Texture Evolution during Cold Rolling and Recrystallisation of IF Steel with a Strong $\{111\}$ Hot Band Texture

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For a new industrially processed, hot rolled Ti+Nb IF steel, finished in the ferrite region and subsequently annealed, the texture development during cold rolling and annealing is investigated. The strong hot band texture is of the $\{554\}\langle 225 \rangle$ peak type. During cold rolling, the absolute maximum shifts from $\{554\}\langle 225 \rangle$ to $\{111\}\langle 112 \rangle$ along the ϵ -fibre and the local maximum along the α -fibre shifts from $\{111\}\langle 110 \rangle$ to $\{223\}\langle 110 \rangle$. During recrystallisation, the weaker α +stronger γ -fibre cold rolling texture is transformed to a very strong γ -fibre annealing texture. Hereby, $\{111\}\langle 110 \rangle$ and, to a much lesser extent, $\{554\}\langle 225 \rangle$ develop at the expense of $\{111\}\langle 112 \rangle$. Orientation Imaging Microscopy (OIM) reveals that γ -fibre grains nucleate and grow in the beginning of recrystallisation by consuming γ -fibre components of the deformed matrix. Already after 23% recrystallisation, the texture of the recrystallised grains is very similar to the one obtained after full recrystallisation. Moreover, $\{111\}\langle 110 \rangle$ grains are found to be larger and more numerous than $\{554\}\langle 225 \rangle$ grains. The former therefore have an advantage over the latter for consuming the partial α -fibre grains during the final stage of recrystallisation and for further development during grain growth. As a result, a very strong $\{111\}\langle 110 \rangle$ peak type annealing texture is obtained in the present material, resulting in extremely good deep drawability.

KEY WORDS: texture; cold rolling; recrystallisation; IF steel; EBSD; OIM.

1. Introduction

The ever developing forming technology used in manufacturing industries makes the application of interstitial free (IF) steels increasingly popular. IF steels not only benefit from a good ductility and resistance to ageing, but are most of all chosen for their excellent deep drawability,¹⁾ which is due to their strong γ -fibre texture ($\langle 111 \rangle // \text{ND}$, i.e. normal direction). In the past, several studies dealt with the texture development during processing of IF steels (e.g. Refs. 1)–13)). Recently, a lot of attention was focused on the improvement of the final product through proper control of the hot rolling parameters.^{3,6–10)} An important result hereby was that hot rolling in the ferrite region, followed by recrystallisation annealing, results in a much stronger γ -fibre than conventional austenite rolling, provided that two conditions are fulfilled.^{4–6)} The first is that little or no solute carbon and/or nitrogen is present during rolling and subsequent annealing.^{5–7)} The second is that lubrication is used to avoid the development of a shear texture near the surface.^{4,5)} Matsuoka *et al.*⁷⁾ further revealed that increasing the strain rate during ferrite rolling considerably enhances the development of the $\{111\}$ recrystallisation texture. This effect is intensified by increasing the Ti-content and by the supplementary addition of Nb. Lowering the rolling temperature between 600 and 850°C also results in higher γ -fibre

intensities.^{3,6,8)}

In the present paper the texture development during cold rolling and annealing is studied for industrially processed hot band of a new Ti+Nb IF steel. The hot rolling parameters and composition were chosen in accordance with the aforementioned results concerning the development of a strong $\{111\}$ hot band texture through ferrite rolling and annealing. The present γ -fibre hot band texture considerably differs from those of ferritically and austenitically rolled IF bands used before for the investigation of the texture development during further processing.^{2,9–12)}

2. Experimental

The starting material used for the present study was an industrially processed, hot rolled, Ti+Nb-added ultra low carbon (ULC) steel, containing 690 ppm Ti, 150 ppm Nb, 18 ppm C, 20 ppm N, 50 ppm S, 90 ppm Si, 90 ppm P, 1200 ppm Mn and 540 ppm Al. The steel was finish rolled in the ferrite region (to 5 mm thickness) and subsequently annealed on the world's first endless hot rolling line in the Chiba Works of Kawasaki Steel Co. Details of the hot rolling process are published elsewhere.¹³⁾ Samples cut from the hot band were 76% cold rolled on a laboratory mill and subsequently soaked for 23 s in a fluidised bed furnace. The annealing temperature was varied between 530 and 810°C. The

Fig. 1. a), b) Orientation maps of the hot band monitored with OIM, highlighting the a) α -fibre and b) γ -fibre grains; c) $\phi_2 = 45^\circ$ -section of the ODF calculated from X-ray pole figures measured in the mid-thickness of the hot band.

samples were cooled from the soaking temperature to room temperature in forced air with a cooling rate of at least 18°C/s above 500°C . Clearly, the annealing cycles used in the present experiments cannot be considered to be iso-thermal, since they include a heating and cooling phase. However, this poses no problems in studying the recrystallisation phenomena because heating and cooling are also part of industrial continuous annealing cycles.

The progress of recrystallisation was monitored using Vickers hardness measurements (500 g) and optical microscopy with point counting to determine the overall recrystallised fraction. Both measurements were performed in the centre of sections defined by the rolling (RD) and normal direction (ND). The etchant used for optical microscopy was essentially the same as the one reported by Saitoh *et al.*,¹⁴⁾ albeit that the HCl-content had to be increased to avoid long etching times. X-ray pole figures were measured at mid-thickness, using a Mo goniometer, to reveal the overall texture change during recrystallisation. The orientation distribution functions (ODF's) were calculated from the measured $\{110\}$, $\{200\}$ and $\{211\}$ pole figures using a discrete method described by Pawlik.¹⁵⁾ Automated Electron Backscattered Diffraction (EBSD) measurements were performed using the Orientation Imaging Microscopy (OIM) apparatus and software of TexSEM Laboratories, coupled with a Field Emission type Scanning Electron Microscope. The OIM samples were carefully prepared using mechanical polishing, electrochemical polishing and finally light etching in the above mentioned solution. All OIM observations were made near the center layer of RD $\times$ ND sections. ODF's were calculated from the measured orientations assuming Gaussian distributions with a half width of 5° around these orientations.

3. Results

3.1. Hot Band and Cold Rolled Condition

In Figs. 1a) and 1b) the orientation maps of the hot

band monitored with OIM are represented. Although the microstructure is fully recrystallised, it contains both equiaxed grains and grains that are somewhat elongated along RD (the average grain diameter is $22\mu\text{m}$, *i.e.* ASTM 8.4). The fraction of near γ -fibre grains (b) is considerably higher than that of near α -fibre grains (a). Taking into account a 15° spread around the ideal orientations, the fractions of the γ - and α -fibre are 0.61 and 0.27, respectively (values are based on a $0.4 \times 0.8\text{ mm}^2$ OIM map, only a part of which is represented in Fig. 1). Of course, the $\{111\}\langle 110 \rangle$ orientations (green grains) are counted in both fibres. In fact, a 15° spread around the latter orientations accounts for a volume fraction of 0.15. This fraction is considerably lower than the 0.33 fraction of grains within a 15° spread around $\{111\}\langle 112 \rangle$ (red grains). The $\phi_2 = 45^\circ$ -section of the ODF calculated from X-ray pole figures measured in the mid-thickness of the hot band, is shown in Fig. 1c). The texture is characterised by a strong (imperfect) γ -fibre and contains almost no intensity on the partial α -fibre (except for $\{111\}\langle 110 \rangle$ which also belongs to the γ -fibre). The maximum intensity of $15 \times$ random intensity (RI) is located in the vicinity of $\{554\}\langle 225 \rangle$, a representative of which lies on the ϵ -fibre (*i.e.* $\langle 110 \rangle // \text{TD}$, *i.e.* transverse direction; the position of the ϵ -fibre is indicated in Fig. 1c)).

During cold rolling (cf. Fig. 2), the maximum intensity along the ϵ -fibre shifts from $\{554\}\langle 225 \rangle$ ($15 \times \text{RI}$) to $\{111\}\langle 112 \rangle$ ($17 \times \text{RI}$) and along the α -fibre from $\{111\}\langle 110 \rangle$ ($5 \times \text{RI}$) to $\{223\}\langle 110 \rangle$ ($11 \times \text{RI}$). On the γ -fibre, $\{111\}\langle 112 \rangle$ is the strongest component, both before ($12 \times \text{RI}$) and after cold rolling ($17 \times \text{RI}$). The $\{111\}\langle 110 \rangle$ intensity is much lower than that of $\{111\}\langle 112 \rangle$, *i.e.* $5 \times \text{RI}$ before and $8 \times \text{RI}$ after rolling. The effect of rolling observed here is in line with experimental results obtained by Inagaki¹⁶⁾ for steel polycrystals. According to these results, the crystal rotations during cold rolling occur along two paths, *i.e.* (A) $\{001\}\langle 100 \rangle \rightarrow \{001\}\langle 110 \rangle \rightarrow \{112\}\langle 110 \rangle \rightarrow \{223\}\langle 110 \rangle$

Fig. 2. Intensity distribution along the α and ϵ -fibre before and after 76% cold rolling reduction.

Fig. 3. Evolution of the Vickers hardness and recrystallised fraction as a function of the annealing temperature.

and (B) $\{110\}\langle 001\rangle \rightarrow \{554\}\langle 225\rangle \rightarrow \{111\}\langle 112\rangle \rightarrow \{111\}\langle 110\rangle \rightarrow \{223\}\langle 110\rangle$. Clearly, for the hot band texture of the present work rotations along path (B) satisfactorily explain the observed texture evolution.

3.2. Recrystallisation

The complementary change of the Vickers hardness and the recrystallised fraction is shown in Fig. 3 as a function of the annealing temperature. The recrystallised fraction only slowly increases between 650 and 700°C, but rises sharply for higher temperatures. Before the start of recrystallisation, recovery phenomena decrease the hardness by about 25 HV. The texture and microstructure development during recrystallisation can be described using Figs. 4 and 5. In Fig. 4, the $\phi_2=45^\circ$ -sections of the ODF's calculated from X-ray pole figures and the corresponding microstructures are presented for 4 samples with various recrystallised fractions. For a more detailed comparison, the intensity distributions along the α , γ and ϵ -fibre are depicted in Fig. 5.

Optical microscopy first of all reveals that the early nucleation primarily occurs at the grain boundaries of heavily etched grains, *i.e.* mainly (near) γ -fibre grains (cf. Ref. 2)). The recrystallised grains become larger and more

Fig. 4. Evolution of the texture and microstructure during recrystallisation; the annealing temperatures and overall recrystallised fractions of the samples selected for this figure are indicated in the ODF plots.

numerous as the soaking temperature increases, but refrain from consuming the lightly etched grains (*i.e.* partial α -fibre grains, cf. Ref. 2)) until the second half of recrystallisation. With respect to the texture development, it is clear that the recrystallisation process transforms the (weaker) α + (stronger) γ -fibre rolling texture in a very strong γ -fibre annealing texture. Hereby, the maximum shifts from $\{111\}\langle 112\rangle$ ($17 \times \text{RI}$ after rolling, Fig. 2) to $\{111\}\langle 110\rangle$ ($23 \times \text{RI}$ for 100% recrystallisation). The intensity of $\{111\}\langle 112\rangle$ is almost the same after cold rolling ($17 \times \text{RI}$) and after recrystallisation ($16 \times \text{RI}$), but clearly lower at an intermediate stage. On the other hand, the intensity of $\{554\}\langle 225\rangle$, which neighbours $\{111\}\langle 112\rangle$, is higher after recrystallisation ($16 \times \text{RI}$) than after rolling ($12 \times \text{RI}$). It can therefore be concluded that the increase of $\{111\}\langle 112\rangle$ during the later stage of recrystallisation is related to the increase of the neighbouring $\{554\}\langle 225\rangle$ intensity (cf. infra).

The evolution of the intensity of the most important texture components during recrystallisation is sum-

Fig. 5. Evolution of the intensity distribution along the α , γ and ϵ -fibre during recrystallisation.

marised in Fig. 6 for all investigated samples. To give an idea of the “individual changes” of $\{111\}\langle 112\rangle$ and $\{554\}\langle 225\rangle$ during annealing, the ϵ -fibre peak was deconvoluted assuming Gaussian distributions for both the $\{111\}\langle 112\rangle$ and $\{554\}\langle 225\rangle$ subpeak. After deconvolution, the intensity of both subpeaks was integrated between $\Phi=40^\circ$ and 75° . The integrated intensities, divided by the integration range, are shown in Fig. 6c) and indicated as “ $\{111\}\langle 112\rangle$ ” and “ $\{554\}\langle 225\rangle$ ”. It is seen in Fig. 6a) that the intensity of $\{111\}\langle 110\rangle$ remains almost constant until about 7% of the steel is recrystallised. The sharp increase of the recrystallised fraction from about 700°C (Fig. 3), is clearly accompanied by a strong increase of the $\{111\}\langle 110\rangle$ intensity, which continues until full recrystallisation. The intensity of “ $\{111\}\langle 112\rangle$ ” steadily decreases between 2 and 61% recrystallisation and then remains almost constant. The “ $\{554\}\langle 225\rangle$ ” intensity continuously increases from 2% to full recrystallisation. The intensities of partial α -fibre orientations between $\{001\}\langle 110\rangle$ and $\{113\}\langle 110\rangle$ only start to decrease after at least 61% recrystallisation. For orientations between $\{112\}\langle 110\rangle$ and $\{223\}\langle 110\rangle$, a slight decrease is already observed between 23 and 61% recrystallised fraction.

Since X-ray measurements cannot distinguish between

Fig. 6. Intensity evolution of the most important texture components during recrystallisation; the intensities shown in (c) were obtained by deconvolution of the ϵ -fibre peak into a “ $\{111\}\langle 112\rangle$ ” and a “ $\{554\}\langle 225\rangle$ ” subpeak, which were subsequently integrated and normalised as is explained in the text.

the contribution of recrystallised and deformed grains, the texture analysis was further refined using OIM. Hereby, a criterion was needed to automatically separate the recrystallised grains from the deformed matrix. Although recrystallised grains generate EBSD patterns of much higher quality than those of deformed grains, the image quality cannot be used as a separation criterion since some non-recrystallised parts of the deformed grains are also found to result in high quality patterns. This is especially true for deformed grains with a low stored energy (*i.e.* partial α -fibre grains) and heavily recovered, but still not recrystallised grains. To identify the recrystallised grains, the following method was adopted in the present work. Since the orientation spread within a recrystallised grain was found to be less than 1° , the misorientation angle defining a grain was chosen to be 1° . With this choice, the deformed grains were split up into their “subgrains”. The recrystallised grains were then separated from these subgrains using a size criterion that defined the minimum number of OIM grid points a grain has to contain to be “recrystallised”. Clearly, this criterion is conservative in the sense that the smallest recrystallised grains are considered to be subgrains, since their sizes are located in the tale of the subgrain size distribution.

As a first example, the overall OIM texture and the one of the recrystallised grains are shown in Fig. 7 for the sample annealed at 725°C (*i.e.* 23% recrystallised). In this case, the size criterion defining a recrystallised grain was that it had to contain at least 7 points of the

Fig. 7. $\phi_2 = 45^\circ$ -sections of the ODF's calculated from OIM data of the sample annealed at 725°C (23 % recrystallised); a) overall texture and b) texture of the recrystallised grains (d =grain diameter).

hexagonal OIM grid, which for a step size of $3\mu\text{m}$ corresponds to a minimum average diameter of $7\mu\text{m}$. In this way, about 300 grains were selected for the ODF calculation. Comparison with Fig. 4d) shows that after 23 % recrystallisation, the texture of the recrystallised grains resembles the final texture, consisting of a strong γ -fibre with an intensity maximum near $\{111\}\langle 110\rangle$ and some deviation from the perfect γ -fibre towards $\{554\}\langle 225\rangle$. A minor feature of the early recrystallisation texture is the presence of some orientation density near $\{001\}\langle 100\rangle$, which was also observed before.²⁾ Comparison of the overall OIM ODF (Fig. 7a)) with the X-ray ODF (Fig. 4b)) further learns that although the general appearance of the ODF's is the same, there may be some differences. For instance, whereas the position of the maximum intensity is the same for both cases, *i.e.* $\{111\}\langle 112\rangle$, the intensity itself is significantly higher for the X-ray than for the OIM measurement. This is mainly because deformed grains generate patterns of lower quality due to the high dislocation density. This introduces uncertainty in the automatic interpretation of the EBSD patterns and therefore somewhat randomises the overall OIM texture. It is also seen that the $\{001\}\langle 110\rangle$ intensity in the OIM ODF is high compared to the maximum intensity. Apart from the statistical error, this is also due to the fact that $\{001\}$ grains have a low stored energy and therefore result in qualitatively better EBSD patterns. This increases their contribution in the ODF with respect to those of deformed grains containing a higher stored energy, *e.g.* the γ -fibre grains.

OIM analysis of the 61 % recrystallised specimen (Fig. 8) further reveals that a systematic texture change is correlated with the size distribution of the recrystallised grains. The textures of 3 size classes of recrystallised grains are presented in Figs. 8b)–8d). In total, the orientations of about 700 recrystallised grains were determined. The 3 size classes were chosen so that the total area fraction of the grains in a given class was constant. For all 3 classes, the texture only consists of γ -fibre orientations. The strongest texture is obtained for the coarsest grains (d). In this case, the intensity is maximum for $\{111\}\langle 110\rangle$ ($\approx 16 \times \text{RI}$) and this intensity is much higher than that of $\{111\}\langle 112\rangle$. A weaker texture (maximum $\approx 9 \times \text{RI}$) is obtained for the 13 to $17\mu\text{m}$ large grains (c). Also, the intensity difference between

Fig. 8. $\phi_2 = 45^\circ$ -sections of the ODF's calculated from OIM data of the sample annealed at 740°C (61 % recrystallised); a) overall texture and b)–d) texture of the recrystallised grains.

Fig. 9. Grain size distributions of grains with orientations included in a 15° spread around $\{111\}\langle 110\rangle$ and $\{554\}\langle 225\rangle$.

$\{111\}\langle 110\rangle$ and $\{111\}\langle 112\rangle$ has decreased. For the finest grains (b), the texture is still somewhat weaker and the intensity distribution along the γ -fibre has become almost homogeneous. However, it should be noted that the intensity along the ε -fibre does not peak at $\{111\}\langle 112\rangle$, but at $\{554\}\langle 225\rangle$. The size distribution of recrystallised grains located within 15° around the ideal $\{111\}\langle 110\rangle$ and $\{554\}\langle 225\rangle$ orientations, is represented in Fig. 9. The total number of near $\{111\}\langle 110\rangle$ grains is higher than that of near $\{554\}\langle 225\rangle$ grains. Moreover, grains with a diameter smaller than $14\mu\text{m}$, have the same probability to be located near $\{111\}\langle 110\rangle$ or near $\{554\}\langle 225\rangle$. Coarser grains, however, are more likely to have a near $\{111\}\langle 110\rangle$ than a near $\{554\}\langle 225\rangle$ orientation.

4. Discussion

It is clear from Sec. 3.2 that the texture development during recrystallisation is mainly characterised by the increase of the intensity of $\{111\}\langle 110 \rangle$ and, to a lesser extent, $\{554\}\langle 225 \rangle$. This increase is accompanied by a decrease of the $\{111\}\langle 112 \rangle$ intensity in the first half of recrystallisation and by the consumption of the partial α -fibre components of the deformed matrix in the second half. With respect to the consumption of the partial α -fibre, the OIM results (Figs. 8 and 9) suggest that the texture change is governed by size-driven growth of the available recrystallised grains. Since these grains have mainly near γ -fibre orientations, the γ -fibre of the overall texture intensifies during growth (Fig. 5). Hereby, the coarser grains (*i.e.* those with orientations that—on average—are closer to the ideal γ -fibre, cf. Fig. 8) have a growth advantage over finer grains (*i.e.* those with orientations that—on average—are further away from the ideal γ -fibre). Moreover, it is clear from Fig. 6 that the $\{111\}\langle 110 \rangle$ component increases much more in intensity during the last 40 % of recrystallisation than $\{554\}\langle 225 \rangle$. This can also be explained by the growth advantage of coarser grains over finer grains. In fact, after 61 % recrystallisation, the coarsest grains are more likely to have a near $\{111\}\langle 110 \rangle$ than near $\{554\}\langle 225 \rangle$ orientation (Fig. 9), enabling the former orientation to develop more than the latter. The preceding discussion suggests that the consumption of the α -fibre orientations is not governed by the presence of special grain boundaries—as in oriented growth (cf. *infra*)—but by the preferential growth of coarser grains at the expense of finer grains (as in normal grain growth). Moreover, the $\{111\}\langle 110 \rangle$ intensity can be expected to further develop strongly during grain growth after recrystallisation is completed. As shown in another paper,¹⁷⁾ this is indeed the case, *e.g.* after annealing at 890°C the $\{111\}\langle 110 \rangle$ intensity exceeds $50 \times \text{RI}$, resulting in an extremely good deep drawability with an r_m -value of about 3.

The mechanism that dominates the texture development during the first half of recrystallisation, is harder to isolate. Of course, the observation that the strong γ -fibre of the cold rolling texture results in a very strong (near) γ -fibre after full recrystallisation and the fact that the ODF of the recrystallised grains after 23 % recrystallisation is similar to the one obtained for full recrystallisation, support the oriented nucleation theory.^{11,18,19)} The latter theory states that the high stored deformation energy of γ -fibre grains is responsible for the observation that γ -fibre orientations nucleate and grow in deformed γ -fibre grains from the start of recrystallisation, whereas α -fibre grains remain unrecrystallised until the later stage of recrystallisation.^{1,2,11)} The question, however, why $\{111\}\langle 110 \rangle$ and $\{554\}\langle 225 \rangle$ develop at the expense of $\{111\}\langle 112 \rangle$ remains to be considered.

First of all, it can be noted that the shift of the maximum intensity along the γ -fibre from $\{111\}\langle 112 \rangle$ after cold rolling to $\{111\}\langle 110 \rangle$ after recrystallisation has often been observed for low carbon steels (*e.g.* Ref. 16,20)). In IF steels, however, an opposite rotation

generally occurs during recrystallisation.^{2,9,10,12)} Only two stage cold rolling with intermediate annealing was reported to show the $\{111\}\langle 112 \rangle$ to $\{111\}\langle 110 \rangle$ shift during the second annealing.²¹⁾ The rotation from $\{111\}\langle 112 \rangle$ after rolling to $\{111\}\langle 110 \rangle$ after recrystallisation was explained by Inagaki¹⁹⁾ by the interaction between $\{111\}\langle uvw \rangle$ and $\{111\}\langle 110 \rangle$ grains. He assumed that the latter are far more stable during cold rolling than the former, so that the compatibility of stress and strain forces the grain boundary region in a $\{111\}\langle uvw \rangle$ grain—neighbouring a $\{111\}\langle 110 \rangle$ grain—to rotate into the $\{111\}\langle 110 \rangle$ orientation. Since nucleation during recrystallisation is expected to occur mainly in the most severely strained regions, *i.e.* the grain boundary region of deformed $\{111\}\langle 112 \rangle$ grains neighbouring $\{111\}\langle 110 \rangle$ grains, the recrystallisation texture is characterised by an increase of the $\{111\}\langle 110 \rangle$ intensity at the expense of that of $\{111\}\langle 112 \rangle$. The experimental evidence given by Inagaki¹⁹⁾ for systematic lattice rotations towards the grain boundary, could not be confirmed in bicrystal experiments reported by Hutchinson.²²⁾ Nevertheless, also in the latter experiment a $\{111\}\langle 112 \rangle$ oriented crystal neighbouring a $\{111\}\langle 110 \rangle$ crystal, resulted in a $\{111\}\langle 110 \rangle$ texture after cold rolling and annealing. Hutchinson argued that cold rolling introduces a wide orientation spread in the deformed crystals by forming subgrains. Recrystallisation nuclei are thought to have their orientation inside this spread. However, only nuclei that are sufficiently misoriented from the bulk of the matrix have a substantial possibility to grow. Since the minimum misorientation angle for significant mobility is assumed to be 15–20°, the 30° $\langle 111 \rangle // \text{ND}$ rotation between the deformed matrix and the recrystallised grains was explained by Hutchinson²²⁾ as a compromise between the frequency of available nuclei and their respective growth velocities.

On the other hand, the development of $\{554\}\langle 225 \rangle$ during the recrystallisation of IF steels has often been explained by oriented growth. Hereby, $\{554\}\langle 225 \rangle$ is thought to develop at the expense of $\{112\}\langle 110 \rangle$, in accordance with the hypothesis that the boundary between a $\{554\}\langle 225 \rangle$ nucleus and a $\{112\}\langle 110 \rangle$ matrix has a significantly higher mobility than other boundaries. Support for this is thought to be found in bicrystal experiments,²³⁾ geometrical and segregation considerations about coincidence site lattice (CSL) boundaries^{12,24)} and a model⁹⁾ in which the nuclei whose directions of minimum elastic modulus are parallel to the maximum internal stress direction, grow preferentially. Recent EBSD experiments,²⁾ however, have shown that the nucleation texture already contains a significant amount of $\{554\}\langle 225 \rangle$ oriented grains after only 5 % recrystallisation, thus long before the partial α -fibre in general and $\{112\}\langle 110 \rangle$ in particular are consumed. A similar conclusion can be drawn from Fig. 7 of the present work. The view that $\{554\}\langle 225 \rangle$ is not due to oriented growth is also supported by experiments of Inagaki,¹⁹⁾ leading to the conclusion that the origin of $\{554\}\langle 225 \rangle$ nuclei lies in severely strained $\{554\}\langle 225 \rangle$ grains. The strong deformation in the latter grains is attributed to the effect of Ti and Nb, which—mainly as solutes—

Fig. 10. a) Image quality and b) grain boundary map of a partially recrystallised section measured with OIM; c) pole figure indicating the orientation spread of the deformed matrix and the orientations of the recrystallised grains.

hinder the rotation of $\{554\}\langle 225 \rangle$ toward $\{111\}\langle 112 \rangle$. Although the steel used here contains a very high amount of solute Ti and Nb, it is not clear whether the aforementioned rotation is severely hindered or not. However, the fact that the intensity of $\{554\}\langle 225 \rangle$ only decreases 20% by cold rolling (*i.e.* from $15 \times \text{RI}$ to $12 \times \text{RI}$) may be interpreted in favour of Inagaki's hypothesis.¹⁹⁾

It is clear that none of the aforementioned recrystallisation mechanisms is able to explain the simultaneous development of $\{111\}\langle 110 \rangle$ and $\{554\}\langle 225 \rangle$ during the initial stage of recrystallisation in a fully satisfying way. To add elements to the discussion, more detailed OIM measurements of small sections were performed for samples with a low recrystallised fraction. These measurements revealed that large orientation spreads are present in the deformed γ -fibre grains and confirmed the observation²⁾ that the scatter of the recrystallised γ -fibre grains is even larger than that of the deformed γ -fibre grain from which they originate. As an example, Fig. 10 shows 3 recrystallised grains located inside a (near) " $\{111\}\langle 112 \rangle$ " matrix. The recrystallised grains can readily be recognised because of their much higher image quality as compared to the surrounding matrix (Fig. 10a)). The orientation spread of the deformed matrix (A) and the orientations of the recrystallised grains (1-3) are shown in Fig. 10b) (for the calculation of the iso-intensity lines, a Gaussian half width of only 1° was used to avoid artificial broadening of the orientation spread). In line with Sec. 3.1, the form of the spread in (A) suggests that rotations have occurred along the ε - and the γ -fibre. The orientations of the grains 2 and 3 are closer to the centre of the orientation spread of (A) than that of grain 1. Roughly, 2 and 3 can be considered to be near $\{554\}\langle 225 \rangle$ grains, whereas 1 has an orientation near $\{111\}\langle 110 \rangle$. It is seen that grain 1

is completely surrounded by high angle grain boundaries (*i.e.* $> 15^\circ$), whereas this is not the case for 2 and 3 (cf. boundaries indicated with *). According to Hutchinson's theory,²²⁾ grain 1, being more misoriented from the bulk of the matrix, can be expected to have a higher probability for further growth than the other grains. However, it remains to be noted that the fact that 3 is much less misoriented from the bulk of A, did not prevent this grain from becoming as large as grain 1 at the time of the observation. Of course, it is possible that grain 3 has grown by only consuming sufficiently misoriented parts of the matrix. Nevertheless, it may also be thought that high misorientation is indeed a prerequisite for the growth of recrystallised grains surrounded by other such grains, but maybe not for the growth of nuclei in a deformed matrix. The latter view is actually supported by recrystallisation experiments of Konishi *et al.*²⁵⁾ Another remark that can be made with respect to Fig. 10 and other detailed measurements not shown here, is that near $\{111\}\langle 110 \rangle$ grains in a $\{111\}\langle 112 \rangle$ matrix are not necessarily located at the grain boundary with a deformed $\{111\}\langle 110 \rangle$ grain as in Inagaki's theory.¹⁹⁾ For instance, grain B in Fig. 10 has a near $\{111\}\langle 110 \rangle$ orientation, but the recrystallised grain 1 is observed in the interior of A and not in the grain boundary region. Also, no systematic rotation toward $\{111\}\langle 110 \rangle$ was observed in the grain boundary region of A neighbouring B. Nevertheless, although some of the present observations seem to disagree with assumptions of recrystallisation theories reported before, an important conclusion from Fig. 10 and other detailed OIM measurements remains that (near) $\{111\}\langle 112 \rangle$ deformed grains are at least partially consumed by (near) $\{111\}\langle 110 \rangle$ nuclei during the initial recrystallisation stage. The very strong development of $\{111\}\langle 110 \rangle$ during further annealing is

probably related to the high $\{111\}\langle 112 \rangle$ intensity in the rolling texture. However, more work is needed to refine the knowledge about the initial stage of recrystallisation in general and the (in the present case less important) development of $\{554\}\langle 225 \rangle$ in particular.

5. Conclusions

When a ferritically rolled and subsequently annealed Ti + Nb IF hot band is cold rolled, the absolute maximum shifts from $\{554\}\langle 225 \rangle$ to $\{111\}\langle 112 \rangle$ along the ε -fibre and the local maximum along the α -fibre shifts from $\{111\}\langle 110 \rangle$ to $\{223\}\langle 110 \rangle$. During recrystallisation the strong γ + weaker α -fibre texture transforms to a very strong γ -fibre texture. Hereby, the maximum intensity shifts from $\{111\}\langle 112 \rangle$ to $\{111\}\langle 110 \rangle$. Additionally, some development of $\{554\}\langle 225 \rangle$ is observed. OIM reveals that γ -fibre grains nucleate and grow in the beginning of recrystallisation by consuming deformed γ -fibre grains. Whereas the latter are subject to a wide orientation spread, the spread of the orientations of the resulting recrystallised grains along the γ -fibre is even wider. Nevertheless, the development of mainly $\{111\}\langle 110 \rangle$ and, to a lesser extent, $\{554\}\langle 225 \rangle$ already occurs during the initial stage of recrystallisation, so that the textures of the recrystallised grains obtained after 23 % and after full recrystallisation are very similar. Since the high intensity of near $\{111\}\langle 112 \rangle$ in the cold rolled texture is preferably consumed by near $\{111\}\langle 110 \rangle$ recrystallised grains, the latter are found to be larger and more numerous than $\{554\}\langle 225 \rangle$ grains. Still finer grains deviate even more from the perfect γ -fibre. $\{111\}\langle 110 \rangle$, therefore, has an advantage over the latter orientations for consuming the partial α -fibre grains during the final 40 % of recrystallisation and for further development during subsequent grain growth. As a result, a very strong $\{111\}\langle 110 \rangle$ peak type texture can be obtained, resulting in extremely good deep drawability.

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