Bainite Orientation in Plastically Deformed Austenite

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A. A. Shirzadi, H. Abreu‡, L. Pocock*,
D. Klobčar†, P. J. Withers* and H. K. D. H. Bhadeshia

University of Cambridge, Materials Science and Metallurgy, Cambridge CB2 3QZ, U.K.

‡Universidade Federal do Ceará, Engenharia Metalúrgica e de Materiais, Fortaleza, Brasil

†Faculty of Mechanical Engineering, University of Ljubljana, SI–1000 Ljubljana, Slovenia Slovenia

*School of Materials, University of Manchester, Manchester M1 7HS, U.K.

Abstract

Experiments have been conducted to see whether specific crystallographic variants of bainite form in polycrystalline steel when transformation occurs from plastically deformed austenite which is otherwise free from externally applied stress. It is demonstrated by studying both overall and microtexture that there is no perceptible variant selection as bainite forms. Indeed, the texture is found to weaken on transformation.

1 Introduction

During displacive transformation the crystal structure of the parent phase is changed into that of the product by a homogeneous deformation which does not require diffusion [1–4]. As a result, there is a specific shape deformation which accompanies the formation of the product phase, a deformation which is characterised as an invariant–plane strain. Bainite forms in this way from austenite and the crystallography of each plate of bainite can be described in terms of a mathematically connected set consisting of the habit plane, orientation relationship with the austenite, and the shape deformation [5].

Transformations like these can be treated in the same way as ordinary plastic deformation when it comes to their interaction with an externally applied stress or system of stresses [6, 7]. Those crystallographic variants whose shape deformation complies with the stress will be favoured over others which oppose the stress [8–10]. The favouring of certain variants over others is known as *variant selection*. In this scenario, the interaction between the stress and the transformation is purely thermodynamic and can be treated rigourously by adding a mechanical driving force

ΔG_{MECH} to the chemical term ΔG_{CHEM} which would ordinarily drive transformation even in the absence of applied stress.

The situation regarding variant selection when the austenite is plastically strained prior to transformation is less clear. The dislocation debris due to the deformation will assist transformation through the provision of nucleation sites [11–17]. Whether this also leads to variant selection is an open question. Most studies of strain–induced transformation do not investigate the isolated effect of strain, but rather a combined effect of stress and strain, for example, by monitoring transformation during a tensile test of some sort. It has been demonstrated in some of these cases that the results can be interpreted in terms of stress–affected transformation without taking into account the plastic strain [18–21]. Indeed, the transformation texture has been predicted assuming stress–affected transformation in samples pulled to strains of some 10% in tension [19, 20]. In the context of martensitic transformation, the fraction of martensite induced has been shown to linearly correlate with the applied tensile stress even when the latter is in excess of the yield strength [22].

An elegant study by Bokros and Parker [23] on the formation of bursts of martensite during the cooling of deformed single–crystals of austenite, may possibly be the only case where the influence of plastic–strain alone (i.e., without the superimposed effect of stress), on variant selection has been studied. It was argued that those plates of martensite which grow *across* slip planes are favoured because the plates then manage to avoid much of the intense dislocation debris on the slip planes. In other words, variant selection occurs because of the anisotropic nature of the dislocation debris. It is likely that such anisotropy will be reduced in polycrystalline samples where multiple slip systems must operate in order to maintain continuity.

There is now a technological reason for a deeper understanding of variant–selection criteria during the transformation of deformed austenite. There is a great deal of effort devoted to the design of welding alloys which compensate for the thermal contraction induced stresses as the weld cools [24–33]. At low temperatures, the weld transforms under the influence these stresses and thereby relieves them; it is believed that the extent of this relief is dependent on variant selection [7]. However, at high temperatures some of the contraction is compensated by the plastic relaxation of the austenite, and it is not clear how this plasticity influences variant selection during subsequent transformation.

The purpose of the present work was therefore to study whether plastic strain in the austenite causes variant selection during the bainite transformation in the absence of an applied stress.

2 Experimental Method

The chemical composition of the steel used was Fe–0.79C–1.56Si–1.98Mn–0.002P–1.01Al–0.24Mo–1.01Cr–1.51Co wt%. This is the same alloy as used by Hase *et al.* [19, 34] to study bainite growth in fully annealed austenite under the influence of a stress which is less than that required for yield, and hence maintains the sample in an elastic state throughout transformation. The intention here was to study transformation from plastically deformed austenite which is free from imposed stress. The high-carbon steel used has the advantage of slow transformation and a simple microstructure following transformation, consisting only of bainitic ferrite and retained austenite.

The presence of austenite is extremely useful in order to fix the orientation of the parent phase during the analysis of crystallographic texture. Cylindrical specimens 8 mm diameter and 12 mm length were machined for the experiments, which were carried out in an adapted thermomechanical simulator, Thermecmast0r Z. Details about this equipment are published elsewhere [35] but the processing used here is illustrated in Fig. 1. Note that there was no applied load during the course of transformation at 300°C.

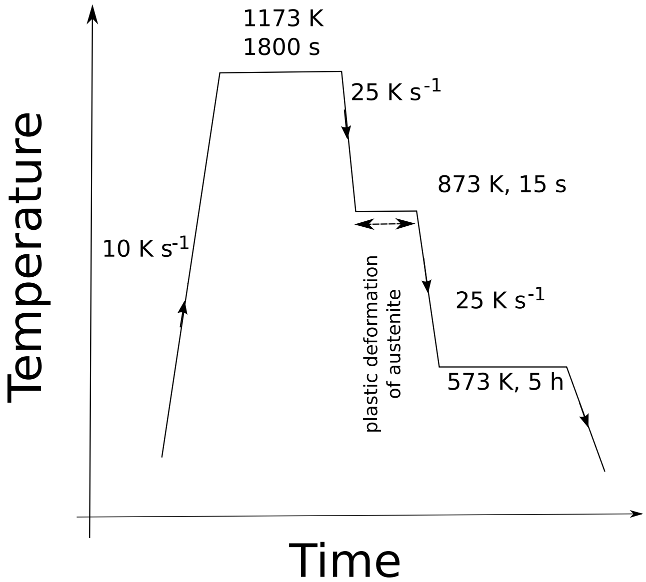


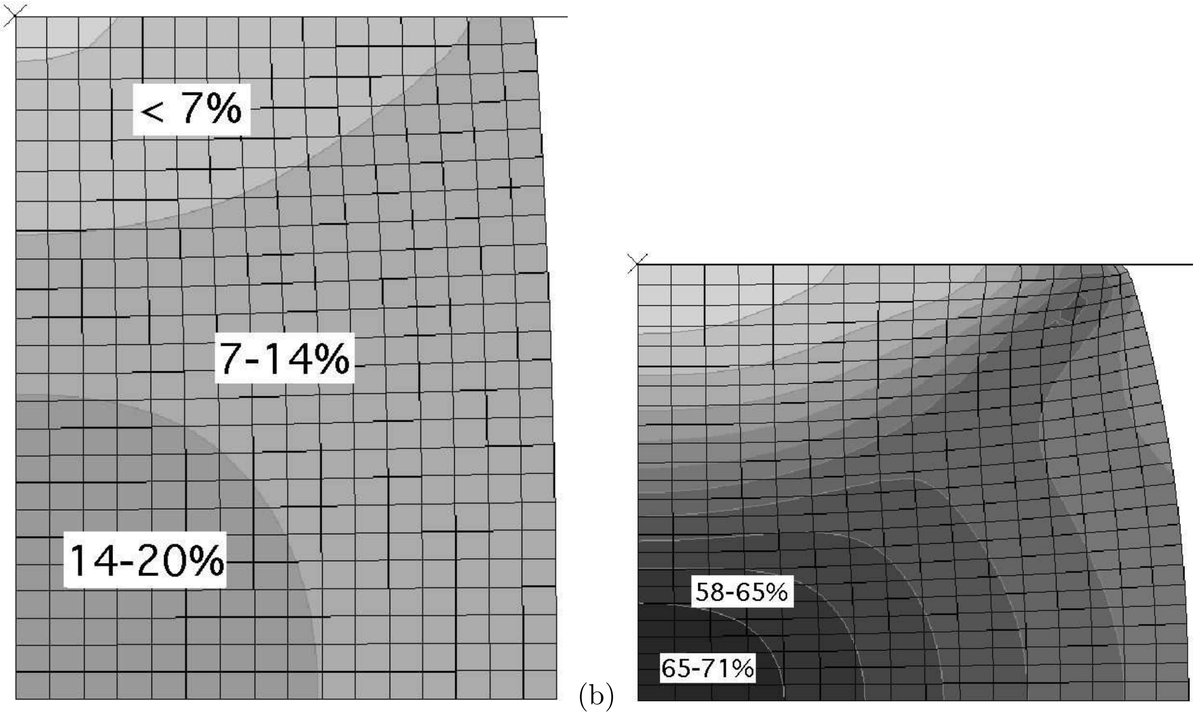
Figure 1: Thermomechanical processing including the formation of baintic ferrite from plastically deformed austenite during isothermal transformation at 300°C for 5 h.

Axial compression of cylindrical samples results in a barrelling effect and a heterogeneous distribution of plastic strain when there is friction between the sample and the platens which apply the compressive force. Finite element calculations were therefore done using an assumed friction coefficient of 0.2 in order to simulate the barrelling. The constitutive equations used to represent the deformation of the austenite are from another steel (Fe–0.15C–0.45Mn wt%) so the results here are indicative. The simulations are illustrated in Fig. 2 and since orientation imaging experiments were conducted at the centre of each specimen, the actual compressive strains are greater than the average deformations of 10% and 33% for the two samples studied.

Specimens for the orientation imaging were mechanically ground and polished, to a final colloidal–silica polish. Crystallographic data were generated by electron backscatter diffraction (EBSD) using a *CAMSCAN Maxim* field emission gun scanning electron microscope at a magnification of $\times 500$ with a step size of $0.5\,\mu$m. Orientation data were acquired from known locations in order to relate the orientation data to strain maps. Subsequent data analysis was carried out using *HKL*'s technology "Channel 5" software.

3 Method and Results

The crystallographic texture was determined using electron backscatter diffraction (EBSD) [36, 37] in the central region of the sample since this had experienced the largest strain. Our aim was to detect variant selection during transformation. Two methods were used to assess this; the



(a) (b)

Figure 2: Finite element analysis estimates of the axial compressive true–strains in the central part of the sample where the orientation imaging was done. The diagrams represent quarter sections of cylindrical samples. (a) 10% average compression. (b) 33% average compression.

first was based on an overall texture due to many austenite grains and the second looking at microtexture development within individual austenite grains as a function of the direction of the applied compressive stress. The overall texture is easily expressed using sections of the orientation distribution function. The microtexture is illustrated on pole figures.

3.1 Overall Texture

The overall transformation texture can be interpreted by noting the strengths of the parent and product textures. Orientation distribution functions obtained from areas $480 \times 360\,\mu$m, typically enclosing 120 austenite grains, are plotted in Fig. 3. Care has been taken to ensure that the contours are on the same absolute scales in the different plots in order to enable comparisons to be made. By noting the density of the contours, it is evident that the transformation texture is much weaker than that of the austenite. This can only happen in the absence of significant variant selection on transformation [19].

3.2 Microtexture

In this section we present an analysis, using representative examples, of the orientations of bainite within individual austenite grains, on the assumption that the applied strain leaves some vestige

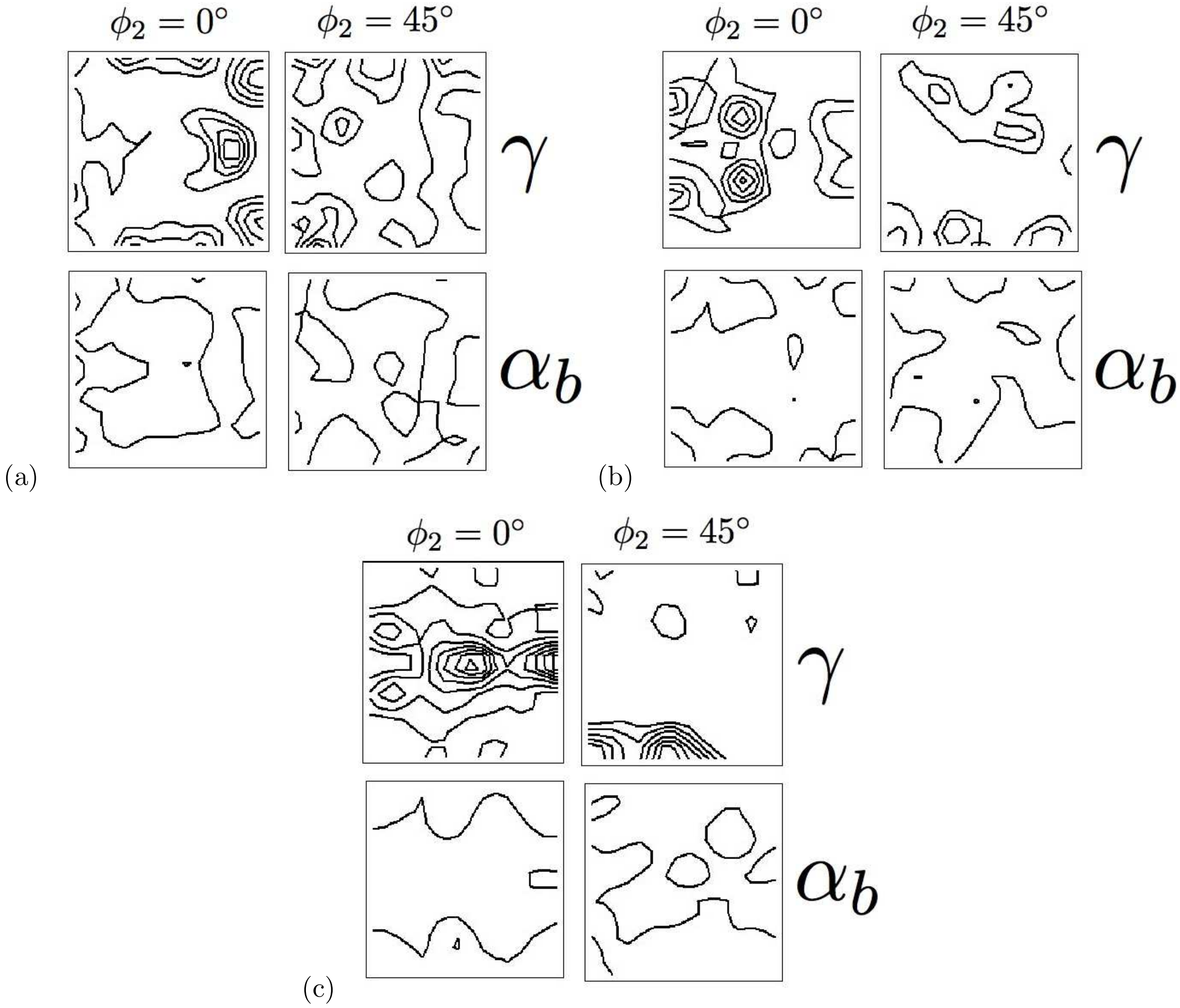


Figure 3: Orientation distribution functions from austenite (γ) and bainitic ferrite (α_b). Only two sections are illustrated in each case for the sake of brevity. (a) Undeformed prior to transformation. (b) Medium deformation prior to transformation. (c) Large deformation prior to transformation.

of the original stress locked within the sample. In other words, do the bainite plates form in a way which complies with the applied stress? The detailed method of analysis is presented elsewhere [19] but the essence of it is as follows.

The phenomenological theory of martensite crystallography gives a complete description of the mathematical connection between the orientation relationship, the habit plane and the shape deformation for each plate that forms by displacive transformation [1, 2, 38–41]. The most common application of this theory is to martensitic transformations but it applies equally well to bainite in steels [42]. Table 1 has the crystallographic set needed to describe each plate of bainite [19]; the terminology used is due to Bowles and MacKenzie [38, 39]. With this information, the measured orientation of the austenite grain, and a knowledge of the stress under whose influence the bainite may form, it is possible to calculate which of the 24 possible crystallographic variants is favoured by the applied stress.

Table 1: Crystallographic data used [19]. $\mathbf{p}_\gamma$ is the habit plane, $(\gamma\ \mathrm{P}\ \gamma)$ is the shape deformation matrix and $(\gamma\ \mathrm{J}\ \ \alpha)$ is the rotation matrix defining the exact orientation relationship between austenite (γ) and bainitic ferrite (α). The orientation relationship is also stated in terms of planes and directions.

$$\mathbf{p}_\gamma = \begin{pmatrix} -0.168640 \\ -0.760394 \\ -0.627185 \end{pmatrix}$$

$$(\gamma\quad \mathrm{P}\quad \gamma) = \begin{pmatrix} 0.992654 & -0.033124 & -0.027321 \\ 0.026378 & 1.118936 & 0.098100 \\ -0.027321 & -0.123190 & 0.898391 \end{pmatrix}$$

$$(\gamma\quad \mathrm{J}\quad \alpha) = \begin{pmatrix} 0.575191 & 0.542067 & 0.097283 \\ -0.550660 & 0.568276 & 0.089338 \\ -0.008610 & -0.131800 & 0.785302 \end{pmatrix}$$

$$[\bar{1}\ 0\ 1]_\gamma || [-0.920611\ -1.062637\ 1.084959]_{\alpha'}$$
$$(1\ 1\ 1)_\gamma || (0.015921\ 0.978543\ 0.971923)_{\alpha'}$$

The effect of this stress on variant selection can be expressed using the Patel and Cohen method [8], in which the interaction of a plate with the stress provides the mechanical driving force U for transformation:

$$U = \sigma_N \zeta + \tau s \tag{1}$$

where σ_N is the stress component normal to the habit plane, τ is the shear stress resolved on the habit plane in the direction of shear and ζ and s are the respective normal and shear strains associated with transformation. The normal and shear stresses on the habit plane can be calculated by resolving the external stress parallel to $\mathbf{p}$ and on to the habit plane in the direction of shear, using the data listed in Table 1, for each possible orientation of an austenite plate. All plates whose shape deformation comply with the compressive residual stress along the direction y are favoured and included here in the calculation of transformation texture.

Figures 4,5 and 6 show representative examples of such an analysis. There is in each case an orientation image which can be qualitatively interpreted to indicate that there is no variant selection in operation since each austenite grain contains many variants and they don't seem to align along particular directions. However, the quantitative interpretation comes from the measured and calculated bainite orientations; the calculated poles are shown in two colours, the red poles consistent with those variants favoured by the applied stress and the black poles those whose shape deformation would oppose the applied stress. It is clear from many such detailed comparisons that both favoured and unfavoured variants are necessary in order to explain the observed pole figures. Indeed, observations in locations of the stereogram where it is possible to avoid the overlap of poles from the favoured and unfavoured variants, show also that the intensities of poles originating from these two varieties are comparable.

It is therefore reasonable to believe that the compression of the specimen prior to transformation does not lead to variant selection via some locked–in vestige of the original applied stress.

4 Summary

Experiments have been conducted in which plastically deformed austenite, which was free from externally applied stress, was transformed into bainite. The austenite prior to transformation was crystallographically textured but phase transformation texture was found to be much weaker. It appears therefore that there is no perceptible tendency for variant selection during transformation. This was confirmed by investigating the development of bainite in individual austenite grains and comparing the experimental data against the possible effects of internal stresses left inside the microstructure due to the original plastic deformation. However, there was no perceptible evidence, either from orientation images or from the examination of pole figures, of the operation of variant selection during transformation.

It is concluded that variant selection does not occur when polycrystalline samples of plastically deformed austenite transform into bainite. This is in contrast to early work on single crystals of austenite in which crystallography resulting from bursts of martensitic transformation seemed to be affected by prior plastic deformation. This observation was attributed originally to the anisotropic distribution of dislocations in the slipped single–crystals. It is possible that this difference with the present observations arises because of the complexity of slip in polycrystalline samples, where a minimum of five slip systems must in general operate in each grain in order to maintain continuity in the sample.

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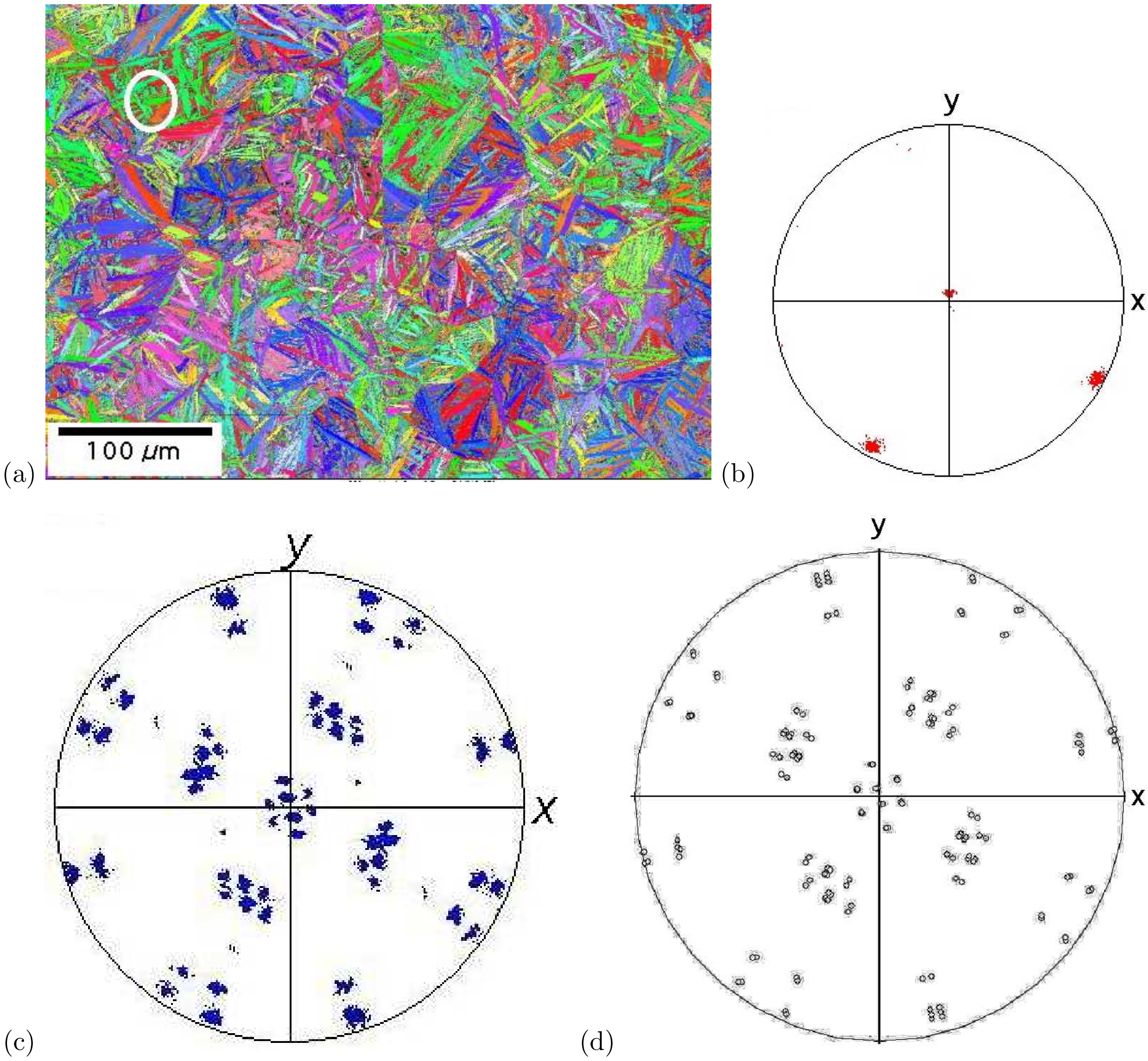


Figure 4: Microtexture data on undeformed specimen. (a) Orientation image with a circle identifying the grain analysed in b–c. (b) Upper hemisphere of the austenite $\{100\}_\gamma$ pole figure. (c) Measured $\{100\}_{\alpha_b}$ pole figure for both hemispheres. (d) Corresponding calculated $\{100\}_{\alpha_b}$ pole figure.

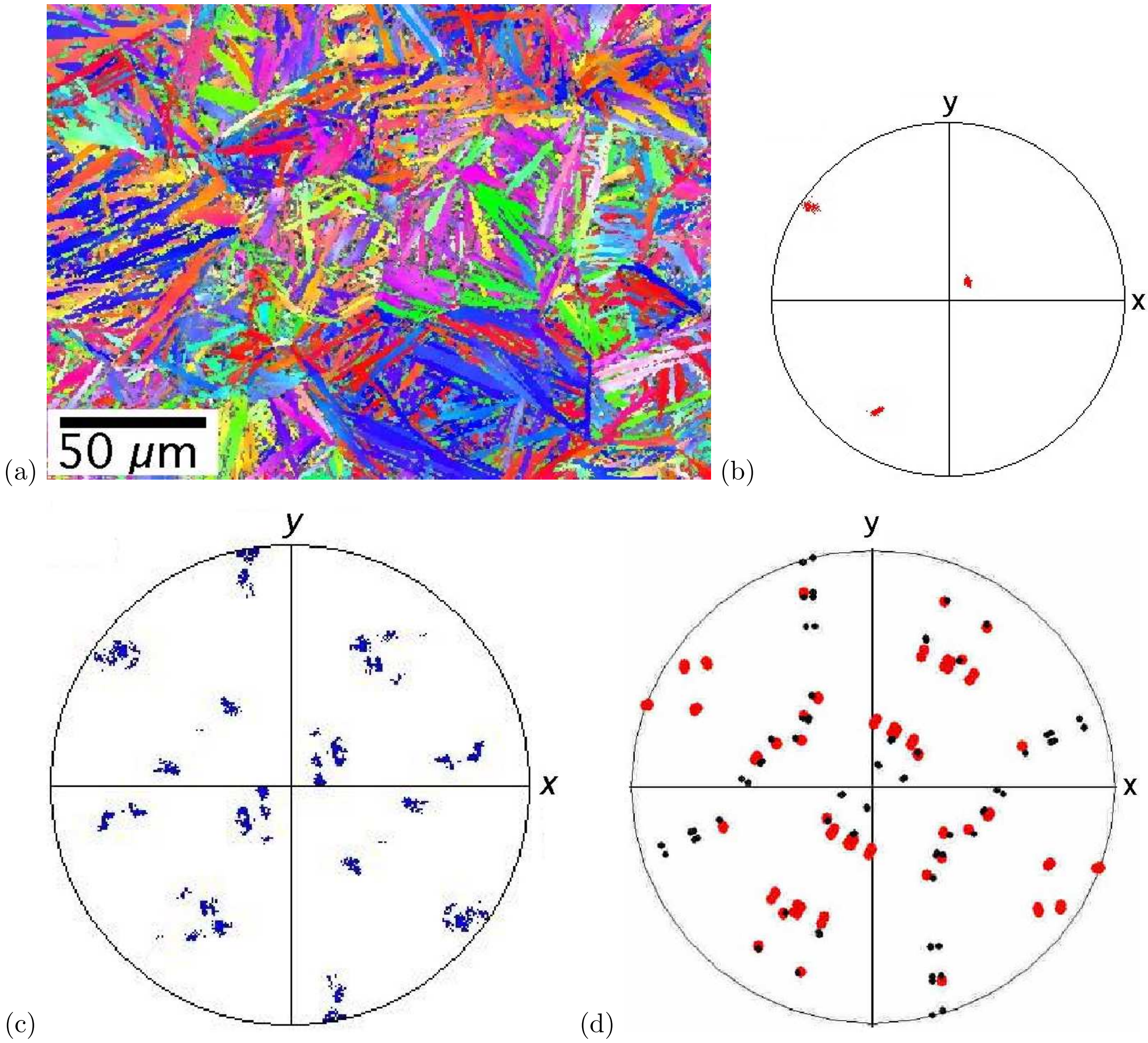
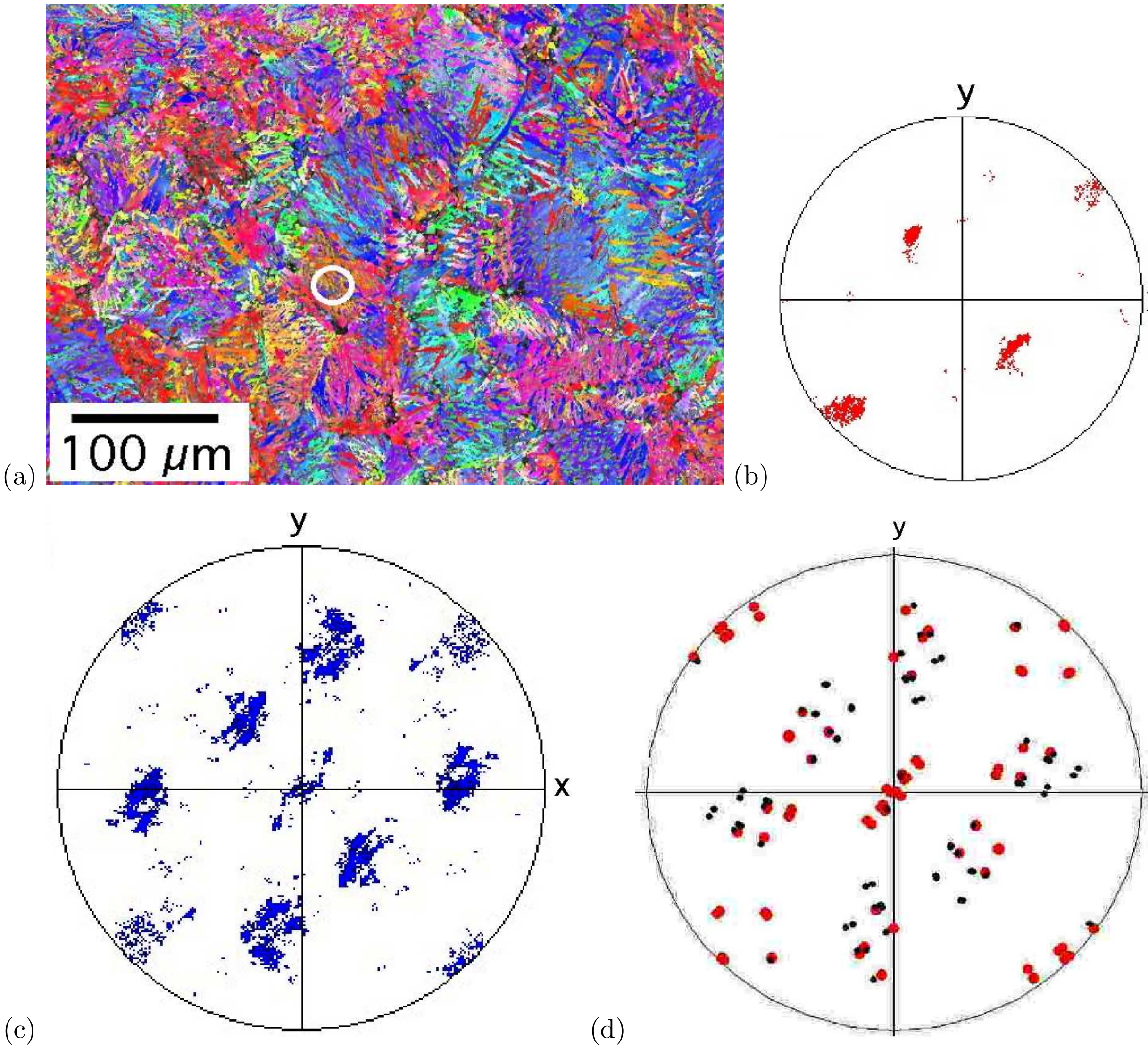


Figure 5: Microtexture data on sample compressed by 10% on average. (a) Orientation image. (b) Upper hemisphere of the austenite $\{100\}_\gamma$ pole figure. (c) Measured $\{100\}_{\alpha_b}$ pole figure for both hemispheres. (d) Corresponding calculated $\{100\}_{\alpha_b}$ pole figure; the red poles are the 'favoured' and the black the 'unfavoured' variants.



(a) (b) (c) (d)

Figure 6: Microtexture data on sample compressed by 33% on average. (a) Orientation image. (b) Upper hemisphere of the austenite $\{100\}_\gamma$ pole figure. (c) Measured $\{100\}_{\alpha_b}$ pole figure for both hemispheres. (d) Corresponding calculated $\{100\}_{\alpha_b}$ pole figure; the red poles are the 'favoured' and the black the 'unfavoured' variants.

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