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The nucleation and growth of martensite that is representative of bulk material is difficult to observe by transmission electron microscopy for the following reasons: A) In thin foils the martensite product may be completely different: Hull (1) reported that martensite in Cu-39 Wt % Zn had formed during electropolishing in the thinnest parts of the foils above the bulk transformation temperature Ms = -60°C and that its structure was severely modified compared to the bulk. Similar results were obtained by Pitsch (2) in Fe alloys which were prepared by vacuum evaporation of the constituents on collodion films. B) Martensite may not form at all even far below Ms for lack of nucleation sites if the volume of the original phase is too small as discussed by Easterling and Miekk-oja (3) who studied the transformation in precipitates of Fe in a Cu-Matrix. C) Otsuka et al (4) were able to induce the transformation in foils of β Ni-Al by cooling the samples inside the electron microscope but could not follow the growth because of the rapid drift of the specimen. Lobodyuk et al (5) avoided the drifting by observing foils of a Cu -14.1 % Al -5% Ni alloy slightly above their Ms temperature. In these samples the transformation was induced by stresses resulting from the local heating by the electron beam. The specimens generally were prepared by first rolling sheets of the alloys and then thinning them electrolytically in a rather homogeneous way.

We employed a different technique and prepared the samples by the thinning of small discs only in their center part. Due to the mechanical stability of the discs the drifting of the samples was reduced when being cooled inside the microscope, and in addition sufficient nucleation sites in the thicker part of the discs remained. In this way we were able to observe by transmission electron microscopy the growth of martensite during the cooling of the samples. The results are reported in this note.

Single crystals of Cu -39.7 and 40.1 at % Zn were grown in sealed

vycor capsules in a helium atmosphere, and subsequently quenched from the β- phase region into ice water. Small platelets of 3mm diameter were spark cut with their faces parallel to $(100)_\beta$. The samples were then thinned down preferentially in the center part by jet polishing at room temperature using a solution of one part orthophosphoric acid and four parts water. In a final step the electropolishing was continued in a bath of one part orthophosphoric acid and one part water at -12°C until the first hole appeared. The discs were inserted as a whole into the cold stage specimen holder and observed in a Philips EM 300 electron microscope. In the Cu -39.7 at % Zn alloy the start of the transformation during cooling manifested itself by the rapid movement of the area under observation: the shaking soon stopped, and if the cooling velocity was sufficiently slow, the growth of single martensite plates could be observed and photographed. In the Cu -40.1 at % Zn alloy the transformation could be induced by cooling the samples to liquid nitrogen temperature (77.3°K), but not when liquid air was used as a coolant. The drift of the samples was much less than in 39.7 at % Zn, probably due to the fact that less volume transformed. The start temperature for the transformation (-173°C) for 39.7 at % Zn and -196°C for 40.1 at % Zn is in close agreement with the temperature M_s at which burst type martensite appears on cooling the bulk material (6). In both alloys the transformation proceeded from the thick regions, rarely from the holes in the foil. The thinnest parts often transformed in a way similar to that reported by Hull (1), therefore we did not study this type in detail. During the transformation two different growth mechanisms could be discerned: a) a spike-like growth of thin sheets similar in appearance to the growth of twin faults in fcc metals. A growth sequence is shown in figure 1. b) a lateral thickening of the thin faults as seen in figure 2. In the Cu -40.1 at % Zn alloy few plates thickened after the first growth step, whereas in the 39.7 at % Zn alloy the thickening occurred often while the tip was growing. Near the tip the interfaces between martensite and matrix generally were not parallel to each other. As can be seen from figure 2, an internal structure inside the martensite exists. The transformation never was complete, and single transformation events continued even after the 39.7 at % alloys had reached liquid nitrogen temperature, probably due to stresses produced by local heating from the beam. When the foils were heated again, the martensite plates disap-

peared in the reverse manner. Occasionally when large areas had transformed, spikes of the β- phase were observed to advance into the martensite. A selected area diffraction pattern is shown in figure 3, and corresponds to a faulted orthorhombic structure that has also been observed for bulk type martensite (7). The orientation of the interface between the transformed region and the matrix has been analysed by the single trace analysis. The results are plotted in the unit stereographic triangle in figure 4 for the thin "faults" and after the martensite had thickened (24 platelets of each kind were analyzed). In the latter case only those martensite platelets have been used for the analysis whose two interfaces were parallel. The results show that the traces all pass through the $(2,11,12)_\beta$ pole but not through $(110)\beta$. The $(2,11,12)\beta$ plane is the commonly observed habit plane for bulk material. (6).

DISCUSSION

The close relationship between the present observation and the martensite formed in β brass bulk material, which manifests itself by the same habit and the similar structure, suggests that the transformation mechanism is essentially the same. The martensite crystallography in β Cu-Zn has successfully been described by the Wechsler-Lieberman-Read (WLR) phenomenological theory (8) and it is generally believed that a shear on a $\{110\}\beta$ $\langle 1\bar{1}0\rangle\beta$ system and an additional "shuffle" of the atoms basically describes the path of the atoms during the transformation. Whereas such a mechanism may explain the thickening of the martensite, it is difficult to see how the thin "faults" grow if only a shear on planes inclined to the fault plane is permitted. Therefore a different model is proposed in which the transformation is decomposed into a shear on the $(110)\beta$ planes that are nearly parallel to the habit plane and an "adjustment" shear on a $\{110\}\beta$ plane that is inclined to the primary plane. The model and its relationship to the WLR theory are discussed in detail elsewhere (9). Summarily it can be described as follows: In a first step partial dislocations with Burgers vector $\frac{a}{8}[1\bar{1}0]$ move on consecutive $(110)_\beta$ planes. During their movement they change over to adjacent planes, since the shear occurs on the average not on a $(110)_\beta$ but a $(2,11,12)_\beta$ plane. In a hard sphere model the shear is associated with a compression of 3.2.% normal to the plane. This transition structure has two $\{110\}_\beta$ plane variants, that are inclined to the shear plane and that have become

close packed planes. An additional shear on these "secondary" planes and a compression of 5.1 % normal to them transforms the structure into a close packed lattice with the same interatomic distance as the original β phase. If the shear plane is denoted by $(111)_{fcc}$ the shear direction is either $\frac{a}{9}[11\bar{2}]_{fcc}$ or $\frac{a}{18}[\bar{1}\bar{1}2]_{fcc}$. Finally a homogeneous distortion adjusts the lattice parameters to those observed. The total adjustment shear γ_2 , which is composed of elemental shears in $\frac{a}{9}[11\bar{2}]_{fcc}$ and $\frac{a}{18}[\bar{1}\bar{1}2]_{fcc}$ is determined by the condition that an undistorted habit plane exists. It can be shown (9) that the present model and the WLR theory predict for β- brass the same habit plane, macroscopic shear and orientation relationship. The total adjustment shear γ_2 on the secondary shear plane for β- brass is small (9). If it were zero, two elemental shears in $-[11\bar{2}]$direction are compensated for by one shear step in $+[\bar{1}\bar{1}2]$ When these shears are distributed evenly, an orthorhombic ABCBCACAB stacking order results. For γ_2 different from zero, additional stacking faults have to be introduced. This orthorhombic structure indeed is the generally observed one in β- brass (7). The use of the rigid sphere model and the decomposition of the homogeneous distortions into three steps has been made for convenience only. Whereas the two shears are likely to occur simultaneously during the growth of the martensite, the first shear may be preceded by the second one during the nucleation stage, in which the intermediate structure is stable until the nucleus reaches a critical size. This type of nucleation by a shear is athermal and similar to the nucleation of martensite in cobalt.

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b)

a)

FIG. 1

Martensite growing into the β matrix. Foil plane parallel $(110)_\beta$
(Dirt spots on the surface of the foil permit to relate the areas of the photos).

b)

ج)

FIG. 2

Thickening of two martensite plates. Foil plane parallel $(110)_\beta$

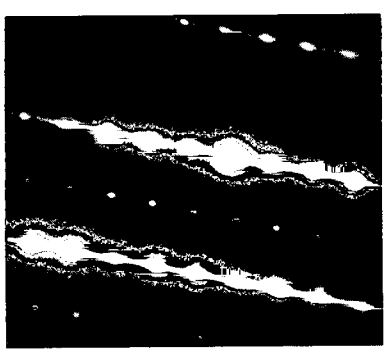

FIG. 3

Selected area difraction of a martensite plate

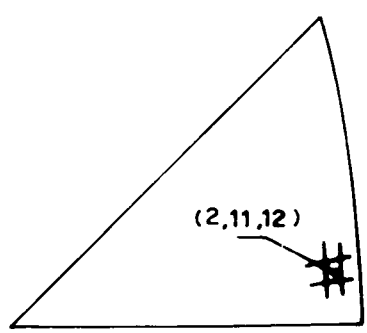

FIG. 4

Experimental habit plane