

Effect of Nanovoid on Fracture Process of Two-Phase $\alpha(\text{TiAl})+\gamma(\text{Ti}_3\text{Al})$ Alloy

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Abstract

The fracture processes of nanocrystalline metallic material is affected by dislocation, nanovoid and other defects. Existing studies of defect evolution in titanium-aluminum alloy cover the case that voids located in single crystals, inside grain in polycrystals and at the grain boundaries. Molecular dynamics simulation was performed to study the evolution of a spherical nanovoid in $\alpha+\gamma$ two-phase titanium-aluminum alloy under uniaxial tension.

Keywords: $\alpha+\gamma$ two phase TiAl alloy; void; molecular dynamics

1. Introduction

TiAl alloy has been used as structural material in aviation industry because its inherent advantages such as low density and self-diffusion rates, high elastic module and high strength [?]. However, single phase γ – TiAl generally brittle at room temperatures and this limits their use in many other fields. Two-phase titanium aluminum alloys with proper phase distribution and grain size exhibit better mechanical performance compared with monolithic constituents $\gamma(\text{TiAl})$ and $\gamma(\text{Ti}_3\text{Al})$ alloy [?].

2. Molecular dynamics simulation analysis

2.1. Model creation of nanocrystalline

γTiAl has a fcc-centered tetragonal with an $L1_0$ structure [?], and αTiAl is hcp structure, the structure of the two initial cell are shown in Fig.??, and the constructing parameters are given by Table.???. The simulation cells of two phase polycrystal with an initially spherical void at different position are shown in figure ???. Periodic boundary conditions are applied along all three directions, that makes polycrystal with periodic nanovoid structures. Each cuboidal model, containing about 4.6 million atoms, The initial dimension of simulation cell is $L_x = nm$, $L_y = nm$, $L_z = nm$. The grain orientation and size were randomly created with Voronoi method with code ATOMSK [?], and resulting in the arbitrary shape and orientation of the grains. Only one spherical void defect was placed intragranularly or intergranularly within each simulation model void within each simulation model. The intragranular spherical void was located in grain interior of the largest grain of the simulation model, as shown in Fig. ???. The intergranular spherical void was at the center of the simulation cell, as shown in Fig. ??.

2.2. Atomic potential

The interaction of particle in the material is determined by interatomic potential. Many reported examples of crack propagation in metal materials were performed with embedded atom method, due to its better accuracy in metal lattice compared with F-S and L/J [?]. The embedded atom method (EAM) potential developed by Zope and Mishin [?] was used in the study, which is submitted by MD simulations with the Large-scale Atomic/Molecular Massively Parallel Simulator (LAMMPS) open-source code [?]. We performed constant-pressure and constant-temperature (NPT) molecular dynamics simulation.

2.3. Analysis method of defect identification.

Table 1
Experimental conditions

Team	α	W	D	L	F	A	Pts
Manchester United	6	4	0	2	10	5	12
Celtic	6	3	0	3	8	9	9
Benefit	6	2	1	3	7	8	7
FC Copenhagen	6	2	1	3	5	8	7

stress calculation. The crystal structure parameters of $\gamma(\text{TiAl})$ phase and $\alpha(\text{Ti}_3\text{Al})$ phase is shown as Table??.

2.4. Molecular Dynamics Model

The crystal structure of γ -TiAl is shown by Fig ??

2.5. Identification of Defects

Table 2
Table caption

Treatments	Response 1	Response 2
Treatment 1	0.0003262	0.562
Treatment 2	0.0015681	0.910
Treatment 3	0.0009271	0.296

$$E = \sum_i^n a_i = 0$$

where

parameters can be given as Table ??:

3. Results and Discussion

3.1. The influence of void on strength of TiAl alloy

The existence of void detract the strength, and the void inside α phase grain have most significant impact on the strength, however the void on the grain boundary have little impact on incipient strength of the material. Detailed observation of specimen with void inside the grain is shown in Figure ??.

1. In many cases the orientation of slip is changed because the crystallographically available slip and directions are not continuous across the interface. This may significantly reduce the Schmid factor and thus impede slip transfer. At the γ/γ interfaces the orientation of the slip plan could change through a relevantly large angle of about 90 degree. Reorientation of slip is always required at the α_2/γ interface; the smallest angle between the corresponding slip planes 111_γ and $10 - 10_{\alpha_2}$ is about 19 degree ref. 2. The core of a dislocation intersecting an interface often needs to be transformed. For example, an ordinary $1/2[110]$ dislocation gliding in one γ grain has to be converted in to a $[101]$ super dislocation with the double Burgers vector gliding in an adjacent γ grain. At the α/γ interface the dislocations existing in the $D0_{19}$ structure have to be transformed into dislocations consistent with the $L1_0$ structure. These core transformations are associated with a change of the dislocation line energy because the lengths of the Burgers vectors and the shear module are different.

3. Dislocations crossing semi-coherent boundaries have to intersect the misfit dislocations, a process that involves elastic interaction, jog formation and the incorporation of gliding dislocations into the mismatch structure of the interface. When the slip is forced to cross α_2 lamellae, pyramidal slip of the α_2 phase is required, which needs an extremely high shear stress.

3.2. Decrease of Strength

4. Conclusion

Fig.??.