Texture Observation and Control in Metal Manufacturing: Theory and Practice

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Abstract. It is generally known that the crystallographic texture of metal alloys is formed during subsequent solid-state transformations, occurring during manufacturing of the product. Applied to the specific case of metal sheet manufacturing, it implies that the texture of the finished product is the result of the crystallographic phenomena involved in hot rolling, cold rolling and annealing. Austenite-to-ferrite phase transformation, plastic deformation, recrystallization and grain growth are the dominant solid-state transformations occurring during thermo-mechanical processing of steel sheet. Moreover, the crystallographic texture ensuing of a solid-state transformation is a characteristic feature of the physical mechanics underlying this transformation and therefore, a precise understanding of the texture formation mechanisms is an important tool in gaining a better understanding of metal physics phenomena. The present paper will give a brief overview of the current state-of-the-art in these issues.

1. Introduction

The crystallographic texture is the microstructural state variable pertaining to the statistical distribution of crystallographic orientations in a polycrystalline material. Though an early monograph on crystallographic textures of metal alloys was published already in 1939 by Wassermann (2nd edition by Wassermann and Grewen [1]), a real boost to the field was given with the publication of the seminal work by Bunge in 1969 (Mathematische Methoden der Texturanalyse [2]). For a more detailed description of the early development of texture research the reader is referred to Bunge [3].

The research carried out within the texture community can be broadly divided in three different categories: (i) research on texture dependent process-structure-property relations for structural and functional materials, (ii) crystallographic aspects of material science phenomena and (iii) crystallographic phase and orientation contrast for microstructural characterization. The present paper will present a brief and kaleidoscopic overview of recent development of the versatile aspects of texture research.
2. Textures in relation to process, microstructure and properties

2.1. Thermomechanical processing of metal sheet through rolling and annealing

It is well understood that texture plays a dominant role for a number of metal sheet properties. Well-known examples are the Lankford R-value (i.e. the so called normal and planar anisotropy) [4], the magnetic properties of electrical steels [5], the earing behavior of can stock aluminum, the roping/ridging behavior of aluminum [6]. This has motivated the detailed study of texture formation during metal sheet manufacturing for a wide variety of alloys. During slab casting, hot/cold rolling and annealing the texture of the finished product is formed and in each processing step the texture inherited from the previous step may be modified. E.g. during cold rolling a texture is developed that is determined by the plastic velocity field that is induced by a specific rolling configuration. When the sheet is recrystallization annealed afterwards, the cold rolling textures will be transferred to a recrystallization texture.

In general, textures induced by plastic deformation depend on the crystallographic structure of the material (which in turn determines the potential slip or twinning systems), on the specific velocity field of the applied plastic deformation and on the initial texture prior to plastic deformation, but to a lesser extent on the precise chemical composition of the alloy. It is quite remarkable that with the rather coarse and approximate assumptions of the Full Constraint Taylor (FCT) model, ignoring phenomena such as strain hardening or strain localization, the cold deformation texture of single phase alloys can be predicted with reasonable accuracy for low and medium strains if the displacement field and the potentially active slip and twinning systems are known.

The reason for the success of the FCT model, and its more advanced successors, is because of the fact that one single texture component (i.e. a crystal orientation) is distributed over many physical grains of the polycrystalline aggregate. Although the local behavior of each individual grain can be very deviatoric with respect to the theoretical prediction, it is apparently true that these local deviations cancel out in the mean field description of the average crystal rotations. The validity of a homogenized mean-field description during cold rolling of a 6xxx aluminum alloy was recently confirmed in a near-insitu experiment reported by Pirgazi and Kestens [7].

Unfortunately the same level of accuracy cannot be obtained for the prediction of recrystallization textures. Despite decades of intense metallurgical research there are as yet no models available that are able to simulate the texture after recrystallization annealing with sufficient precision so as to reliably predict the texture dependent properties mentioned above (assuming there is an accurate property state model available that allows to express the property under consideration as a function of the texture). One of the reasons is that specifically the nucleation stage during recrystallization (i.e. the initial stage of subgrain growth, coalescence or bulging [8] critically depends on the detailed local features of the deformation structure for which the statistical methods of crystal plasticity models do not have sufficient resolution to provide an accurate description. Hence, advanced recrystallization models need to be supplied with detailed experimental data of the deformation structure (e.g. by EBSD or TKD measurements), which are far from trivial to obtain and which represent an approach that does not allow for a through-process simulation. Perhaps a viable alternative is provided by the full-field crystal plasticity finite element models (CPFEM) [9], which, in principle, are capable of reproducing the detailed features of the deformation microstructure. It remains to be verified, though, whether the microstructural resolution and accuracy of such models are enough to allow for an accurate simulation of thermally activated recovery and recrystallization models.

2.2. Texture formation in non-conventional processes: Additive Manufacturing

Additive manufacturing (AM) based techniques for production of near-net-shape metallic materials have been recently in the centre of attention of many researchers in the fields of materials science and
mechanical engineering. AM processes uses computer controlled high energy sources, which layer by layer melts or deposits the metallic powder/feeding wire and produces fully dense components. Texture evolution in the AM processes is based on the directional solidification of metal in the melt-pools with very high cooling rates. Despite the texture study of the rolled sheets in which the crystallographic orientations are considered based on the sample reference frame composed of three reference directions (i.e. RD, TD and ND), crystallographic orientations in the AM products are studied with respect to only one sample reference direction, i.e. the building direction (BD).

Texture of the as-built AM products is a result of (i) the directional solidification in which the preferred solidification direction (i.e. <001> in BCC and FCC crystals [10]) is parallel to the direction of heat flow (i.e. perpendicular to the melt-pool boundaries), and (ii) of the processing parameters and the occurrence of solid state phase transformation during cooling cycles. For example, for the alloys fabricated by selective laser melting (SLM) process, it has been reported that variation of the laser scanning parameters (e.g. laser power, scanning speed) and/or the hatching strategy can change the melt-pool geometry and the resulting crystallographic texture [11-14]. Consequently, different fibre textures such as <001>||BD [15;16], <011>||BD [12] and <111>||BD [17;18] have been observed in BCC and FCC alloys processed by the SLM process.

The occurrence of solid state phase transformation (e.g. FCC to BCC in steels) in alloys with allotropic crystal structures will change the solidification texture in the AM products. Dominant texture components after phase transformation is expected to follow the governing orientation relationship between the two phases. Furthermore, the fast cooling rate of the AM processes results in the formation of variants with equal statistical probability which eventually reduces the overall texture intensity in alloys featuring phase transformation [19;20].

3. Microstructural characterization and data processing based on electron diffraction contrast

In the last decades of the 20th century Electron Back Scatter Diffraction (EBSD) has developed into a powerful and versatile microstructural characterization technique with an optimum combination of reasonable resolution (~30-50 nm in a SEM with Field Emission Gun), large field width (~mm²) and angular resolution of ~1°. The texture community has largely contributed to the development of this technique. Perhaps the most interesting quality of orientation contrast imaging based on EBSD is the fact that each image represents a huge set of crystallographic data that is ideally suitable for post-processing. Given the fact that artificial intelligence and machine learning algorithms will play an increasing role in the future, it appears that EBSD images are particularly suited for this type of quantitative analysis.

EBSD is currently a mature and routine tool employed for microstructural characterization both in industrial and academic research. The present section highlights 2 recent developments in the field of orientation contrast microscopy: 3D-EBSD and Transmission Kikuchi Diffraction. As EBSD allows for concurrently gathering orientation and microstructural information, it renders EBSD maps suitable to provide information on the multi-variate Grain Size Orientation Distribution Functions, which will be introduced below.

3.1. 3D-EBSD by slice and viewing

For decades, conventional microstructural characterization techniques such as electron backscattering diffraction (EBSD), scanning electron microscopy (SEM) and optical microscopy have been extensively used to uncover microstructural details in different length scales. Among these, EBSD has proven its
potential in providing high-resolution digital images in which each pixel contains microstructural and crystallographic information. With the post-processing of these digital images, one can extract tremendous amount of information, regarding for example grain size and distribution, local crystallographic orientation and texture, crystallographic and morphological phases, grain boundary type, etc. The capability of the above-mentioned techniques is limited to the observation from planar surfaces (i.e. two dimensional (2D)). The microstructures, however, are of three-dimensional (3D) nature and a full characterization of many features such as e.g. grain boundary planes and true shape of microstructural constituents demands the utilization of a 3D based characterization technique [21;22]. With the latest progress in equipment design and computational speed, various techniques for 3D characterization of microstructures have been developed. Atom probe tomography [23;24], X-ray tomography [25;26], High-energy X-ray diffraction microscopy (HEDM) [27;28] and 3D-EBSD are the commonly used techniques to uncover the 3D nature of microstructural features at different length scales.

Among these techniques, 3D-EBSD has gotten special attention because on one hand it is an adaptation of the conventional 2D EBSD, and on the other hand it features a balance between high-resolution and large field of observation which is of great interest for characterization of metallic microstructures. 3D-EBSD is the combination of conventional EBSD and tomography based on serial sectioning. The usual method for serial sectioning involves cyclic removal of parallel layers of the sample and then imaging of the planar sections by e.g. the EBSD technique (cf. Fig. 1a) [29;30]. The removal of the material for serial sectioning can be performed with different methods: e.g., mechanical polishing, chemical polishing, Focused Ion Beam (FIB) tomography, Broad Ion Beam tomography (BIB) and laser or electrical discharge ablation [31]. Among these method, FIB tomography and manual serial sectioning are the most frequently used techniques, which have been extensively used to generate 3D micrographs of different metals.

When the serial sectioning experiment is completed and the stack of 2D data are created the data need to be processed. The main step of serial sectioning data processing is data alignment. Data
alignment means to ensure that the spatial registry of each 2D data file is accurate. This step is common for most tomographic methods; however, manually performed experiments usually need more alignment, which must be done before further data processing [32;33]. In the alignment step, the internal features of the microstructure are used for the precise alignment of the collected 3D stack. One generally accepted approach proposed by Lee et al. [34] and further developed by Pirgazi et al. [21;32] consists of a cross correlation procedure based on the minimization of the average misorientation between two adjacent sections. It has been shown that the full alignment procedure produces a more reliable grain boundary plane morphology and continuous orientation gradients inside the grains of a single phase nickel sample with large grains (cf. Figure 1b and c).

The reconstructed grain boundary network extracted from the 3D microstructural volumes have been frequently used to study the five parameter grain boundary character distribution (e.g. [35;36]). Figure 1d represents the distribution of grain boundary plane normal for grain boundaries carrying a S3 crystal misorientation in a polycrystalline nickel specimen. The strong peak at the (111) pole, is an indication of the presence of coherent twin boundaries. This dataset was reconstructed based on 50 EBSD sections spaced by a 7 mm distance to obtain a total volume of 3000x3000x350 mm³ comprising 470 grains.

3.2. Transmission Kikuchi Diffraction
3.2.1 Architecture of the TKD systems and working principle. The principle setup of the TKD system developed at Ghent University is shown in Figure 2a, whereas Figure 2b shows a picture taken inside the Quanta 450 SEM chamber with the sample in TKD condition. The sample holder (5) is pre-tilted in the negative direction up to 45°, allowing some degree of freedom for the researcher to search for the optimum tilt angle for each individual measurement. The stage in the case shown in Figure 2b is tilted 15° in positive direction. These settings define the acquisition angle of -30° which is found to be optimum with respect to position of the diffraction pattern on the phosphorous screen and the indexation reliability. In most of the cases we acquired the diffraction patterns at working distances between 4 to 6mm, but the best results are usually obtained at a working distance of ~5mm. Usually we used the incident electron beam (2) with an energy corresponding to 30 or 20 keV accelerating voltage and a and a beam current in the range between 300 pA and 2.3 nA, focused on the transparent part of the different TEM samples(3)-foils or FIB processed samples beam current in the range between 300 pA and 2.3 nA,
focused on the transparent part of the different samples (3) — foils or FIB processed samples. By these setting we were able to obtain a spatial resolution of in the range of 5 to 10nm in case we use the configuration on Figure 2, known as the off-axis setup, and ~4nm in case when we use the on-axis detector configuration. The angular accuracy was better than 1° in all cases.

3.2.2 Results. The example shown in Figure 3 is from pure aluminum, deformed to a von Misses (vM) true strain of 5.4 by high pressure torsion [37]. The observation is made in a plane perpendicular to the torsion axis.

The acquisition parameters are 30kV accelerating voltage, 2.3 nA probe current, working distance 5 mm and tilt angle -30°. The EBSD scan was performed on a hexagonal scan grid with a step size of

![Figure 3](image1.png)

(a) Forward scattered electrons image; (b) Normal direction inverse pole figure map and (c) a kernel average orientation map with superimposed high angle grain boundaries (>15°) in black, low angle grain boundaries (5-15°) of deformed to $\varepsilon_{vM}=5.4$ by high pressure torsion technically pure Al.

![Figure 4](image2.png)

Figure 4. a) TEM image after ultrafast heating an subsequent quenching b) KAM map of the same area obtained from the TKD analysis. Step size 10nm, 30kV, square grid.
The original data were post-processed by applying only a confidence index standardization procedure. The Kernel average misorientation map (Figure 3c) is calculated for the 3rd neighbor with upper cut off criterion 5°. The KAM map shows clearly the fragmentation of the Al grains in three stages: (i) strain localization inside the existing grains (green and yellow colored regions)(arrow #1); (ii) formation of the low angle grain boundaries in the heavily distorted regions (arrow#2) and (iii) transition from LAGBs to HAGBs in some regions (arrow #3).

Figure 4a shows a TEM bright field image of a steel with 0.2%C, 1.6 %Mn, 1%Al and 0.5%Si (in mass%). The steel was heated with a heating rate of 800°C/s to 860°C, soaked for 0.2 s and subsequently quenched. The microstructure after this treatment consists of ferrite, martensite, bainite and retained austenite with a grains size of ~1.5µm. The KAM map that was gathered on the same sample, covering the same site as the TEM bright field image of Figure 4a, gives an idea of the sensitivity of the TKD to determination of the dislocation density, cf. Figure 4b. It is clearly shown that the zones with KAM values of the order ~1-2° formed a cellular sub-grain structure. The arrow points to a zone with increased dislocation density in a ferrite grain adjacent to a martensite grain. A nearly excellent correspondence can be observed between the dislocation network observed on the TEM image and the TKD-KAM map.

In summary TKD is a very promising characterization technique, which allows effectively and in a relatively simple way to bridge the resolution gap between TEM and classical EBSD. A minimum effort is required to adjust a standard EBSD system to an off axis TKD system. However, the on line TKD system exhibits a number of advantages – faster acquisition and ability to work at better spatial resolution, but they require dedicated hardware.

3.3. The Grain Size Orientation Distribution Function

The grain size and orientation distribution function, or GSODF, combines, in a single continuous function, the probability distributions of crystallographic orientations and grain sizes [38]. The following method has been proposed to find the GSODF of a microstructure: (i) Calculate the grain size distribution of the material; (ii) divide the total number of grains in bins depending on grain size; (iii) Calculate an ODF for each of these bins (using the exponential method of Bunge [39]) and (iv) find a mathematical correlation between the coefficients of the ODFs calculated in the previous step and grain size. Once the C-coefficients of the ODF are expressed as functions of grain size, $d$, an ODF dependent on grain size is obtained: $f_d(g)$ (where $g$ represents a triplet of Euler angles $\varphi_1, \Phi, \varphi_2$). Then, the

Figure 5. (a) Linear dependence of C-coefficients with respect to grain size (the coefficients corresponding are represented in blue, turning into red and finally yellow as the harmonic level increases). The coefficients of the global ODF are represented with black dots; (b) Probability distribution function that show the probability of finding a grain with a certain grain size and orientation in the $\gamma$-fibre.
GSODF is given by the product of the grain size distribution, $p(d)$, and the size dependent ODF, $f_d(g)$:

$$F(d, g) = p(d)f_d(g)$$

Although, a priori, it is not obvious how to find the size dependent ODF, the investigation of several low carbon steels has shown that, for these microstructures, a very clear linear relationship is found between the C-coefficients obtained for different grain size ranges and the logarithm of the grain size (see Figure 5a). When the C-coefficients present this simple relationship with respect to grain size, the GSODF can be succinctly displayed using the grain size distribution function and two ODF plots: the first one, or global ODF, which corresponds to the conventional ODF, calculated for all the grains of the sample, and the second one, or slope ODF, which shows the dependence of intensity on the logarithm of the grain size. In order to obtain the probability density of a material volume element that is part of a grain with a certain size and crystallographic orientation, first the total fraction for that size is found from the size probability function, then it is multiplied by the sum of the intensity for the chosen crystallographic orientation in the global ODF and the correction calculated from the slope ODF for the given size.

The GSODF offers a large number of possibilities for the analysis of microstructures. To begin with, it includes the grain size and orientation distribution functions, which can be calculated as marginal distributions of the GSODF. Moreover, it can be used to study the correlation between both properties. As an example, Figure 5b shows how the GSODF can be used to compare two dual-phase steels. Each of the graphs show the grain size distributions of the grains with orientations along the $\gamma$-fibre ($<111>/ND$). The concept of grain size dependent ODFs can also be applied in modelling. Several examples of how the GSODF is applied to the generation of RVEs for full-field models and in mean-field simulations can be found in [38] and [40].

4. Texture control in non-oriented electrical steels

For decades already texture control in Non-Oriented (NO) electrical steel is the subject of intense study [41]. Unlike for Grain-Oriented (GO) electrical steel there is not a generally accepted manufacturing procedure that guarantees the optimum $<001>/ND$. Various procedures have been developed on a laboratory scale [42;43], but proofed to be difficult to upscale for industrial processing. Unlike for FCC alloys the $\{001\}<100>$ cube component is not a common recrystallization component in low-carbon steel after cold rolling and annealing. Nguyen-Minh et al have discovered, however, that the cube component appears in localized micro shear bands in rotated Goss grains $\{110\}<110>$ [44].

Strain localization in the form of micro shear bands can be accounted for by geometric softening of heavily deformed polycrystals. The change of the deformation geometry is to soften the high hardening stage of crystals. The condition for this change of deformation configurations is proposed by Dillamore et al [45;46], based on crystal plasticity theory. The transition from homogeneous to local plastic flow of materials is usually marked by a plastic instability point ($d\sigma/d\varepsilon = 0$) on the stress-strain ($\sigma$-$\varepsilon$) curve. Beyond this point, geometric softening ($d\sigma/d\varepsilon < 0$) due to local strain accommodation is dominant, and thus grains in shear bands are deformed.

Besides the well-documented case of SBs appearing in the hard $\{111\}<112>$ grains [47], also the rotated Goss ($\{110\}<110>$) oriented crystals, in fact, exhibit the highest Taylor factor value and thus can be considered as the hardest orientations of BCC structured materials under the plane strain
compression of the conventional rolling. Geometric softening by micro shear banding, therefore, should be the most prevalent in rotated Goss oriented grains [48]. An investigation by electron backscatter diffraction (EBSD) technique, therefore, has been carried on samples of an Fe-1.2 wt.% Si alloy after the cold rolling of 75% thickness reduction. The rotated Goss oriented grains in cold rolled steel sheet, in general, are not very frequently observed because the {110}<110> orientation is just a minor component of the rolling texture. The resistance of these grains to plane strain compression is very high, but the grains are merely present as a minor transformation product of recrystallization Cube ({001}<100>) austenite grains. Nevertheless, by EBSD measurements on large samples areas (~1 mm$^2$) with small scanning step sizes (~0.1-0.5 μm), it is possible to find some rotated Goss grains having micro shear bands. Figure 6 shows the presence of Cube oriented crystals in these micro shear bands, which has been predicted by crystal plasticity calculation [44]. Recently, cross rolling and annealing experiments have been implemented on strongly Goss oriented samples [49]. Results of these experiments clearly show the presence of the Cube ({001}<100>) oriented bands in the rotated Goss samples. Moreover, retention and development of the Cube texture are observed in annealed samples [49].

5. Conclusions

In the present paper a very broad overview was given of current state-of-the-art and recent developments in a number of aspects of applied metallurgical research pertaining to texture control in metal manufacturing. It was shown that both in conventional metal sheet manufacturing and in innovative processing technologies such as Additive Manufacturing, a detailed analysis of the texture formation can lead to a better understanding of these processes in terms of their physical mechanisms, which in turn may lead to a better texture control and ultimately to improved material properties.

In another register the crystallographic nature of most metal alloys and the preference of texture allows for advanced microstructural characterization by EBSD. It was shown how the added value of orientation contrast microscopy can be further augmented by gathering 3D-EBSD images through a slice-and-view approach and how the resolution of conventional EBSD can be increased down to less than 10 nm by Transmission Kikuchi Diffraction (TKD).

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