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DISSOCIATION OF DISLOCATIONS IN $Al_{67}Mn_8Ti_{25}$ DEFORMED AT AMBIENT AND HIGH TEMPERATURES

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Introduction

The $L1_2$ structured Al_3Ti-based alloys are now of considerable interest because they have shown appreciable compressive ductility at room temperature. It has been determined that the slip systems in $L1_2$ Al_3Ti alloys are ⟨110⟩{111}, but studies on the dislocation structures in the deformed alloys give conflicting results [1-8]. Earlier observations by transmission electron microscopy (TEM) on the room temperature deformed $Al_{67}Ni_8Ti_{25}$ [1,2] or Al_3Ti_3Fe [3] samples suggested that the a⟨110⟩ dislocations moving on {111} planes are undissociated. Further works on the Fe-modified $L1_2$ alloy have shown that at room temperature the a⟨110⟩ dislocations tend to dissociate into a/2⟨110⟩ partials on {111} planes separated by antiphase boundary (APB)[4-6], while others cited that the dissociation is of the a/3⟨112⟩-type with the superlattice intrinsic stacking faults (SISF) between two partials on {111} planes [7,8]. TEM studies on the dislocations in Al_3Ti-based $L1_2$ alloy deformed at elevated temperatures are limited. Lerf and Morris [3] reported that the majority of the dislocations in the Fe-modified alloy after 500℃ deformation are dissociated as pairs of a/2⟨110⟩ on {111} planes with APB between, and the results of Inui et al [8] showed, also, the APB-coupled a/2⟨110⟩ superpartials at 600℃ but on {001} planes.

In the present study, the dislocation structures and dissociation modes of superdislocations in the $Al_{67}Mn_8Ti_{25}$ alloy deformed at ambient and elevated temperatures were investigated by TEM. The temperature dependence of mechanical properties of this alloy are discussed in relation to the observed dislocation structures.

Experimental

An intermetallic alloy with nominal composition $Al_{67}Ti_{25}Mn_8$ (at%) was prepared by arc melting in argon on a water cooled copper hearth. The button ingots were homogenized at 1373K for 60h. Samples with dimensions 4mm × 4mm × 7mm were cut from the homogenized ingots and deformed under compression to ~2% plastic strain at room temperature, 673K, and 873K respectively. Thin disks of 3mm in diameter were spark cut from the deformed samples and then mechanically polished to ~100um in thickness. Final preparation of the thin foils for TEM observation was accomplished by twin jet polishing in a mixture of perchloric acid, butanol, and methanol (30:175:300 by volume) at 233K. Conventional and weak beam techniques were performed in TEM study.

Results

The general features of dislocation structure deformed at room temperature, 673K, and 873K are shown in Fig.1 respectively. In contrast to the long, straight dislocations observed in the room temperature deformed alloy (Fig.1(a)), the dislocations in 673K deformed sample are curved, and more dipoles are seen (Fig.1(b)), while the 873K deformed sample shows random arrangement of short, curved dislocation segments and dipoles (Fig.1(c)).

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Detailed analyses were done to determine the slip system and dissociation mode of the superdislocations in the above mentioned samples. The dislocations in the sample deformed at room temperature consisting of paired partials, labeled A and B in Fig.2, were chosen for analysis. Trace analysis through a series of dislocation images taken with different beam directions determined that both A and B are lying on the $(\bar{1}11)$ plane. The Burgers vectors of the dislocations were examined by diffraction contrast analysis using various reflections. Table I gives the $\mathbf{g}\cdot\mathbf{b}$ values for the partials under different reflections and among them typical micrographs are shown in Fig.2. It was found that the superdislocations formed during compressive deformation at room temperature are dissociated into two $a/3\langle 112\rangle$ superpartials on $\{111\}$ planes, with SISF between them. For the dislocations A : $a[110] \rightarrow a/3[211] + \mathrm{SISF} + a/3[12\bar{1}]$, and for B: $a[0\bar{1}1] \rightarrow a/3[\bar{1}\bar{2}1] + \mathrm{SISF} + a/3[1\bar{1}2]$.

TABLE I. $\mathbf{g}\cdot\mathbf{b}$ Values for the Dislocations Formed by Room Temperature Deformation under Different Reflections

g	A				B				Fig.
	1 ($\mathbf{b}=1/3[211]$)		2 ($\mathbf{b}=1/3[12\bar{1}]$)		1 ($\mathbf{b}=1/3[\bar{1}\bar{2}1]$)		2 ($\mathbf{b}=1/3[1\bar{1}2]$)		
	obser.	$\mathbf{g}\cdot\mathbf{b}$	obser.	$\mathbf{g}\cdot\mathbf{b}$	obser.	$\mathbf{g}\cdot\mathbf{b}$	obser.	$\mathbf{g}\cdot\mathbf{b}$	
220	V.	2	V.	2	V.	−2	I.V.	0	2(a)
$02\bar{2}$	I.V.	0	V.	2	V.	−2	V.	−2	2(b)
$31\bar{1}$	V.	2	V.	2	V.	−2	I.V.	0	
$3\bar{1}1$	V.	2	I.V.	0	I.V.	0	V.	2	
$1\bar{3}1$	I.V.	0	V.	−2	V.	2	V.	2	
$\bar{1}\bar{1}1$	R.C.	−2/3	V.	−4/3	R.C.	4/3	V.	2/3	2(c)

V. : visible, I.V. : invisible, R.C. : residual contrast

Analyses of the sample deformed at 673K determined that the superdislocations are dissociated in the same mode as those in the room temperature deformed sample.

The dislocations in the sample deformed at 873K are apparently dissociated, but the contrast analysis shows that both partials are in contrast or out of contrast simultaneously for all the operating reflections. This suggests that the dissociation is of the $a/2\langle 110\rangle$ type. Table II lists the $\mathbf{g}\cdot\mathbf{b}$ values and the weak-beam images used for the analysis are shown in Fig.3. The Burgers vector of the superdislocation shown in the micrograph was determined by the "invisibility criterion" as $a[\bar{1}01]$, and the dissociation is $a[\bar{1}01] \rightarrow a/2[\bar{1}01] + \mathrm{APB} + a/2[\bar{1}01]$. The dissociation plane was determined by the widest separation of the dissociated pairs through tilting of the sample. The result shows that the APB is on the (111) plane (see Fig.3(b)), and the measured separation is ~17nm. The APB plane is checked further by observing the change of dislocation configuration when viewed in different orientations. The possible dissociation planes for the $[\bar{1}01]$ superdislocation are (111), $(1\bar{1}1)$ of $\{111\}$ and (010) of $\{001\}$. Since the curved dislocations did not appear straight when the sample was tilted to a beam direction [001] or [011], the dislocation plane must not be (010) or $(1\bar{1}1)$ respectively.

TABLE II. $\mathbf{g}\cdot\mathbf{b}$ Values for the Dislocations in 873K Deformed Sample

g	$\mathbf{b}=\mathbf{b}_1=\mathbf{b}_2=1/2[\bar{1}01]$		Fig.
	obser.	$\mathbf{g}\cdot\mathbf{b}$	
020	I.V.	0	
$1\bar{3}1$	I.V.	0	3(a)
$11\bar{1}$	V.	−1	
$\bar{2}20$	V.	1	
$\bar{2}02$	V.	2	3(b)
$\bar{3}11$	V.	2	3(c)

V. : visible, I.V. : invisible

Discussion

The TEM studies of dislocation structures in the deformed $Al_{67}Mn_8Ti_{25}$ indicated that at ambient and intermediate temperatures, e.g. 673K, the a⟨110⟩ superdislocations dissociate into a/3⟨112⟩ superpartials bounding SISF on {111} planes, and at higher temperatures, e.g. 873K, they dissociate into a/2⟨110⟩ bounding APB on {111} planes. It has been shown by atomistic modeling [9] that the dislocation core structure of a/3⟨112⟩ superpartials is nonplanar and is, therefore, sessile. Their motion requires a thermally activated process, so the yield stress will decrease rapidly with increasing temperature. But in the $Al_{67}Mn_8Ti_{25}$ alloy, the yield stress declines slowly from room temperature to 523K, and then shows a plateau with increasing temperature (Fig.4) [10]. This phenomenon may be related to the strong tendency of dipole formation with increasing temperature in this alloy (Fig.1). Thus, a high stress is required to break these dipoles and a plateau temperature dependence of the $L1_2$ Al_3Ti alloy results.

Another interesting feature of the $Al_{67}Mn_8Ti_{25}$ alloy is the rapid increase of compressive ductility with temperature above 773K, while the yield stress is almost constant (Fig.4) [10]. This behavior can be explained by the dissociation mode of superdislocations at higher temperatures. It has been generally accepted that the symmetry of the $L1_2$ structure assures the stability of a/2⟨110⟩ APB on {001} but not on {111}, and, therefore, cross slip of a/2⟨110⟩ screw dislocations from {111} to {001} leads to the anomalous temperature dependence of yield stress in some $L1_2$ alloys such as Ni_3Al. In the present case, the a/2⟨110⟩ superpartials are on {111} planes, and no {001} cross slip was observed by the weak-beam electron microscopy. It is also shown by atomistic modeling [9] that when a/2⟨110⟩ superpartials with APB are on the (111) plane, one of the three core configurations of a/2⟨110⟩ is glissile, with the core spreading in the APB plane. Thus a⟨110⟩ superdislocations can move on the {111} planes and contribute to the plastic deformation of $Al_{67}Mn_8Ti_{25}$ at higher temperatures, and the ductility of this alloy is greatly improved.

Conclusions

In the $L1_2$ $Al_{67}Mn_8Ti_{25}$ alloy deformed at room temperature and 673K, the a⟨110⟩ superdislocations dissociate into two a/3⟨112⟩ superpartials with SISF between them on {111} planes, while at 873K, the dissociation mode changes into a/2⟨110⟩ partials separated by APB on {111} planes. The temperature dependence of mechanical properties can be explained adequately by the dislocation structures and dissociation modes at different temperatures.

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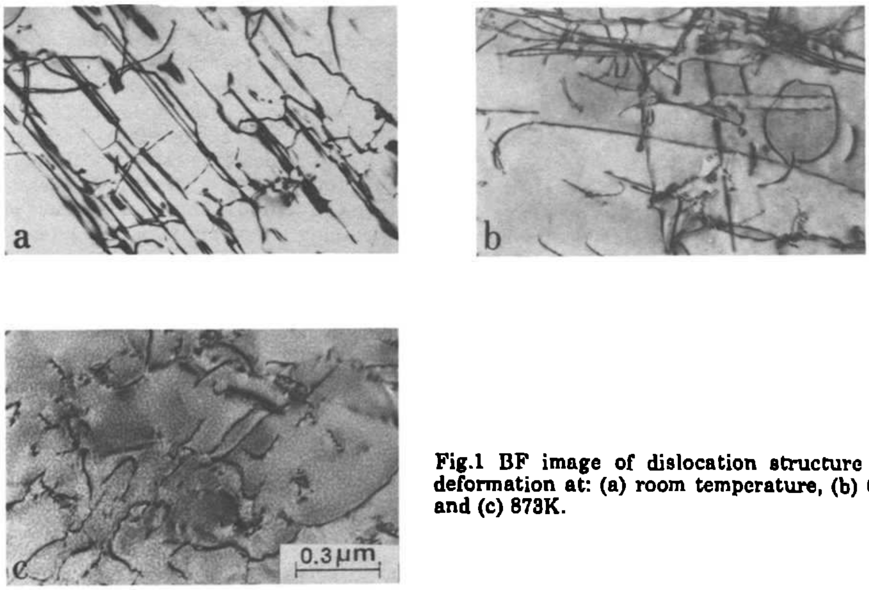

Fig.1 BF image of dislocation structure after deformation at: (a) room temperature, (b) 673K, and (c) 873K.

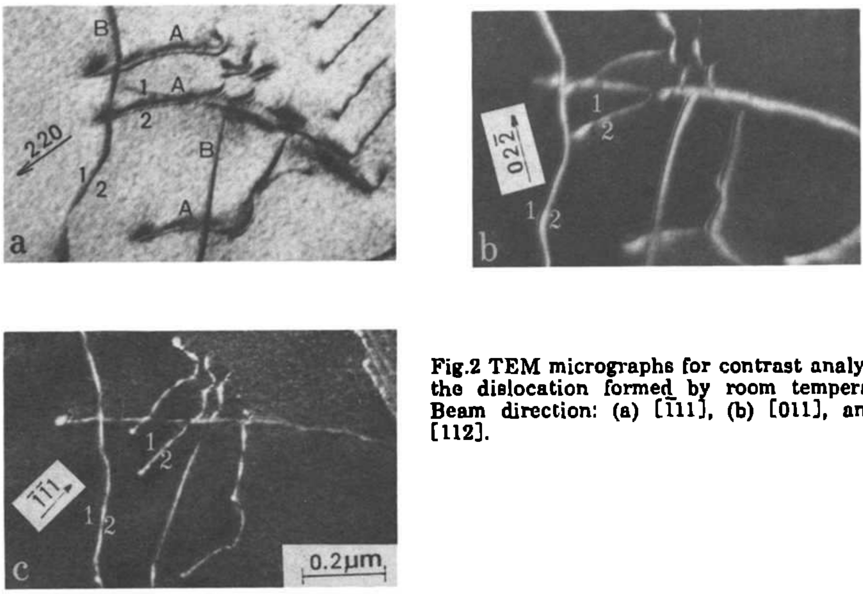

Fig.2 TEM micrographs for contrast analysis of the dislocation formed by room temperature. Beam direction: (a) $[\bar{1}11]$, (b) [011], and (c) [112].

Fig.3 TEM micrographs for contrast analysis of the dislocations formed at 873K deformation. Beam direction: (a) [112], (b) [111], and (c) [112].

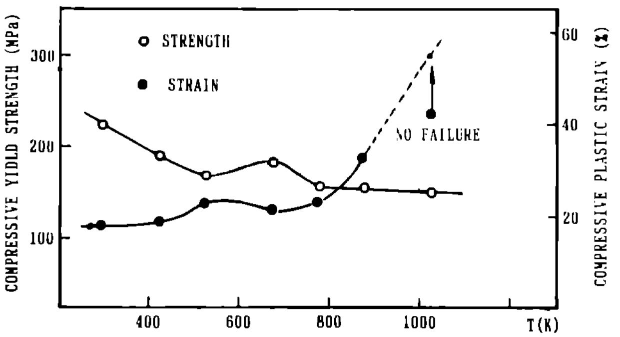

Fig.4 Temperature dependence of compressive yield stress and compressive ductility for $Al_{67}Mn_8Ti_{25}$