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Structure and mobility of martensite variant interfaces in a Cu-Zn-Al shape memory alloy

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Abstract. The structure and mobility of martensite variant interfaces have been investigated using TEM and HREM observations, and then the relations of interface structure to its mobility are revealed. It is shown that the A/C interface is straight, well-defined and perfectly coherent. The A/B interface is irrational, coherent and gradually curved. The A/D interface is stepped and the interface steps are the basal planes of the two variants. The intervariant-group interfaces are curved and blurred and some variants are tapered at the intervariant-group interface. The A/C and A/B interface have obvious mobility and may move smoothly and remain straight during moving. The mobility is not effective for A/D interface, and even worse for intervariant-group interfaces. The morphology, structure and mobility of interplate boundary are all related to the degree of self-accommodation and the misorientation of the twin boundary.

1. INTRODUCTION

It is well known that the mobility of various inter-crystalline boundaries in the martensitic microstructure plays a key role in regard to the shape memory effect and pseudoelasticity [1-3]. Thus, an understanding of the structure of these boundaries and their mobility are vital to a further understanding of the shape memory behavior. Three types of twin are seen in CuZnAl martensite variants, viz., the habit plane, $(\bar{1}28)$ and (1010) twins, termed A/B, A/C and A/D types [4], respectively. However, the structures and mobility of these boundaries in the same group or among different groups have been short of a clear and unambiguous explanation up to now.

In the present work, some TEM and HREM observations have been performed to elucidate the exact structure and mobility of martensite variant interfaces in a CuZnAl alloy, and then the relations among the interface structures and their mobility are revealed.

2. EXPERIMENTAL

A Cu-20.4Zn-5.6Al (wt%) alloy was prepared by melting the elements ((99.9% mass%) in an induction furnace followed by casting in an iron mold. After homogenization at 850℃ for 12h, the ingot was rolled into thin plate of 1mm in thickness approximately. The samples were held at 850℃ for 10min, quenched in boiling water, aged at 100℃ for 30min, and then air cooled to room temperature. The transformation temperatures determined by DSC were Ms=78℃, Mf=40℃, As=50℃, Af=86℃. The specimens were mechanically thinned to 50μm and then punched into 3mm diameter discs. Thin foils were prepared by electrolytic polishing with an automatic jet polishing apparatus in a solution of $HNO_3:CH_3OH=1:2$. TEM and HREM observations were carried out at room temperature in H-800 and JEOL 2000EX II electron microscope operated at 175kV and 200kV, respectively.

3. RESULTS AND DISCUSSION

3.1 TEM and HREM observations on the intervariant boundaries

Fig.1 shows a two-dimensional lattice image of A/C boundary plane. The A/C type boundary is quite straight and a coherent matching of atoms can be seen on the $(\bar{1}28)_{18R}$ boundary plane.

Fig.2 shows a two-dimensional lattice image of the type II twin boundary (A/B) taken along the $[210]_A//[\bar{2}92]_B$ direction. The boundary of Type II twin boundary is gradually and randomly curved with slight strain contrast. There is neither ledge nor step structure at the boundary, indicating that a strain around the boundary is elastically relaxed by gradually displacement of the atoms, although the quantitative estimation of the displacement could not be made.

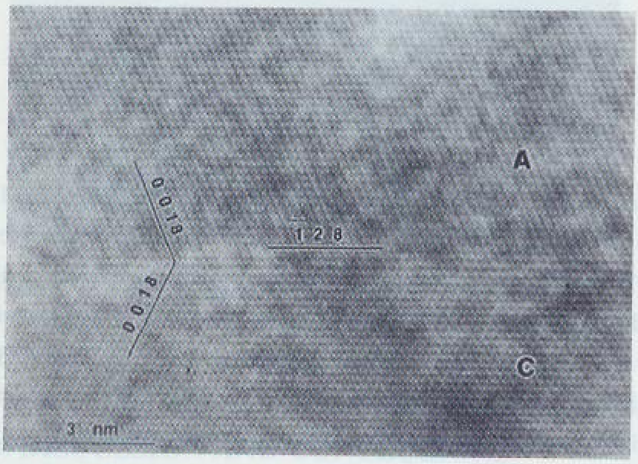

Fig.1 Lattice image of A/C type twin boundary

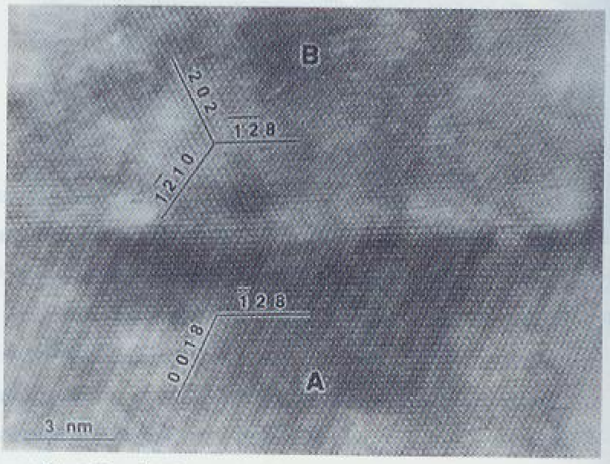

Fig.2 Lattice image of A/B type twin boundary

Fig.3 is a lattice image showing the $(1010)_{18R}$ A/D type twin boundary taken along the $[010]_{18R}$ zone axis. The upper variant appears misoriented which may be due to the fact that the $[010]_{18R}$ direction of the two variants are not exactly parallel (based on a calculation [5], the angle between $[010]_A$ and $[010]_D$ is 2.9°). Microscopically (on an atomic scale), the A/D type boundary consists of irregularly serrated steps, with the facet parallel to the basal plane of one or the other variant. Along the facet there is a distorted area where the facet deviates from its corresponding basal plane (shown by arrows B,C), i.e. A basal plane stacking fault can be identified in martensitic variant D. Unfortunately, the other stacking faults cannot be distinguished from the stacking faults sequence due to the misorientation. There is no significant difference in the boundary orientation, no matter whether the stacking sequence in the martensite variant is perfect or a single fault is present.

3.2 TEM observations on the interplate group boundaries

An example of a typical plate group combination is presented in Fig.4. Compared with straight and well-defined interplate boundary in the same group, the interplate group boundary is blurred and some strain contrast appears along the interplate group boundary. We now call attention to the interplate group junction plane between the 1 and 2 groups in Fig.4, which can be clearly delineated as parallel to the A/C or A/B boundary. In most observed cases, the interplate group boundary was parallel to either the A/B or A/C boundary of the impinged plate group which is also near $\{110\}_{\beta 1}$ planes.

3.3 Mobility of the various interplate boundaries in 18R martensite

In-situ observation [6] shows that A/B and A/C have obvious mobility. Fig.5 shows an example of motion of the A/B type boundary under tensile stress. During the initial stage, the boundary is not strongly

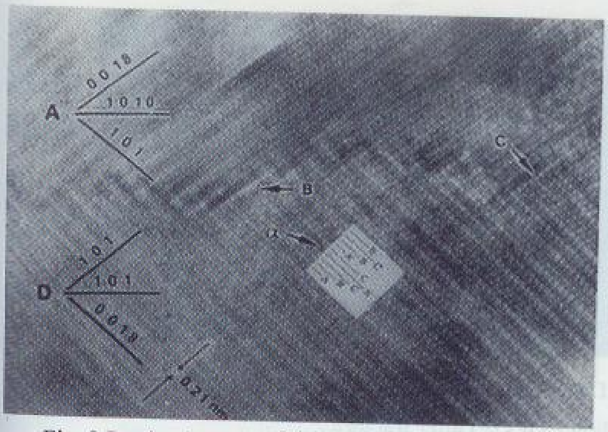

Fig. 3 Lattice image of A/D type twin boundary

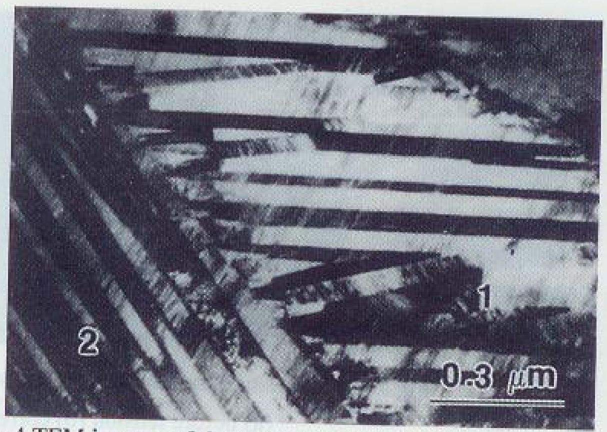

Fig. 4 TEM images of the combination of plate groups 1 and 2

activated by the external tensile stress; the observed phenomenon is only due to the adjustment of stacking fault contrasts. When the strain is ~1.5%, the boundary starts moving; while the strain is ~3%, the width of variant A between two arrows decreases by 0.1μm. At a low stress level, its movement was very smooth; when the stress rises, the reaction of the boundary is different from segment to segment. That is, certain segments of the boundary would move easier than others, resulting in a wandering boundary.

Fig.6 and 7 give the evolution of an A/D type boundary and different group intervariant boundaries in response to the applied stress, respectively. Though the A/D type boundary has been observed to move with lattice imaging technique [7] or by common TEM [8], we could not find substantial displacement during tension. When the strain reaches ~3%, the boundary in some sites starts to intrude from one side into the other side, as shown by arrows. With the increase of strain, the motion of the boundary was very jerky and irregular, as shown by arrows in Fig.6(d). The different group intervariant boundaries appear to have no reaction to the applied stress, and the increase of strain only results in big strain contrasts at the boundaries [denoted by arrows in Fig.7(c)]. The results show that these boundaries are immobile.

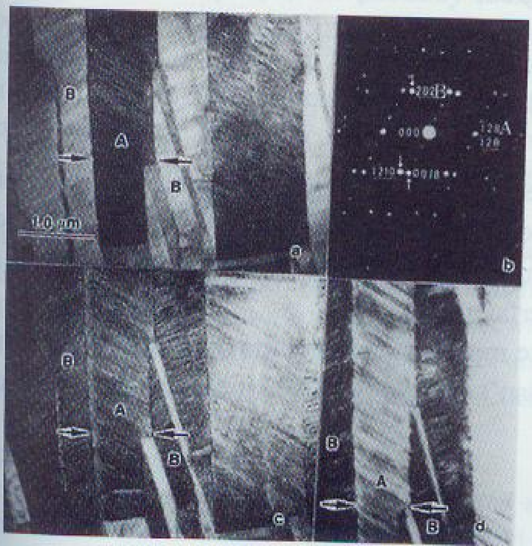

Fig.5 TEM in-situ observation of the motion of A/B boundary due to tensile strain (a) 0% (b) corresponding EDPs (c) ~1.5% (d) ~3%

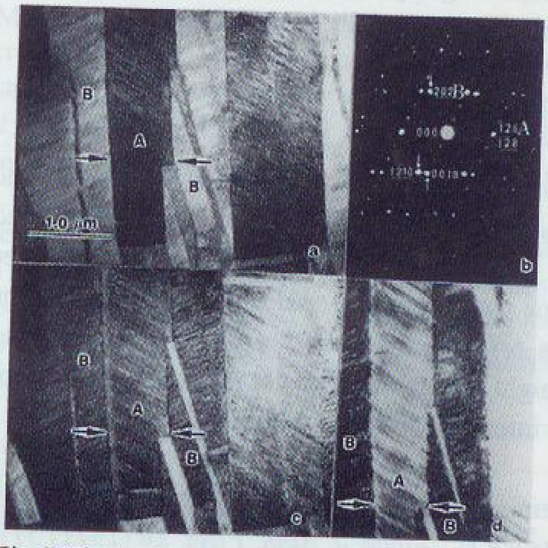

Fig.6 TEM in-situ observation of the motion of A/D boundary due to tensile strain (a) 0% (b) corresponding EDPs (c) ~3% (d) ~4.5%

The foregoing observations show definitely that different kinds of boundary in CuZnAl 18R martensite give different reaction under the applied stress. Though the A/C and A/B type boundary has weak Schmid factor, they have substantial reaction to the tensile stress. On the other hand, the A/D type boundary or different group intervariant boundaries have considerable favorable Schmid factor, but they

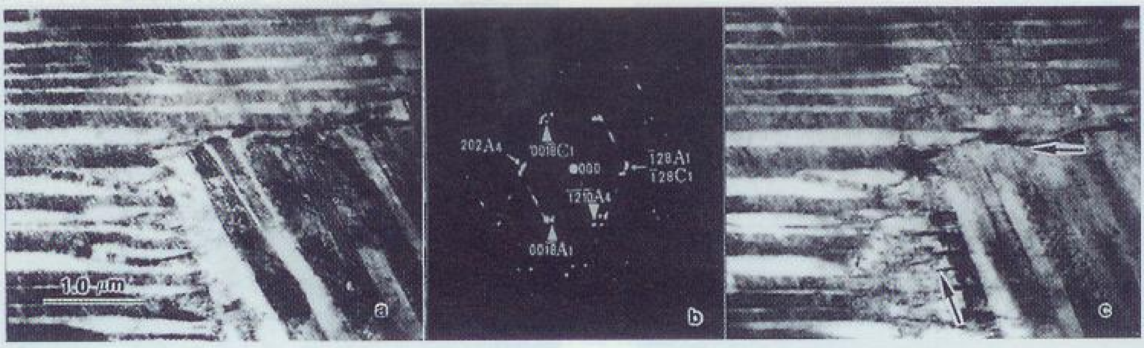

Fig.7 TEM in-situ observations of the motion of intervariant due to external tensile strain
(a) before tension (b) corresponding EDPs (c) after fracture

have a weak reaction or even have no reaction. Previous work on A/D type boundary has revealed its certain movement, though the movement cannot to be compared with that of A/C type. In our present situation, the A/D type boundary does not give obvious displacement. There are two reasons to be considered: one is that the A/D type boundary is not in a site of most preferable Schmid factor, and the other is that the tensile strain is not high enough to activate the boundary and the sample breaks before reaching a large strain. So it is with the different group intervariant boundaries.

On the basis of the analysis of the boundary morphology, we think that the mobility may be controlled by two factors: one is the degree of self-accommodation, and the other is the misorientation of the twin plane or matching plane. Nearly perfect self-accommodation and little misorientation results in the fact that the A/B and A/C boundaries have good mobility. Though A/D pair can be compared with A/B pair in self-accommodation, its big misorientation of the matching planes resists to movement of the boundary. As to the interplate group boundaries, bad or non-self-accommodation of variant-variant combination and big misorientation of the matching planes are enough to make these kind of boundary to have behavior like a small angle grain boundary or even a large angle grain boundary. Therefore, the interplate group boundaries are essentially immobile.

4. CONCLUSIONS

1. The A/C interface is straight, well-defined and perfectly coherent. The A/B interface is irrational, coherent and gradually curved. The A/D interface is stepped , and the boundary steps are composed of the basal planes of the two variants. The intervariant-group interfaces are curved and blurred and some variants are tapered at the intervariant-group interface.
2. The A/C interface and the A/B interface have effective mobility and may move smoothly and exhibit straight during moving. On the contrary, the mobility isn't effective for A/D interface, and even worse for intervariant-group interfaces.

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