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Interfacial Control of Deformation Twinning in Creep-Deformed TiAl/Ti3Al Nanolaminate

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Interfacial Control of Deformation Twinning in Creep-Deformed TiAl/Ti_3Al Nanolaminate

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Objectives

- Exploit thermomechanical-processing techniques to fabricate TiAl/Ti_3Al nanolaminate composites with the size of lamella width down to nanometer length-scales.
- Characterize microstructure and elevated-temperature creep resistance of the nanolaminate composites.
- Investigate the fundamental interrelationships among microstructures, alloying additions, and mechanical properties of the nanolaminate composites so as to achieve the desired properties of the composites for high-temperature structural applications.

Approach

- In-situ laminate composites with nominal compositions of Ti-47Al-2Cr-2Nb, Ti-46Al-3Nb-1W-0.1B, and Ti-46Al-3Nb-2W-0.1B (at.%) were employed for the study. The in-situ laminate composites were fabricated at Oak Ridge National Laboratory by hot-extrusion of cast alloys at 1350 $^{\circ}$C.
- Creep tests were conducted in a dead-load creep machine with a lever arm ratio of 16:1. Tests were performed in air in a split furnace with three zones at 760 $^{\circ}$C and 815 $^{\circ}$C.
- The microstructures of creep-deformed samples were examined using a JEOL-200CX transmission electron microscope.

Accomplishments

- Collaborated with ORNL (Dr. C.T. Liu) to fabricate TiAl/Ti_3Al laminate composites with high W content using hot-extrusion processing techniques.
- Characterized and measured the effect of tungsten addition on creep resistance of the TiAl/Ti_3Al laminate composites.

Future Direction

- Continue to collaborate with ORNL to fabricate the oxidation- and heat-resistant class of TiAl/Ti_3Al laminate composites with high Nb content (>10 at.%) using hot-extrusion processes.
- Continue to investigate the effects of alloying addition and deformation (mechanical) twinning on the microstructural stability and creep resistance of the nanolaminate composites at elevated temperatures up to 850 $^{\circ}$C.

Introduction

One of the unique deformation substructures of TiAl/Ti_3Al laminate composite is the formation of deformation twins (DT) within γ lamellae. The twinning phenomena have found to be significantly promoted within ultrafine lamellar TiAl as a result of refined lamellar spacing presumably because the increment of lamellar interfaces provides even more nucleation sites for twinning. Although it has been well known that the deformation twinning can be activated by the homogeneous glide of $1/6\langle 11\bar{2}\rangle$ twinning dislocations on the {111} planes, yet 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 deformation twinning mechanisms in TiAl/Ti_3Al nanolaminate.

Approach

A TiAl/Ti_3Al nanolaminate composite was fabricated by a hot extrusion process, which involves hot-extrusion of a cast TiAl alloy at 1350 °C. After extrusion, the alloy was stress-relieved at 900 °C in a vacuum (~ 10^{-4} Pa) for 2 h. Creep tests were conducted in a dead-load creep machine with a lever arm ratio of 16:1. For the current study, the deformation substructures of specimen [tested at 760 °C, 138 MPa (creep strain: 0.25%), 760 °C, 518 MPa (creep strain: 3.6%), and 815 °C, 420 MPa (creep strain: 1.7%)] were investigated. TEM foils were prepared by twinjet 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 tested samples were examined using a JEOL-200CX transmission electron microscope equipped with a double-tilt goniometer stage. Images of dislocations were mostly recorded using weak–beam dark field (WBDF) imaging techniques under ***g** (3**g**)* two–beam diffraction conditions with the deviation factor $\omega\ (= \xi_g\, s) > \sim 1$, where ξ_g is the extinction distance and s is the deviation distance from the exact Bragg position. The $\boldsymbol{g \bullet b}$ invisibility criteria used for determining the Burgers vector of Shockley partials are described as follows [1]: (a) Invisible if $\boldsymbol{g \bullet b} = 0$ or $\pm 1/3$. (b) Invisible if $\boldsymbol{g \bullet b} = -2/3$ but visible if $\boldsymbol{g \bullet b} = +2/3$ provided the deviation factor $\omega > \sim 1$. (c) Invisible if $\boldsymbol{g \bullet b} = +4/3$ but visible if $\boldsymbol{g \bullet b} = -4/3$ provided the deviation factor $\omega > \sim 1$.

Results

Microstructure

Figure 1 is a bright-field TEM image showing a typical edge-on microstructure within a TiAl (γ)-Ti_3Al (α_2) nanolaminate. In general, the material contains two types of interfaces [2]: (1) The γ/α_2 interphase interface which has a usual orientation relationship $(0001)_{\alpha 2} \parallel (111)_{\gamma}$ and $<11\bar{2}0>_{\alpha 2} \parallel <1\bar{1}0>_{\gamma}$. (2) The γ/γ twin-related interface which includes true-twin (180° rotational) and pseudo–twin (60 and/or 120° rotational) interfaces. Here, the width of α_2 layers ranges from 10 to 50 nm, and that of γ layers ranges from 150 to 300 nm. Figure 2 is a weak-beam dark-field (WBDF) TEM image showing a typical dislocation substructure within the nanolaminate. Both lattice dislocations (*LD* hereafter) within γ layer and a high density of interfacial dislocations (*ID* hereafter) on inclined interfaces can be clearly seen. The density of *ID* is much greater than that of *LD*, and the *LD* are primarily threading dislocations which terminate their two ends at the interfaces. While the *ID* on semi-coherent γ/α_2 and γ/γ pseudo-twin interfaces are *1/6*<112> or *1/3*<112> type misfit dislocations [3], the on γ/γ true-twin interface are mainly *1/6*$[11\bar{2}]$ type twinning dislocations or geometry necessary dislocations for accommodating the departure of true-twin interface from the exact (111) twin plane.

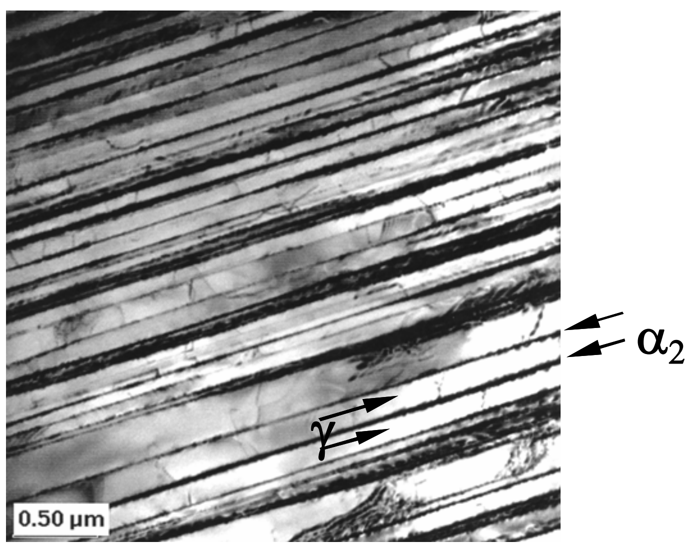


Fig. 1. Bright-field TEM image showing a lamellar grain viewing from an edge-on orientation.

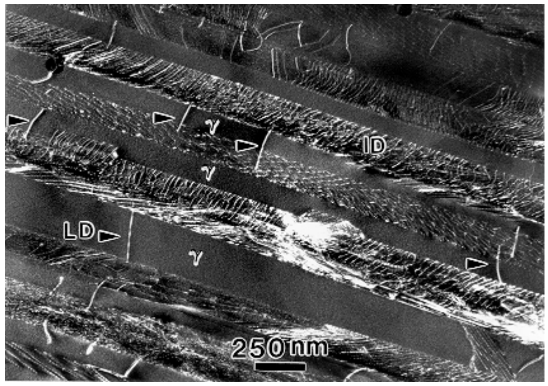
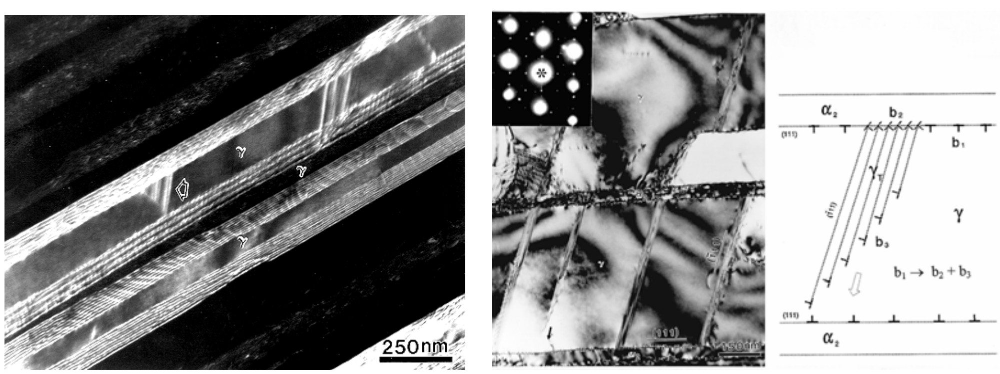


Fig. 2. Weak-beam dark-field (WBDF) TEM image showing a typical dislocation structure of TiAl nanolaminate.

Deformation twinning and proposed mechanisms

When the nanolaminate was creep deformed at 760 °C and 518 MPa, a deformation substructure associated with deformation twins (*DT* hereafter) within γ layers was developed. Typical examples of the formation of $(\bar{1}11)$ [211]-type *DT* within the nanolaminate are shown in Figs. 3 (a) and (b). It is noted that one of the twin lamellae was still growing between two lamellar interfaces, and its growth would be eventually blocked by the lower interface. This observation suggests that the interfaces are preferred nucleation sites for *DT*, presumably resulting from the high local stresses caused by the pileup of *ID*. Accordingly, it is proposed that deformation twinning in the TiAl-Ti_3Al nanolaminate can be viewed as a stress relaxation process to relief the local stress concentration caused by the pile-up of interfacial dislocations during deformation. The effective stress (τ_e) at the tip of the pile-up of n dislocations can be evaluated by $\tau_e = n\tau_i$ [4], where τ_i is the resolved shear stress acting on the interface. To relieve the stress concentration, deformation twinning in γ layers is therefore taking place by a dislocation reaction based upon a stair-rod cross-slip mechanism [5, 6]. As for an example of the $(\bar{1}11)$-type *DT* formed in the nanolaminate, the corresponding dislocation reaction (dissociation) is proposed to be $1/6[\bar{1}2\bar{1}]_{(111)} \rightarrow 1/6[011]_{(100)} + 1/6[\bar{1}1\bar{2}]_{(\bar{1}11)}$. The $(\bar{1}11)$-type *DT* is accordingly formed by a successive cross-slip of the twinning dislocations $1/6[1\bar{1}2]$ on the $(\bar{1}11)$ 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 layer and *DT*. This is schematically illustrated in Fig. 3 (c).

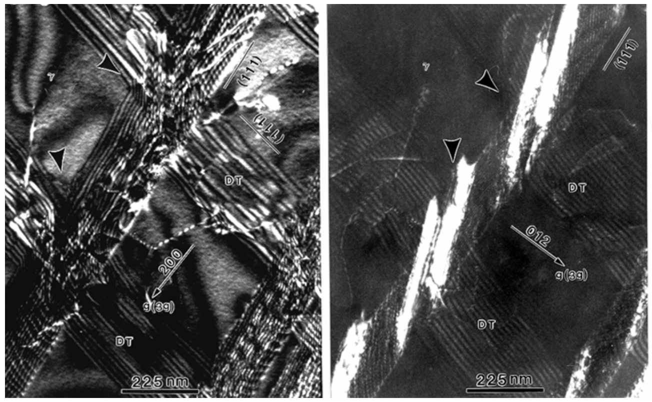


(a) (b) (c)

Fig. 3. (a) Dark-field and (b) bright-field TEM images showing several $(\bar{1}11)$ type deformation twins formed growing process toward another interface. (c) Schematic illustration of the nucleation of a $(\bar{1}11)$ type *DT* from a γ/α_2 interface, where b_1, b_2, and b_3 denote the interfacial, stair-rod, and twinning dislocations, respectively.

The formation of stair-rod dislocations at the intersections between the *DT* and α_2 layer is evidenced in Fig. 4, where the array of $1/6[011]$ stair-rod dislocations become invisible [Fig. 4(a)] or visible [Fig. 4(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 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 (F) 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 faults. Among them the repulsive force (F) between the interfacial (Shockley) and stair-rod dislocations is considered to be rate controlling. That is, a critical (minimum) stress (τ_C) is required to activate the dissociation reaction for twinning.



(a) (b)

Fig. 4. Paired WBDF images demonstrating the existence of the array of *1/6*[011] stair-rod dislocations at the intersections (indicated by arrows) between the $(\bar{1}11)$-type *DT* and α_2 layer. (a) Invisible at $\boldsymbol{g} = 200$ ($\boldsymbol{g} \bullet \boldsymbol{b} = 0$), (b) visible at $\boldsymbol{g} = 021$, **Z** (zone axis) $\approx [0\bar{1}2]$.

Conclusion

The role of interfaces in deformation twinning of TiAl/Ti_3Al nanolaminate has been investigated. Since the multiplication of lattice dislocations within both γ and α_2 lamellae becomes very limited at a low stress level, the motion of interfacial dislocations (i.e. interface sliding) becomes an important deformation mode. Impinged lattice dislocations were observed to impede the movement of interfacial dislocations, which move 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 laminate deform at high stress level, deformation twinning becomes a predominant deformation mode. The deformation twinning is suggested to be a stress relaxation process for the concentration of stress at the head of each dislocation pile-up. An interface-controlled twinning mechanism based upon a stair-rod cross-slip dislocation reaction is proposed and verified.

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2. L. M. Hsiung, T.G. Nieh, "Microstructures and Properties of Powder Metallurgy TiAl Alloys," *Mater. Sci. & Engrg.* **A364** (2004), p. 1.
3. A. Hodge, L. M. Hsiung, T. G. Nieh, "Creep of Nearly Lamellar TiAl Alloy Containing W," *Scripta Mater.* **51** (2004), p. 411.