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Recrystallized {3 1 1}⟨1 3 6⟩ orientation in ferrite steels

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The role of deformed ⟨1 1 0⟩/RD orientations for the nucleation of the {3 1 1}⟨1 3 6⟩ component observed in recrystallization textures of ferrite steels is investigated. For the purpose of characterizing the plastic behaviour of ⟨1 1 0⟩//RD grains, a 60% cross-rolled IF steel was examined by high-resolution orientation scanning microscopy. The minute but relevant amounts of crystallite volume that rotated towards the {3 1 1}⟨1 3 6⟩ component in the deformed ⟨1 1 0⟩//RD grains revealed the occurrence of a fragmentation process that produced locally potent nucleation sites of recrystallization.

Keywords: {3 1 1}⟨1 3 6⟩ orientation; Recrystallization texture; Nucleation mechanism; Ferrite steel

Current developments in materials science technology aim to contribute to a more sustainable management of finite resources through the reduction of production costs and energy consumption, in compliance with environmental regulation policy. An increased understanding of the fundamental mechanisms through which the microstructure has formed provides a good strategy to achieve excellence in materials manufacturing. In line with that, the development of crystallographic textures has been the subject of research for many decades, motivated by the critical role texture plays in the mechanical and functional anisotropy of materials. In the particular case of low-carbon steels for deep-drawing applications, significant success in terms of product development has been achieved after intensive investigation of the effects of chemistry and the processing conditions [1,2]. This has resulted in the development of steels with an intense ⟨1 1 1⟩//ND fibre texture, which is of crucial importance for deep-drawing properties [1–4]. After annealing of severely rolled low-carbon steel, the {3 1 1}⟨1 3 6⟩ recrystallization texture appears [5–7]. The origin of this component is not yet known. Quadir and Duggan [8] observed the {4 1 1}⟨1 4 8⟩ component (7° away from the {3 1 1}⟨1 3 6⟩ component) in the annealing texture of a 95% cold rolled IF steel with a cold rolling texture dominated by the {0 0 1}⟨1 1 0⟩ component. They associated the appearance of the {4 1 1}⟨1 4 8⟩ component with a deformation banding process during the cold rolling. Gobernado et al. [9] reported that, after cold rolling reductions larger than conventional practice (>95%), a deformation texture characterized by a strong rotated cube component is formed in an Fe–3% Si alloy. After annealing, a recrystallization texture appeared, with an intensity maximum at the {3 1 1}⟨1 3 6⟩ orientation (in Euler angles corresponding to $\varphi_1 \approx 20$, $\Phi \approx 25$ and $\varphi_2 = 45°$). A similar behaviour was observed in IF steel deformed by accumulated roll bonding [10]. After a reduction of 99.90%, the material showed a texture dominated by a sharp rotated cube orientation (of >30× random levels), which, after the annealing treatment, transformed to the {3 1 1}⟨1 3 6⟩ orientation. In both examples the {3 1 1}⟨1 3 6⟩ orientation was accompanied with the {4 1 1}⟨1 4 8⟩ orientation. These orientations are part of the {h 1 1}⟨1/h, 1, 2⟩ fibre reported by Homma et al. [11], which in the $\varphi_2 = 45°$ section visually appears parallel to the well-known α-fibre of the body-centred cubic (bcc) cold rolling texture. Hereinafter, the range of orientations {3 1 1}⟨1 3 6⟩–{4 1 1}⟨1 4 8⟩ will be referred to as the representative component (R component) of the recrystallization texture of severely rolled bcc metals.

Verbeken et al. [12] claimed to have discovered the selective growth origin of the R component. In their work, the deformation texture of an ultra-low-carbon steel cold rolled to 95% reduction was dominated by the {2 1 1}⟨1 1 0⟩ orientation. After annealing, the

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recrystallization texture was characterized by two intensity maxima, one at the R component and another at the $\{5\,5\,4\}\langle 2\,2\,5\rangle$ orientation. Based on the $\langle 1\,1\,0\rangle 26.5°$ misorientation between the deformed $\{2\,1\,1\}\langle 1\,1\,0\rangle$ and both recrystallization maxima, they concluded that the enhanced boundary mobility associated with the $\langle 1\,1\,0\rangle$ 26.5° (Σ19a CSL boundary) was responsible for the recrystallization texture development in this case. A similar conclusion was obtained by Nakamura and Homma [14], who investigated the plastic behaviour of a $\{2\,1\,1\}\langle 1\,1\,0\rangle$-oriented single crystal after 70% cold reduction and subsequent recrystallization. Alternatively, Homma and co-workers [11] associated the appearance of the R component with an oriented nucleation mechanism. In their work, the R component could be identified in the vicinity of the grain boundary of a $\{1\,0\,0\}\langle 0\,0\,1\rangle$ bicrystal after 70% cold rolling. A minority $\{3\,1\,1\}\langle 1\,3\,6\rangle$ component was also observed by Hayakawa and Kurosawa [13] in the fully annealed textures of a sample that was cross-rolled to a reduction of 77%. It is the purpose of the present work to investigate the plastic instability associated to $\langle 1\,1\,0\rangle$//RD orientations in relation with the appearance of the R component after recrystallization.

A Ti-stabilized IF steel with composition 0.002% C, 0.204% Mn, 0.030% Si, 0.015% P, 0.010% S, 0.004% N, 0.045% Ti (wt.%) was hot rolled to 70% reduction and slowly cooled to room temperature to simulate the slow cooling of the industrial coiling process. For the purpose of ensuring a significant fraction of rotated cube grains in the cold deformed metal, the material was cold deformed to a total reduction of 60% by cross-rolling. In the cross-rolling experiment the rolling direction (RD) and transverse direction (TD) are alternatively changed after each pass in 12 subsequent rolling passes (with an average reduction per pass of 7%). For the representation of the texture results the x-axis of the sample reference system coincides with the rolling direction of the hot band. The macrotexture of the cold worked specimen was gathered on the rolling plane (RD–TD section) by conventional X-ray diffraction and the orientation distribution function (ODF) was calculated with the software developed by Van Houtte [15]. The partial recrystallization annealing was performed in a salt bath at 700 °C with a holding time of 3 min. Extensive orientation imaging microscopy was performed on the deformed material and the partially recrystallized material. Textural data were acquired with a FEI Nova 600 dual-beam microscope coupled with an EDAX-TSL® electron backscatter diffraction facility.

The macrotexture of the cross-rolled sample in the $\varphi_2 = 45°$ section is shown in Figure 1a. The latter shows a relatively weak $\langle 1\,1\,1\rangle$//ND fibre (3× random) and an intensity maximum (5.6× random) located at the rotated cube ($\{0\,0\,1\}\langle 1\,1\,0\rangle$) orientation. Figure 1b shows the inverse pole figure (IPF) map of the longitudinal section, i.e. the section perpendicular to the TD of hot rolling. The texture corresponding to the orientation scan shows a maximum at the rotated cube orientation and a certain spread of intensity from the rotated cube towards the R component, as indicated by the arrow in Figure 1c. The term "upper α-fibre orientations" is introduced to account for the set of $\langle 1\,1\,0\rangle$//RD orientations with

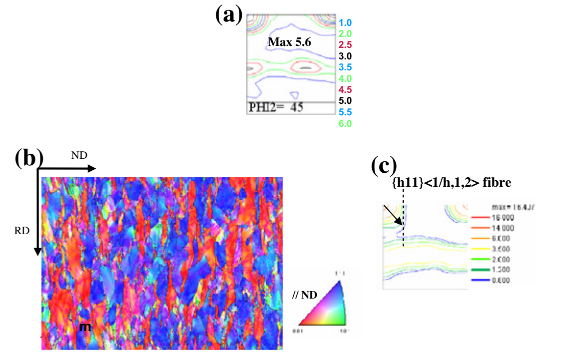

Figure 1. Textural data for IF steel 60% cross-rolled. Macrotexture (XRD) (a), longitudinal section inverse pole figure (IPF) map (b) and local texture corresponding to the scanned area (c).

$0 < \Phi < 20°$. The development of strain heterogeneities, related to the sub-structure development within the grains as a response to the plastic deformation, is revealed by the non-uniform colour in the IPF map. In the present context, the question arises whether or not such local strain heterogeneities bear any connection with the presence of R component orientations. The deformed grains in Figure 1b are qualified according to their orientation as upper-α-grains (with a maximum at the rotated cube orientation) and γ-fibre-oriented grains (cf. Fig. 2a and b, respectively). The distribution of R orientations is evaluated separately in each partition by the crystal orientation map (overlaid on the image quality (IQ) map). The volume fraction of the R component calculated with a tolerance of 15° with respect to the ideal orientations is highlighted on the IQ map and calculated by the EDAX-TSL® software. The textural analysis reveals the presence of the R component preferentially in upper-α-fibre grains. When a partition is considered of α-fibre grains, approximately 35% of these grains exhibit a fragmented substructure, which contains some R component volume fraction (within a tolerance angle of 15°). The volume fraction of R component orientations (Fig. 2a) is found to be 25% of the partition data points belonging to the α-fibre. In contrast, the R component is entirely absent in deformed grains of the γ-fibre type.

For the purpose of investigating the plastic instability in upper α-fibre grains, a representative pair of grains

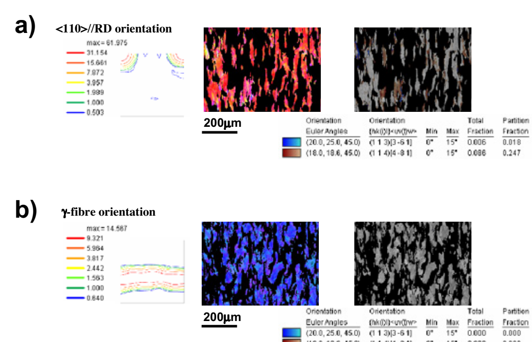

Figure 2. Deformed grains grouped according to orientation: $\langle 1\,1\,0\rangle$//RD grains (a) and $\langle 1\,1\,1\rangle$//ND fibre grains (b). The crystallographic texture is represented in the $\varphi_2 = 45°$ section and in the IPF map. The volume fraction of R component orientations (tolerance 15°) is highlighted on the IQ map.