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ON THE FORMATION OF COMPLEX MARTENSITES IN QUENCHED Cu–Al AND Cu–Al–Ni ALLOYS

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Although the structures of martensites formed in β–Cu_3Al and Cu –14.2 Al –4.3 Ni (wt.%) upon fast cooling are fairly well known, there is still some controversy regarding the transformation sequences, as well as the mechanisms operating during formation of these martensites. It is believed[1–4], in general, that both martensites, the β_1'(3R) and the γ_1'(2H) are formed directly from the intermediate β_1 phase which has an ordered b.c.c. structure (1), although it is formally more convenient to derive the product structures from the ordered f.c.c. (DO_{22}) which is uniquely related to the β_1 phase (2). Both transformation sequences (1) and (2) are, however, equivalent so far as the correspondence between the original (parent) and final structure (product) within one plate is concerned. If, as will be shown in the present work, the relative orientational relationship between two adjacent martensite plates (the compound plate) which originate from a single crystal of the parent structure, is taken into account, some additional insight into the transformational peculiarities can be obtained.

Fig. 1 represents a transmission electron micrograph of two martensite plates in a bulk specimen which was suitably heat treated and electrothinned before examination in an electron microscope (a); and the corresponding diffraction pattern (b). A pair of martensite plates is shown in an orientation where the interface between two plates and the (001)β_1' planes in each plate are parallel to the electron beam. From the corresponding diffraction pattern, which is a superimposition of two symmetrical patterns of the 3R structure (periodicity 1/3 in reciprocal space), it is also evident that the interface is a coherent boundary of the twin type. The angle of 129° between close packed planes of the $\{001\}\beta_1'$ type is indicated.

Fig. 2, which was published by Otsuka and Shimizu[2], similarly represents a transmission electron micrograph of the so-called "small" martensite in the Cu –14.2 Al –4.3 Ni alloy, where again the interface between two plates and the (001)β_1' close packed planes are parallel to the electron beam. The electron diffraction pattern exhibits a periodicity of 1/2, which indicates hexagonal stacking (2H) of the close packed planes. In this case an angle of 123.4° is measurable between the basal planes of two plates*.

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*The transition to the 2H structure is usually observed to occur in very thin foils[1,2,3], i.e. in the absence of constraints in the surrounding matrix.

In the discussion which follows some kind of simplified phenomenological analysis will be applied, i.e. the phase transitions will be shown schematically through their influence onto the butterfly shaped piece of material limited by the $\{110\}_1$ planes which are parallel to the $[111]\beta_1$ direction. Apart from the generally accepted crystallographic features which characterize martensitic transitions[5], two additional assumptions will be applied:

i) the relative atomic arrangement of the ordered parent structure is retained during the martensitic transition into a close packed complex product;

ii) The complex product formed in bulk must have a self-accommodating shape.

The first assumption is common to all analyses mentioned above, and the second was justified by Taš et al.[4] for Cu_3Al alloy.

Let us consider first the transition (2), i.e. $\beta_1(DO_3) \rightarrow \alpha_1'(DO_{22}) \rightarrow \beta_1'(3R)$ which is shown schematically in Fig. 3a. The influence of the homogeneous (Bain) and the influence of the inhomogeneous (lattice invariant shear) parts of the martensitic transition are shown separately, because in self-accommodating martensites they have to cancel each other ii). Note that in this case condition ii) is fulfilled because the net effect of the lattice invariant shears in both plates is to diminish the shape deformation caused by the Bain deformations in each plate. I denotes the correspondence where the prominent axis of the new structure (extension) is along the $[100]\beta_1$ and the other two axes (contraction) along the $[010]\beta_1$ and the $[001]\beta_1$, respectively. Similarly II means that the prominent axis of the new structure in this variant is along the $[010]\beta_1$. In Fig. 3b it is clearly demonstrated that direct transition from the b.c. cubic to the orthorhombic structure does not bring the close packed planes in both plates to the position where they would form an angle of 129° as is observed experimentally. Also, it has to be noted that condition ii) is not satisfied, i.e. the lattice invariant deformation acts in the same sense as the homogeneous one. On the other hand it is evident from Fig. 3c that the direct transition from b.c. cubic to the orthorhombic structure fits the experimentally observed microstructure shown in Fig. 2. In this case the lattice invariant deformations do not change the macroscopic shape at all, but just shear the orthorhombic structure in opposite directions to form the hexagonal stacking. It is interesting to note that this sequence was mentioned as possible by Kajiwara[1], who considered a b.c.c. → h.c.p. transformation and found that the correspondence between these two structures is somewhat unusual.

Let us now consider the transformations (1) and (2) in the frame of the stress–temperature diagram proposed recently by Otsuka et al.[6] (Fig. 4). One can see that both sequences (1) and (2) considered above are quite plausible. The existence of the intermediate phases, the orthorhombic in case (1) and the α_1' in the case (2), respectively, indicates that, at least in the first stage of the transformation, the nucleation of the new phase is accompanied by constraints in the surrounding matrix. However, in case (2) these constraints are so high that the transformation path passes through the region of the α_1' phase.

Consider now transitions (1) and (2) with respect to condition i). It has been already shown that the lattice invariant shear which occurs in this alloy does not disturb the ordered structure[4]. There is only one direction (out of three) in each close packed (or basal) plane which satisfies this condition. It is also easy to show that both lattice invariant shears in variant I and in variant II (Fig. 3a) which originate from the $[0\bar{1}1]\,(011)\beta_1$ and the $[011]\,(0\bar{1}1)\beta_1$, shear systems satisfy condition i). Hence, it has to be proved that the irregularity along the interface between the two plates I and II is of the same type as the irregularity created by the lattice invariant shears mentioned above. Let us consider the relative orientational relationship between I and II. One can see that II can be created from I by the twin producing shear which originates from the $[110]\,(1\bar{1}0)\beta_1$ shear system (primary shear). It is easy to check now that glide in this shear system belongs to the same set of shears which does not disturb the ordered structure. Hence the complex structure of the compound plate is formed by the trinity of shears which satisfies condition i), although the role of the so-called primary shear is somewhat different.

The results presented above prove that on the basis of the relative orientational relationship between adjacent martensite plates which originate from a single crystal parent some valuable conclusions can be made.

A more detailed account of this work will be published elsewhere.

References

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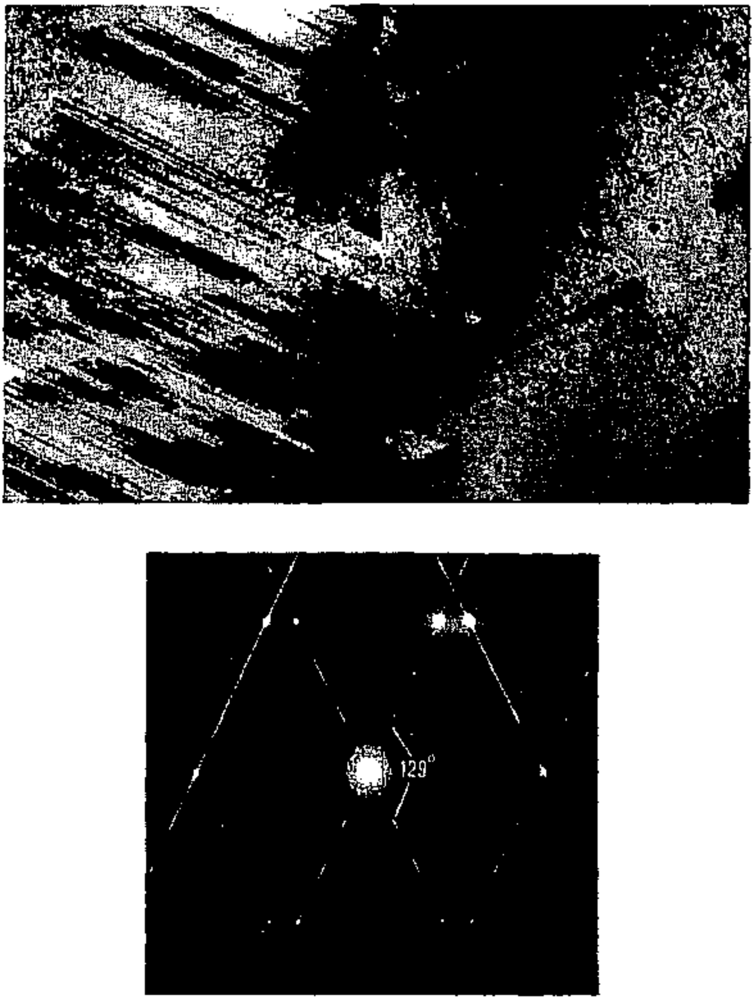

Fig 1

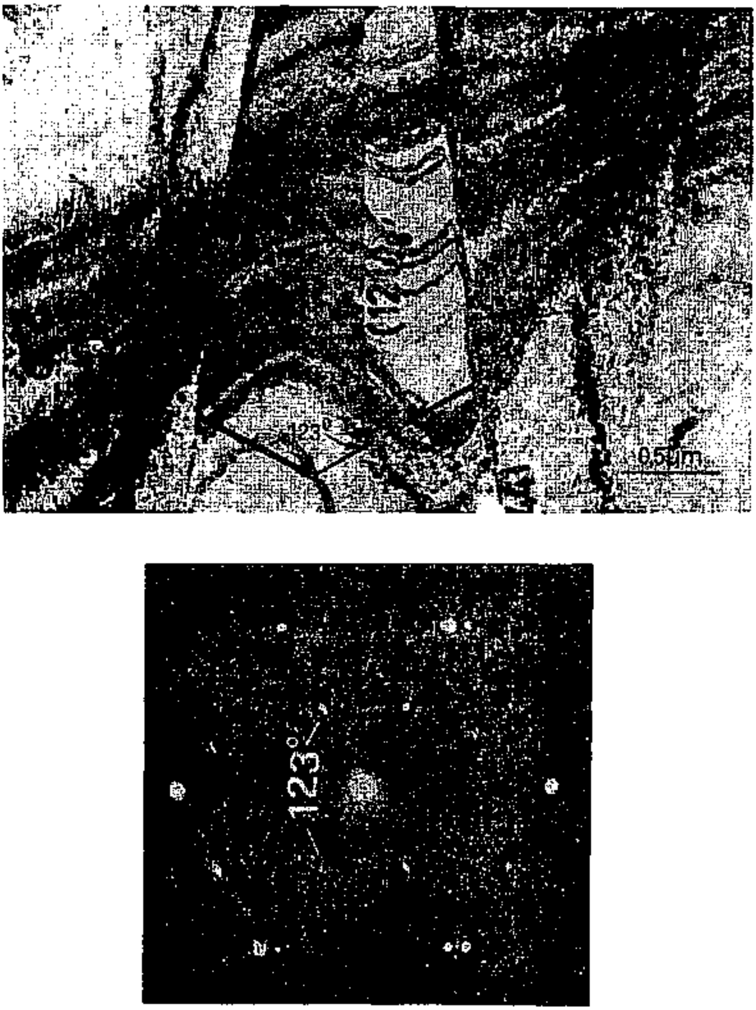

Fig 2

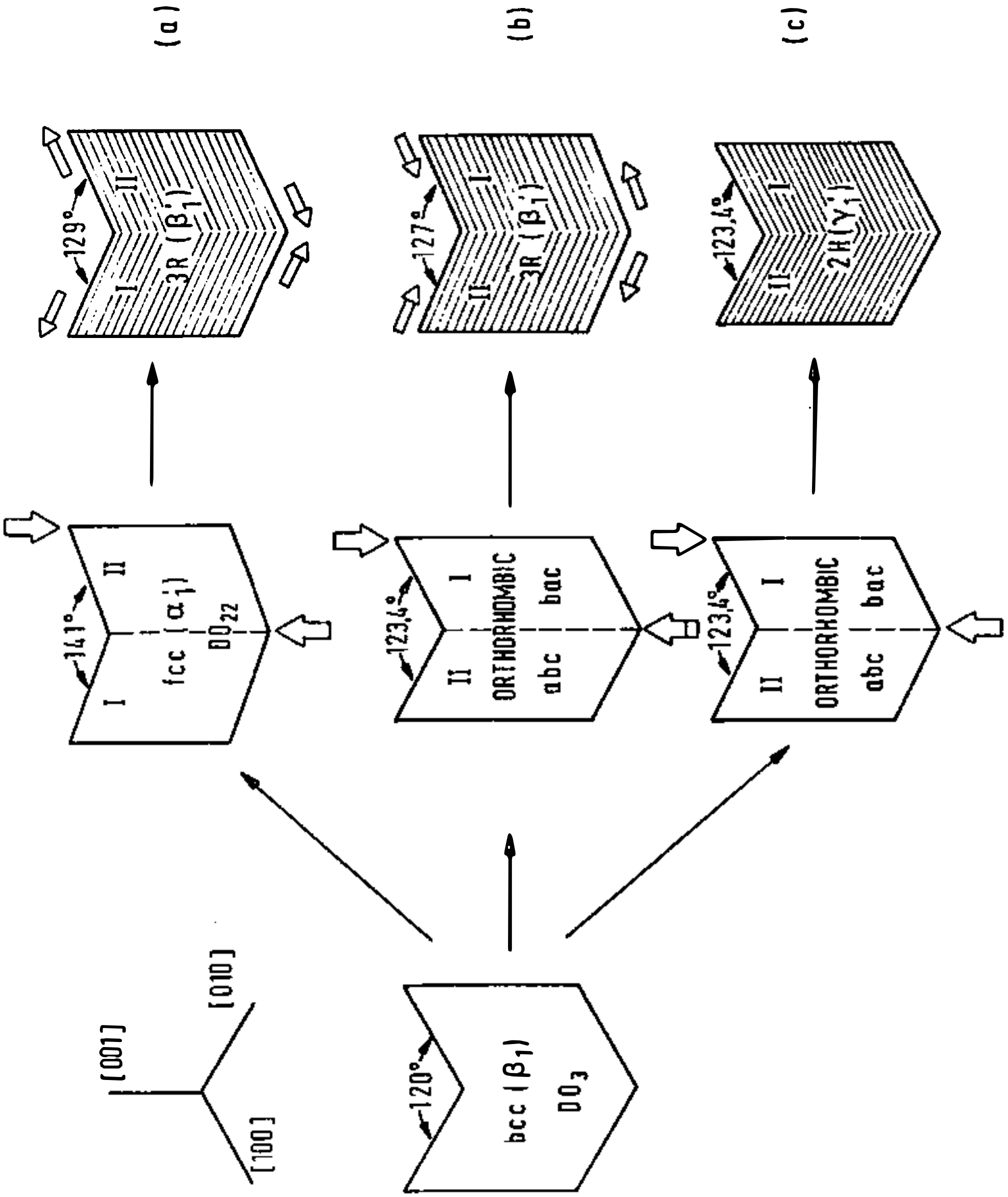
(a)
(b)
(c)
129°
I
II
3R (β'1)
127°
II
I
3R (β'1)
123,4°
II
I
2H (γ'1)
141°
I
II
fcc (α'1)
DO22
123,4°
II
I
ORTHORHOMBIC
abc
bac
123,4°
II
I
ORTHORHOMBIC
abc
bac
120°
bcc (β1)
DO3
[001]
[010]
[100]

Fig3

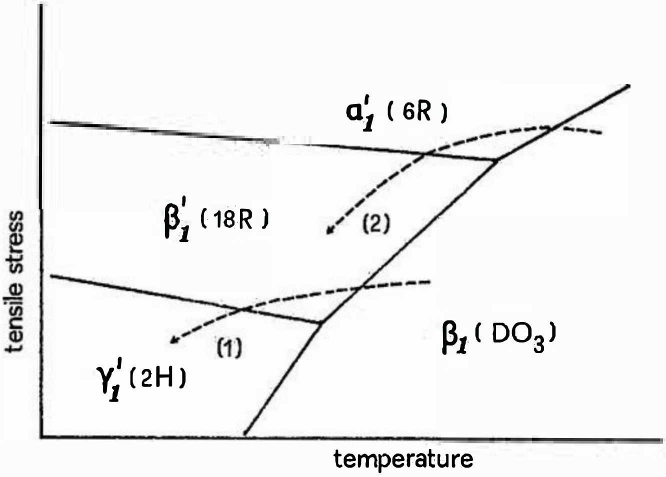

Fig 4