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THE CHARACTERIZATION OF PRETRANSFORMATION MORPHOLOGIES : PERIODIC STRAIN MODULATIONS

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Abstract. - The parent cubic phases of alloys which undergo martensitic transformations exhibit various pretransformational effects including the development of fine-scale periodic strain modulations. This diffuse microstructure, revealed by electron microscopy and diffraction, is identified as a dense array of incommensurate shear strains of {110} <1$\bar{1}$0> type. This behavior, which results in an incipient loss of cubic symmetry, appears to be related to the concurrent parent lattice softening (i.e. the decrease of C' with decreasing temperature approaching M_s) that is observed in most of the same alloys.

Solid state phase transformations proceed by a variety of mechanisms involving either thermally activated diffusional atomic rearrangements or athermal diffusionless atomic displacements. Multiple transitions are also possible, occurring sequentially or concurrently depending upon the interplay of thermodynamic and kinetic factors established by material treatment conditions. In the course of such treatment cycles, parent phases often develop fluctuations in atomic order or concentration and/or lattice strain and these tend to anticipate critical characteristics of the impending transition(s). Of these, fluctuations in strain are usually associated with the displacive mechanisms and in particular, a periodic modulation of shear strains has been found to be common to most of the various and diverse alloy systems that undergo martensitic transformations [1,2]. A representative, though far from an all inclusive listing of the systems in which these effects have been observed is given in table 1.

The strain modulations are readily detected by electron microscopy and electron and x-ray diffraction. Transmission electron micrographs exhibit a fine -scale, diffuse, striated microstructure which is commonly referred to as "tweed" and is shown in figures 1a,b and 3 [3]. The striations, typically of 3-6 nm periodicity, lie parallel to {110} traces of the parent cubic phases, where the parent structure may be variously disordered bcc, disordered fcc or ordered B2, DO_3, $L1_2$, etc. (see table 1). The corresponding diffraction patterns reveal diffuse rel-streaks along <110>* at each Bragg reflection as seen in figure 1c as well as schematically in figure 2 [3(a),(b)]. The unique features of the diffuse scattering are as follows: streaking is not present at 000, whereas selected streak absences occur systematically at each reflection, and the length of streaks increases with increasing order of reflection. The diffuse intensity, I_d, about each Bragg reflection is related to strain by

$$I_d \sim \bar{\varepsilon}_k \cdot \bar{g}$$

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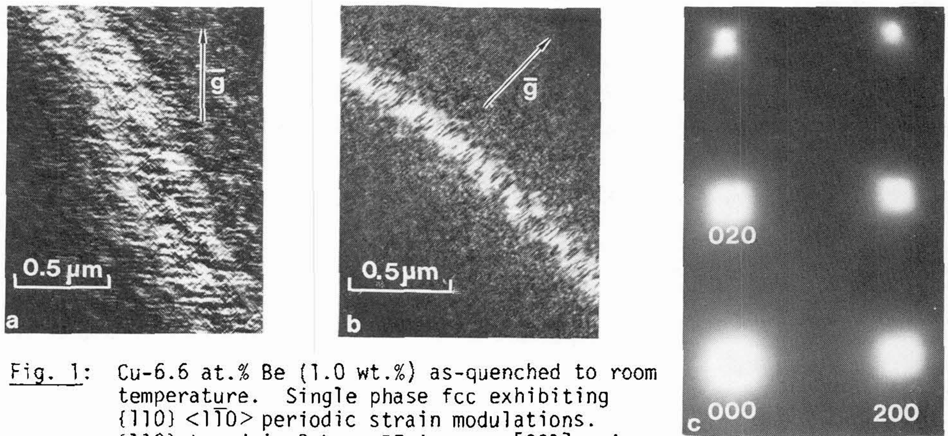


Fig. 1: Cu-6.6 at.% Be (1.0 wt.%) as-quenched to room temperature. Single phase fcc exhibiting {110} <1$\bar{1}$0> periodic strain modulations. {110} tweed in 2-beam BF images, [001] orientation, (a) $\bar{g}$=[002]*, (b) $\bar{g}$=[220]*, (c) [001]* SADP showing <110>* rel-streaks.

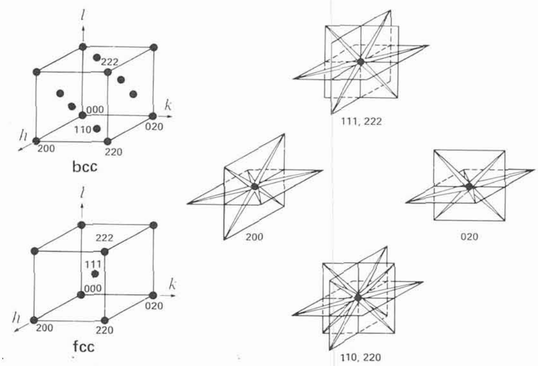
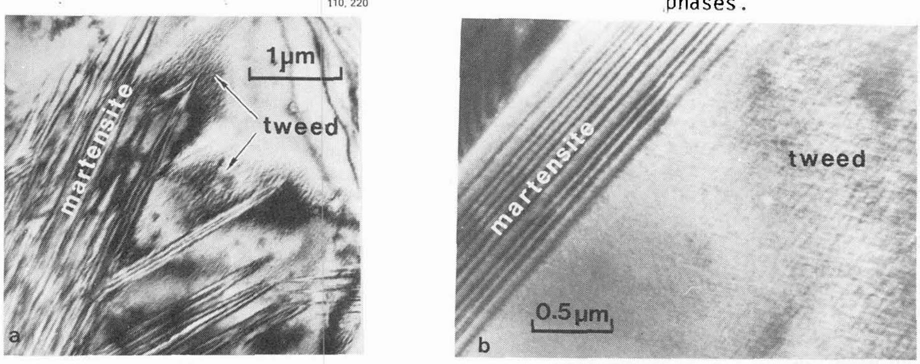


Fig. 2: Schematic representation of the <110>* strain diffuse scattering in either bcc or fcc parent phases.



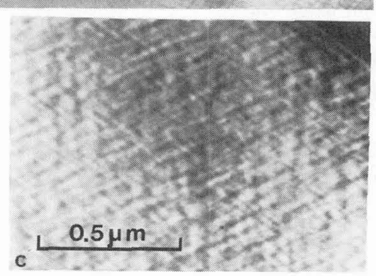


Fig. 3 Tweed strain contrast due to {110} <1$\bar{1}$0> periodic strain modulations within parent cubic phases of (a) Mn-26 at.% Cu at ∿180°K (from K. Shimizu), (b) Fe - 30.6 at.% Pd at 213°K (from R. Oshima) and (c) Ni - 36.8 at.% Al at room temperature (from A. Lasalmonie).

where lattice strains are represented by a displacement wave of vector $\bar{\varepsilon}_k$ and $\bar{g}$ is a reciprocal lattice vector [4,5]. Streaks are therefore absent when $\bar{\varepsilon}_k \cdot \bar{g} = 0$. Note that the same criterion also applies to the microstructure; that is, tweed striations absent in an image are those which would lie along traces normal to the directions of missing rel-streaks in a corresponding diffraction pattern [3(a)]. The foregoing uniquely identifies the strains as shears of $\{110\}\langle 1\bar{1}0\rangle$ type or equivalently modelled as transverse displacement waves of $\langle 110\rangle$ wave vector and $\langle 1\bar{1}0\rangle$ polarization vector.

The interaction of such waves can be shown to give rise to a modulated array of incommensurate displacements (static and/or dynamic). The result is an incipient loss of cubic symmetry which translates the parent phase toward tetragonal symmetry as the transformation is approached [5]. This behavior can be related to the concurrent temperature dependent "softening" of the parent lattice since the appropriate shears are involved. That is, these cubic phases generally experience an anomolous decrease of the elastic constants $C' = 1/2(C_{11}-C_{12})$ with decreasing temperature approaching M_s (see table 1) [6]. In addition, there is mounting evidence that $\{110\}\langle 1\bar{1}0\rangle$ shears are characteristic of the initial step in the formation martensite in most systems [1,2,6,10]. These apparent links in structure and behavior seem more than coincidental and thus suggest an intimate relationship between the pretransformational variations in state and the martensitic transformations themselves. Proper elucidation of the nature and significance of the foregoing will require further experimental investigation, as well as theoretical modelling. Our approach to the former is to carry out structural studies utilizing high resolution electron microscopy. As to the latter, several new theories incorporating pretransformational effects, localized soft phonon mode concepts and/or non-classical nucleation are currently being developed [7-10].

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table 1 MARTENSITIC SYSTEMS EXHIBITING PRETRANSFORMATION {110} <1$\bar{1}$0> PERIODIC STRAIN MODULATIONS

Alloy System	Parent Phase Structure	C' Softening Observed	Reference
In-Tl	fcc	✓	LASALMONIE, A., et al, Proc. ICOMAT-79 (MIT, Camb.) 1979, p. 538. NITTONO, O., et al, Trans. Jap. Inst. Met., 23 (1982) 285.
Mn-Cu	fcc	✓	SHIMIZU, K., et al, Trans. Jap. Inst. Met., 23 (1982) 53. VINTAIKIN, E. Z., et al, Sov. Phys. Dokl., 26 (1981) 617.
Mn-Ga	fcc	no data	VLASOVA, E. N., et al, Sov. Phys. Dokl., 22 (1977) 765.
Fe-Ni	fcc	✓	KONDRAT'YEV, V. V., et al, PMM, 45, No. 4 (1978) 82.
Fe-Pd	fcc	✓	OSHIMA, R., Scripta Met., 15 (1981) 829. SATO, M., et al, J. Phys. F (1982) in press.
Fe-Pt	$L1_2$	✓	FOOS, M., et al, Proc. ICOMAT-79 (MIT, Camb.) 1979, p. 485. FOOS, M., et al, Acta Met., 29 (1981) 1091.
Cu-Be	fcc,bcc,B2	no data	AUVRAY, X., et al, Scripta Met., 8 (1974) 995. SVANIDZE, L. S., et al, FMM, 51, No. 5 (1981) 883.
U-Nb	bcc	✓	GIRAUD-HERAUD, F., Mem. Sci. Rev. Met., 71 (1974) 37. BLAKE, D., Phys. Met. of U Alloys, (Brk. Hill, Bos.) 1976, p.189.
V-O	bcc	no data	HIRAGA, K., et al, Jap. J. Appl. Phys., 19 (1980) 397.
Cu-Zn	B2	✓	OTSUKA, K., et al, Met. Trans., 12A (1981) 595.
Ni-Al	B2	✓	ENAMI, K., et al, Scripta Met., 10 (1976) 879. LASALMONIE, A., Scripta Met., 11 (1977) 527.
Au-Cd	B2	✓	MUKHERJEE, K., Shape Mem. Effects in Alloys(Plenum, NY) 1975 p.177.