Orientation Dependence of Martensitic and Reverse Transformations in a Textured Fe-30%Ni Alloy

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Introduction

It is well known that martensitic transformation may be affected by cold working in the austenite state before transformation. This means that martensitic transformation is very sensitive to the stresses and the defect structures existing in the parent phase. After cold working the texture in the parent phase is also changed. Since the grains in different orientations possess different stress states and different defect structures, it is not difficult to imagine that many properties of martensite transformation may be orientation dependent. Some of the orientation-dependent properties, such as the variant selection (1,2) and the orientation-selective transformation(3), have been studied by many investigators. The results show that if martensite transformation is carried out in a textured material, the influence of orientation must be taken into account.
In order to determine the orientation-dependent properties, the textures of parent and product phases must be determined precisely. The best material for this purpose is the alloy Fe-30%Ni although it has been studied extensively for many years and the textures in this material have also been measured using the pole figure method. Because of the limitations of the pole figure method it did not allow to determine the orientation-dependent properties of transformation. This can only be done by means of the Orientation Distribution Function method (ODF) which is a quantitative method for texture analysis(4,5, 6,7).
In the present work the textures of the parent Austenite, the Martensite and the retained Austenite in an Fe-30%Ni alloy were measured and calculated. From these results the volume fractions of retained Austenite and the martensite start temperatures as a function of orientation were determined.
In the alloy Fe-30%Ni the reverse transformation from Martensite to Austenite on heating is also carried out through a martensitic mechanism(8) if the material is heated to only a lower temperature. At higher temperature the reverse transformation is a diffusion-controlled process. The transition temperature between these two mechanisms can be determined from texture measurement. The results indicate that the mechanism-transition temperature of reverse transformation depends upon the orientation.
Finally, the variant selection of the reverse transformation at lower-temperature stage was confirmed by comparing the measured and calculated textures.

Experimental

In order to determine the relationships between the orientations and the volume fractions of retained Austenite, a commercial hot-rolled sheet of Fe-30%Ni was annealed and then cold rolled to 60% and 80% reductions in thickness.
After cold rolling the samples were quenched to different temperatures i.e. 233, 223, 213, 193, 173, 143, 113 and 77K to obtain different volume fractions of retained Austenite and Martensite.
The pole figures of the parent Austenite and the retained Austenite were measured from the samples before and after quenching to different temperatures

using an X-ray texture goniometer. From three pole figures the complete ODFs were calculated using the series expansion method(6).
The average volume fractions of the retained Austenite (Tab.1) were obtained by integrating the measured X-ray intensities over the partial pole figures and comparing the integrated intensities of the parent and the retained Austenites. Since the samples after cold rolling contain 8-15% deformation-induced Martensite, the volume fractions of retained Austenite were calculated relative to that of the parent Austenite not to that of the whole sample.

Rolling Reductions	Quenching Temperatures K						
	233	223	213	173	143	113	77
60%	0.95	0.90	0.74	0.50	0.40	0.32	0.28
80%	0.97	0.93	0.81	0.52	0.42	0.34	0.27

Table 1: Average volume fractions of retained Austenite after cooling to different temperatures.

For the reverse transformation the textures of re-transformed Austenite and Martensite, which has a texture inherited from that of cold-rolling Austenite, were measured and calculated from the samples after cold rolling to 80% reduction, quenching to 77K and re-heating to 723, 743, 763 and 813K respectively. The volume fractions of re-transformed Austenite and Martensite were measured with the same method mentioned above. They are listed in Tab.2.

Heating Temperatures K	Volume Fractions	
	Austenite	Martensite
After cold rolling	0.85	0.15
After quenching to 77K	0.22	0.78
723	0.52	0.48
743	0.68	0.32
763	0.84	0.16
813	0.95	0.05

Table 2: Volume fractions of Austenite and Martensite after 80% cold rolling, quenching to 77K and re-heating to different temperatures.

Results and discussion

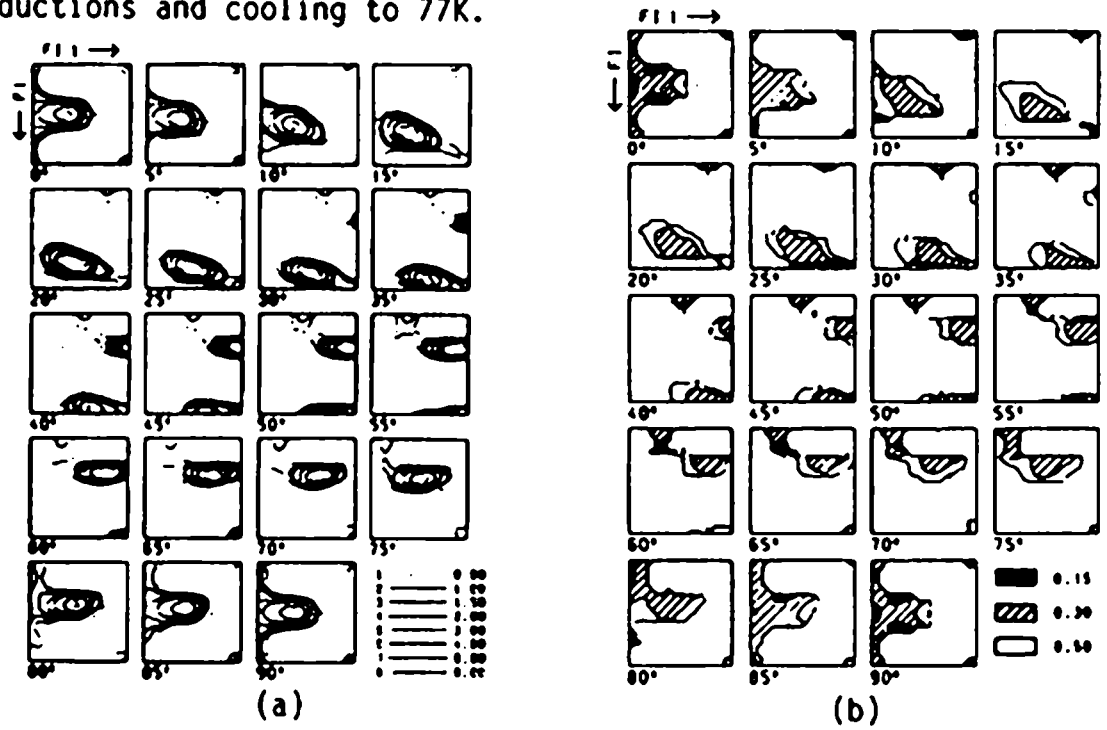
Retained Austenite function

The volume fractions of retained Austenite in different orientations can be calculated from the texture of parent Austenite $f^A(g)$, the texture of retained Austenite $f^{RA}(g)$ and the average volume fractions of the parent Austenite V^A and the retained Austenite V^{RA}. Let $V^A = 1$ then:

$$R(g) = V^{RA} f^{RA}(g) / f^A(g) \qquad \text{Eq. 1}$$

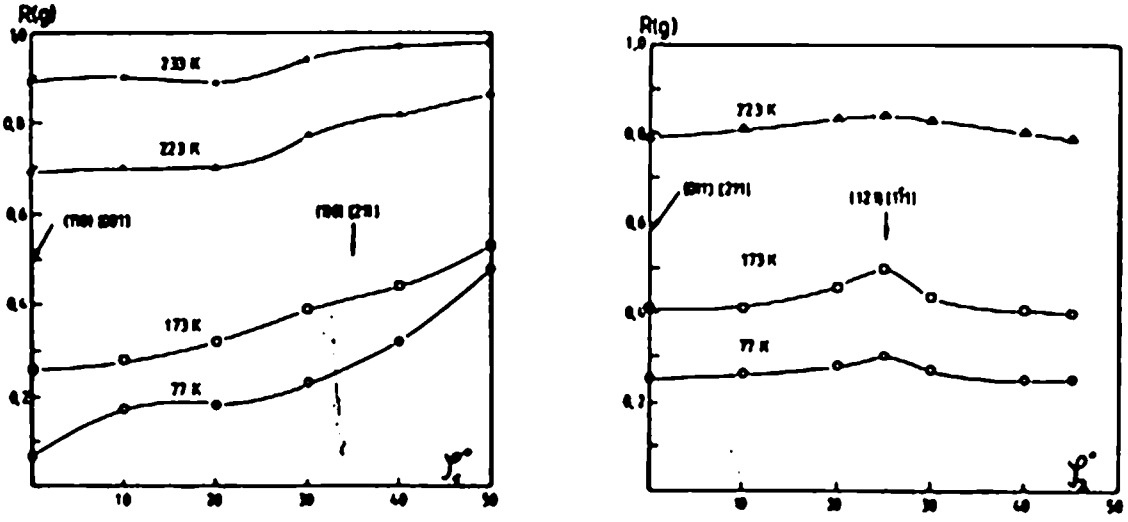
where R(g) is the relative quantity of retained Austenite as a function of the orientation g which can be represented in Euler space as φ_1, Φ, φ_2.

If we calculate the R(g) according to Eq.1 over all orientations, a retained Austenite function in Euler space is obtained as is shown in Fig.1. The function in Fig.1 was obtained from the sample after cold rolling to 60% reduction and cooling to 77K. It was calculated over the range of $f^{A}(g) \geqslant 1$.
Fig.2 and 3 show the values of the retained Austenite function along the section $\Phi = 45°$ $\varphi_2 = 0°$ and along the skeleton line for the samples after 60% and 80% reductions and cooling to 77K.



(a) (b)

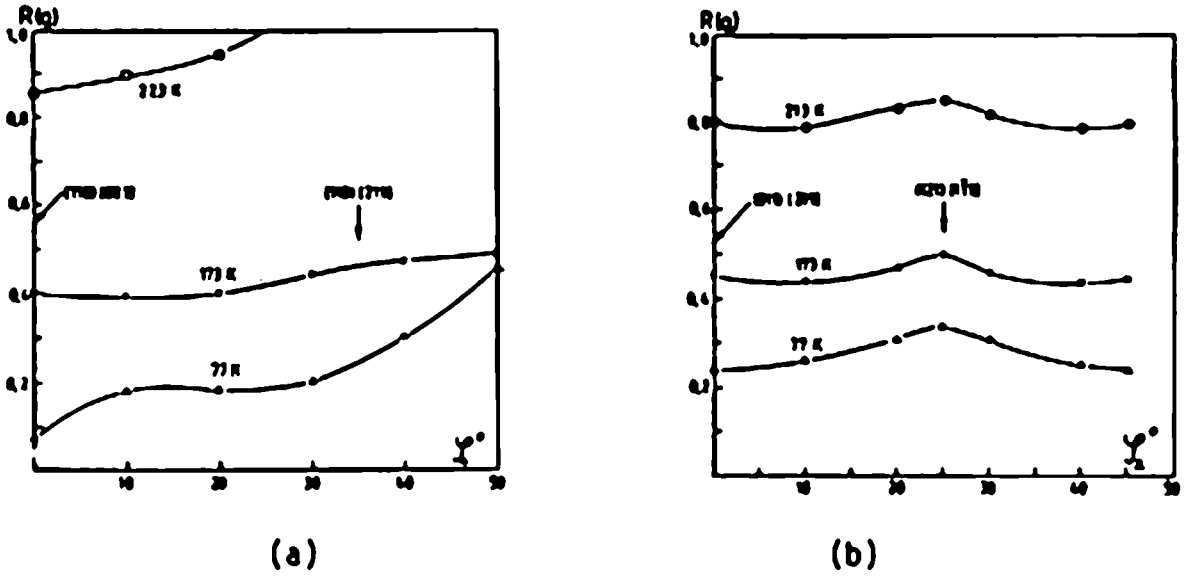
Fig.1 The texture of Fe-30Ni after 60% cold-rolling reduction (a) and the retained Austenite function R(g) (b)



(a) (b)

Fig.2 The values of R(g) along the section $\Phi = 45°$ $\varphi_2 = 0°$ (a) and along the skeleton line (b) for 60% reduction and cooling to different temperatures.

If we compare the values of R(g) it can be seen that for both reductions the R(g) has low values in the orientations near {110}<100> ($\varphi_1 = 0°$, $\Phi = 45°$, $\varphi_2 = 0°$ in Euler space). This means that after cooling to 77K most of the Austenite in



(a) (b)

Fig. 3 The values of R(g) along the section $\Phi = 45°, \varphi_2 = 0°$ (a) and along the skeleton line (b) for 80% reduction and cooling to different temperatures.

these orientations has transformed into Martensite. In other orientations, such as {211}<111>, the function takes a higher value, more retained Austenite thus exists in these orientation.
This result can also be seen from the metallographic analysis. Fig.4 shows the microstructure of Fe30%Ni after 60% cold rolling and cooling to 77K. The distribution of retained Austenite is not regular. In some places in this photo nearly no retained Austenite exists and the plates of Martensite are thicker and longer than in the other places.
Since the habit planes of Martensite in this material are {3,10,15} and the normals of habit plane must be perpendicular to the traces of the martensite plate, the orientations of the parent Austenite can be calculated uniquely if the directions of three non-parallel Martensite traces in a parent-Austenite grain have been measured. With two non-parallel traces of Martensite plate the orientation of the parent Austenite can only be estimated. Fig.5 shows the normals of two habit planes represented in a (011) projection which were measured from the martensite plates located in the circle in Fig.4. Let the 100 direction in the projection be parallel to the rolling direction shown in the photo, then the normals of these two habit planes run just through {3,10,15} pole. That is, the orientation of the parent Austenite might be {110}<100>. In this orientation R(g) just takes a low value as is comfirmed by the texture measurement mentioned above.
The different values of R(g) should be related to the cold-rolling process. The low R(g) values of the orientation {110}<100> may be attributed to the geometrical shape of the twin lamellae (9). The twin structure can be assumed to be favourable for martensite transformation.

Martensite starting temperatures

It can be seen that before martensitic transformation $V^{RA} = V^{A} = 1$ and $f^{RA}(g) = f^{A}(g)$. Hence according to Eq.1 the function R(g) must be equal to unity. Therefore, the martensite starting temperature M_S can be defined by the lowest value of T for which R(g,T) = 1, where T is the temperature, and M_f is defined by R(g,T) = 0. When the functions at different temperatures are known, then M_S (g) and M_f (g) can be determined by extrapolating the R(g,T) curves to

Fig.4 Microstructure of Fe-30Ni after 60% rolling reduction and cooling to 77K.

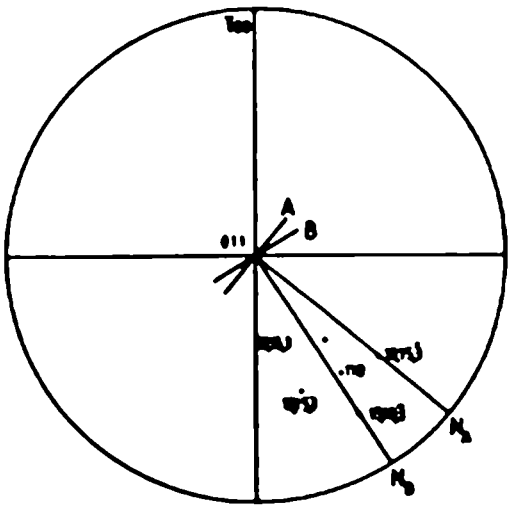


Fig.5 The normals of two habit planes. A,B: traces of Martensite. N_A , N_B : The normals of A and B.

R(g,T) = 1 and to R(g,T) = 0 respectively.
From the average volume fractions of retained Austenite (Tab.1) and the textures measured from the samples after cooling to different temperatures, the retained Austenite function R(g,T) can be calculated. The R(g,T) values for the three main texture components are shown in Fig.6. The Ms temperatures for these three orientations are given by the points R(g,T) = 1 which depend upon the orientation. They are in the temperature range of 233 - 223K.
For both rolling reductions the parent Austenite in the orientation {110}<100> possesses a higher Ms value and that in the orientation {211}<111> has a lower one.
At temperatures near M_f only a small volume fraction of retained Austenite exists. Because of the limited accuracy of texture measurement and calculation the M_f temperature as a function of orientation can only be determined with limited accuracy.

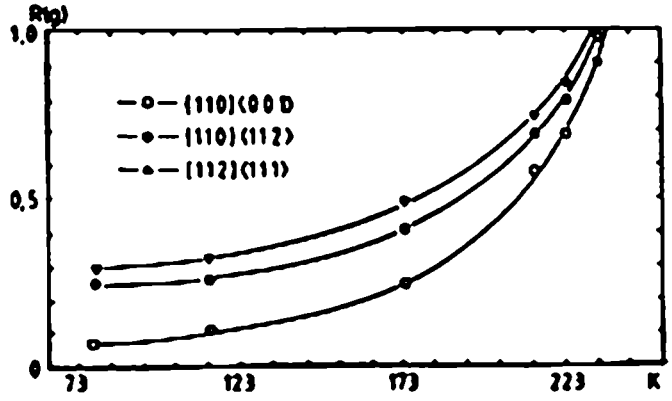
Mechanism - transition temperature during reverse transformation

The structure change during the α-γ reverse transformation in Fe-30%Ni can be separated into three temperature ranges. At lower temperatures the reverse transformation is carried out through a shear-mechanism as that of martensitic transformation. At higher temperatures the process is diffusion-controlled. Between these two temperature ranges there exists a " transformation-stop" temperature range in which the transformation is impeded by defects and/or concentration change (8). Thus, the transformation-stop temperature can be seen as the mechanism-transition temperature. The transformation-stop temperatures for different orientation can be determined from the difference ODF of the re-transformed Austenite.
The difference ODF can be calculated as follows:

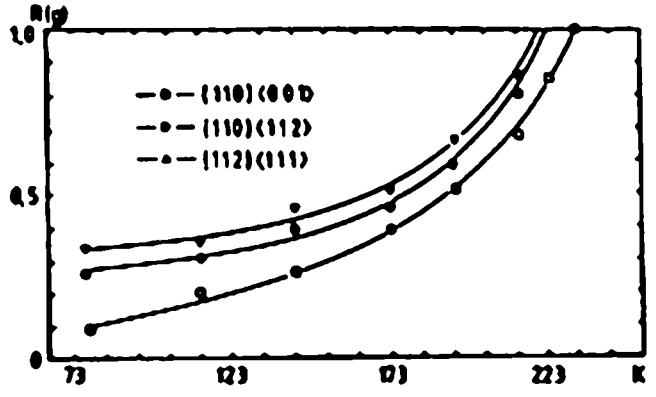
$$f^{D}(g) = [V^{T2}f^{T2}(g) - V^{T1}f^{T1}(g)] / (V^{T2} - V^{T1}) \qquad \text{Eq.2}$$

where V^{T2} and V^{T1} are the volume fractions of Austenite measured after heating to the temperatures T_2 and T_1 respectively, $f^{T2}(g)$ and $f^{T1}(g)$ are correspending ODFs and $f^{D}(g)$ is the difference ODF which represents the true orientation densities of Austenite re-transformed in the temperature range from T_1 to T_2. Fig.7 shows the difference ODF calculated according to Eq.2 in the re-transforming temperature range 723 - 743K. In this temperature range the $f^{D}(g)$ value for the orientation {110}<100> drops nearly to zero. This means that for this orientation the mechanism-transition begins earlier than that for the other orientations.

Comparing the orientation densities for the three main texture components in the difference ODF (Fig.8) we can see that the re-transformed Austenite in the orientation {011}<100> has the lowest mechanism-transition temperature whereas it has the highest transition temperature in the orientation {211}<111>. The transition temperature for the orientation {011}<211> has not been measured. It may have occurred in a very narrow temperature range or the two temperature ranges are superimposed.



(a)



(b)

Fig.6 The R(g) values of three main texture components as a function of T for 60% (a) and 80% (b) rolling reductions.

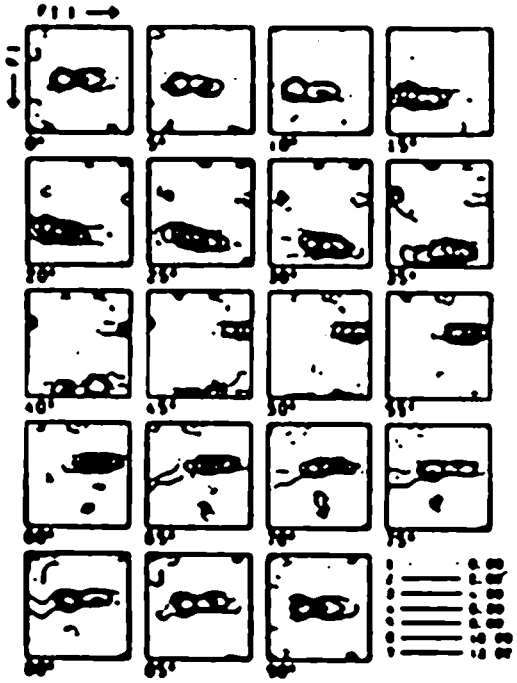


Fig.7 Difference ODF calculated according to Eq.2 in the temperature range 723-743 K

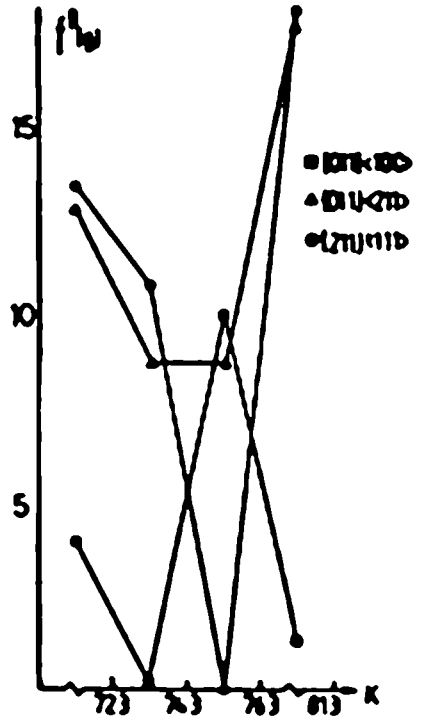
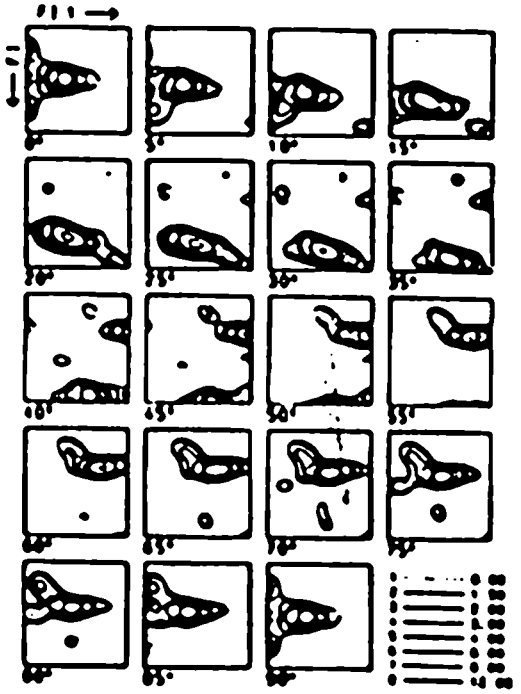
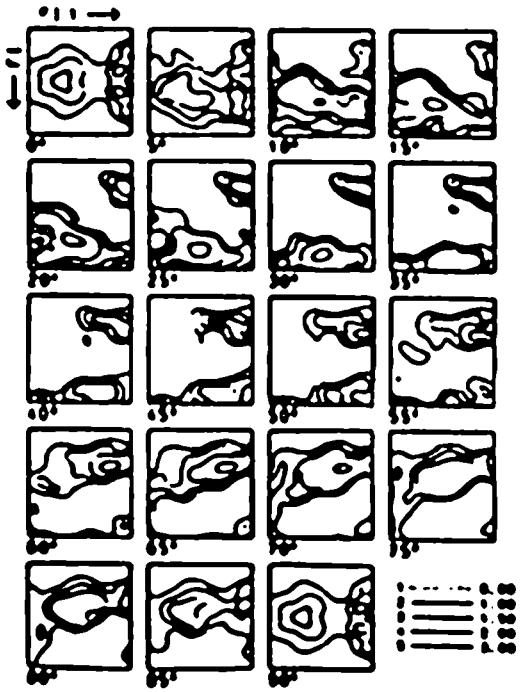


Fig.8 Orientation densities in difference ODF as a function of T
$T = 1/2\ (T_1 + T_2)$.



(a)



(b)

Fig.9 Measured (a) and calculated (b) textures of retransformed Austenite.

The mechanism-transition temperature reflects the facts that during reverse transformation the Martensite can not transform completely into Austenite through shear mechanism. There must be some "retained Martensite" similar to the retained Austenite after martensitic γ-α transformation. Therefore, the transformation must be stopped at certain temperature. For some orientations of the re-transformed Austenite, such as {110}<100>, the transformation from Martensite to Austenite may be easier and hence it can be achieved at lower temperature with lower driving force.

Variant selection during reverse transformation

Since at lower-heating temperatures Martensite transforms into Austenite through a diffusionless process and since the orientation relationship is similar to that of martensitic transformation (10), the variant selection may also occur.
In order to determine the variant selection from texture measurement, we have to compare the austenite texture measured from the sample after heating to 743K which is the average mechanism-transition temperature (Fig.8) with the austenite texture calculated theoretically from the texture of parent Martensite supposing no variant selection. These two textures are shown in Fig.9.
Comparing these two textures we can see that the experimentally measured texture is similar to the texture of retained Austenite, the calculated texture is much weaker and many new orientations appear which do not exist in the measured texture. Therefore, variant selection can be confirmed.
Since the texture of re-transformed Austenite is very similar to that of retained Austenite, the variant selection during reverse transformation may be caused by oriented nucleation of re-transformed Austenite on the retained Austenite.

Acknowledgement

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References

(1) H.J.Bunge, M.Humbert and P.I.Welch, Textures and Microstructures 6(1984) 81.
(2) R.M.Bateman and G.J.Davies, in The proceedings of the sixth International Conference on Texture of Materials, ICOTOM 6, Tokyo 1980 690.
(3) W.P.Liu and H.J.Bunge, in Theoretical Methods of Texture Analysis, DGM Informationsgesellschaft, Oberursal 1987.
(4) H.J.Bunge, Z. Metallkunde 56(1965) 872.
(5) R.J.Roe, J. Appl. Phys. 36(1965) 2024.
(6) H.J.Bunge, Texture Analysis in Material Science, Butterworths London 1982.
(7) H.J.Bunge and C.Esling ed. Quantitative Texture Analysis, DGM Informationsgesellschaft, Oberursal 1986.
(8) H.Kessler and W.Pitsch, Arch. Eisenhüttenw. 38(1967) 321, 469.
(9) G.Wassermann, Z. Metallkunde 54(1963) 61.
(10) S.Shapiro and G.Krauss, Trans. AIME 239(1967) 1408.