STRENGTHENING THE {111} TEXTURE IN STEEL SHEET BY INCREASING THE LEVEL OF SOLUTE CARBON DURING COLD ROLLING

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A series of cold rolling and annealing experiments were conducted to examine the influence of solute carbon during cold rolling on the annealing texture. Hot band samples with grain sizes of 8 and 47 μm were used and rolling was carried out to reductions of 70% and 85%. Two levels of solute carbon were obtained prior to cold rolling by quenching and over-aging. Recrystallization textures were measured following single and two step isothermal annealing treatments. In the latter, the level of solute carbon is lowered prior to the commencement of recrystallization. For the single step treatment, the {111} texture intensity was lower in all the samples that had a high level of solute carbon present during rolling. However, when the two step annealing treatment was employed a different trend emerged. Under these circumstances, the fine grain size high rolling reduction sample showed a *stronger* {111} texture after annealing when the solute carbon level was *high* during rolling.

Keywords: Cold rolling; Annealing; Low carbon steel; Recrystallization;
Texture; Shear bands

INTRODUCTION

The recrystallization textures of steel are very sensitive to the presence of solute carbon during annealing. High levels of solute carbon impair the formation of textures that give steel its ability to be deep drawn

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(Ray *et al.*, 1994). This fact has been verified in many experiments and is an important consideration in the design and processing of sheet steels.

Reviews written in this (Ray *et al.*, 1994), and in the previous (Hutchinson, 1984) decade concur that solute carbon present during cold rolling is also detrimental to the formation of deep drawing textures. However, the earlier paper concedes:

"The fact that carbon can operate in two different ways, i.e. during deformation and during annealing, and that both affect the texture in the same way is highly confusing".

The most recent review also points out that controversy still exists over the respective contributions of carbon present during rolling and that in solution during annealing. One of the reasons for this is the difficulty in separating these two effects experimentally.

A number of workers have examined the influence of high levels of solute carbon during cold rolling by quenching samples from high temperatures prior to rolling. Steels processed in this manner generally show a weak deep drawing texture after annealing. That is, the intensities of orientations with a $\{111\}$ plane parallel to the plane of the sheet – $\{111\}$ textures* – are low (Huh *et al.*, 1995; Lavigne *et al.*, 1981). However, in these experiments the effect of carbon during rolling is confounded with that during annealing.

Instead of using quenching, Okamoto and Takahashi (1981) employed a complex series of decarburizing and carburizing experiments to vary the solute carbon level. Some of their results are reproduced in Fig. 1. Although there is, by the authors' admission, a reasonable degree of error in the data, the final $\{222\}$ intensity is clearly sensitive to the stage at which the carbon was introduced during processing.

There are two interesting points worthy of notice in these results. One is the increase in $\{222\}$ intensity that occurs with an increase in carbon content, irrespective of when the carbon was introduced. The other is the fact that the $\{222\}$ intensities are higher in the sample that has carbon in solution during rolling.

The possibility that solute C present during rolling may, under certain circumstances, enhance the formation of $\{111\}$ annealing textures is not at all in contradiction with recent studies on the nucleation of

* These orientations are often termed ND or α fibre textures.

FIGURE 1 Influence of solute carbon on the {111} texture after annealing – data obtained from Okamoto and Takahashi (1981).

recrystallization. One of the present authors (Barnett, 1998a) and Duggan *et al.* (1997) have shown that the nucleation of {111} textures can take place in shear bands. And it has been established that shear bands are more prevalent when there are higher levels of carbon in solution during rolling (Inagaki, 1990a). Until these recent studies, shear band nucleation in steel had been chiefly linked with the formation of {011}{100} or “Goss” textures (Ushioda *et al.*, 1981, Ushioda and Hutchinson, 1989).

The possible link between solute C during rolling and stronger {111} recrystallization textures is explored further in the present study. This is achieved using cold rolling and annealing tests where the level of solute carbon is contrived to be high during rolling but low during annealing, an experimental permutation that seems to have escaped previous attention.

EXPERIMENTAL

A high degree of carbon in solution during rolling was achieved by quenching a low carbon steel from 700°C to room temperature prior to processing. The solute C level was lowered prior to annealing by employing a low temperature heat treatment to precipitate the carbon as Fe_3C . Subsequent annealing was then conducted at a rapid heating rate to minimize the dissolution of these precipitates (Hutchinson and Ushioda, 1984; Ushioda *et al.*, 1986). The textures formed using this procedure were compared with those formed under other conditions.

The material used in this study was a commercial low carbon grade steel (0.06% C, 0.25% Mn, 0.04% Al, 0.003% N, 0.01% S, 0.01% P, 0.01% Si). The samples were received in the hot band state with a grain size of 8 μm and were 2.5 mm in thickness. One sample was heated to 950°C for 30 min and slowly cooled in insulation to produce a grain size of 47 μm . All the samples were subjected to a 1 h heat treatment at 700°C. This had negligible impact on the grain size and allowed near equilibrium levels of solute carbon to be established. It is also expected to have caused the complete precipitation of AlN (Wilson and Gladman, 1988). The samples were then quenched into a solution of brine. After quenching, the level of solute carbon is estimated to be in the order of 100 ppm (Lavigne *et al.*, 1981).

The samples were then subjected to a different series of rolling and heat treatments. Half were reheated to 450°C and held for 4 h followed by furnace cooling in an air oven. This was employed to precipitate the solute carbon as Fe_3C on existing particles and at grain boundaries. The solute carbon level in these samples is estimated to be in the order of 5 ppm (Lavigne *et al.*, 1981). These, and the as-quenched samples, were then cold rolled, some to 85% reduction and others to 70% reduction.

Next, two different annealing treatments were employed. One involved a salt bath treatment at 750°C for 30 s. The other incorporated a two step treatment, the first stage of which involved holding the samples at 450°C for 4 h in an air oven followed by furnace cooling. The second stage was a salt bath anneal at 750°C for 30 s.

The single step treatment meant that recrystallization took place with a similar, or greater, solute carbon level to that existing during rolling. In the two stage treatment, recrystallization of the samples that had high solute C levels during rolling occurred in the presence of considerably

lower levels of solute carbon, due to the precipitation of Fe_3C during the first annealing step. Precise measurements of the solute carbon level during annealing were beyond the scope of the present work but it may be surmised[†] that recrystallization in the low solute carbon samples began in the presence of between 20 and 50 ppm solute carbon.

In all, three groups of samples were obtained. One group had a starting grain size of $8\text{ }\mu\text{m}$ and was cold rolled to 70% reduction. Another had a starting grain size of $8\text{ }\mu\text{m}$ and was cold rolled to 85% reduction. The other group had a starting grain size of $47\text{ }\mu\text{m}$ and was cold rolled to 85% reduction. In each group, four different heat treatment combinations were used. These four conditions are illustrated in Fig. 2. Conditions three and four are different in that the two step anneal, while not impacting on solute carbon, is expected to impact on the degree of recovery occurring prior to recrystallization.

The superficial Rockwell hardness (HR30T) of the rolled samples were measured before and after the first step of the two step anneal. The textures were measured after rolling and after annealing using a Siemens goniometer. Texture software developed by Van Houtte (1995) was employed to calculate the $\varphi_2 = 45^\circ$ ODFs. A schematic showing the main orientations in this nomenclature is given in Fig. 3. The grain size was also measured after annealing.

RESULTS

Optical Microscopy

Optical microscopy of the deformed samples (see Fig. 4) revealed the existence of diagonal markings within the elongated grains. These were observed in all samples and were accompanied, in many cases, with a kink in the grain boundary. This kink suggests that localized flow had occurred along the diagonal band during rolling. Markings with a similar appearance have been observed by other authors (Huh *et al.*, 1995; Duggan *et al.*, 1997; Inagaki, 1990a; Ushioda *et al.*, 1981) and

[†] The present samples are expected to be 1% recrystallized by the time they reach 650°C in the salt pot (Barnett, 1998b). At this temperature Hutchinson and co-workers suggest that between 20 and 50 ppm C will be in solution (Hutchinson and Ushioda, 1984; Ushioda *et al.*, 1986).

FIGURE 2 Experimental conditions used in the present work.

FIGURE 3 ODF representations of the main steel texture components ($\varphi_2 = 45^\circ$ ODF section).

FIGURE 4(a)–(c)

FIGURE 4(d)–(f)

FIGURE 4 Example of in-grain shear bands seen in the present samples following cold rolling (light Nital etch – 2% solution, 15 s). Shear displacements can be seen where the bands contact the grain boundary. (a) 8 μm grain size, rolling reduction 70%, high solute carbon, (b) 8 μm grain size, rolling reduction 70%, low solute carbon, (c) 47 μm grain size, rolling reduction 85%, high solute carbon, (d) 47 μm grain size, rolling reduction 85%, low solute carbon, (e) 8 μm grain size, rolling reduction 85%, high solute carbon, (f) 8 μm grain size, rolling reduction 85%, low solute carbon.

TABLE I Percentage of grains with in-grain shear bands intercepting a linear trace imposed along the normal direction of a transverse section micrograph (average of 5 traces $\sim 160 \mu\text{m}$ in length)

	<i>Low solute C (%)</i>	<i>High solute C (%)</i>
8 μm grain size, 70% reduction	9	17
47 μm grain size, 85% reduction	19	29
8 μm grain size, 85% reduction	10	13

identified as shear bands, micro bands, striations or in-grain shear bands. Because of its descriptive nature, the present authors prefer the last of these terms.

Comparison of the frequency with which each sample displayed in-grain shear bands was hampered by the difficulty in etching and identifying the bands, particularly in the heavily rolled fine grain samples. A rough measure of the percentage of grains exhibiting in-grain shear bands was nevertheless obtained. This was achieved by counting the number of grains with diagonal markings that intersected a line imposed along the normal direction. The results are given in Table I.

A comparison of the data for high and low carbon levels reveals that in-grain shear bands are more prevalent when the solute carbon levels are high during cold rolling. It is also clear that the coarse grain size sample displays considerably more in-grain shear bands than the other samples.

Hardness Results

The superficial hardness values measured before and after the first step of the two step annealing treatment are given in Fig. 5. There are two main points to be made. One, the high solute carbon samples are noticeably harder than the low carbon specimens immediately following rolling. Two, the difference in hardness between the two sets of samples is virtually eliminated following the first stage of the two step anneal. The later demonstrates that over-aging occurred during the heat treatment at 450°C , as desired.

Rolling Textures

The $\varphi_2 = 45^\circ$ ODF sections of the rolling textures are presented in Fig. 6. Generally speaking, the textures formed in the samples with a high level

FIGURE 5 Hardness values for the present samples prior to annealing at 750°C.

of solute carbon are slightly weaker than their counterparts. It is also evident that increasing the rolling reduction in the 8 μm samples increased the texture sharpness, as would be expected.

The main rolling texture orientations in steel fall along the α fibre (or RD fibre). This texture fibre is defined in terms of Euler angles as $\varphi_2 = 45^\circ$ and $\varphi_1 = 0^\circ$. It corresponds to the left edge of the $\varphi_2 = 45^\circ$ ODF section shown in Fig. 3. The intensities along this fibre are plotted in Fig. 7 for each condition. It can be seen that the textures corresponding to 85% rolling reduction in the presence of high levels of solute carbon are sharper, particularly in the range of $\Phi = 0^\circ - 40^\circ$ (i.e. orientations approximately between $\{001\}\langle 110 \rangle$ and $\{112\}\langle 110 \rangle$). In the samples rolled to 70% reduction, the one with a high solute carbon level displayed an increased texture intensity in the region of $\{112\}\langle 110 \rangle$ but a lower intensity in the region of $\{001\}\langle 110 \rangle$.

Recrystallized Grain Size

After annealing, the average grain size was measured using a circle intercept method (see Fig. 8). The effects of initial grain size and rolling

FIGURE 6 ODF representation of the rolling textures ($\varphi_2 = 45^\circ$ ODF sections, intensity levels: 1, 2, 3, ...).

FIGURE 7 "Skeleton line" representation of the RD or α fibre for the rolling textures. These plots correspond to the data on the left edge of the ODF diagrams used in Fig. 6.

FIGURE 8 Recrystallized grain sizes measured using the circle intercept method.

reduction are clearly evident in all the samples; the grain size is decreased when the initial grain size is reduced and when the rolling reduction is increased.

In the sample with a 47 μm grain size, and in the sample rolled to 70% reduction, the presence of high levels of solute carbon during rolling correlates with a slight decrease in the grain size after annealing. However, the opposite is true for the sample with a starting grain size of 8 μm that was rolled to 85% reduction.

Recrystallization Textures

The recrystallization textures are presented in Figs. 9–11 as $\varphi_2 = 45^\circ$ ODF sections. Textures corresponding to the four experimental conditions outlined in Fig. 2 are grouped according to the initial grain size and rolling reduction. These are considered in turn below.

The sample subjected to a rolling reduction of 70% (Fig. 9) shows a weaker recrystallization texture for the material with a high level of

FIGURE 9 ODF representation of the recrystallization textures for the steel with a $8\mu\text{m}$ grain size rolled to 70% reduction ($\varphi_2 = 45^\circ$ ODF sections, intensity levels: 1, 2, 3, ...).

solute carbon present during rolling, for both annealing conditions. This effect is considerably stronger in the samples subjected to the single step annealing treatment. Also, the textures corresponding to the two step (i.e. low solute carbon) anneal show a peak near $\{111\}\langle 112 \rangle$ for the sample with a high solute carbon level during rolling. For the same annealing treatment, the sample with a low carbon level during rolling displays a peak close to $\{111\}\langle 110 \rangle$. The Goss component ($\varphi_2 = 45^\circ$, $\varphi_1 = 90^\circ$, $\Phi = 90^\circ$ or $\{001\}\langle 110 \rangle$) does not seem to be affected significantly by the level of solute carbon during rolling.

Similar trends are evident in the textures of the sample with a starting grain size of $47\mu\text{m}$ (Fig. 10). The $\{111\}\langle 112 \rangle$ component is more

FIGURE 10 ODF representation of the recrystallization textures for the steel with a $47\mu\text{m}$ grain size rolled to 85% reduction ($\varphi_2 = 45^\circ$ ODF sections, intensity levels: 1, 2, 3, ...).

prominent in the samples rolled with high levels of carbon in solution. The “Goss” component is also strong in all of these textures.

The recrystallization textures of the fine grain size ($8\mu\text{m}$) heavily rolled (85% reduction) sample shown in Fig. 11 display a similar dependency on processing conditions to the other samples for the *single step* anneal. For the *double step* annealing treatment, however, the steel with a high solute C level during rolling displays a sharper texture after annealing than the low carbon sample. This is most marked in the intensities of orientations near $\{111\}\langle 112 \rangle$. It is also worth pointing out that the “Goss” component is effectively absent in all of these samples.

FIGURE 11 ODF representation of the recrystallization textures for the steel with a $8\mu\text{m}$ grain size rolled to 85% reduction ($\phi_2 = 45^\circ$ ODF sections, intensity levels: 1, 2, 3, ...).

DISCUSSION

The current experiments differ from previous studies that have used quenching to examine the role of solute carbon on rolling and annealing in two ways. Firstly, and most importantly, two step annealing treatments were employed in the present work to lower the solute carbon level during recrystallization. Secondly, the finer of the two grain sizes of the current materials ($8\mu\text{m}$) seems to be considerably smaller than those corresponding to steels used in previous research (Huh *et al.*, 1995; Lavigne *et al.*, 1981; Inagaki, 1990a; Ushioda *et al.*, 1987). Despite these differences, there is considerable overlap between this and other studies to corroborate a number of findings and reinforce the veracity of the new results.

Deformed State

The trends in in-grain shear bands with changing processing conditions seen here were in agreement with those reported in the literature. That is, the degree of banding is increased when the grain size is raised and when the solute carbon level is increased. It is known that the heterogeneity of deformation increases with increasing grain size (e.g. Ridha and Hutchinson, 1982). The same holds true for the occurrence of in-grain shear bands. An increased frequency in in-grain shear bands at higher levels of solute carbon during rolling has been observed by a number of different researchers (Huh *et al.*, 1995; Inagaki, 1990a).

It has been shown that dynamic strain aging of carbon and nitrogen occurring during rolling at around 300–400°C increases the frequency of in-grain shear banding (Barnett and Jonas, 1997b; Ushioda *et al.*, 1981). It is possible that a similar effect occurs locally in steels with a high solute carbon level during *cold* rolling (Barnett and Jonas, 1997b; Ushioda *et al.*, 1987). The action of dynamic strain aging serves to decrease the strain rate sensitivity. This in turn facilitates the formation of shear flow localizations, which are regions of higher than average strain and strain rate.

The hardness measurements and the deformation textures agree well with expectations based on previous work. The samples with a high level of solute carbon present during rolling displayed a high hardness reading and a decreased rolling texture intensity for orientations between {001}<110> and {112}<110>. The latter observation has been made in studies of material with a starting grain size of 60 µm rolled to 70% and 90% reduction (Inagaki, 1990b). It has also been seen in material cold rolled to 80% reduction, with a hot band grain size of 30 µm (Huh *et al.*, 1995).

Inagaki (1990a,b) has suggested that this effect arises because solute carbon atoms lead to dense dislocation tangles that suppress normal lattice rotations. He also points out that the presence of solute carbon leads to an increase in the occurrence of shear bands and grain fragmentation. In the light of recent work (Barnett and Jonas, 1997b), in which a clear increase in in-grain shear banding was correlated with a significant drop in texture intensity, the present authors feel that the latter mechanism is probably the dominant one.

Recrystallized State

The influence of solute carbon on the frequency of in-grain shear bands can also explain the slightly finer recrystallized grain sizes seen in the samples rolled to 70% reduction and those with a 47 μm initial grain size. In-grain shear bands provide additional sites for the nucleation of recrystallization (Lavigne *et al.*, 1981; Inagaki, 1990a; Barnett and Jonas, 1997a). Generally, the more numerous the nucleation, the finer the final grain size (Barnett and Jonas, 1997a).

However, a breakdown of this generalization can be seen in the recrystallized grain sizes of the 8 μm material rolled to 85% reduction. In these samples, the material with a low degree of solute carbon present during rolling gave the finer grain size after annealing. This may be due to the fact that increased fragmentation resulting from the presence of solute carbon during rolling also causes an increase in the rate of growth of recrystallized grains. Under certain conditions, it is quite conceivable that a higher rate of growth of newly formed grains leads to a reduction in the activation of other nucleation sites. This can occur whenever potential sites are consumed by new grains before they give rise to a nucleus. Such an effect would be enhanced in the fine-grained heavily rolled sample because of the proximity of grain boundary nucleation sites to each other.

The intensities of the Goss ($\{011\}\langle 100 \rangle$) component in the recrystallized textures are summarized in Fig. 12. As noted above, the intensity of the Goss texture is greater on average in the 47 μm material compared to the 8 μm grade. This trend has been observed by other workers and has been explained in terms of shear band nucleation (Hutchinson *et al.*, 1990). Both single and poly-crystal studies have shown that Goss orientated grains nucleate in in-grain shear bands (Ushioda *et al.*, 1981; Ushioda and Hutchinson, 1989). The larger the grain size, the higher the frequency of in-grain shear bands and the higher the degree of Goss nucleation.

The decrease in Goss texture with increasing rolling reduction seen here has also been observed previously (Hutchinson *et al.*, 1990; von Schlippenbach and Lucke, 1988). In these cases, however, no explanation was offered. It is possible that the effect of rolling reduction is somewhat analogous to that of grain size. According to this view, thinning the grains by increasing the reduction serves to constrain the

FIGURE 12 ODF intensities of "Goss" component ($\{011\}\{100\}$) after annealing.

formation of in-grain shear bands. Although this is not unambiguously manifested in terms of shear band volume fraction in the present results (see Table I), the constraint placed on shear bands by nearby grain boundaries may be sufficient to prevent Goss orientations from forming in the bands. Of course, it is also probable that nucleation at grain boundaries is enhanced at higher strains, because of the increase in grain boundary area. This will lessen the contribution made by shear band nucleation to the overall texture.

As for the influence of solute carbon level during rolling on the Goss intensity after recrystallization, only a slight effect is evident in the current results. For the sample rolled to 70% reduction, the Goss intensity was sharper in the samples that had the higher carbon levels during rolling. This is in agreement with literature observations (Okamoto and Takahashi, 1981; Ushioda *et al.*, 1986; 1987; Hutchinson *et al.*, 1990) and is presumably due to an increase in in-grain shear banding. However, the situation for the large grain size sample is not so clear. In this material, the Goss texture is higher in the sample rolled with a low solute carbon level, after two step annealing. The trend in the

FIGURE 13 Average ODF intensities of the $\{111\}$ orientations after annealing.

samples annealed in the single step treatment is, however, consistent with previous findings.

The intensities of the $\{111\}$ components are summarized in Fig. 13. The $\{111\}$ intensity in this figure was calculated by taking the average of the ODF intensities along the ND fibre. Most of the samples with high levels of solute carbon present during rolling yielded a weaker $\{111\}$ texture after annealing. This is in agreement with previous studies (Ray *et al.*, 1994; Huh *et al.*, 1995; Ushioda *et al.*, 1986). However, the sharp $\{111\}$ texture (particularly the $\{111\}\langle 112 \rangle$ component) evident in the $8\ \mu\text{m}$ grain size material rolled to 85% reduction is a notable exception.

It is suggested that in this sample the presence of a high level of solute carbon does indeed increase the proclivity of the material to form shear flow localizations during rolling. This effect is, however, counter-balanced by the fine grain size and the high reduction. As a consequence, the shear bands that do form are not able to bear sufficient strain to produce Goss orientated regions (Barnett, 1998a). Instead, they cause fragmentation around the orientation of the grain in which they exist (Huh *et al.*, 1995; Barnett, 1998a). This, in turn, is likely to favour the nucleation during recrystallization of orientations that contain in-grain

shear bands. In steel, the orientation that seems to form the most in-grain shear bands is {111}<112> (Huh *et al.*, 1995; Barnett, 1998a; Ushioda *et al.*, 1981; 1986; Ushioda and Hutchinson, 1989).

This line of reasoning explains the appearance of a strong near {111}<112> recrystallization texture component in the sample with a fine grain size and high level of solute carbon that was subjected to a high cold rolling reduction and a two step annealing treatment. That the nucleation of near {111}<112> recrystallized grains in in-grain shear band structures is possible at all is supported by a number of recent electron microscopic studies (Barnett, 1998a; Duggan *et al.*, 1997). The detrimental impact of solute carbon during annealing on the nucleation of {111} textures explains why this effect is not observed after single step annealing.

The reasons why shear bands apparently cause the nucleation of Goss textures under certain conditions (i.e. coarse grain size, relatively low rolling reduction and high solute carbon levels during rolling and annealing) and {111} orientations under others (fine grain sizes, high rolling reductions, low solute carbon during annealing and high solute carbon during rolling) are not yet totally understood. A number of workers have suggested that considering the stable orientations for shear deformation inclined between 20° and 35° to the rolling plane can help understand the textures in shear bands (Barnett, 1998a; Ushioda and Hutchinson, 1989). These explanations are, however, not fully convincing and more work is required for a comprehensive explanation of this phenomenon.

CONCLUSION

The {111} texture intensity after annealing can be increased by raising the level of solute carbon present during rolling. This was accomplished here by employing a high rolling reduction, a finite grain size and an overaging treatment prior to annealing.

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References

- Barnett, M.R. (1998a). *ISIJ International*, **38**, 78.
- Barnett, M.R. (1998b). Unpublished work, BHP Steel.
- Barnett, M.R. and Jonas, J.J. (1997a). *ISIJ International*, **37**, 706.
- Barnett, M.R. and Jonas, J.J. (1997b). *ISIJ International*, **37**, 697.
- Duggan, B.J., Ning, H. and Zhang, L.X. (1997). *Int. Conf. on Thermomechanical Processing of Steels and Other Materials II*, Eds. Chandra, T. and Sakai, T., TMMMS, Warrendale, p. 2305.
- Huh, M.-Y., Raabe, D. and Engler, O., (1995). *Steel Research*, **8**, 353.
- Hutchinson, W.B. (1984). *International Metals Reviews*, **29**, 25.
- Hutchinson, W.B., Nilsson, K.-I. and Hirsch, J. (1990). *Metallurgy of Vacuum-Degassed Steel Products*, Ed. R. Pradhan, TMMMS, Warrendale, p. 109.
- Hutchinson, W.B. and Ushioda, K. (1984). *Scand. J. Metallurgy*, **13**, 269.
- Inagaki, H. (1990a). *Z. Metallkde.*, **81**, 474.
- Inagaki, H. (1990b). *Z. Metallkde.*, **81**, 15.
- Lavigne, J.-J., Suzuki, T. and Abe, H. (1981). *Proc. 6th Int. Conf. of Textures of Materials*, ISIJ, Tokyo, p. 749.
- Okamoto, A. and Takahashi, M. (1981). *Proc. 6th Int. Conf. of Textures of Materials*, ISIJ, Tokyo, p. 739.
- Ray, R.K., Jonas, J.J. and Hook, R.E. (1994). *International Materials Reviews*, **39**, 29.
- Ridha, A.A. and Hutchinson, W.B. (1982). *Acta Metall.*, **30**, 1929.
- Ushioda, K. and Hutchinson, W.B. (1989). *ISIJ International*, **29**, 862.
- Ushioda, K., Hutchinson, W.B., Agren, J. and von Schlippenbach, U. (1986). *Materials Science and Technology*, **2**, 807.
- Ushioda, K., Oshone, H. and Abe, M. (1981). *Proc. 6th Int. Conf. of Textures of Materials*, ISIJ, Tokyo, p. 829.
- Ushioda, K., von Schlippenbach, U. and Hutchinson, W.B. (1987). *Textures and Microstructures*, **7**, 11.
- Van Houtte, P. (1995). The MTM-FHM Software System, version 2, Users Manual.
- von Schlippenbach, U. and Lucke, K. (1988). *8th Int. conf. on Textures of Materials*, Eds. Kallend, J.S. and Gottstein, G., TMS, Warrendale, p. 861.
- Wilson, F.G. and Gladman, T. (1988). *International Materials Reviews*, **23**, 222.