formation of dislocation nodes due to the reaction between impinged LD (stronger contrast) and intrinsic ID (weaker contrast).
Fig 5 (a) A bright-field TEM image showing several (111) type deformation twins formed growing process toward another lamellar interface. (b) Schematic illustration of the nucleation of a (111) type DT from a γ/α₂ interface, where b₁, b₂, and b₃ denote the interfacial, stair-rod, and twinning dislocations, respectively.

The formation of stair-rod dislocations at the intersections between DT and an α₂ lamella is evidenced in Fig. 6, where the array of 1/6[011] stair-rod dislocations become invisible [Fig. 6(a)] or visible [Fig. 6(b)] when a reflection vector (g) 200 or 021 is used for imaging. It is noted that the individual stair-rod dislocation is not resolvable because of a narrow distance (0.25 nm) between two stair-rod dislocations. The existence of the stair-rod dislocations can also be indirectly evidenced in Figs. 7(a) and 7(b). Here, a pair of approximately aligned DT is originally formed on the opposite side of an α₂ lamella. These two aligned DT eventually coalesce at the joints adhered to the α₂ lamella. This coalescence is considered to be resulted from the annihilation of stair-rod dislocations (with opposite Burgers vectors) through a thermally activated process. A schematic representation of a pair of approximately aligned DT on the opposite side of an α₂ lamella and the formation of stair-rod dislocations with opposite Burgers vectors is shown in Fig. 8. The significance of the proposed mechanism is to reveal that there are several barriers to be overcome in order to activate the twinning reaction. These barriers include (1) the repulsive force between the interfacial (Shockley) and stair-rod dislocations, (2) the increase of line energy due to the dislocation dissociation, and (3) the increase of interfacial energy due to the formation of twin (stacking) faults. The resolved shear stresses on both the interfacial and cross-slip planes and a local stress concentration due to the pile-up of interfacial dislocations are considered to be important factors for the formation of deformation twins in lamellar TiAl/Ti₃Al.

Fig. 6 Paired WBDF images demonstrating the existence of the array of 1/6[011] stair-rod dislocations at the intersections (indicated by arrows) between the (111) type DT and an α₂ lamella. (a) Invisible at g = 200 (g·bν = 0), (b) visible at g = 021, Z (zone axis) ≈ [012]
Fig. 7 WBDF images showing the faulting of $\alpha_2$ lamellae [indicated by an arrow in (a)] resulted from the coalescence of two approximately aligned DT at the joints adhered to a $\gamma$ lamella [indicated by an arrow in (b)]. Images generated from the same region using $\gamma$ and DT reflections, respectively.

Fig. 8 A schematic representation of a pair of approximately aligned DT formed on the opposite side of a $\alpha_2$ lamella and the existence of stair-rod dislocations (with opposite Burgers vectors) at the interceptions between DT and a $\alpha_2$ lamella.

Summary

Creep mechanisms and the role of interfaces in deformation twinning of a two-phase lamellar TiAl alloy have been investigated. Since the multiplication of lattice dislocations within both $\gamma$ and $\alpha_2$ lamellae becomes very limited at a low stress level, the gliding of interfacial dislocations (i.e., interface sliding) becomes an important deformation mode. Impinged lattice dislocations are observed to impede the movement of interfacial dislocations, which glide in a cooperative fashion along the lamellar interfaces. The impediment of dislocation motion subsequently causes a dislocation pile-up in front of the obstacle as creep strain accumulates. When the alloy deform at high stress level, deformation twinning becomes a predominant deformation mode. The deformation twinning in the lamellar TiAl/Ti$_3$Al crystals is suggested to be a stress relaxation process for the concentration of stress at the head of each dislocation pile-up. An interface-assisted twinning mechanism based upon a stair-rod cross-slip dislocation reaction is proposed.

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References
