

Slip-Induced Directional Order in Fe-Ni Alloys. I. Extension of the Chikazumi-Suzuki-Iwata Theory

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In the present paper, calculations have been made for the magnetic anisotropy developed in cold-worked iron-nickel alloys near the 75% Ni composition. The treatment is based on the slip-induced directional order theory of Chikazumi *et al.* [*J. Phys. Soc. Japan* **12**, 1259 (1957)]. The theory states that in an ordered ferromagnetic solid solution (long- or short-range order, or clustering), slip results in the creation and preferential alignment of distinct type(s) of atom pairs (e.g., Fe-Fe) which lead to an induced magnetic anisotropy. The important conclusion from this theory is that the direction of the induced easy axis of magnetization depends sensitively on the orientation of the deforming crystal and the geometry of the deformation process, as well as the type and degree of ordering in the alloy. The present treatment extends the original analysis for the case of rolling to single crystals of several new orientations: (110)[$\bar{1}12$], (110)[$\bar{1}10$], (111)[$\bar{1}12$], (112)[$\bar{1}10$], and (112)[$\bar{1}11$]. These orientations, plus those studied by Chikazumi *et al.* ((100)[001], (100)[011], and (110)[001]), contain the most likely operating slip systems whose slip directions are symmetrically disposed about the rolling direction and are hence more stable than others during rolling. Some of these orientations are important components of texture in recrystallized or cold-rolled polycrystalline face-centered cubic materials. In addition to the rolling analysis, calculations have been made for the anisotropy resulting from drawing of wires with (111) and (001) axial orientations. These two orientations are likewise prominent components of polycrystalline wire textures.

The present work shows that drawing of a (001) wire induces a hard magnetic direction along the wire axis, while this axis becomes an easy direction if the orientation is (111). For rolling, except for the cases of (001)[100] and (110)[001] studied by Chikazumi *et al.*, where the induced easy axis lies along the transverse direction, most other orientations generally place the easy axis along the rolling direction. This suggests a plausible explanation for the observation that the rolling direction becomes an easy axis of magnetization when a randomly oriented polycrystalline aggregate is rolled. This work also predicts that rolling of {112}(111) and {110}(112) textures (symmetrical variants included), which are prominent texture components in rolled Permalloy, results in an easy axis along the rolling direction. Suggestions for experimental testing of the theory are discussed.

INTRODUCTION

THE phenomenon of cold-work induced magnetic anisotropy in iron-nickel alloys has been studied in detail in connection with the early development of Isoperm, a 50% Fe-50% Ni alloy.¹⁻³ It was found that rolling of a cube-textured alloy induces an "easy" axis of magnetization along the transverse direction. The induced easy axis lies along the rolling direction, however, if the initial grain orientation were random. This phenomenon was found to exist throughout the face-centered cubic range (at room temperature) of the Fe-Ni system and was unrelated to magnetostriction or crystalline anisotropy. Bunge and Mueller⁴ and Chikazumi, Suzuki, and Iwata⁵ (hereafter called CSI) proposed a slip-induced directional order theory to account for this peculiar anisotropy. Briefly, the theory states that in a ferromagnetic alloy containing some degree of chemical order (long or short range), there are more unlike nearest-neighbor atom pairs than in a random solid solution. During plastic deformation, like-atom pairs are created at the expense of unlike

pairs.⁶ Due to the crystallographic nature of slip, these induced like-atom pairs are distributed asymmetrically. Such asymmetrical distribution results in a magnetic anisotropy since it is expected from the theory of magnetic annealing^{7,8} that the pseudodipolar magnetic coupling energy of an atom pair in a solid solution of *A* and *B* atoms depends on the identity of the pair; that is *AA*, *BB*, or *AB*. In this manner, Bunge and Mueller accounted qualitatively for the directional dependence of slip-induced anisotropy in rolling both randomly oriented and cube-textured materials. The treatment of CSI was more complete. They considered the difference in slip behavior between long- and short-range-ordered structures and analyzed the slip systems operating to accommodate the rolling deformation. The calculations based on rolling single crystals of Permalloy (76% Ni-24%Fe) of several orientations were found in satisfactory agreement with observed magnetic data. Later work on Fe₃Al⁹ and Ni-Co alloys¹⁰ again produced satisfactory results.

⁶ The theory is equally applicable to the case of clustering in which atoms prefer like neighbors. Plastic deformation would then result in unlike atom pairs at the expense of like ones.

⁷ L. Neel, *J. Phys. Radium* **15**, 225 (1954).

⁸ S. Taniguchi and M. Yamamoto, *Sci. Rept. Res. Inst. Tohoku Univ.* **A6**, 330 (1954).

⁹ S. Chikazumi, K. Suzuki, and H. Iwata, *J. Phys. Soc. Japan* **15**, 250 (1960).

¹⁰ N. Tamagawa, Y. Nakagawa, and S. Chikazumi, *J. Phys. Soc. Japan* **17**, 1256 (1962).

¹ W. H. Six, J. L. Snoek, and W. G. Burgers, *De Ingenieur* **49**, E 195 (1934).

² H. W. Conradt, O. Dahl, and K. Sixtus, *Z. Metallk.* **32**, 231 (1940).

³ G. W. Rathenau and J. L. Snoek, *Physica* **8**, 555 (1941).

⁴ H. J. Bunge and H. G. Mueller, *Z. Metallk.* **48**, 26 (1957).

⁵ S. Chikazumi, K. Suzuki, and H. Iwata, *J. Phys. Soc. Japan* **12**, 1259 (1957).

It should be pointed out that Neel⁷ has also interpreted the cold-work induced anisotropy on the basis of directional order theory. He suggested, however, that the ordering is a result of the large applied stresses coupled with diffusion enhanced by plastic deformation. This theory was criticized by Bunge and Mueller on grounds that the Bauschinger effect had to be invoked to explain the difference in the directional dependence of the induced easy axis between randomly oriented and cube-textured polycrystalline materials.

This paper extends the CSI analysis to rolling crystals of several new orientations of FeNi₃ composition, with particular emphasis on those orientations encountered in textured polycrystalline materials. In addition, since slip-induced directional order need not be restricted to rolling deformation, analyses were made for the case of wire-drawing. The present aim is twofold. First, since the CSI analysis predicts different directions for the induced easy axis depending on the crystal orientation and the type of deformation processing, study of a wide range of orientations and different types of deformation processing will further test the theory. Secondly, commercial fabrication of magnetic alloy components often entails a series of complicated thermal and mechanical treatments. As different textures may be developed by these treatments, their effects on the magnetic anisotropy of the finished product can be overriding and must not be ignored.

CHIKAZUMI-SUZUKI-IWATA ANALYSIS

The magnetic energy density of an alloy crystal of *A* and *B* atoms, due to creation of *BB* atom pairs induced by slip, can be written as

$$E = l \sum_i N_{BBi} \cos^2 \varphi_i, \quad (1)$$

where $l = l_{AA} + l_{BB} - 2l_{AB}$, with l_{AA} , l_{BB} , and l_{AB} as the coefficients of pseudodipolar coupling of *AA*, *BB*, and *AB* atom pairs. N_{BBi} is the number of *BB* pairs per unit volume created by the slip on slip system *i*, and φ_i is the angle between the local magnetization vector and the induced *BB* pair direction due to the *i*th slip system.

In calculating the distribution of induced *BB* pairs, CSI distinguished two types of deformation: (1) deformation of a long-range-ordered lattice with slip confined within the ordered domains and (2) deformation of a short-range-ordered lattice, or of a long-range-ordered lattice with slip extending beyond the domain boundaries. In the following these will be denoted as L.F. and S.C. types, respectively, in accordance with the original notation. The essential difference between the two types is that, for alloys of face-centered cubic structure which usually slip on {111} planes and along <110> directions, the induced *BB* pair direction (for a given slip system) in L.F. deformation is that <110> direction perpendicular to the slip direction. In the S.C. deformation, the slip plane normal becomes the effective induced *BB* pair direction. The expressions for

the induced magnetic energy density according to these two types of deformation have been derived in the original paper and will not be repeated here. The results for a Ni-25%Fe alloy are

$$E_{LF} = \frac{1}{8} K_{LF} \sum_i |S_i| (\alpha_1 \beta_{1i} + \alpha_2 \beta_{2i} + \alpha_3 \beta_{3i})^2, \quad (2)$$

$$E_{SC} = \frac{1}{16} K_{SC} \sum_i |S_i| (n_{2i} n_{3i} \alpha_2 \alpha_3 + n_{3i} n_{1i} \alpha_3 \alpha_1 + n_{1i} n_{2i} \alpha_1 \alpha_2), \quad (3)$$

where

$$K_{LF} = N l p_0 p' s^2$$

$$K_{SC} = N l p' \sigma$$

N = number of atoms per unit volume.

$Nl \sim 3.1 \times 10^8$ erg/cc for Fe-Ni alloys.⁵

p_0 = probability that a dislocation will not be paired with another to form a superdislocation.

p' = probability of one dislocation passed per atomic (slip) plane.

s = Bragg and Williams long-range-order parameter.

σ = Bethe short-range-order parameter.

$\alpha_{1,2,3}$ = direction cosines of the local magnetization vector with respect to the cubic axes of the crystal.

$\beta_{1i,2i,3i}$ = direction cosines of the induced *BB* pair direction for slip system *i*.

$n_{1i,2i,3i}$ = direction cosines of the slip plane normal for slip system *i*.

S_i = "slip density," or effective number of dislocations passed per atomic (slip) plane; proportional to the macroscopic glide-shear produced by system *i*. See Appendix.

The terms inside the parentheses are merely the expanded expressions for the $\cos^2 \varphi_i$ term in Eq. (1). As to the summation, it is carried over all operating slip systems *i*. For face-centered cubic alloys, there are a maximum of twelve {111} <110> slip systems. Which of these systems will operate depend on the crystal orientation and on the geometry of the deformation. Generally these are the systems which have the largest value of the factor for resolving the applied stresses on the slip plane and in the slip direction (Schmid factor), and which act in such a way so as to accommodate the imposed external shape change of the material. This shape change is expressed as six components of the macroscopic strain tensor ϵ_{ij} . Each of these components in turn can be expressed as a linear combination of glide-shears γ_i in the active slip systems. The values γ_i are then solved in terms of ϵ_{ij} . This practice follows the procedures of Taylor¹¹ and Bishop and Hill,^{12,13} who made calculations on a polycrystalline material by assuming that the shape change of an individual grain is the same as that of the aggregate.

It may be noted that for an arbitrary macroscopic shape change, five of the six strain components are

¹¹ G. I. Taylor, J. Inst. Met. **62**, 307 (1938).

¹² J. F. W. Bishop and R. Hill, Phil. Mag. **42**, 1298 (1951).

¹³ J. F. W. Bishop, Phil. Mag. **44**, 51 (1953).