Microstructural and Texture Development during Multi-Pass Hot Deformation of a Stabilized High-Chromium Ferritic Stainless Steel

Saara MEHTONEN,1)* Eric PALMIERE,2) Devesh MISRA,3) Pentti KARJALAINEN1) and David PORTER1)

1) Centre for Advanced Steels Research, University of Oulu, P.O. Box 4200, 90014 University of Oulu, Finland. 2) Department of Materials Science and Engineering, The University of Sheffield, Sir Robert Hadfield Building, Mappin Street, Sheffield, S1 3JD, UK. 3) Center for Structural and Functional Materials, University of Louisiana at Lafayette, P.O. Box 44130, Lafayette, LA 70504-4130 USA.

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With the ultimate target of improving the deep drawability of a dual-stabilized 21%Cr ferritic stainless steel, the evolution of the flow stress, microstructure, texture, dislocation structures and precipitation during multi-pass hot deformation were studied. Plane strain compression in three passes with 0.4 – 0.5 pass strains and 20 s inter-pass times was employed together with scanning electron microscopy combined with electron backscatter diffraction (SEM-EBSD) and transmission electron microscopy (TEM). The temperature of the final pass was varied between 1 223 K and 923 K and the final cooling took place either by water quenching to room temperature or water cooling to 923 K followed by cooling at 0.33 K/s to room temperature. At 1 223 K, static recrystallization was almost complete during the 20 s inter-pass times and this randomized the texture. When the deformation temperature was lowered to 1 073 K or 923 K, in-grain shear bands were formed in the grains belonging to the γ fibre. The deformation temperature of the third pass had only a minor effect on the deformation texture intensity maxima. The final dislocation structure was not changed by the cooling rate from 923 K, but slow cooling enabled precipitation to occur. The results indicate that although the deformation conditions affect the deformed microstructures and dislocation structures, the effect of the deformation temperature on the texture was insignificant.

KEYWORDS: hot deformation; deformation temperature; cooling rate; recrystallization; texture; dislocation structure.

1. Introduction

Dual Ti–Nb stabilized high-chromium ferritic stainless steels with over 20% Cr are a rather new and promising steel group. They can be used in many general stainless steel applications such as in kitchen appliances or in construction materials. The key requirement is often good formability, especially deep drawability, and surface quality. Unfortunately, the high-chromium ferritic stainless steels may develop an undesirable surface defect in deep drawing known as roping and the appearance of which is closely related to the texture of the steel, e.g.1,2)

The composition of the stabilized high-chromium ferritic stainless steels is such that the austenite – ferrite phase transformation, which would affect the texture, does not occur. Therefore, in the course of hot rolling, the texture of the hot band can only be changed from the solidification texture, which is generally considered harmful, by deformation and recrystallization. The textures in the final cold rolled and annealed products are known to be inherited from the hot rolling textures3) and therefore, in order to achieve good surface quality and formability, the texture of the hot rolled sheet should be optimized regarding the later processing stages.

Because of extensive recovery in ferritic steels, recrystallization may not even take place or is not fully completed during the inter-pass times at conventional hot rolling temperatures. Since recovery does not remove the unfavorable texture structures,4) the effect of lowering the hot rolling temperature into the warm rolling temperature range, i.e. below ~970 K, on the microstructure and texture of IF and low-carbon steels has been studied extensively, e.g.5–8) in order to find ways to promote static recrystallization (SRX) and the formation of favorable textures during the course of hot deformation. It is generally agreed that lowering the hot rolling temperature leads to flow localization within individual grains and thereby to the formation of in-grain shear bands, which act as nucleation sites for SRX. In-grain shear bands not only enhance the SRX kinetics but they also increase the number of recrystallized grains oriented with <111>//ND, which helps achieve high r-values and good deep drawability.7)

The positive effects of lowering the hot rolling finishing temperature to ~1 020 K has also been observed in stabilized 21% Cr9) and 17% Cr10–12) ferritic stainless steels. As with
low-carbon and IF steels, the positive effect of lowering the hot rolling finishing temperature was caused by the formation of in-grain shear bands.\textsuperscript{9–11} However, Sawatani et al.\textsuperscript{12} suggested that the positive effect was caused by different precipitation behavior at different temperatures instead of formation of in-grain shear bands.

In the above-mentioned studies the positive effect of lowering the finishing rolling temperature on SRX kinetics, texture evolution and on the deep drawability of the final product has been investigated using various hot rolling simulations. However, a more detailed study, which would combine the microstructural evolution during the first passes and inter-pass times together with lowering the hot deformation finishing temperature needs to be carried out in order to identify the actual phenomena taking place during the various stages of multi-pass hot deformation.

In the current study, multi-pass hot deformation is studied in detail by simulating Steckel mill rolling conditions. Similar simulations using plane strain compression tests on AISI 430 ferritic stainless steel have been carried out previously by Hinton and Beynon.\textsuperscript{13} However, the behavior of AISI 430 differs significantly from that of the currently investigated steel because it undergoes austenite – ferrite phase transformation, which affects both the restoration mechanisms and texture formation.

The final aim of the current work is to study the microstructure and texture development during multi-pass hot deformation and to identify the optimal finishing temperature range, which would lead to favorable microstructures and textures regarding later processing stages and ultimately to an improved formability of the end product.

2. Experimental

The behavior of a stabilized high-Cr ferritic stainless steel during hot deformation and the influence of the final pass temperature were studied by simulating an industrial Steckel mill rolling process with the aid of three-pass plane strain compression (PSC) tests. The tests were carried out at the Institute for Microstructural and Mechanical Process Engineering, IMPPETUS at the University of Sheffield, UK by employing the TMC thermomechanical compression machine. The test material was a Ti–Nb stabilized ferritic stainless steel supplied as a transfer bar. The initial grain size was 72 μm with an initial random texture. The chemical composition of the steel in weight percent is listed in Table 1. 10 mm thick, 30 mm wide and 60 mm long specimens were cut from the middle of the hot band with their normal axis parallel to the normal direction of the hot band and longitudinal axis parallel to the rolling direction. The width of the compression anvil was 15 mm. No lubrication was used, however the effect of friction was taken into account when calculating the isothermal stress, which is the measured stress compensated for deformation heating, as proposed in Ref. 14).

The specimens were reheated at the rate of 10 K/s to 1373 K for solution treatment, held for 120 s and then cooled at 10 K/s to 1223 K where the first two reductions were applied with a strain of 0.4 or 0.5 at a strain rate of 20 s\textsuperscript{-1} and an inter-pass time of 20 s. The process parameters are comparable to the deformation conditions in the industrial Steckel mill rolling.

The third hit after an inter-pass time of 20 s was applied with an equivalent strain of about 0.45 at the true strain rate of 20 s\textsuperscript{-1} at different temperatures of 1223, 1073 or 923 K. The cooling from 1223 K to 1073 or 923 K during the second inter-pass time was assisted and controlled by a water - air mist, so that the temperature of the third pass was achieved within 20 s. The cooling was started immediately after the second compression and the cooling rates from 1223 to 1073 or 923 K were 7.5 K/s and 15 K/s, respectively. After the third deformation, the cooling rate was also varied. The specimens were either water quenched to room temperature or to 923 K from where they were cooled slowly over 1 200 s to 523 K at a constant cooling rate of approximately 0.33 K/s.

In order to study the deformed structures after the first, second and third pass, some specimens were water quenched right after the deformation hit. Other specimens were water quenched after the inter-pass period for studying the static recrystallization kinetics and recrystallization textures. The microstructures and textures were examined using the electron backscatter diffraction (EBSD) method on a Zeiss Ultra Plus FEG-SEM with an HKL EBSD detector. The accelerating voltage was 15 kV and step size either 0.5 μm or 0.2 μm at magnifications of 300× and 1 000×, respectively. For obtaining the orientation distribution functions (ODF), a lower magnification of 300× was employed in order to measure a sufficient number of grains (minimum of ~100). For EBSD studies, the specimen surfaces were polished using a diamond suspension down to 1 μm after which a chemical polishing was applied using a 0.05 μm colloidal silica suspension. The structural features were examined in a transmission electron microscope (Hitachi H7600) operated at 120 kV. Thin foils were prepared from 3 mm disks punched from the specimen, by using twin-jet electropolishing in an electrolyte containing 10% perchloric acid in acetic acid.

3. Results

3.1. Flow Curves

The measured flow stress curves during the three-pass deformation are presented in Fig. 1. True stress represents the measured flow stress and isothermal stress shows the flow stress compensated for deformation heating. The isothermal stress is slightly higher than the measured true stress, except at 1073 K or 923 K, where the difference is more noticeable. During the first two passes carried out at 1223 K, the flow stress appears to be at the same level at the both passes (~160 MPa) (Fig. 1(a)). This indicates that during the inter-pass period, the microstructure undergoes significant SRX that reduces the dislocation density created during the deformation leading to complete softening of the material. Also after an initial work hardening stage, a steady state is reached as seen from the flow stress curves.

| Table 1. Chemical composition of the investigated steel (in wt%). |
|-----------------|-----------------|-----------------|-----------------|-----------------|-----------------|-----------------|-----------------|-----------------|
| C               | N               | Cr              | Ni              | Ti              | Nb              | Cu              | Si              | Mn              |
| 0.02            | 0.02            | 20.8            | 0.2             | 0.15            | 0.25            | 0.3             | 0.6             | 0.35            |

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As expected, lowering the temperature of the third pass leads to higher flow stress values, as seen in Figs. 1(b) and 1(c). The maximum flow stresses measured during the third pass were 160, 300 and 400 MPa at 1 223, 1 073 and 923 K, respectively. During the third pass at the lower temperatures of 1 073 and 923 K, a steady state is not achieved, and work hardening continues till the end of the deformation pass. Work hardening is especially extensive at 923 K, which indicates significant dislocation storage in the microstructure.

3.2. The Evolution of Microstructures and Textures during the First Two Passes

The microstructural evolution at 1 223 K during and after the first two passes of the Steckel rolling simulation is presented in Fig. 2. After the first pass and immediate water quench (Fig. 2(a)), the microstructure consists of deformed and recovered grains. Numerous low-angle grain boundaries are formed inside the original grains. However, it is evident that the amount of substructure varies significantly between grains. After the first pass (Fig. 2(a)), grain A contains minimal substructure whereas there are numerous low-angle grain boundaries in grain B. Further, many of the subboundary in grain B seem to be aligned in an angle of approximately 35° to the rolling direction, and they seem to form a “banded” substructure.

New small grains with a diameter up to ~10 μm can be seen at the original grain boundaries (shown by arrows in Fig. 2(a)). In the TMC machine, the deformation takes place inside a furnace and therefore the water quench cannot be carried out in situ, but is delayed by approximately 0.4 s. Therefore, these grains are likely to have formed statically during the delay. Based on the microstructure observations, it is evident that the deformation conditions at 1 223 K (0.4 strain, 20 s\(^{-1}\) strain rate) enable dynamic recovery (DRV) to occur, which can be seen as substructure, cells and subgrains formed during the first pass. Further, some static recovery has probably occurred during the delay between the deformation pass and quench. It is unlikely that the recrystallized grains had formed dynamically during the deformation.

Almost complete SRX took place during the 20 s inter-pass time after the first pass (Fig. 2(b)). The grain size and the recrystallized fraction after the first pass were determined to be 39 μm and 92%, respectively.

The deformation microstructure after the second pass is similar to that after the first pass, i.e. the amount of substructure varied between the grains and a banded substructure was present within some grains, as seen in Fig. 2(c). The bands were aligned at ~35° to the rolling direction. Further, small (max ~10 μm) recrystallized grains were detected at the original grain boundaries, as indicated by the arrows in Fig. 2(c), and around coarse TiN particles as a result of particle stimulated nucleation (PSN) as shown in Fig. 3, where a part of Fig. 2(c) is shown in a higher magnification. A TiN particle is shown in grey color and pointed out with a black arrow. The recrystallized grains around the TiN particle are pointed out with white arrows. As with the first pass, these grains have most likely formed through SRX during the delay between the end of deformation and the quench.

The SRX after the second pass resulted in a slightly larger grain size (54 μm) than after the first inter-pass period (39 μm). However, the recrystallized fraction (97%) was marginally higher than the 92% achieved after the first pass. The recovered grain (marked with a black arrow in Fig. 2(d)) had orientation close to \{001\}\langle110\rangle, which is an orientation known to have low Taylor factor, low dislocation density, and thus a low driving force for recrystallization, and is prone to undergo only recovery and is therefore the last to recrystallize.\(^{1,15,16}\)

The \(\phi_2=45^\circ\) sections of the orientation distribution functions (ODF) representing the texture evolution between the first two passes are presented in Fig. 4. After the first pass...
and water quench (Fig. 4(a)), there was a strong texture component in the $\alpha$ fibre at $\{223\}<110>$ and a smaller intensity peak at the beginning of the $\varepsilon$ and $\alpha$ fibres, at $\{001\}<110>$. Also small texture components between $\{111\}<132>$ and $\{111\}<011>$ in the $\gamma$ fibre were present. The maximum texture intensity was 9.49.

The SRX during the inter-pass time weakened the texture and the maximum texture intensity was measured to be 3.85 (Fig. 4(b)). However, in terms of the locations of the intensity maxima, the recrystallization texture was very similar to the deformation texture. A maximum is detected at the end of the $\alpha$ fibre at $\{111\}<110>$. Also, as before SRX, there is a small maximum in the $\gamma$ fibre at $\{111\}<011>$. The inter-pass SRX also introduces a small maximum at $\{111\}<112>$, seen in the middle of the $\varepsilon$ fibre and the end of the $\gamma$ fibre.

The deformation texture after the second pass consists of both $\alpha$ and $\gamma$ fibres with the intensity maxima along the $\alpha$ fibre at $\{001\}<110>$ and $\{111\}<110>$, and along the $\gamma$ fibre at $\{111\}<123>$ and $\{111\}<112>$. Also there was a maximum at the beginning of the $\varepsilon$ fibre at $\{001\}<110>$. Overall,
the texture formed after the second pass was weaker than after the first pass: the maximum texture intensity was 5.71 after the second pass compared to 9.49 after the first pass. This is probably due to the texture randomizing effect of the SRX in the inter-pass time prior to the second pass.

3.3. The Effect of the Third Pass Temperature on Microstructure and Texture

The microstructures of the deformed specimens after the third pass and water quenching are presented in Fig. 5. It is evident that the temperature of the third deformation pass strongly affects the number of sub boundaries and the degree of dynamic restoration. At 1223 K (Fig. 5(a)) the microstructure is dynamically recovered whereas after deformation at 1073 or 923 K (Figs. 5(b) and 5(c), respectively) the microstructure is still much in a deformed state.

The number of low-angle grain boundaries (pictured as yellow and white lines) increases significantly as the deformation temperature is lowered from 1223 to 923 K.

Based on grain morphology and orientation, three different types of grains were identified, marked by A, B and C in Fig. 5. Type A grains, which belonged to the α fibre (<110>//RD), were deformed but contained a lower number of low-angle grain boundaries than type B grains. Type B grains, which belonged to the γ fibre (<111>//ND), had numerous low-angle grain boundaries, and in many cases even high-angle grain boundaries, aligned at 35° to the rolling direction. These aligned boundaries are termed in the literature as in-grain shear bands, e.g. (5,7,17). The third type of grains, type C, had quite an equiaxed shape and in most cases had some substructure indicating that the grains were deformed. Further, they grains were of random orientations.

After the third pass at 1223 K, the microstructure was recovered. The cooling rate did not have an apparent effect on the microstructure and therefore it seems evident that marked recovery took place already during the deformation. However, since the average intercept length in the normal direction between low-angle grain boundaries, i.e. boundaries of the misorientation between 2° and 15°, was slightly higher after slow cooling regimen than after water quench (6.5 μm compared to 5 μm), it can be concluded that some static recovery took place during the slow cooling treatment. Small equiaxed grains, which resembled type C grains but had no substructure, were also detected at the original grain boundaries in both specimens deformed at 1223 K regardless of the cooling schedule. The grains have most likely nucleated statically during the unavoidable waiting time between the end of the deformation and the quench. Further, no in-grain shear bands were detected in the specimens deformed at 1223 K.

When the deformation temperature of the third pass was decreased to 1073 K, the microstructures gained features associated with cold rolling, i.e. no visible effect of DRV and a banded structure within the grains (Fig. 5(b)). The A and B grains were heavily pancaked but some small equiaxed type C grains were also detected (marked with C in Fig. 5(b)). Since the type C grains were deformed and had substructure, it can be concluded that they were formed statically during the cooling from 1223 K to 1073 K before the final deformation pass, and were deformed during the final pass. A slower cooling rate did not have any apparent effect on the microstructure after the final deformation pass at 1073 K, except for the fact that the in-grain shear bands, which were detected in type B grains, seemed to be more severe after water quenching compared slow cooling which is likely due to slight static recovery taking place during the slow cooling. Texture studies revealed that Type A grains had the orientation of {001}<110>, where as type B grains had orientations of {111}<112>, {111}<121> and {111}<231>.

Lowering the deformation temperature further from 1073 K to 923 K led to a significant increase in the number of low-angle grain boundaries, as seen in Fig. 5(c). Also, the grains were heavily flattened. However, the cooling path from 923 K to room temperature had no clear effect, which indicates that 923 K or below is too low a temperature for static recovery to occur effectively.

The three different types of grains were also present after

![Fig. 5. EBSD IPF-Z and grain boundary maps after the third pass at a) 1223 K, b) 1073 K c) 923 K followed by immediate water quenching. The white and yellow lines represent low-angle grain boundaries with misorientation of 2° – 5° and 5° – 15°, respectively. The black colored boundaries have misorientation greater than 15°. (Online version in color.)](image-url)
deformation at 923 K (marked with A, B and C in Fig. 5(c)). The number and direction of subgrain boundaries varied between type A and type B grains. Within the type B grains, numerous bands had formed in order to accommodate the high amount of deformation with few chances for because of the low deformation temperature and high strain rate. No visible effect of recovery on the microstructures was detected. The type C grains contained substructure and therefore their formation mechanism was also likely to be the same as with deformation at 1 073 K. Further, the orientations of type A, B and C were the same as after deformation at higher temperatures.

The $\varphi_2=45^\circ$ cross sections of the ODFs after the third pass are presented in the Fig. 6. It is evident that the temperature of the third pass and the following cooling path did not have any significant effect on the resultant textures. Regardless of the deformation temperature or the cooling rate, $\alpha$ and $\gamma$ fibres were present as was an additional maximum at the beginning of the $\varepsilon$ fibre. Only slight variations in the locations of the individual maxima along the texture fibres and in the texture sharpness were detected. The measured maximum texture intensities were relatively low varying between 4.26 and 9.25.

After the third pass at 1 223 K and water quenching, the highest intensity of 6.68 was measured along the $\gamma$ fibre at $\{111\}<121>$ and at $\{111\}<112>$. Other intensity maxima were detected along the $\alpha$ fibre and at beginning of the $\varepsilon$ fibre at $\{001\}<110>$. The slow cooling path weakened the texture and it decreased from 6.68 to 4.97.

When the deformation temperature of the third pass is lowered to 1 073 K, unexpectedly, the texture is weaker and more randomized than after deformation at 1 223 K and water quenching. The intensity maximum was reduced from 6.68 to 4.26. According to the SEM-EBSD examinations, the grains seemed more heavily deformed and pancaked at 1 073 K than at 1 223 K due to less DRV, and it could be assumed that this would result in stronger texture. The locations of the intensity maxima were, however, similar to those observed after deformation at 1 223 K. Also for at 1 073 K deformed specimens, the texture was sharper in the slowly cooled specimen (9.25) compared to the water quenched specimen (4.26).

After deformation at 923 K, the slow cooling path had a slight intensifying effect on the texture: the measured intensity maxima were 5.79 and 7.43 after water quenching and slow cooling, respectively. In both cases, however, the intensity maximum is detected at $\{001\}<110>$ in the $\varepsilon$ fibre, while the $\gamma$ fibre is relatively weak, and the second intensity maximum is along the $\alpha$ fibre.

### 3.4. The Effect of the Third Pass Temperature on the Dislocation Structures

Dislocation structures and possible precipitation were examined using TEM. Typical TEM bright field images of dislocation structures are presented in Fig. 7 and precipitation structure in Fig. 8. The highest deformation temperature during the third pass, 1 223 K, resulted in dynamically recovered dislocation structures, as seen in Fig. 7(a). The dislocation density was relatively low, although it varied between grains. The dislocation structures in the water quenched and slowly cooled specimen were similar. Because of the high stacking fault energy, the annihilation of dislocations is easy during the deformation and therefore the number of dislocations in dislocation tangles is quite low. In the slowly cooled specimen after deformation at 1 223 K numerous small precipitates were detected both in the matrix and at dislocations and grain boundaries, as shown in Fig. 8(a).

After the third pass at 1 073 K, the microstructure had undergone less DRV than at 1 223 K, as evident from the higher dislocation density (Fig. 7). Dislocation tangles adjacent to grain boundaries were detected, and also some subgrains had formed. The cooling stage had a minor effect on the dislocation structures, as seen in Figs. 7(c) and 7(d). Some subgrains were relatively free of dislocations, whereas some had them inside – this is the case especially after the fast cooling regimen (Fig. 7(c)).

Likewise as after deformation at 1 223 K, the most significant effect of the different cooling paths was on the precipitation. After slow cooling, precipitates had formed at grain boundaries and dislocations (Fig. 8(b)), but no precipitates were apparent in water-quenched specimens, indicating that the precipitation had occurred during the slow cooling stage from 923 K to room temperature.

The highest dislocation densities were found at the lowest deformation temperature of the third pass of 923 K. At this
temperature, DRV had not taken place, and numerous dislocations were in tangles forming cell walls (Figs. 7(e) and 7(f)). There was also a high number of dislocations inside the cells. This dislocation storage is in accordance with the high amount of work hardening observed in the flow stress curves (Fig. 1(c)). Similarly as with the higher third pass temperature, the slow cooling did not affect the dislocation structures but it did result in precipitation (Figs. 7(f) and 8(c)).

4. Discussion

4.1. Microstructure and Texture Evolution

The SRX, which took place during the inter-pass times, was not able to remove all recovered grains from the microstructure. The slowing down of SRX kinetics towards the completion of the recrystallization process in ferritic stainless steels has been previously reported by the present authors and also by Sinclair et al. At 1223 K, DRV is intense as evident from the flow stress curves in Fig. 1(a), and therefore it effectively reduces the stored energy and therefore also the driving force for SRX. Since few unrecrystallized grains were still present after the first and second inter-pass times, it is likely that recrystallization annealing after the third pass at the conventional hot deformation temperatures would not be able to remove all the deformed grains. The unrecrystallized grains are would have a detrimental effect on texture development and therefore also on the deep drawability of the final product.

Lowering of the deformation temperature of the third pass to 1073 and 923 K resulted in a higher flow stress and a higher amount of work hardening, as seen in Figs. 1(b) and 1(c), which indicates that higher dislocation densities were created during the deformation, i.e. the amount of stored energy was increased when the DRV was less effective at lower deformation temperature. This is likely to result a higher driving force and for SRX and accelerated SRX kinetics during annealing at a sufficiently high temperature. The increased dislocation storage was also confirmed in the TEM examinations (Fig. 7). It has been shown in a previous paper that lowering the deformation temperature increases the SRX kinetics significantly during the subsequent hot band annealing. After deformation at 1223 K at 1 s\(^{-1}\) to strain of 0.5 and recrystallization annealing at 1223 K for 30 s, the recrystallized fraction remained very low, ~5%. However, when the deformation temperature was lowered to 1073 K, 30 s recrystallization annealing at 1223 K led to almost 90% of recrystallization and lowering the deformation temperature further below 1023 K led to complete recrystallization during the 30 s annealing.

It was evident that lowering the deformation temperature of the third pass already to 1073 K would increase the driving force for SRX during the post hot rolling annealing. However, lowering the deformation temperature further to 923 K would have an even more significant effect on the SRX kinetics during post-hot rolling annealing as shown previously.

The temperature of the third pass and the cooling regimen had no significant effect on the resultant textures, as seen in Fig. 6, which is in accordance with previous results Sawatani et al. who concluded that the hot rolling temperature has little effect on the textures of hot rolled and annealed sheet – the difference will appear after cold rolling and subsequent annealing.

On the contrary to the study by Sawatani et al. and to the current results, Gao et al. observed that the deformation temperature indeed had a significant influence on the deformation textures, and further, the hot deformation tex-
tures affected the formation of texture structures during the later processing stages - especially the uniformity of the $\gamma$ fibre in the final cold rolled and annealed material was observed to be affected by the hot deformation temperature.\(^{11}\)

The uniformity of the $\gamma$ fibre texture determined the deep drawable of the final product, and a non-uniform $\gamma$ fibre was identified as especially harmful.\(^{11}\)

According to Gao et al.\(^{11}\) when the hot deformation temperature was decreased, the location of the intensity maximum along the $\alpha$ fibre was shifted from mainly $\{001\}<110>$ towards $\{111\}<110>$ because lattice rotations were increased, which promoted the development of texture towards stable end orientations. Due to oriented nucleation and preferred growth of certain orientations, the different deformation textures led to the formation of different recrystallization textures. Therefore, after hot deformation and hot band annealing, the warm rolled specimen displayed a texture maximum at $\{111\}<121>$ along the $\gamma$ fibre, where as conventionally hot rolled materials displayed intensity maximum along the $\alpha$ fibre at $\{001\}<110>$, $\{112\}<110>$ and $\{223\}<110>$.

In the current study, the intensity maximum was shifted towards $\{111\}<110>$ along the $\alpha$ fibre when the deformation temperature was decreased, but the differences in the deformation textures were much less significant as proposed by Gao et al.\(^{11}\) However, the slight shift of the locations of the intensity maxima might eventually lead to a positive texture development.

Nevertheless, the main issue with the current steels is likely to be the presence of $\{001\}<110>$ oriented grains, which were detected regardless of the deformation temperature and cooling rate. These $\{001\}<110>$ oriented grains have a pronounced tendency for recovery due to their low Taylor factor and they are therefore the last grains to recrystallize.\(^{15,16}\)

Removing the $\{001\}<110>$ oriented grains by recrystallization will be difficult at high conventional hot deformation temperatures due to intense dynamic and static recovery. The $\{001\}<110>$ orientation is known to turn into $\{111\}<121>$ during recrystallization; however, the transformation is slow.\(^{5}\) Therefore it is likely that the $\{001\}<110>$ orientation will be present after hot rolling and hot band annealing in the currently investigated materials and they will possibly be inherited by the final product, which would have a negative effect on the deep drawable.

4.2. The Effect of In-grain Shear Bands on Texture Evolution

In-grain shear bands have been found to form as a result of flow localization during warm deformation in IF and low-carbon steels e.g.\(^{5,7,17}\) and also in high-chromium ferritic stainless steels.\(^{9,10}\) The formation of in-grain shear bands during hot deformation is dependent on the deformation temperature and in-grain shear bands are only generated when the deformation temperature is lowered below the conventional hot rolling temperature range.\(^{9,17}\) Indeed, in the current study, in-grain shear bands were formed at 1 073 and 923 K, but not at the conventional hot rolling temperature of 1 223 K.

The in-grain shear bands are known to form in grains of high Taylor factor orientations,\(^{5,7,17}\) i.e. the in-grain shear bands appear in grains oriented with $\langle 111 \rangle$//ND, which was also the case in the current study (grains marked with B in Fig. 5). The in-grain shear bands cause grains to split up, which weakens the texture and also the matrix rotations are slower in the presence of in-grain shear bands.\(^{9,11,17}\) Zhang et al.\(^{18}\) argued that flow localization in the form of in-grain shear bands might prevent grain rotations into the stable and undesirable end orientations of $\{001\}<310>$. In-grain shear bands offer suitable nucleation sites for SRX and during annealing, numerous $\langle 111 \rangle$//ND i.e. $\gamma$ fibre grains, nucleate in the in-grain shear bands. Therefore the presence of in-grain shear bands increases the amount of $\{111\}$ recrystallization texture.\(^{9}\)

Since in-grain shear banding was observed in the current study when the deformation temperature of the third pass was lowered to 1 073 or 923 K, it is likely that recrystallized grains of $\gamma$ fibre would nucleate in grains with in-grain shear bands thereby promoting the formation of $\gamma$ fibre texture after hot band annealing and also in the final cold rolled and annealed sheet.

4.3. The Effect of Precipitation on Microstructure and Texture Evolution

The slow cooling from 923 K to room temperature did not cause any significant differences in the dislocation structures as seen in the TEM micrographs in Fig. 7. However, the slower cooling from 923 K to room temperature did allow precipitation to occur as seen in Fig. 8. The possible composition of these precipitates is discussed below.

Ti and Nb are added to stainless steels in order to stabilize the microstructure by preventing sensitization, i.e. precipitation of Cr containing $\text{M}_2\text{C}$ $\text{Co}$ carbides. However, they themselves cause precipitation in the form of various (TiNb)(CN) particles. Ti has a tendency to form coarse TiN and Ti(CN) particles already in the molten steel.\(^{21}\) And indeed, coarse TiN were also present in the currently investigated steel. Because of the stabilization, most of the C and N are removed from the matrix at the beginning of the hot deformation stage.\(^{22}\) However, there might be traces of C and N left in the matrix regardless of the stabilization. $\text{M}_2\text{C}$ ($\text{Fe}_3\text{Nb}_3\text{C}$) carbide precipitates at 1 173 K and coarsens rapidly in 19Cr–0.8Nb ferritic stainless steel.\(^{22}\) It is likely, though, that the amounts of C and N are too low for carbides or nitrides to precipitate during hot deformation.

Also intermetallic phases, such as Laves, Chi ($\chi$) and Sigma ($\sigma$), can precipitate in the course of hot rolling depending on the chemical composition of the steel. The formation temperature for these phases is generally considered to be below 1 123 K,\(^{23}\) however with the absence of Mo, as in the current steel, the precipitation would take several hours at 923 K.\(^{24}\) Therefore, the temperature – time window of the current study is too narrow for most of the intermetallic phases to precipitate.

The steel investigated here is also alloyed with 0.3 wt.% Cu, which can precipitate as $\epsilon$-Cu in ferritic stainless steel in the temperature range of 773 – 1 073 K.\(^{25}\) The time for $\epsilon$-Cu precipitation is relatively short; $\epsilon$-Cu particles or hardening caused by the precipitation has been detected within minutes of ageing in 16 – 17% Cr ferritic stainless steels albeit with 1.17 – 1.65% Cu.\(^{25,27}\) Since the present steel contains Cu, although less than in the antibacterial ferritic stainless steels, the precipitation of $\epsilon$-Cu may be possible at
undeformed steel, the precipitation occurred at both in dislocations and in the matrix. The phase equilibrium calculations using ThermoCalc for the present place at both dislocations and in the matrix. The phase TEM investigation (Fig. 8) revealed that precipitation took place at both dislocations and in the matrix. The phase equilibrium calculations using ThermoCalc for the present composition revealed (Fig. 9) that there is a driving force for ε-Cu precipitation at the temperatures concerned even though it only contains 0.3% Cu.

Precipitates, when present in the microstructure, affect recovery, recrystallization, and texture formation.\(^4\) Particles can have either accelerating or slowing effect on the SRX kinetics depending on their size. Large, non-deformable particles enhance nucleation of new grains during primary recrystallization by particle stimulated nucleation (PSN),\(^{4,25}\) which was detected in the present study around coarse TiN particles (pointed out with arrows in Fig. 2(c)). Small, closely spaced particles can influence moving boundaries by the Zener drag mechanism, which is generally observed as a longer incubation period for SRX giving more time for recovery and a lower driving force for SRX.\(^5\) In the case of Zener drag, the precipitates would also affect the formation of recrystallization texture since the effect of the Zener-drag depends on the type of the boundary and not all boundaries are affected in the same way.\(^5\) Because of the stabilization of the currently investigated steel, small Ti and Nb carbonitrides, which are present in the microstructure, are likely to have an effect on SRX kinetics and on the texture development during hot deformation through Zener drag. A solution annealing was carried out at a sufficiently high temperature for Ti and Nb carbonitrides to dissolve, however some precipitation might have occurred during hot deformation and annealing, which in turn would affect the texture formation during the later processing stages.

However, the effect of the precipitates formed during the slow cooling regimen on the texture formation and SRX kinetics during hot rolling stages is likely to be much less significant. If indeed these precipitates are ε-Cu particles, their solution temperature is below the hot deformation temperature and therefore they do not affect the dynamic restoration processes during hot deformation stage. However, it is likely that if similar precipitation takes place during cooling of an actual industrial coil, it will affect the restoration mechanisms and texture evolution during the later processing stages.

### 4.4. Optimization of the Final Pass Temperature

In the previous studies by Gao et al.,\(^{10,11}\) the effect of the deformation temperature on the texture structures was quite significant already in the deformed state. In the current study such effect was not detected, and therefore the significance of lowering the deformation temperature on positive texture development can be questioned. Although, in this study only the deformed state was studied and the effects might have appeared in a greater scale if the experiments would have been continued to later processing stages.

However, greater differences in the flow stress, microstructures and dislocation structures were observed. Microstructural features, especially in-grain shear bands, together with higher dislocation density achieved when the deformation temperature was lowered, would enhance the SRX kinetics and also affect beneficially the texture development. In-grain shear bands were detected in the specimens deformed at 1 073 to 923 K but they were completely absent after deformation at 1 223 K. The number of grains containing in-grain shear bands did not seem to increase significantly as the deformation temperature was lowered from 1 073 to 923 K. Therefore it can be concluded, that the number of γ fibre SRX grains would increase as well when the deformation is carried out below the conventional hot deformation temperatures as a result of nucleation of in-grain shear bands.

Based on these results, it is evident that lowering of the hot deformation finishing temperature is necessary in order to achieve favorable microstructures and also texture structures regarding the deep drawability of the end product. Lowering the finishing temperature to 1 073 K is likely to accelerate the SRX kinetics and have some positive effects on the texture formation, although, the effects would be greater if the temperature would be lowered to temperatures such as 923 K. On the other hand, 923 K might not be achievable in the industrial hot rolling mill because of issues relating to production technology. Therefore, it can be concluded that adequate results are achieved using finishing temperatures of 1 073 K and below, which are still achievable in the industrial scale.

### 5. Summary

The hot deformation behavior of a dual Ti–Nb stabilized 21%Cr ferritic stainless steel has been investigated using simulated multi-pass hot deformation conditions. The effects of the temperature of the final pass on the resultant flow stress, microstructure, texture, dislocation structures and precipitation were studied for different post-deformation cooling rates. The main results can be summarized as follows:

1. Almost complete static recrystallization took place during the 20 s inter-pass times at 1 223 K after pass strains

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Fig. 9. Equilibrium mass-fractions of various phases against temperature.
of 0.4–0.5. The recrystallization together with static recovery led to a complete softening of the material, so that flow stress remained at the same level in all passes.

(2) The static recrystallization during the inter-pass times had a randomizing effect on the texture, even though the conventional $\alpha$ and $\gamma$ fibres together with $\varepsilon$ fibre were always present.

(3) Lowering the temperature of the third i.e. final pass from 1 223 K to 1 073 K and further to 923 K led to a significant increase in the flow stress from 160 MPa to 325 MPa and further to 400 MPa.

(4) Lowering of the deformation temperature of the third pass to 1 073 and 923 K increased the number of sub-grain boundaries and caused the formation of in-grain shear bands in grains belonging to the $\gamma$ fibre.

(5) During slow cooling from 923 K precipitation of fine particles, possibly $\varepsilon$-Cu, occurred regardless of the deformation temperature of the third pass.

(6) Lowering of the deformation temperature of the third pass did not have any influence on the texture fibres present, though the locations of the individual intensity maxima in the fibres were affected.

(7) Lowering of the deformation temperature of the third pass did lead to an increase in the stored energy and therefore would result in accelerated static recrystallization kinetics during annealing subsequent to hot rolling.

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