Effect of Chemical Composition and Thermomechanical Processing on Texture in Hot Bands of Ti and Ti+Nb Containing Ultra-low Carbon Steels

S. YIM, P. WRAY, K. M. TIITTO, C. I. GARCIA and A. J. DEARDO

Basic Metals Processing Research Institute (BAMPRI), Department of Materials Science and Engineering, University of Pittsburgh, Pittsburgh, PA 15261 USA. Also at Pohang Iron & Steel Co. Ltd (POSCO), 1 Koedong-dong, Pohang, Kyongsangbuk-do 790-785, Korea. 1) Basic Metals Processing Research Institute (BAMPRI), Department of Materials Science and Engineering, University of Pittsburgh, Pittsburgh, PA 15261 USA. (Received on February 10, 2002; accepted in final form on April 9, 2003)

The effect of carbon, Ti and Nb on the formation of hot band texture in ultra-low carbon steels was studied. The chemical compositions were selected so that two of the four steels were fully stabilized with respect to carbon and the other two were expected to have some carbon in solution under equilibrium conditions. The slab reheating temperature ranged from 1 200 to 1 280°C. The first deformation was applied with a 50 % reduction to simulate the roughing pass at 1 150°C. The samples were then deformed with another 50 % reduction at either 1 050 or 920°C. The crystallographic orientations of the resulting ferrite were presented in the form of so-called skeleton plots along the RD-, TD- and ND-fibers. The main texture found was the cube-on-corner, (111)(110), component. The rotated cube component, (001)(110), was also present but its intensity was always lower than the intensity of the cube-on-corner component. The presence of the cube-on-corner texture was explained by the large grain size in the starting as-cast ingots and by the heavy reductions per pass. The combination of low reheating temperature and low finishing temperature generated the highest ratio of (111)(100). Decreasing the carbon content from 35 to 17 ppm and adding 90 ppm Nb to a Ti-alloyed ULC steel further increased the (111)(100) ratio.

KEY WORDS: ultra-low carbon steel; thermomechanical processing; hot rolling; texture; average strain ratio.

1. Introduction

The production of Ultra-Low Carbon (ULC) and Interstitial-Free (IF) steels has been increasing year by year thanks to the progress in vacuum degassing technology and the strong demand from the automotive industry. Good formability and moderate yield strength of these steels are responsible for their worldwide applications in the structural parts of the automotive vehicles. ULC steels contain low amounts of carbon and nitrogen, typically below 50 ppm. Adding strong carbide and nitride forming elements such as titanium and/or niobium can further reduce the level of interstitial elements in solution. The elimination of interstitial elements from the matrix results in good strain aging and deep drawability properties.

Variations in the thermomechanical processing (TMP) can have a profound effect on the microstructure and mechanical properties of ULC and IF steels. Numerous studies have been published on the effect of composition, slab reheating temperature (SRT), hot rolling finishing temperature (FT), coiling temperature, amount of deformation during hot and cold rolling, and annealing temperature on the mechanical properties of IF steels. There are, however, some disagreements in the results presented in the literature. Inconsistent results have been given for the effect of SRT and FT on the average strain ratio, $\bar{\tau}$. Some have reported the $\bar{\tau}$-values to be insensitive to SRT and FT, while others have shown a slight decrease in $\bar{\tau}$-values with decreasing FT. Other additional studies have shown little or no effect of FT on the ductility of the final product.

The work of Furano et al. indicates that there is a large decrease in ductility with decreasing FT. In another study, a slight increase in ductility with decreasing FT was reported. An interactive influence of SRT, FT and coiling temperature that results in a large effect of finishing temperature has also been reported.

The texture control of ULC and IF steels is of great importance since the average strain ratio, $\bar{\tau}$, and hence the deep drawability and springback are directly related to the presence of certain crystallographic orientations. The appropriate favorable texture to attain high $\bar{\tau}$-values is one in which a high proportion of grains is oriented with {111} planes parallel to the sheet plane, while {332} and {311} are also desirable. Strong {111} textures are also beneficial in reducing the in-plane variation in spring-back. Other texture components are relatively unfavorable, in particular those with {100} planes oriented near to the sheet plane. A high ratio of the texture intensity, $\{111\}/\{100\}$, is found to correlate well with $\bar{\tau}$-values. Several studies on Ti- and Ti+Nb alloyed ULC steels have...
found that the texture in the hot band consists mainly of a partial RD-fiber. This fiber starts at (001)(110) and the intensity decreases towards (111)(110). For high finishing temperatures, the intensity of (110)(111) may also be important. Ruiz-Aparicio et al. studied Ti- and Nb-Ti alloyed steels in the hot band condition and after cold rolling and annealing, and found a strong (111) texture for the hot band, which had undergone heavy deformation through a single pass reduction in the austenite region. They proposed that the origin of this texture is the presence of the austenitic Goss component, (110)(100), in the shear bands formed by heavy deformation on coarse austenite grains. This hot band texture was shown to lead to the formation of favorable textures during subsequent cold rolling and annealing.

Hutchinson et al. proposed two mechanisms for the effect of Ti and Nb on the formation of texture. The first is based on the scavenging effect whereby stable precipitates remove carbon and nitrogen from the solution. The highest r-values are obtained at carbon content of approximately 0.001%. An increase beyond this amount of carbon decreases r-values. The second effect is that of elements in solid solution on the beneficial texture formation. Especially Nb in solution was recognized to yield high r-values. The effect of Nb- and Ti-bearing precipitates on texture has been studied by many researchers who have emphasized the effect of these precipitates on recrystallization and grain growth. Especially Nb precipitation can inhibit the recrystallization of austenite during hot rolling and lead to sharp rolling textures in austenite.

The role of hot band texture has been emphasized because hot rolled ULC steels are increasingly being used without further processing. Since the presence of various texture components in the hot band of ULC steels seems to be unclear, the present work was conducted to study the effect of composition, mainly carbon, Ti and Nb, and thermomechanical processing variables on the formation of hot band texture in ULC steels. The main processing variables studied were the slab reheating temperature and the finishing temperature of the deformation in the austenite region.

2. Experimental Procedure

2.1. Materials and Processing

The materials prepared were 100 kg vacuum-melted ingots, the chemical composition of which is given in Table 1. The first three steels are alloyed with Ti and Nb, whereas steel 4 has Ti only. The locations of the steels in the stabilization maps developed by Hua et al. are shown in Fig. 1. The compositions of the first two steels are above the stoichiometric line and therefore the steels are fully stabilized with respect to carbon. The compositions of steels 3 and 4 fall below this line and these steels are expected to have some carbon in solid solution under equilibrium conditions.

Small specimens with a size of 25x50x100 mm were cut out of the as-cast ingots and austenitized for 1 h at either 1280 or 1200°C in a controlled atmosphere furnace. Hot rolling simulation was composed of two passes. The first deformation was applied with a 50% reduction to simulate a roughing pass at 1150°C. The samples were then deformed with another 50% reduction at either 1050 or 920°C. These temperatures are representative of the complete recrystallization and non-recrystallization (pancake) regions of austenite. After hot rolling, the strips were water cooled to the coiling temperature of 700°C. From 700°C they were cooled in a furnace at a cooling rate of 29°C/h to simulate the industrial coiling process.

2.2. Metallographic Analysis

Standard metallographic procedures were applied to prepare samples for optical microscopy as well as for transmission and scanning electron microscopic studies. Samples for texture analysis were obtained from square sections parallel to the rolling direction. Texture measurements were conducted by X-ray diffraction using a Philips X'PERT system with an open cradle and copper Kα radiation. Orientation distribution functions (ODF) of the global structure were calculated using the popLA algorithm based on the series expansion method proposed by Bunge and Roe from three incomplete pole figures of (200), (110), and (211), which were measured up to an azimuth of 85 degrees. Experimental background and defocusing corrections were applied to the pole figures by measuring the texture of a pure random iron sample. To better illustrate the presence of various textures so-called skeleton plots were constructed from the ODF data. These plots describe the preferred crystallographic orientations in the rolling direction (RD//<110>), normal direction (ND//<111>) and transverse direction (TD//<110>).

The presence of interstitial elements was assessed using internal friction by studying the value of the Snoek peak. Deconvolution of C and N from the overlapping Snoek peaks was conducted by using a computer program.

3. Results

3.1. Microstructure

Optical microscopy indicated the presence of equiaxed polygonal ferrite grains for all the hot rolled strips independent of chemical composition and thermomechanical processing. The average ferrite grain size ranged from 24 to 59 μm depending on the composition and process, steel 1 being most sensitive and steel 4 least sensitive to changes in the process. Table 2. Grain refinement by lowering the austenitizing temperature from 1280 to 1200°C was very small, whereas lowering the finishing temperature from 1050 to 920°C decreased the grain size considerably for all steels. For the finishing temperature of 920°C, the grain size is similar regardless of steel chemistry. On the other hand, finishing at 1050°C shows a trend for decreasing

| Steel | 1 | 2 | 3 | 4 |
|-------|---|---|---|---|
| C     | 17| 35| 35| 37|
| S     | 70| 70| 71| 72|
| N     | 27| 34| 30| 31|
| Ti    | 320| 290| 220| 300|
| Nb    | 210| 210| 90| 0 |
| Mn    | 1220| 1300| 1200| 1300|
| P     | 50 | 60| 60| 50|
| Si    | 40 | 70| 70| 70|
| Al    | 400| 520| 500| 519|
grain size from steel 1 towards steel 4. As a summary, the effect of finishing temperature on grain refinement is more pronounced than the effect of the slab reheating temperature. By providing more sites for ferrite nucleation, a low finishing temperature leads to fine ferrite grains. Therefore, tight control of the finishing temperature will yield a narrow range of grain size variation. Combining a low finishing temperature with a low slab reheating temperature yields further benefits for obtaining a fine uniform ferritic grain structure.

Transmission and electron microscopy showed the presence of TiN, Ti₄C₂S₂ and free standing MC precipitates for all steels independent of the processing condition. Using the mass balance provided by the stabilization maps, it was possible to calculate the amount of precipitates for the equilibrium condition. The results are given in Table 3. The observed presence of TiN, Ti₄C₂S₂ and MC precipitates is in a good agreement with the calculations for steels 2, 3 and 4. For steel 1, in contrast, the mass balance calculations predict no independent MC precipitation. The MC precipitates were primarily distributed in the matrix, while a few precipitates were found along the ferrite grain boundaries. In some cases, the MC precipitates were segregated. In addition to the above precipitates, sandwich-like precipitates, TiS–Ti₄C₂S₂–TiS, were also found. Most of these sulfide-bearing precipitates were randomly distributed in the matrix. The presence of these precipitates indicates an incomplete transformation of the TiS phase to the Ti₄C₂S₂ phase. Water spraying after finish rolling in the austenite did not allow enough time for the complete transformation of TiS to Ti₄C₂S₂.

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The amount of C, Ti and Nb in solid solution calculated based on the mass balance of the stabilization maps is given in Table 4. As mentioned above, steels 3 and 4 have a small
amount of carbon in solution. In these steels all titanium and niobium is tied in precipitates, while steels 1 and 2 have substantial amounts of these elements in solution.

3.2. Internal Friction

Studies of the Snoek peak showed the presence of interstitial elements only for steels 3 and 4, which according to the stabilization maps of Fig. 1 are not fully stabilized with respect to carbon. The total amount of carbon and nitrogen in the solid solution was measured to be between 1.5 to 1.8 ppm depending on the processing parameters. The presence of interstitial elements in steels 3 and 4 and their absence in steels 1 and 2 are in agreement with the predictions of the stabilization maps.

3.3. Texture

The skeleton plots following different thermomechanical treatments are given in Figs. 2(a)–2(d) for all steels studied. A characteristic feature of these results is the strong texture intensity in the ND fibers, extending from \{111\}(110) to \{111\}(112) with an intensity of 4 to about 7 times that of a random intensity. The strongest intensity in the RD fiber turns out to be the cube-on-corner component, \{111\}(110). The RD fiber also has a moderate intensity of the rotated cube component, \{001\}(110). This intensity seems to be higher when the rolling is finished at 1050°C than at

Fig. 2. Intensities f(\(\phi\)) of RD-, ND- and TD-fibers for steels 1 to 4. (a) SRT = 1,280°C, FT = 1,050°C, (b) SRT = 1,280°C, FT = 920°C, (c) SRT = 1,200°C, FT = 1,050°C, (d) SRT = 1,200°C, FT = 920°C.
920°C. The presence of the rotated cube component is generally observed in ULC steels, whereas the presence of the cube-on-corner texture in a hot band of ULC and IF steels has rarely been reported in the literature. The present results are in agreement with those of Ruiz-Aparicio et al.\textsuperscript{15,16} who also found the cube-on-corner texture component for Ti and Ti/Nb ULC steels using similar reheating and hot deformation parameters on an as-cast material as employed in this work. The TD fiber has a strong \{111\}(112) component followed by a strong \{554\}(225) and a relatively weak \{332\}(113) component. Other typical transformation texture components such as \{113\}(110), \{332\}(113) and \{110\}(110) are less prominent.

It can readily be observed in Fig. 2 that the same texture components having similar intensities are present for all steels. There are, however, some variations in the intensities between the different steels and the different processing conditions. To study the effect of composition and processing in more detail, the \(I\{111\}(110)/I\{001\}(110)\) intensity ratio was calculated for the present steels as shown in Fig. 3. This figure indicates the importance of the finishing temperature since the highest ratios are obtained when the finishing is conducted in the non-recrystallized region of austenite. Lowering the reheating temperature from 1280 to 1200°C further increases this ratio except in two special cases. The effect of this temperature, however, is minimal if the finishing temperature of the deformation is high.

The highest intensity ratios in Fig. 3 are obtained for steels 1 and 3 when both the reheating and finishing temperatures are low. Both these steels are alloyed with Nb and Ti. Their carbon contents are, though, substantially different. According to the stabilization maps of Fig. 1, steel 1 is fully stabilized, whereas steel 3 was found to have some carbon left in solution. This result would indicate that the absolute carbon content and whether it is in solution or tied in precipitates is not of paramount importance. Comparing steel 1 to 2 after low temperature austenitizing and finish rolling leads to a different conclusion. These two steels have essentially the same chemical composition except for the carbon content. Although both steels are expected to be fully stabilized with respect to carbon and although neither of them showed the presence of interstitial elements with the internal friction measurement, the intensity ratio for steel 1 with the lower carbon content is 30% higher than that for steel 2.

The effect of Nb can be seen by comparing steels 3 and 4 in Fig. 3. For the high finishing temperature of 1050°C, Nb seems to have no significant effect. The beneficial effect of Nb can readily be observed when the deformation is finished in the non-recrystallization region. Adding as little as 90 ppm of Nb can increase the intensity ratio by about 50%.

4. Discussion

The skeleton plots of Fig. 2 indicate that the \{111\} texture in the ND fiber is the most significant texture component for the steels studied. The detrimental \{001\}(110) component is present but its intensity is low compared to that of any \{111\} component. This observation is independent of steel composition and thermomechanical processing. The textures in the present work differ from those reported by several other authors,\textsuperscript{12–14} who also investigated austenitically hot-rolled Ti- and Ti+Nb-alloyed ULC steels. In all of these studies, the hot band texture consisted mainly of a partial RD-fiber, see Fig. 4. This fiber started at \{001\}(110) and the intensity decreased towards \{111\}(110). For a high finishing temperature, the intensity of \{001\}(110) was also found. The \{001\}(110) component was always strongest and its intensity was higher than that of the ND-fiber. The present results are, however, in agreement with

![Fig. 3. Intensity ratio of \{111\}(110)/\{001\}(110) for steels 1 to 4.](image)

![Fig. 4. Examples of RD (\(\alpha\)) and ND (\(\gamma\)) fibers, as reported in the literature. Temperatures in the legend are finishing temperatures. E, \(E'\) = \{111\}(110), F = \{111\}(122), H = \{001\}(110), L = \{011\}(110).](image)
the work of Ruiz-Aparicio et al.15,16 who also found a strong {111} texture in the ND fiber and a low intensity for the rotated cube component, {001}//{110}. Reasons for these observations can be found by comparing the condition of the starting material and the thermomechanical processing parameters to those applied in the earlier works. In the present study, as-cast slabs were hot rolled using only two passes with a 50% reduction per pass. The same kind of rolling schedule was used by Ruiz-Aparicio, whereas those investigators who found the unfavorable {001}//{110} as the main texture component, employed several passes on transfer bars or laboratory cast steels with a reduction of 20 to 30% per pass. Reheating the as-cast slabs in the austenitic region generates a much larger grain size than reheating the transfer bars or laboratory cast steels. Ruiz-Aparicio15 found that the ferritic grain size of a Ti-alloyed IF steel can be 15 times larger in the as-cast condition than in the transfer bar. The implication is that the prior austenitic grain size was also much larger in the as-cast condition. Heavy deformation for sheet plane orientations near {110} exists with the non-recrystallization region having a greater effect on the formation of shear bands.36–38 Shear bands develop with a non-crystallographic geometry when the normal crystallographic slip is strongly inhibited.34 In the case of ULC non-crystallographic geometry when the normal crystallographic slip is strongly inhibited.34 In the case of ULC steel, shear band formation has been observed in warm rolling and cold rolling.35,36 Hutchinson et al.36 have shown that the intensity of the Goss component, {110}/(001), in fcc materials with a low stacking fault energy increases as the volume of shear bands increases. They also showed that even when there is a wide scatter in the orientation of individual crystals within the shear bands, a preference for sheet plane orientations near {110} exists with the orientation {110}/(001) being most favored. The significance of the Goss component is obvious since it transforms into the cube-on-corner, {111}//{110}, orientation according to the Kurdjumov–Sachs relationship.39 It can be concluded that the presence of this orientation {111}//{110} orientation in the present steels is due to the large initial grain size in the as-cast material and to the heavy passes in the rolling schedule.

A general observation in Fig. 3 is that the highest {111}//{001} intensity ratio is obtained for the combination of low slab reheating temperature and low finishing temperature, although the variations in this intensity ratio are rather small. The grain size for both high and low slab reheating temperatures is expected to be large enough to generate shear bands under the heavy 50% deformation. The variation of the slab reheating temperature did not change the unique texture of the present study, since the Goss mechanism plays the dominant role in the texture development in shear bands. In contrast, decreasing the finishing temperature from the complete recrystallization region to the non-recrystallization region has a greater effect on the {111}//{001} intensity ratio than the slab reheating temperature. By lowering the finishing temperature from 1050 to 920°C increases the intensity of the cube-on-corner component, {111}//{110}, at the expense of the intensity of the rotated cube component. Recrystallized austenite is known to bring about a higher intensity of the rotated cube component, whereas finishing in the non-recrystallized austenite region favors the generation of the beneficial cube-on-corner component. The presence of the rotated cube compon-

tent for the finishing temperature of 920°C indicates that recrystallization still occurs to some degree before transformation to ferrite.

According to the internal friction measurements, steels 1 and 2 are fully stabilized with respect to carbon and nitrogen, whereas steels 3 and 4 have 1.5 to 1.8 ppm of carbon and nitrogen in solution, in agreement with the stabilization maps. The beneficial effect of lowering the carbon content is obvious for the combination of low slab reheating temperature and low finishing temperature by comparing steel 1 to steel 2 in Fig. 3. Precipitation in the austenite region can delay the onset of recrystallization and grain growth. On the other hand, the lower total content of precipitates in steel 1 than in the other steels (Table 3) means that there can be some Nb left in solid solution hindering recrystallization and grain growth. Non-recrystallized austenite could lead to the generation of a larger proportion of the beneficial {111} texture component, as seen in Fig. 3.

5. Conclusions

(1) The main texture found for the Ti and Ti+Nb alloyed ULC steels studied was the cube-on-corner, {111}/(110), component. The rotated cube component, {001}//{110}, was also present but to a lower degree. The strong presence of the beneficial cube-on-corner texture was explained by the large grain size in the starting as-cast material and by the heavy reductions per pass, both of which favor the formation of shear bands.

(2) The highest intensity ratio of {111}/(100) was found for the combination of low slab reheating temperature and low finishing temperature.

(3) Decreasing carbon content from 35 to 17 ppm increased the intensity ratio of {111}/(100) by 30% when both the slab reheating temperature and finishing temperature were low.

(4) Adding 90 ppm niobium to a Ti-alloyed ULC steel increased the {111}/(100) intensity ratio by 50%.

REFERENCES

1) T. A. Bloom and R. W. Nunke: Proc. of 32nd MWSP Conf., ed. by L. G. Kuhn, ISS-AIME, Warrendale, PA, (1990), 229.
2) I. Gupta and D. Bhattacharya: Proc. of the Int. Symp. on Metallurgy of Vacuum-Degassed Steel Products, ed. by R. Pradhan, TMS, Warrendale, PA, (1990), 43.
3) I. Gupta, T. Parayil and L. T. Shiang: Hot and Cold Rolled Sheet Steels, ed. by R. Pradhan and G. Ludkovsky, TMS-AIME, Warrendale, PA, (1988), 139.
4) S. Satoh: KSC Report, (1985), 36.
5) Y. Funno: Tetsu-To-Hagane, 71 (1985), 1362.
6) S. Satoh, T. Obara, M. Nishida and T. Irie: Technology of Continuously Annealed Cold Rolled Sheet Steel, ed. by R. Pradhan, TMS-AIME, Warrendale, PA, (1985), 151.
7) S. Sayanagi: Tetsu-To-Hagane, 71 (1985), 1361.
8) S. Satoh, T. Obara, M. Nishida and T. Irie: Trans. Iron Steel Inst. Jpn., 24 (1984), 838.
9) H. Katoh, H. Takechi, N. Takahashi and M. Abe: Technology of Continuously Annealed Cold Rolled Sheet Steel, ed. by R. Pradhan, TMS, Warrendale, PA, (1985), 37.
10) W. B. Hutchinson: Int. Mater. Rev., 29 (1984), No. 1, 25.
11) M. P. Renaivkar, P. J. Wray, C. I. Garcia and A. J. DeArdo: Proc. of 39th Mechanical Working and Steel Processing Conf., ed. by M. A. Baker, ISS, Warrendale, PA, (2001), 211.
12) M. P. Butoiu-Guillien and J. J. Jonas: Proc. of the Int. Forum for Physical Metallurgy of IF Steels, ISIJ, Tokyo, (1994), 123.
13) Y. B. Park, D. N. Lee and D. Gottstein: *Acta Mater.*, 44 (1996), 3421.
14) K. Eloot: Internal Report, BAMPRI, University of Pittsburgh, (1997).
15) L. J. Ruiz-Aparicio: Ph. D. Thesis, School of Engineering, University of Pittsburgh, (1998).
16) L. J. Ruiz-Aparicio, C. I. Garcia and A. J. DeArdo: IF Steels 2000 Proc., ISS, Warrendale, PA, (2000), 85.
17) W. B. Hutchinson, K.-I. Nilsson and J. Hirsch: Proc. of the Int. Symp. on Metallurgy of Vacuum-Degassed Steel Products, ed. by R. Pradhan, TMS, Warrendale, PA, (1990), 109.
18) R. E. Hook, A. J. Heckler and J. A. Elias: *Metall. Trans.*, 6A (1975), 1683.
19) R. E. Hook and H. Nyo: *Metall. Trans.*, 6A (1975), 1443.
20) S. Satoh, T. Ohara and K. Taniyama: *Trans. Iron Steel Inst. Jpn.*, 26 (1986), 737.
21) S. V. Subramanian, M. Priyky, B. D. Gaulin, D. D. Clifford, S. Benincasa and I. O’Reilly: *ISIJ Int.*, 34 (1994), No. 1, 61.
22) O. Kwon, G. Kim and R. W. Chang: Proc. of the Int. Symp. on Metallurgy of Vacuum-Degassed Steel Products, ed. by R. Pradhan, TMS, Warrendale, PA, (1990), 215.
23) S. Hashimoto, T. Yukushiji, T. Tashima and K. Hosomi: Proc. of Int. Conf. on Physical Metallurgy of Thermomechanical Processing of Steels and Other Metals, Vol. 2, ISIJ, Tokyo, (1988), 652.
24) T. Senuma and H. Yada: Proc. of Int. Conf. on Physical Metallurgy of Thermomechanical Processing of Steels and Other Metals, Vol. 2, ISIJ, Tokyo, (1988), 636.
25) T. Nakamura and K. Esaka: Proc. of Int. Conf. on Physical Metallurgy of Thermomechanical Processing of Steels and Other Metals, ISIJ, Vol. 2, Tokyo, (1988), 644.
26) M. Hua, K. Eloot, M. V. Phadke, P. J. Wray, C. I. Garcia and A. J. DeArdo: TOOLBOX 1-B, University of Pittsburgh, (1997).
27) H. J. Bunge: *Z. Metallkd.*, 56 (1965), 872.
28) R. J. Roe: *J. Appl. Phys.*, 36 (1965), 2024.
29) K. Brown: *J. Inst. Met.*, 100 (1972), 313.
30) S. Nourbaksh: Ph. D. Thesis, University of Leeds, (1978).
31) D. J. Willis and M. Hatherly: Proc. of the 5th Int. Conf. on Texture of Materials, Vol. 1, Springer Verlag, Berlin, (1978), 48.
32) P. S. Mathur and W. A. Backofen: *Metall. Trans.*, 4 (1973), 643.
33) J. G. Sevillano, P. Houtte and E. Aernoudt: *Scr. Metall.*, 11 (1977), 581.
34) J. Hirsch, K. Lucke and M. Hatherly: *Acta Metall.*, 36 (1988), 2905.
35) M. R. Barnett: *ISIJ Int.*, 38 (1998), No. 1, 78.
36) D. O. Wilshinsky-Dresler, D. K. Matlock and G. Karuss: Int. Forum for Physical Metallurgy of IF Steels, ISIJ Tokyo, (1994), 13.
37) W. B. Hutchinson, B. J. Daggan and M. Hatherly: Copper ’77, Met. Soc., *Met. Technol.*, 6 (1979), No. 10, 398.
38) G. Kurdjumov and G. Sachs: *Z. Phys.*, 64 (1930), 225.
39) L. Burgelman and S. Claessens: unpublished result, Universiteit Gent, Belgium, (1996).