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TEXTURAL STUDIES OF THERMOMECHANICALLY TREATED EUTECTOID STEEL

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Abstract: The important class of metallurgical treatments known as thermomechanical treatment (TMT) is especially appropriate for textural studies because the crystallographic disposition of the matrix largely determines the final properties of the material. The current work utilized regular and inverse pole figure measurements of the ferrite matrix in TMT eutectoid steel to monitor the effects of various processing steps such as degree of cold rolling and the temperature and time of rapid annealing. Textural results provided the key to understanding fatigue life differences between longitudinally and transversely oriented sheet samples. The most favored ferrite slip system was not well oriented to act in the transverse samples. This in turn retarded the slip which nucleates fatigue cracking and hence extended the total fatigue life of transverse samples.

1. INTRODUCTION

When a polycrystalline metal is plastically deformed particular grains tend to rotate toward a common orientation depending upon the amount of deformation.[1,2] Although the effect is most accentuated after cold work, it may also be observed after annealing.[3,4] The type and degree of preferred orientation depends on the mode, magnitude and temperature of deformation[5,6] and the subsequent annealing treatment,[7] all of which influence crystallographic texture.

Lankford et al.[7] found that plastic anisotropy governed the mechanical properties of sheet steel. This plastic anisotropy has its origin in the texture of the sheet.[8,9] Steel textures usually are multicomponent in nature[1,10] and the superposition of the poles in a regular pole figure (RPF) can make quantitative analysis difficult.[11] This difficulty can

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be partly alleviated by utilizing the inverse pole figure (IPF) technique[12] to complement the RPF information.

2. EXPERIMENTAL PROCEDURES

2.1 General

Regular and inverse pole figures were determined using 2.54 cm square samples taken from the center of each sheet of TMT eutectoid steel. Samples were mechanically polished and then etched in an aqueous solution[13] of 2% phosphoric acid and 49% peroxide to remove the disturbed metal. Finally the samples were electropolished in a solution of 6% perchloric acid in glacial acetic acid at 40 volts for two to three minutes.

2.2 Inverse Pole Figures

The diffracted intensities from a textured sample were compared with those calculated for an ideal non-textured sample by calculating an index called the texture coefficient (TC)[12,15] which is defined as:

$$TC = \frac{I_{(hk\ell)}/I_{o(hk\ell)}}{1/n \sum_{o}^{n} I_{(hk\ell)}/I_{o(hk\ell)}}$$

where I is the measured integrated intensity of a given (hkℓ) reflection, I_o is the calculated intensity for the same (hkℓ) reflection as would be produced by a sample with an equal number of grains oriented in all directions and n is the total number of reflections measured. The calculated TC values were then plotted within a unit stereographic triangle. The reflections used were (110), (200), (211), (310), (321), (411,330), (332), (521), (510,431) and (530,433). Zr-filtered Mo Kα radiation was used with a DIANO horizontal diffractometer.

2.3 Regular Pole Figures

The technique used was that of Schwartz[17] and the instrument was a Siemens Pole Figure Goniometer. Intensities were rated relative to those from a random carbonyl iron sample, i.e., "times random."

2.4 Materials

Consumable electrode vacuum arc remelted (CEVAR) eutectoid steel of composition listed in Table I was used.

The methods used to produce the random coarse pearlite (RCP) and oriented pearlite were those of Grange[18] and of Cairns and Charles,[19] for a similar eutectoid steel. The as-received material was austenitized at 1090°C for 10 minutes in an argon atmosphere and then transformed at 705°C for 17 hours and furnace cooled.

TABLE I

Chemical Composition of the Steel (wt %)

C	Mn	P	S	Si	Mo	V	Ni	Cr
0.85	0.77	0.002	0.008	0.17	0.2	0.002	0.02	0.05

Ti	Cu	Sn	Al
0.001	0.03	0.002	0.005

Samples containing random coarse pearlite were cold rolled to a 75% reduction in thickness which was sufficient to orient the cementite lamellae in the rolling direction. Lesser reductions were studied[20] but only samples based on the 75% reduction will be reported here since that reduction was critical to the TMT.

The cold work in the ferrite matrix was removed by a rapid anneal or up-quench above the eutectoid temperature. The rapid annealing was performed by up-quenching in a lead bath over a range of temperatures between 730°C and 820°C for periods of time from 5 to 60s followed by air cooling. The complete schedule for the thermomechanical treatments is shown in Table II.

TABLE II

Thermomechanical Treatment Schedule for the Eutectoid Steel*

Nomenclature of the TMT**	Reduction by Cold Rolling %	Rapid Annealing Temperature (RAT), °C	Rapid Annealing Time (RAT) s
75-730-t	75	730	5,10,15,20,30,40,60
75-740-t	75	740	5,10,15,20,30,40,60
75-750-t	75	750	5,10,15,20,30,60
75-760-t	75	760	5,10,15,20,30,60
75-780-t	75	780	5,10,15,20,30,60
75-800-t	75	800	5,10,15,20,30,60
75-820-t	75	820	5,10,15,20,30
70-730-t	70	730	10
70-740-t	70	740	10,15
70-760-t	70	760	10
60-730-t	60	730	10
50-730-t	50	730	10
40-730-t	40	730	10

*All the rectangular-shaped slices were austenitized at 1090°C for 10 min. and then isotransformed at 705°C for 17 hours before the cold rolling.

**To allow for easy identification the TMT specimens will hereafter be referred to with the following nomenclature: TMT (% deformation by cold rolling-rapid annealing temperature-rapid annealing time).

3. RESULTS

3.1 Inverse Pole Figures

Figures 1 and 2 contain the IPF of the random coarse pearlite, the cold rolled and the TMT conditions. All the

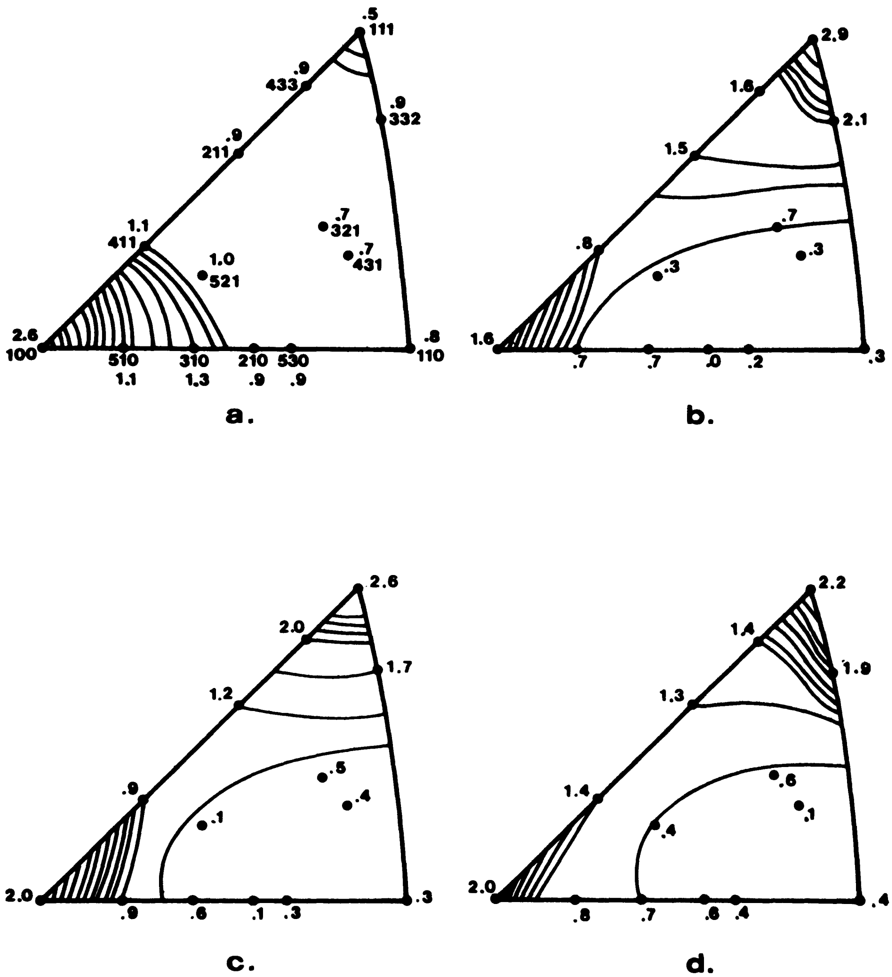

Figure 1. Inverse pole figures taken with respect to sheet normal direction (ND). (a) Random coarse pearlite, (b) 75% cold-rolled, (c) TMT-75-730-10, (d) TMT-75-740-10.

inverse pole figures were taken in a normal direction (ND) to the rolling plane of the sheet.

Figure 1a contains the IPF of the random coarse pearlite. In most of the stereographic triangle the pole density was constant with some decrease around (111) and an increase near (100). These results indicate that the RCP had a cube-on-face texture. The IPF for the 75% cold rolled pearlite is shown in Figure 1b. The TC were high near (111), relatively unchanged near (110) and low near (100). This suggests a

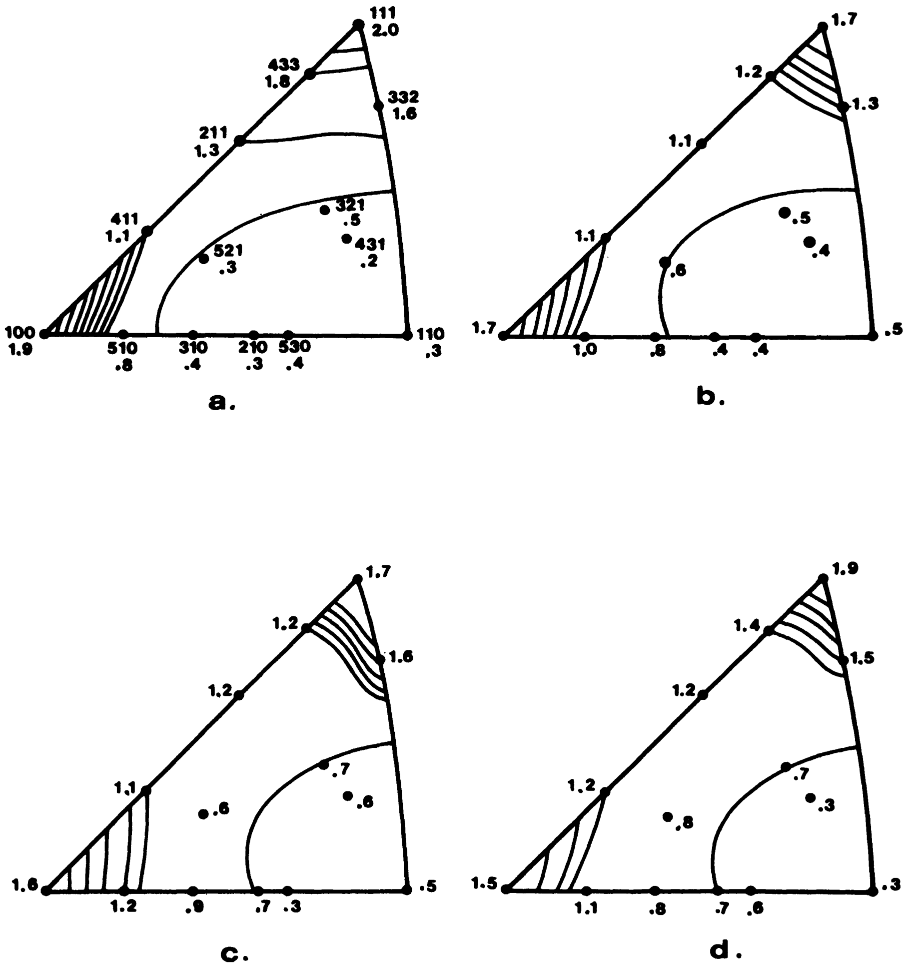

Figure 2 Normal direction inverse pole figures for (a) TMT-75-750-10, (b) TMT-75-760-10, (c) TMT-75-800-10, (d) TMT-75-820-10.

cube-on-corner texture in agreement with other workers.[11,20,21]

Figure 3 shows the influence of the temperature of rapid annealing on the texture coefficients of TMT specimens. The TC near (111), which is a maximum for 75% cold rolling, began to decrease as the RAT increased, reaching the lowest value at 760°C. That value was the same up to 800°C. The texture coefficient increased slightly at 820°C. The TC near (100) were higher at lower RAT and then decreased up to 760°C. The changes in TC near (110) were not so large but reached slightly higher values in the range of 760°C to 800°C.

The set of TC closest to the ideal random value was for the TMT between 760°C and 800°C for 10s. In that temperature range the specimens develop a duplex <100> and <111> fiber texture normal to the rolling plane of the sheet.

The effect of rapid annealing time on the TC of the TMT at 740°C is presented in Figure 4. The (111) pole density decreased, the (100) pole density increased and the TC around

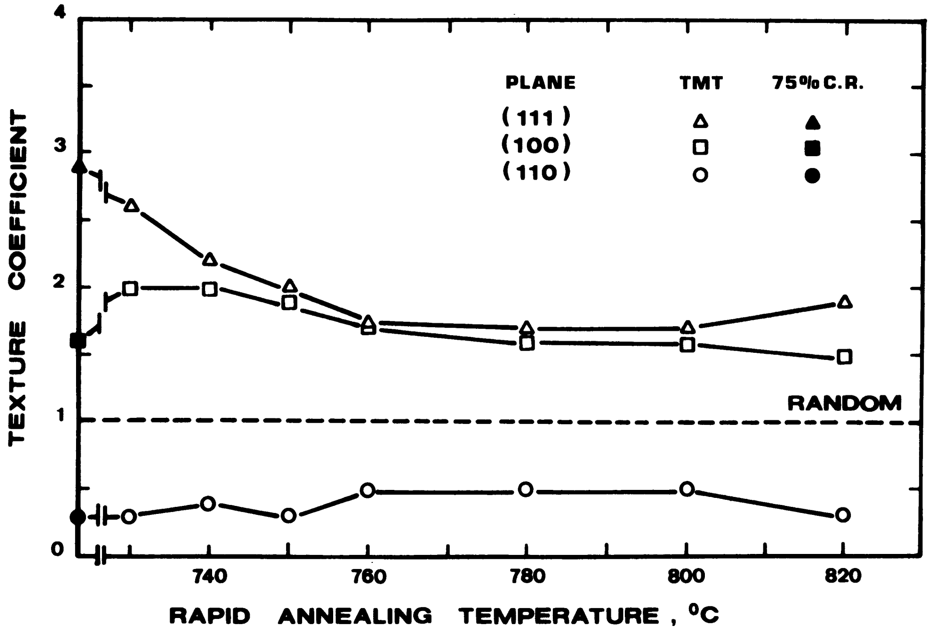

Figure 3. Texture coefficients for the <111>, <100> and <110> corner locations of the unit stereographic triangle for TMT specimens as a function of rapid annealing temperature. TC for the 75% cold-rolled condition appear along the y axis.

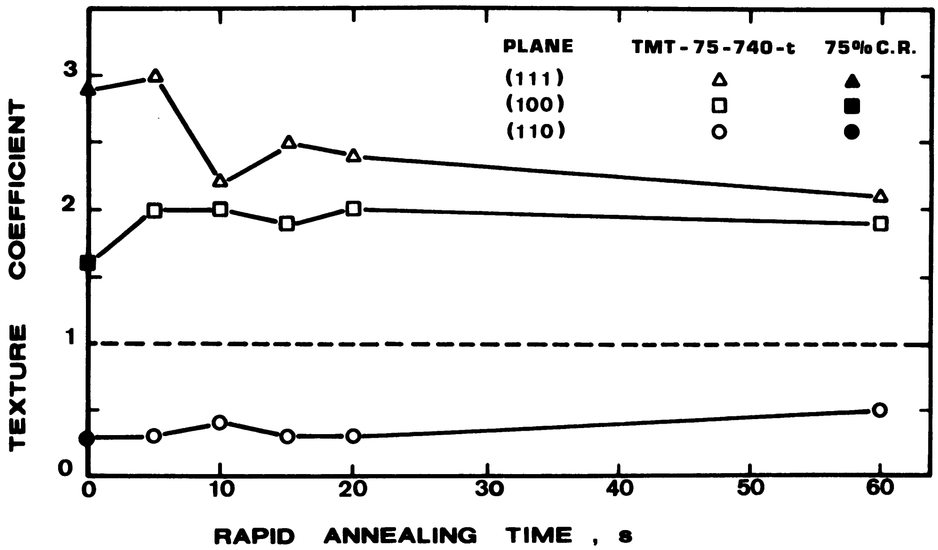

Figure 4. Texture coefficients for the <111>, <110> and <110> corner locations of the unit stereographic triangle for TMT specimens as a function of rapid annealing time.

(110) remained largely unchanged with increasing time at 740°C. The set of TC closest to the ideal random orientation were for 10s and 60s.

Figure 5 represents changes in the IPF due to (a) cold rolling and (b) rapid annealing 10s at 740°C. Figures 1 to 5 indicate that the TMT oriented pearlite sheets inherited some crystallographic texture from the cold rolling process. That texture was closer to the rolling texture at lower RAT.

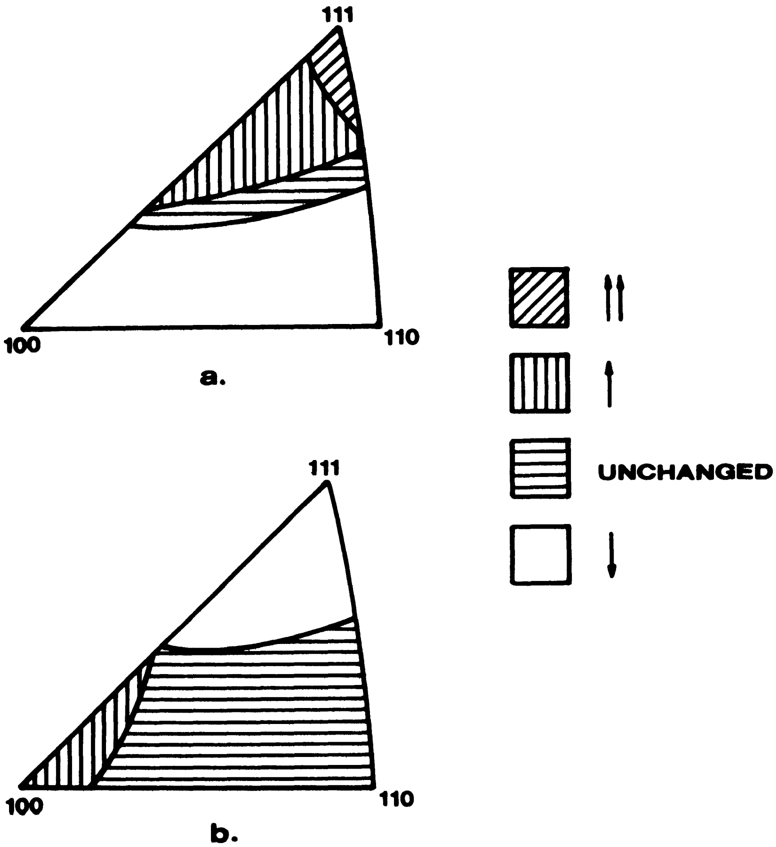

Figure 5. Schematic changes in inverse pole figure texture coefficients due to: (a) cold rolling, (b) rapid annealing at 710°C for 10s.

Figure 4 shows how the rolling texture is inherited. In cold rolling the random material many (111) planes became parallel to the sheet rolling plane at the expense of other planes. The axis density moves largely from (100) and (110) to the (111) corner. By up-quenching the samples only a small fraction of the (111) poles change orientation and the axis density near (100) and (110) increased only slightly.

3.2 Regular Pole Figures

Figure 6 shows the (110) pole figure of the ferrite in the 75% cold rolled condition. The {112}<110>, {111}<112> and {111}<110> orientation are consistent with the highest intensity regions of the pole figure. The {110}<110> component can be considered as minor. The component {554}<225> may exist, but it is so close to the {111}<112> orientation that

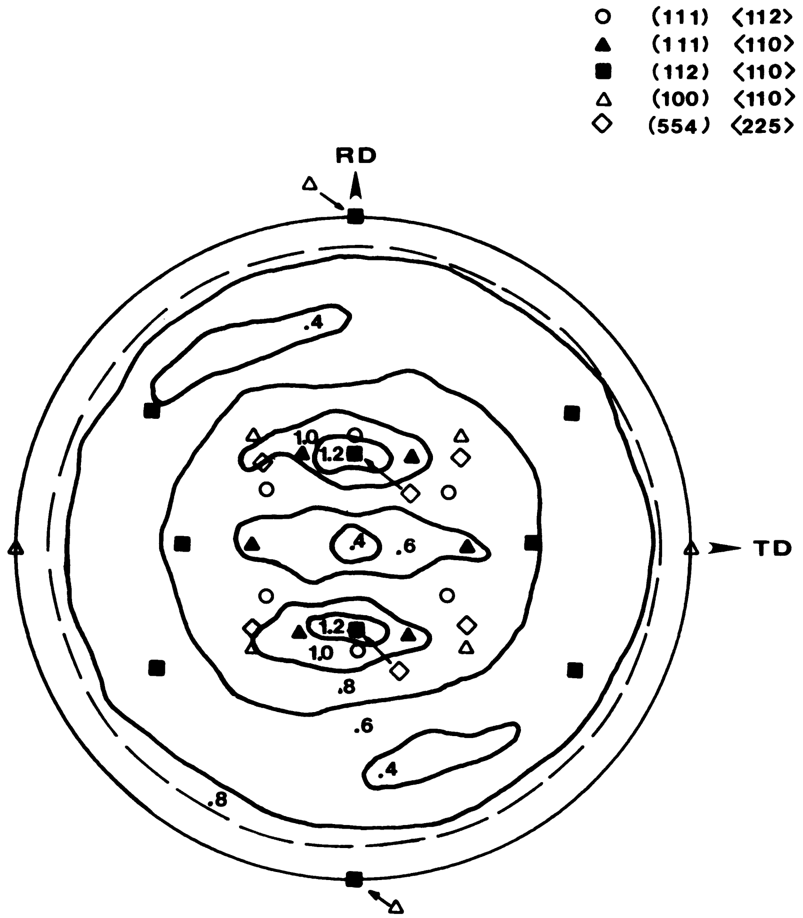

Figure 6. (110) pole figure for 75% cold rolled material.

it is difficult to distinguish very well between the two. This was also true for (200) pole figures, which are not shown in order to conserve space.

The (100) pole figures of the ferrite in the TMT oriented pearlite sheets are shown in Figures 7 through 10. The rolling texture persisted in the TMT specimens. As the temperature of up-quenching increased the (110) pole figures lost sharpness. This became more accentuated when the up-quench time was increased as is shown in Figure 10. The various (110) TMT pole figures clearly show the {111}<112> and {112}<110> components. The preferred orientation {112}<110> from the cold rolling was always in the highest intensity region.

4. DISCUSSION

The high axis density in the ND inverse pole figure for the cold rolled condition (Figure 1) suggests that the preferred orientation obtained by cold rolling the coarse pearlite must be mainly {111}<uvw>, {211}<uvw> and {100}<uvw> components. The low (110) TC values mean that almost all the (110) poles

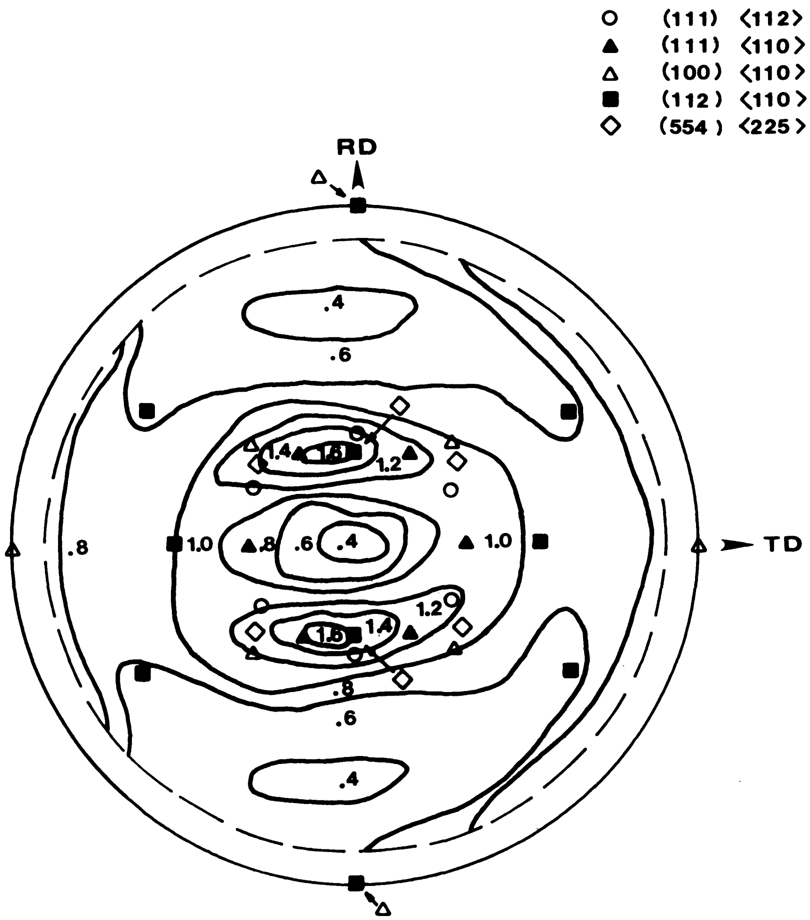

Figure 7. (110) pole figure for TMT-75-730-10.

are lying in the rolling plane. Based just on the IPF results one can postulate that the 75% cold rolled pearlite texture can be described as a <110> fiber axis in the rolling direction with components of {111}<110>, {112}<110> and {100}<110>, together with a <111> fiber axis normal to the surface of the sheet. The results obtained from the (110) regular pole figures indeed showed that the {111}<112>, {112}<110>, {100}<110> and {111}<110> orientations are consistent with the highest-intensity regions of the pole figures. Considering the two types of pole figures together one finds confirmation of a <110> fiber axis in the rolling direction and a <111> fiber axis normal to the sheet surface. In addition one has the following four textural components: {111}<112>, {112}<110>, {100}<110> and {111}<110>. This description is in agreement with results of other workers[22,23,24] for low carbon steels in the same range of deformation. No references on eutectoid steel textures were found in the literature.

The textures of the TMT involving lower rapid annealing temperatures showed more retention of the cold rolled texture. The shorter times of rapid annealing at those lower temperatures

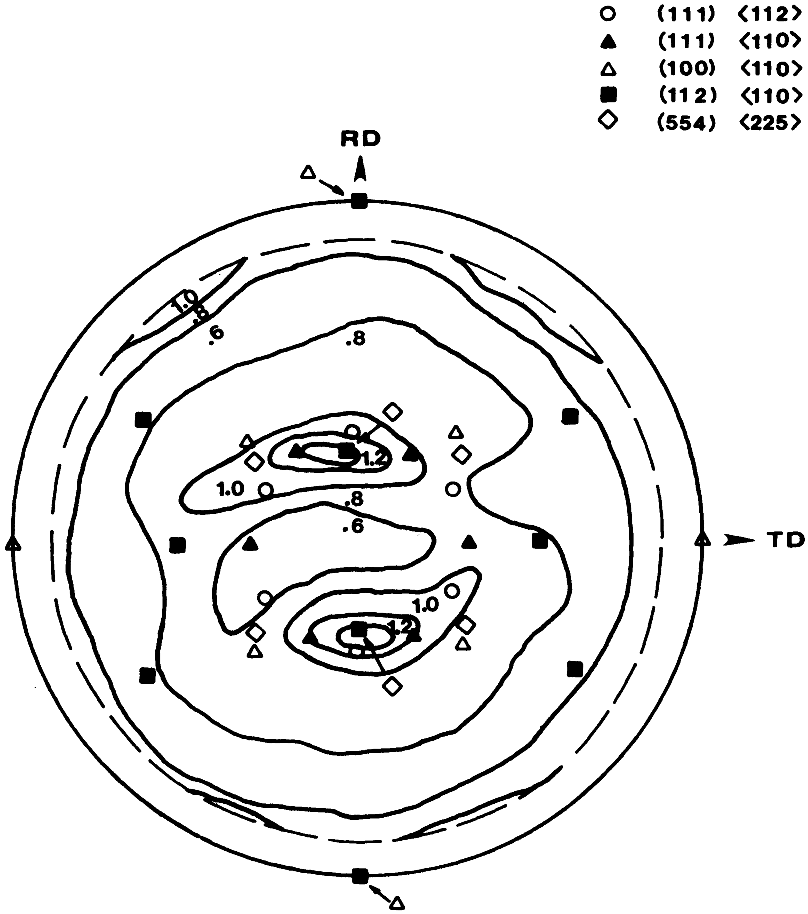

Figure 8. (110) pole figure for TMT-75-760-10.

produced microstructures which still showed deformed grains, hence no general grain growth of the austenite occurred. The IPF in Figure 1c illustrates how the texture is still cube-on-corner in the TMT-75-730-10 sample. The evidence of some sharpening of the texture in Figure 7 may be partially caused by the removal of the line broadening of cold work by the rapid anneal in the TMT. Line broadening has the effect of lowering the relative intensities from cold worked specimens.[25] Subsequent annealing apparently sharpens the texture because line broadening has been reduced.

When the temperature of annealing increased, the texture became less sharp; however, the preferred orientations are mainly the same as the rolling texture. Only at relatively high temperature (760°C) and for the holding time of 60s did diffuse changes in the shapes of the high-intensity contours become evident. This treatment produced complete softening, partial destruction of the mechanical fibering and slight grain growth.

An important feature was the stability of the (112) orientation parallel to the rolling surface as is shown in Figures

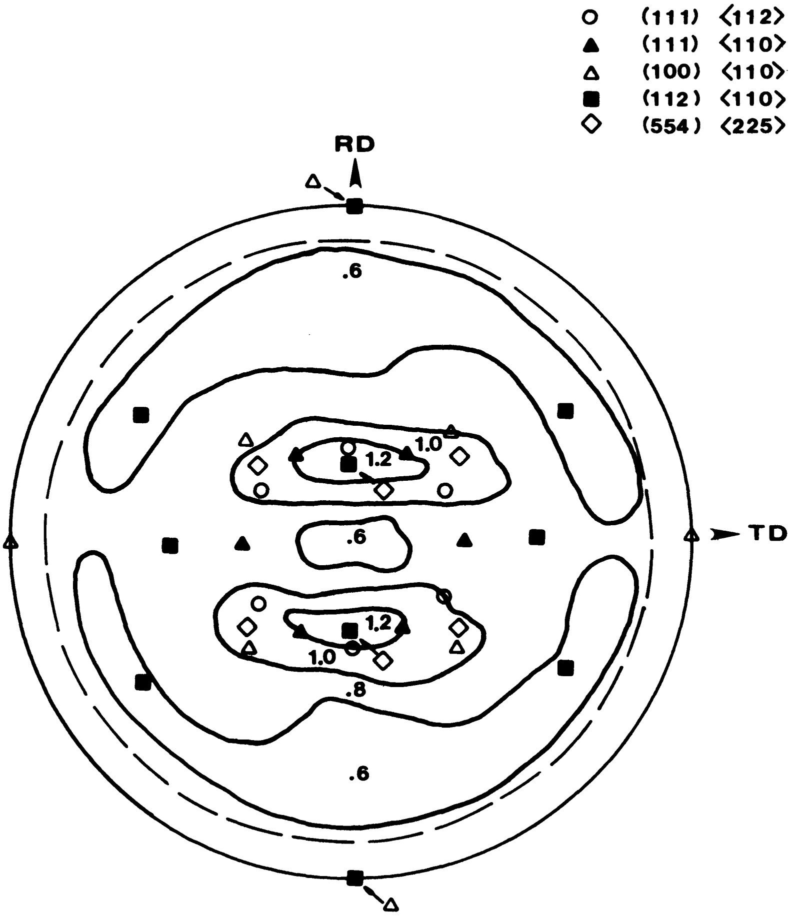

Figure 9. (110) pole figure for TMT-75-800-10.

1 and 2. This is consistent with the presence of {112}<110> components in the highest-intensity contours of the regular pole figures of the reheated specimens.

The texture of the specimens with TMT using lower rapid anneal temperatures can be described as the same as that of cold rolled pearlite. The textures of samples annealed from 760°C to 800°C for 10s can be described as having a duplex <100> and <111> fiber axis perpendicular to rolling surface and a <110> fiber axis in the rolling direction. The remaining components are the same as for the rolled material.

Figure 7 may be used to explain the fact that transverse fatigue samples for this TMT had superior fatigue resistance to longitudinal samples.[26] For bending fatigue of longitudinally cut (i.e. long dimension parallel to the rolling direction) samples, the bending is about an axis transverse to the rolling direction. This causes slip on {110} planes oriented optimally for such shear as indicated in Figure 7 by their pole concentrations along a north-south line at intermediate positions between the sheet normal and the north and south directions respectively. If slip on these systems is

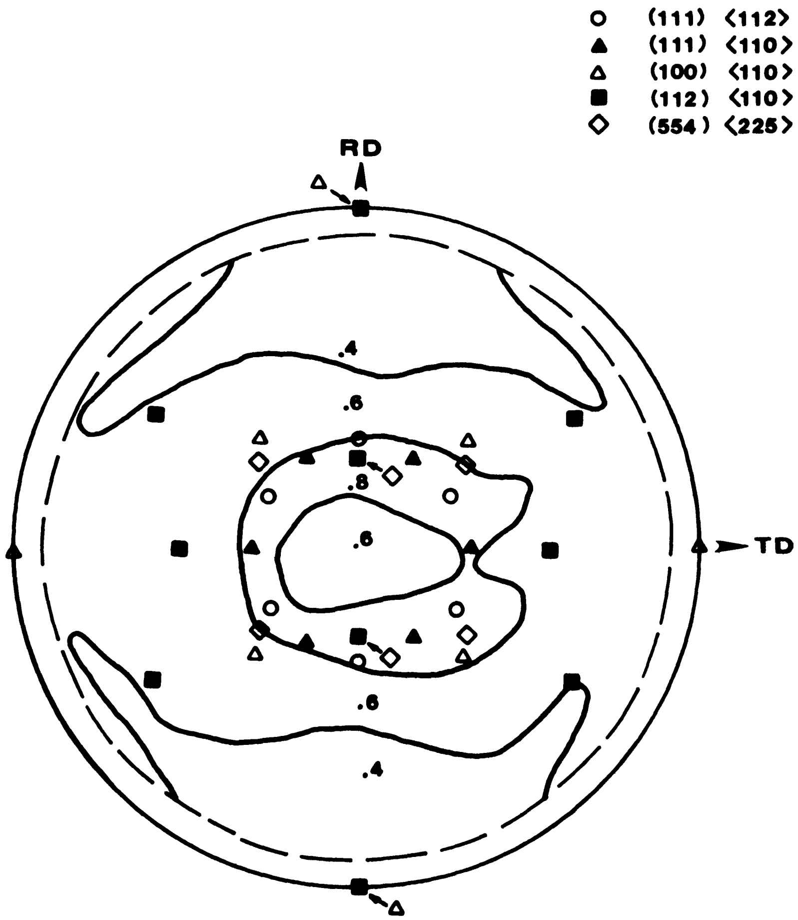

Figure 10. (110) pole figure for TMT-75-760-60.

encouraged slip induced fatigue crack initiation will be earlier than in a sample with a more random (110) pole distribution. Conversely, in transversely cut samples, with a north-south flexure axis, Figure 7 shows that no (110) pole concentrations exist along the east-west axis in positions appropriate to slip under maximum applied shear during bending. If slip on the main slip system is discouraged fatigue crack initiation will be retarded and this may well account for the higher total observed fatigue life of transverse TMT samples.[26]

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REFERENCES

1. J. Bennewitz, *Arch. Eisenh.*, 33, 393 (1962).
2. P. K. Koh and C. G. Dunn, *J. of Metals*, 7, 401 (1955).
3. I. L. Dillamore and W. T. Roberts, *Acta Met.*, 12, 281 (1964).
4. C. A. Stickels, *Trans. AIME*, 233, 1550 (1965).
5. D. S. Hoddinott and G. J. Davis, *J. Iron and Steel Inst.*, 210, 624 (1972).
6. P. R. Evan, H. Bitcon and I. F. Hughes, *J. Iron and Steel Inst.*, 207, 81 (1966).
7. W. T. Lankford, S. C. Snyder and J. A. Bauscher, *Trans. ASM*, 42, 1197 (1950).
8. W. F. Hosford and W. A. Backofen, in *Fundamentals of Deformation Processing*, Proc. 9th Sagamore Conf., (Syracuse Univ. Press, Syracuse, N.Y., 1964), p. 259.
9. R. L. Whiteley and D. E. Wise, in *Flat Rolled Product III*, (Interscience, New York, 1962), p. 47.
10. T. R. Thompson, *J. Austr. Inst. Met.*, 18, 14 (1973).
11. H. Takechi, H. Kato and G. Nagashima, *Trans. AIME*, 242, 56 (1968).
12. E. F. Sturken and W. G. Duke, *Metals, Ceramics and Materials*, U.S. Atomic Energy Comm., TID-4500 Rep. (1961).
13. G. L. Kehl, *Principles of Metallographic Laboratory Practice*, (McGraw-Hill, New York, 1943).
14. *ASM Metals Handbook*, 8, 30 (1973).
15. G. B. Harris, *Phil. Mag.*, 43, 113 (1952).
16. B. D. Cullity, *Elements of X-Ray Diffraction*, (Addison-Wesley, Reading, Mass., 1967), p. 132.
17. M. Schwartz, *J. Appl. Phys.*, 26, 1507 (1955).
18. R. A. Grange, *Second Intl. Conf. on the Strength of Metals and Alloys*, (ASM, 1970), p. 861.
19. R. L. Cairns and J. A. Charles, *J. Iron and Steel Inst.*, 205, 1044 (1967).
20. G. F. Melloy and J. D. Dennison, *The Microstructure and Design of Alloys*, (The Metals Society, London, 1963), p. 101.
21. A. A. Vostrikov, V. P. Gubchevsky and E. D. Nenkina, *Fiz. Met. Metalloved*, 35, 832 (1975).
22. P. Parniere, *Les Memoires Sc. Rev. Met.*, 70, 445 (1973).
23. A. A. Vostrikov, V. P. Gubchevsky and E. D. Nenkina, *Fiz. Met. Metalloved*, 35, 645 (1975).
24. I. L. Dillamore, C. J. Smith and T. W. Watson, *Met. Sci. J.*, 1, 49 (1967).
25. C. A. Stickels, *Trans. AIME*, 233, 1550 (1965).
26. A. Querales and J. G. Byrne, Third Intl. Conf. on Mech. Behavior of Materials, Extension Seminar, Sheffield, England, Aug. 27-29, 1979.