Evolution of recrystallization texture in non-oriented electrical steels during final annealing – influence of shear stress after cold rolling

B Bacroix\textsuperscript{1}, J Schneider\textsuperscript{2}, A Franke\textsuperscript{3}

\textsuperscript{1}LSPM – CNRS, UPR3407, Université Paris 13, 99, Av. J.B. Clément, 93430 Villetaneuse, France.
\textsuperscript{2}Institute of Metal Forming, Technische Universität Bergakademie Freiberg, Bernhard-von Cotta-Str.4, D-09596 Freiberg, Germany.
\textsuperscript{3}Stahlzentrum Freiberg e.V., Leipziger Straße 34, 09599 Freiberg, Germany.

Abstract. The fabrication process of non-oriented electrical steels comprises hot rolling, annealing of the hot band (optional), cold rolling and final annealing. Starting from different hot band, quite distinct deformation substructures have been observed after cold rolling of ferritic FeSi steels. The recrystallization texture after annealing of the cold rolled material depends on the deformation structure obtained after cold rolling. In the present paper, a simple method to evaluate the energy stored during cold deformation of Fe-Si steels is presented. An estimate of the stored energy has then been made for various deformation modes (in-plane compression superimposed with shear stress). It is shown that, by adding a shear component, a more appropriate explanation of the experimental observed recrystallization textures may be obtained. The existence of shear components produces an inversion of the soft and hard orientations with respect to the case of a pure in-plane strain compression.

1. Introduction

It is now well established that the magnetic properties of the electrical steels strongly depend on their crystallographic texture and microstructure. The existence of a strong texture determines their magnetization behaviour (and especially at high values of applied external magnetic field) as well as the remanent induction value. Large values of grain size are also necessary to reach low values of the specific magnetic losses. These two features (strong texture and large grain size) are obtained in FeSi steels through a fabrication process, which comprises several steps: reheating of the slab, pre-rolling, final hot rolling, annealing of the hot band (optional), cold rolling and final annealing. The microstructure after cold rolling reflects mostly a complex deformation structure across the whole thickness, which is quite different from those of conventional steels. The ongoing structural changes during the final annealing step then strongly depend on the deformation structure obtained after heavy cold rolling. In this case, one usually observes in the very same sample, zones composed of deformation bands (which generally correspond to highly homogeneously deformed grains during cold rolling in plane strain compression) and zones between the deformation bands containing shear bands. Sometimes these shear bands are regions, where slip may be concentrated within one grain [1]. However, for large amounts of cold rolling, like for the regarded ferritic FeSi steels, the bands can cross several grains [1]. When hot rolling is supplemented by an additional heat treatment, either
immediately after hot rolling or as a separate processing step before cold rolling, the amount of zones with shear bands after cold rolling is higher than for a material subjected to hot and cold rolling without intermediate annealing stage [2]. As a high amount of zones with shear bands has also been observed to lead to the final development of the good magnetic texture and to enhanced grain growth [2], it is important to understand the influence of the shear bands on the nucleation and growth process at recrystallization during final annealing.

It is well recognized that the difference in the energy stored during plastic strain in various crystallographic orientations presenting different microstructures is one of the major factors affecting the behaviour of each orientation during recrystallization [3,4]. This stored energy (SE) strongly depends on the local strain mode and exhibit an orientation dependence [5]. It is generally assumed that cold rolling can be regarded as a plane strain compression deformation mode. However, if the deformation is considered to be homogeneous plane strain compression through the sample, one cannot explain the complete resulting figure for the texture for the regarded feritic FeSi steels [2,3]. Therefore, it becomes important to study the effect of shear bands in addition to in plane strain compression on the stored energy and its orientation dependence. In the present paper, we will describe the features of the microstructure after cold rolling by means of two typical examples. We will then calculate, with the simple Taylor model, the effect of adding one shear component to uniform plane stress for specific texture components, and, using the Taylor factor as estimate of stored energy, we will propose a possible scenario for active recrystallization mechanisms during final annealing and their effect of the final textures.

### 2. Microstructure after cold rolling

Figure 1 shows some typical orientations maps measured by EBSD on the section perpendicular to TD after cold rolling, on two FeSi 2.4 samples. Details of the fabrication of the hot rolled and cold rolled materials are given in [2,6]. Sample 1 was rapidly quenched immediately after hot rolling, whereas sample 2 experienced a thermal treatment (TT) immediately after the last hot rolling pass. These two hot bands with a thickness of 2mm were then cold rolled to the same final thickness of 0.50 mm.

It is noticed that in these steels (unlike conventional steels), a rather inhomogeneous microstructure containing deformation and shear bands is generally obtained after rolling. Regions with low IQ, associated with sheared grains, alternate with regions of high IQ, comprising the deformation bands. But, although the two samples have experienced the same amount of hot and cold rolling, the zones with shear bands are broader in sample 2, which may be due to grain growth during the additional TT before cold rolling. The EBSD map for sample 1 shows well-defined deformation bands with very small zones containing shear bands in between, whereas in sample 2, we see quite broad zones containing shear bands. Some deformation bands are also present, also thicker than in sample 1.

From the inspection of these EBSD figures and of the associated data file, we can conclude that for both samples, the orientations associated with deformation bands generally belong to the $\theta$ fiber (see Figure 2 for definition) with a maximum of rotated cube oriented grains (in red in Figure 1). The orientations associated with shear bands belong mainly to the $\alpha^*$ fiber (with a spread of colour between red and blue), and secondarily to the $\gamma$ fiber (in blue), essentially close to the $\{111\}<112>$ orientation. A high orientation spread is also present in these shear-banded zones, especially in sample 2, adding minor components to the description of the texture in these zones. We may now consider that the actual deformation of the deformation bands is closer to plane strain compression, while the sheared zones were additionally subjected to a shear component.

### 3. Estimation of the Taylor Factor for varying boundary conditions

In the case of different microstructures in differently oriented grains, we should choose, in order to evaluate the stored energy as a function of orientation, a model capable of predicting, from given macroscopically imposed boundary conditions, different states of stress and strain in each grain of a polycrystalline sample. For this purpose, we can use e.g. Crystal Plasticity Finite Element modelling, which can take into account the polycrystalline nature of the material and the initial texture, but which
does not treat satisfactorily the question of different grain sizes or of the fragmentation of grains due to the formation of dislocation structures at very large strains. This limitation reduces significantly their predictive capacity and renders the associated computing efforts useless.

Figure 1: EBSD orientations maps obtained for the two samples (left sample 1 and right sample 2) after cold rolling (CR). The color indicates the orientation of the normal direction of the sample, according to the color code indicated by the triangles.

We thus adopt here a more direct approach, based on the use of the simple Taylor model and the recognition that the influence of the orientation on the stored energy can be indirectly assessed from the calculation of the Taylor factor from the final orientations after cold rolling. It has indeed been shown that, for a given model, these two quantities (stored energy and Taylor factor) evolve in the same way with crystal orientation [5]. The Taylor model is based on the main hypothesis that each grain within a polycrystalline sample will be subjected to the very same strain rate tensor as the one macroscopically imposed. As a consequence, the strain rate tensor becomes totally imposed at the level of the grain, and thus, each grain can then be treated separately. It becomes then easy to study the influence of varying boundary conditions imposed to each grain on the response of differently oriented grains, and to reproduce in a simple way the fact that within the same sample, different grains experience different strain tensors (with more or less shear) according to their direct environment and orientation – dependent resistance.

If we assume that cold rolling does correspond to pure plane strain compression (PSC), the strain rate tensor \( \dot{\varepsilon}_g \) imposed to each grain labelled \( g \) can then be written in the macroscopic reference frame linked to the rolling process (RD, TD, ND)\(^1\) as:

\[
\dot{\varepsilon}_g = \dot{\varepsilon}_0 \begin{pmatrix} 1 & 0 & 0 \\ 0 & 0 & 0 \\ 0 & 0 & -1 \end{pmatrix}
\]

In this equation, \( \dot{\varepsilon}_0 \) is the macroscopic strain rate along RD. But, if an additional shear component is added to account for the observations of in-grain shear zones, this strain rate tensor becomes

\[
\dot{\varepsilon}_g = \dot{\varepsilon}_0 \begin{pmatrix} 1 & 0 & K \\ 0 & 0 & 0 \\ K & 0 & -1 \end{pmatrix}
\]

\(^1\) RD = Rolling Direction, TD = Transverse Direction and ND = Normal Direction (normal to the rolling plane).
with $K = \dot{\varepsilon}_{13}/\dot{\varepsilon}_{11}$, the ratio characterizing the relative amount of additional shear (imposed along the rolling direction in the rolling plane, as observed). Plastic strain within each grain then takes place by slip on several systems, each labelled by index $s$. If we adopt a rate-dependent formalism to describe plastic slip, a widely accepted phenomenological viscoplastic power law relates within each grain $g$ the shear rate $\dot{\gamma}_{g}^s$ on each slip system to the associated resolved shear stress $\tau_{g}^s$ as

$$\dot{\gamma}_{g}^s = \dot{\gamma}_{0}^s \left( \frac{\tau_{g}^s}{\tau_{0}^s} \right)^n$$

(3)

In this expression, $n$ characterizes the material rate sensitivity. The parameter $\dot{\gamma}_{0}^s$, usually called a reference shear rate, will be considered the same for all systems in all grains (i.e. $\dot{\gamma}_{0}^s = \dot{\gamma}_{0}$). The reference shear stress of slip system $s$ in grain $g$, $\tau_{0}^s$, normally evolves with strain [7]. Also, $\tau_{g}^s = \sigma_{g} \cdot R_{g}^s$ is the associated resolved shear stress on system $s$, $\sigma_{g}$ being the symmetrical deviatoric stress tensor derived from the stress tensor $\sigma_{g}$ and $R_{g}^s$ the orientation tensor of system $s$ in grain $g$. The components of this last tensor read

$$R_{g}^{s}_{ij} = \frac{1}{2} \left( n_{g}^{s} b_{g}^{s} + n_{g}^{s} b_{g}^{s} \right)$$

(4)

with $n_{g}^{s}$ and $b_{g}^{s}$ characterizing respectively the slip plane normal and the slip direction for system $s$ in grain $g$. The components of the strain rate tensor $\dot{\varepsilon}_{g}$ in grain $g$ are then expressed as:

$$\dot{\varepsilon}_{g}^{ij} = \sum_{s} \dot{\gamma}_{0}^{s} \left( \frac{\tau_{g}^{s}}{\tau_{0}^{s}} \right)^n R_{g}^{s}_{ij} = \sum_{s} \dot{\gamma}_{0}^{s} \left( \frac{\sigma_{g} \cdot R_{g}^{s}}{\tau_{0}^{s}} \right)^n R_{g}^{s}_{ij}$$

This equation represents a unique relationship between strain rate and deviatoric stress tensors at the level of grain $g$, which allows to determine both tensors $\dot{\varepsilon}_{g}$ and $S_{g}$ from the boundary conditions imposed to the considered grain. In this framework, the Taylor factor is then defined as

$$M_{g} = \frac{\dot{\varepsilon}_{g} S_{g}}{||\dot{\varepsilon}_{g}|| \tau_{0}^{s}} = \frac{\dot{W}_{g}}{||\dot{\varepsilon}_{g}|| \tau_{0}^{s}} = \frac{\sum_{s} \dot{\gamma}_{0}^{s} \left( \frac{\sigma_{g} \cdot R_{g}^{s}}{\tau_{0}^{s}} \right)^n S_{g} \cdot R_{g}^{s}}{||\dot{\varepsilon}_{g}|| \tau_{0}^{s}}$$

(6)

In other words, this factor is equal to the plastic work rate $\dot{W}_{g}$ divided by a normalization strain rate $||\dot{\varepsilon}_{g}||$ and by a normalization reference stress $|\tau_{0}^{s}|$ (taken here as the average on all systems of the references shear stresses $\tau_{0}^{s}$ reached after severe cold rolling). This normalization renders $M_{g}$ dimensionless, which allows the influence of orientation on the capacity of deforming easily or not to be assessed, for a given model and given boundary conditions. Thus, when it is calculated at the end of the cold rolling process, it does not depend much on the strain path, on the actual strain rate or strain tensor or on the hardening state of the orientation [5].

In order to calculate it, two families of slip systems are selected to account for the usually mentioned slip capacities of carbon steels, the 12 \{110\} <111> and 12 \{112\} <111> systems and value of the initial (annealed state) reference stress $\tau_{0}^{s}$ is taken equal for the 24 systems in all grains, and as its evolution with strain is neglected, this stress simply becomes

$$|\tau_{0}^{s}| = \tau_{0}$$

(7)

\(^2\) By using only odd values of the exponent $n$, we avoid in the formulation the use of absolute values. Thus, the more classically chosen value of 20 will be replaced here by 21.
The reference strain rate $|\dot{e}_g|$ is taken here equal to the von Mises equivalent strain rate

$$|\dot{e}_g| = \dot{e}_{VM} = \frac{2}{3} \sum \dot{e}_{ij} \dot{e}_{ij} = \dot{e}_0 \sqrt{\frac{4}{3} (1 + K^2)}$$

Then, by taking $\dot{e}_{VM} = \tau_0 = 1$ and $n = 21$, the Taylor factor has been calculated for a discrete set of orientations describing the $\phi_2 = 45^\circ$ section of the Euler space, which contains all the main orientations composing the rolling and annealing textures. Some typical values are presented in Figure 2, in the $\phi_2 = 45^\circ$ section, which contains most of the information related to the rolling and annealing textures.

![Figure 2](image)

Figure 2. Predicted $M_g$ values in the $\phi_2 = 45^\circ$ section for (a) $K = 0$ (plane strain compression), (b) $K = 1$, (c) $K = 5$ and (d) principal orientation fibers of interest: $\alpha = \{hkl\}<110>$, $\alpha^* = \{h,1,1\}<1/h,1,2>$, $\gamma = \{111\}<uvw>$, $\theta = \{100\}<uvw>$.

In the case of non-oriented electrical steels, a low intensity of the $\alpha = \{hkl\}<110>$ and $\gamma = \{111\}<uvw>$ fibers and a high intensity of the $\theta = \{100\}<uvw>$ fibre texture (and especially the Cube or Rotated Cube components) are desirable. Additionally, a high intensity of the $\alpha^* = \{h,1,1\}<1/h,1,2>$ fiber is also beneficial for a better magnetization behaviour [8].

When adding more and more shear to plane strain compression, there is a shift of the position of the maximal and minimal values in the Euler Space. For plane strain compression ($K = 0$), the orientations associated with a maximal value of the Taylor factor, i.e. the “hard” orientations, are the $\{110\}<110>$ orientation (Euler angles (0,90,45)), which is generally absent from the rolling or
annealing textures of steels, and also orientations located along the $\gamma$ fiber. The softest components are the rotated Cube (B) and Goss (C) which are not always present in the deformation textures. When $K \geq 5$, there is a complete inversion, since the rotated Cube (B) and Goss (C) orientations become the hardest, whereas the $\{110\}<110>$ orientation and the $\gamma$ fiber become the softest components.

Between these two extreme cases, if we consider only the components of interest for magnetic properties, it is clear from Figure 3, which represents the evolution of $M_g$ with $K$ along the fibers of interest, that there is a shift of hard and soft orientations for $K$ between 0.5 and 1. It is interesting to note, that although the maximal value of $M_g$ is reached for the Cube orientations for $K > 1$, the components of the $\alpha^*$ fiber are indeed very close.

Now, if we accept that there is a link between Taylor factor and energy stored during cold deformation, it may be hypothesized that recrystallization occurs by nucleation and growth after deformation in the orientations associated with a high Taylor factor [4,5] and by Strain Induced Boundary Migration (SIBM) in the orientations associated with a low Taylor factor [4]. Of course, this simple reasoning does not allow to predict which one will be more active when two possible mechanisms are simultaneously expected. The relevance of this hypothesis can now be investigated by taking a closer look at the textures obtained after deformation and annealing and more particularly at the components with high shear.

4. Discussion
We thus examine some experimental data obtained after cold rolling and final annealing for the two FeSi2.4 samples, and especially the recrystallization texture at the very first steps of the annealing process, i.e. before extensive grain growth appears. The ODFs calculated from EBSD measurements performed on maps of 500 * 500 µm (representing the total thickness of the samples) after cold rolling and annealing at two different temperatures during 20s are presented in Figure 4 for the two samples.

Before checking the validity of our hypothesis on these data, it is worth recalling that the CR textures (although associated with different microstructures) measured before annealing were quite similar for both samples and mainly composed of two strong fibers $\alpha = \{hkl\}<110>$ and $\gamma = \{111\}<uvw>$ quite classical for these steels [2]. The $\alpha^*$ components found within the shear bands can be considered as part of the spread around these fibers. It has also been shown that the evolution of the microstructures during recrystallization was quite different within the deformation bands (in which recovery was seen first) and within the sheared zones (in which nucleation was much more active than in deformation bands) [2]. This has been shown clearly for sample 2, in which the shear zones are large and the process of nucleation more active. On the contrary, due to the small size of these shear bands in sample 1, nuclei were not visible after annealing at 700°C during 20s. For annealing at 760°C during 20s, new grains were visible in both samples, even in deformation bands in which these new grains can be produced by recovery and growth of subgrains. These already published observations, as
well as the textures presented in Figure 4 are indeed consistent with the expected mechanisms listed in Table 1.

![Figure 4](image)

**Figure 4.** ODF φ₂=45° section for samples 1 and 2 after annealing at 700°C for 20s (top) and at 760°C for 20s (bottom); (total cross section).

**Table 1.** Expected favoured recrystallization mechanisms in different zones of the CR samples 1 and 2, as deduced from the Taylor factor calculations.

| Sample | 1 | 2 |
|--------|---|---|
| Deformation bands (K = 0) | Rot. Cube (Low M₉) = > SIBM | Rot. Cube (Low M₉) = > SIBM |
| Sheared zones (K ≳ 1) | γ fiber (Low M₉) = > SIBM | (1) α* fiber (High M₉) = > Nucleation |
| | | (2) γ fiber (Low M₉) = > SIBM |

The observed evolution of texture for samples 1 and 2 after annealing at 700°C for 20s and at 760°C for 20s (see figure 4) can be summarized as follows:

- for annealing at 700°C, we observe a limited texture evolution in sample 1 with an increase of the Rotated Cube component. This can be explained by SIBM around this component during annealing for 20s. SIBM, also expected in the small {111} shear zones, will necessarily imply smaller changes of the ODF value due to the spread of orientations along the fibre. Nucleation, hardly visible in sample 1, could be completely absent at this stage, as deduced from our analysis.

- For sample 2, we observe an increase of the rotated Cube component which could again be due to SIBM in deformation bands, whereas nucleation of the α* components within the shear bands could be responsible for the increased spread of orientations around the γ and α* fibers.
• for annealing at 760°C, nucleation of minor α’ components within the sheared zones occurs in sample 1. The texture observed in sample 2 confirms the growth of the α’ nuclei.

• the Goss component becomes also visible within the texture of both samples, especially after annealing at 760°C. This could also be due to SIBM around this minor component, as already observed in some other non-oriented Fe-Si steels [4].

It is worth noting that the annealing conditions (annealing temperature and time) may also have a significant effect on the occurrence of both mechanisms (nucleation, SIBM). This is not considered in this case. The present analysis has also neglected the effect of temperature and strain gradients during processing, which also play an important role on the developing microstructures. Especially, additional shear is generally more present within the surface region of the sample. Fluctuations of Taylor factor across the thickness of the samples should also be considered to interpret more precisely the recrystallization processes. This will be addressed in a forthcoming paper.

5. Conclusions and perspectives
The present paper has presented a simple method to evaluate the distribution of stored energy due to cold deformation in cold rolled FeSi steel. Starting from different hot bands, quite distinct rather inhomogeneous deformation substructures have been observed after cold rolling. An estimate of the stored energy has then been made for various deformation modes (in-plane compression superimposed with shear stress). It has been shown that adding a shear component produces an inversion of the soft and hard orientations with respect to plane strain compression. By considering that the deformations bands were deformed by in-plane strain compression while the sheared zones were subjected to an additional shear component, it has been possible to explain the texture evolution at low annealing temperatures, where recrystallization is dominant and grain growth has not started. Recrystallization seems to occur by different processes: SIBM associated with low values of $M$ within the deformation bands (subjected to PSC) and nucleation associated with high values of $M$, within the shear bands (subjected to PSC + Shear). Further research work is then desirable to clarify the mechanisms and driving forces during normal grain growth in these ferritic FeSi Steels and especially the interplay between recrystallization and grain growth, which appears across the thickness of the sheets.

References
[1] W B Hutchinson, Deformation Substructures and Recrystallisation. Materials Science Forum, 2007. 558-559: p. 13-22.
[2] A Franke, J Schneider and R Kawalla. Evolution of Texture at Final Annealing by Recrystallization and Grain Growth in Non-Oriented Ferritic FeSi. in WMM18, 8th International Conference on agentism and Metallurgy. 2018. Dresden, Germany.
[3] J J Sidor, K Verbeken, E Gomes, J Schneider, P R Calvillo and L A I Kestens, Through process texture evolution and magnetic properties of high Si non-oriented electrical steels. Materials Characterization, 2012. 71: p. 49-57.
[4] F Gregori, K Murakami and B Bacroix, The influence of microstructural features of individual grains on texture formation by strain-induced boundary migration in non-oriented electrical steels. Journal of Materials Science, 2014. 49(4): p. 1764-1775.
[5] B Bacroix, A Miroux and O Castelnau, Simulation of the orientation dependence of stored energy during rolling deformation of low carbon steels. Modelling and Simulation in Materials Science and Engineering, 1999. 7(5): p. 851-864.
[6] J Schneider, A Stöcker, R Kawalla and A Franke, Evolution of Optimum Grain Size for Low Loss Ferritic FeSi Steels. Steel Research International, 2017. 87.
[7] B Bacroix and R Brenner, A phenomenological anisotropic description for dislocation storage and recovery processes in fcc crystals. Computational Materials Science, 2012. 54: p. 97-100.
[8] J Schneider, A Franke, A Stöcker and R Kawalla. Quantification of the Effect of Texture on the Magnetization Behavior. in Conference Magnetic Measurements. 2017. Prague.