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ROLLING TEXTURE DEVELOPMENT IN NITROGEN ALLOYED FERRITIC-AUSTENITIC STEEL OF DUPLEX TYPE

ROZWÓJ TEKSTUR WALCOWANIA W FERRYTYCZNO-AUSTENITYCZNEJ STALI TYPU DUPLEX Z DODATKIEM AZOTU

The present research concerns the rolling texture formation in the X2CrNiMoN22-6-3 steel, which chemical composition corresponds to the commercial mid-alloy duplex steel grade SAF 2205. The applied thermo-mechanical treatment included hot-rolling and subsequent solution annealing at the temperature 1150°C. After the preliminary treatment the two-phase structure consisted of about 40% of austenite volume fraction within the ferrite matrix. Subsequently the steel was cold-rolled within the range up to 85% of reduction, perpendicularly to the direction of hot-deformation.

Analysis of the rolling texture formation in both constituent phases indicates that the examined steel shows tendency to develop the textures close to those in single phase steels. In both phases however the process proceeds in different way and the final textures display certain important differences in comparison to the typical ferrite and austenite textures. In the case of ferrite the major component of the initial texture remains stable up to high deformations and the final texture essentially does not exhibit the fibrous character. On the other hand a gradual formation of the texture close to an alloy-type texture is observed in austenite and the orientation changes proceed over a wide deformation range. It is concluded that the band-like morphology of the ferrite-austenite structure and the initial orientations resulting from the change of rolling direction strongly influence the formation of the final rolling textures.

Keywords: duplex stainless steel, rolling schedule, ferrite and austenite rolling textures, two-phase morphology, band-like structure, orientation distribution

Prezentowane badania dotyczą rozwoju tekstury walcowania w stali X2CrNiMoN22-6-3, której skład chemiczny odpowiada średnio-stopowej stali duplex w gatunku SAF 2205. Zastosowana obróbka cieplno-mechaniczna obejmowała walcowanie na gorąco oraz przesycanie z temperatury 1150°C. Po obróbce wstępnej struktura dwufazowa składała się z około 40% austenitu w osnowie ferrytu. Następnie stal poddano walcowaniu na zimno w zakresie do 85% deformacji, prostopadle do kierunku odkształcenia na gorąco.

Analiza tworzenia się tekstur walcowania w obu składowych fazach wskazuje, że badana stal wykazuje tendencje do rozwoju tekstur zbliżonych do tych w stalach jednofazowych. Jednak w obu fazach proces ten zachodzi w odmienny sposób a końcowe tekstury wykazują pewne istotne różnice w porównaniu do typowych tekstur ferrytu i austenitu. W przypadku ferrytu, główna składowa tekstury początkowej pozostaje stabilna do wysokich odkształceń a tekstura końcowa zasadniczo nie wykazuje charakteru włóknistego. Z kolei w austenicie obserwuje się stopniowy rozwój tekstury zbliżonej do tekstury typu stopu a zmiany orientacji zachodzą w szerokim zakresie odkształceń. Stwierdzono, że pasmowa morfologia ferrytu i austenitu oraz orientacje początkowe wynikające ze zmiany kierunku walcowania wywierają istotny wpływ na tworzenie się końcowych tekstur odkształcenia.

1. Introduction

From a number of experimental data it results that the anisotropy in mechanical properties of cold-rolled duplex stainless steels is not only the effect of a specific morphology of the (α/γ) two-phase structure but also depends on the crystallographic textures of the constituent phases (e.g. [1-3]). Since rolling deformation is predominantly one of the first manufacturing steps, the factors affecting a development of rolling textures in (α/γ) stainless steels are the subject of increasing interest [3-8].

Many reported investigations indicate that a development of rolling textures in ferritic-austenitic steels strongly depends on the chemical composition, which determines a phase composition of the steel as well as the stability of the austenitic γ-phase in the course of deformation. However, deformation mechanisms governing

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the formation of ferrite and austenite rolling textures in duplex steels are supposed to change in comparison to single phase steels (e.g. [9, 10]), first of all due to the interaction of both phases as well as the formation of the ferrite-austenite band-like structure, which develops in the course of deformation [3-8]. This specific two-phase morphology (so-called pancake-structure) considerably constrains lattice rotations upon deformation and hence affects the formation of rolling textures of both phases. That is why the initial morphology and the starting orientation distributions of the constituent phases after thermo-mechanical pre-treatment become very important factors, which should be taken into account when analysing texture formation of cold-rolled duplex type steels [3-8]. Since both, hot-deformation and subsequent solution treatment produce preferred orientation distributions of the constituent phases, thus conditions of both processes are very important not only from the view point of phase composition and initial morphology but also starting textures for further cold-rolling.

The main purpose of the present study was to analyze the effect of the ferrite-austenite band-like morphology and the initial orientation distributions resulting from the change of rolling direction on a development of the ferrite and austenite deformation textures in duplex type steel sheets. Cold-rolling deformation was conducted after solution annealing within the wide range of strains up to about 85% of reduction, perpendicularly to the hot-rolling direction. The chemical composition of the steel and the volume fractions of the constituent phases after thermo-mechanical treatment were also taken into account. The results were compared with those for one-phase ferritic and austenitic steels [9, 10].

2. Material and experimental procedure

The material examined in the present research was the ferritic-austenitic stainless steel of duplex type X2CrNiMoN22-6-3, with nitrogen addition of 0,19%. Its chemical composition (see Table I) corresponds to the commercial mid-alloy steel SAF 2205 having the PREN factor of about 35, which is to-day the most frequently produced duplex steel grade.

TABLE

The chemical composition of duplex type steel under examination (in weight %)

C	Cr	Ni	Mo	N	Mn	Si
0.02	22.28	6.05	2.95	0.19	1.72	0.51
Cu	Al	Nb	V	P	S	PREN
0.12	0.03	0.02	0.11	0.014	0.005	35

The preliminary thermo-mechanical treatment included hot-rolling and subsequent solution annealing. The material was received in the form of industrially hot-rolled sheets with the thickness h_0 = 12.5 mm. After the solution treatment at the temperature 1150°C for 3 hours the flat shapes cut of the sheets were subjected to reversed rolling at room temperature on laboratory mill (Φ = 160 mm) within the range up to about 85% of thickness reduction ($\varepsilon_t \approx 2$). Cold-rolling was conducted perpendicularly to the direction of hot deformation (see Fig. 1), i.e. $RD_C \perp RD_H$.

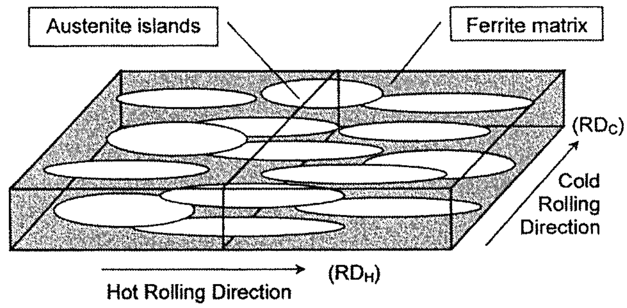

Fig. 1. Schematic illustration of the ferrite-austenite morphology and geometry of the applied rolling schedule ($RD_C \perp RD_H$)

The applied reductions per pass ensured the least strain gradient throughout the cross section, with the ratio $(l_c/h_m) > 1.0$; where $l_c = \sqrt{r\Delta h}$ – the length of the arc of contact, r – radius of rolls, Δh – thickness reduction per pass and h_m – the mean thickness of the sheet. It is assumed that for such rolling conditions the compressive strain penetrates throughout the whole thickness of the sheet and a deformation rate of the centre layer, parallel to the rolling plane, is similar to that of the subsurface layers [11].

The microstructure observations of the initial morphology of two-phase structure after hot-rolling and subsequent solution treatment as well as the ferrite-austenite band-like structure formed in the course of cold-rolling were conducted by means of optical microscopy (Neophot-2 and Axiovert-200 MAT microscopes). The phase composition of the steel and the ferrite-austenite morphology, were determined by means of the quantitative metallographic analysis employing program MET-ILO, version V.9.11 [12, 13]. The volume fractions of ferrite (V_V^F) and austenite (V_A^V) after the solution treatment at 1150°C were estimated at about 60% and 40% respectively. Thus the ferritic α-phase was more continuous and constituted the matrix with islands of the austenitic γ-phase. That is way the ferritic-austenitic steel under examination had not a typical duplex structure.

X-ray investigations were conducted by means of Bruker diffractometer D8 Advance, using Co Kα radiation ($\lambda_{K\alpha}$ = 0.179 nm). X-ray examination included the texture measurements and the phase analysis from the centre layers of the rolled sheets, for the initial state and after selected rolling reductions. Texture analysis was performed on the basis of the orientation distribution functions (ODFs) calculated from experimental (incomplete) pole figures recorded of three planes for each of the constituent phases, i.e. {110}, {100} and {211} planes for the bcc α-phase and {111}, {100} and {110} planes for the fcc γ-phase. The values of the orientation distribution functions f(g) along the typical orientation fibres α_1 = <110>||RD, γ = <111>||ND, ε = <001>||ND for ferrite and α = <110>||ND, η = <001>||RD, τ = <110>||TD for austenite were examined. Additionally, changes of the ferrite and austenite rolling textures resulting from the change of the rolling direction ($RD_C \perp RD_H$) were analyzed on the basis of calculated and experimental pole figures after the selected rolling reductions.

3. Results and discussion

3.1. Initial structure and texture

The microstructure after a preliminary thermo--mechanical treatment of a ferritic-austenitic steel with a given chemical composition usually results from the applied type of hot-deformation process and the conditions of subsequent solution annealing, which determine the phase composition and the morphology of two-phase structure [5, 13].

After hot-rolling of duplex steel under examination the austenitic γ-phase formed islands on the background of the ferritic α-phase, which constituted the matrix. The (α/γ) two-phase structure exhibited significant heterogeneity, which concerned the distribution and arrangement as well as the size and shape of areas (grains) of both component phases. On the longitudinal sections (ND_H-RD_H) of the sheet the regions with considerable degree of banding predominate. Nearly continuous ferrite and austenite bands strongly elongated parallel to the rolling direction RD_H having different thickness were mainly observed (Fig. 2a). The microstructure observed on the cross-sections (ND_H-TD_H) of the hot-rolled sheet also showed considerable heterogeneity, however significantly lower degree of banding, i.e. much smaller elongation along the transverse direction TD_H.

After solution annealing at the temperature 1150°C for 3 hours the morphology of two-phase structure was considerably changed. Pronounced directionality of the ferrite-austenite structure is still observed on the longitudinal (ND_H-RD_H) sections (Fig. 2b). The areas of the γ-phase remained elongated along the RD_H, however are much less continuous. On the other hand the microstructure observed on the cross-sections (ND_H-TD_H) shows much less structural anisotropy. The exact description of the microstructure after thermo-mechanical treatment, employing quantitative image analysis, is given by the present authors elsewhere [13].

Texture measurements conducted after hot-rolling revealed relatively week textures in both constituent phases. The major component of the austenite texture is cubic orientation {100}<001> with the intensity f(g) = 2.3, whereas the ferrite texture shows partially fibrous character with orientations {001}<100-110> and the maximum intensity f(g) = 3.2 corresponding to the ~{001}<320> orientation (Fig. 3a).

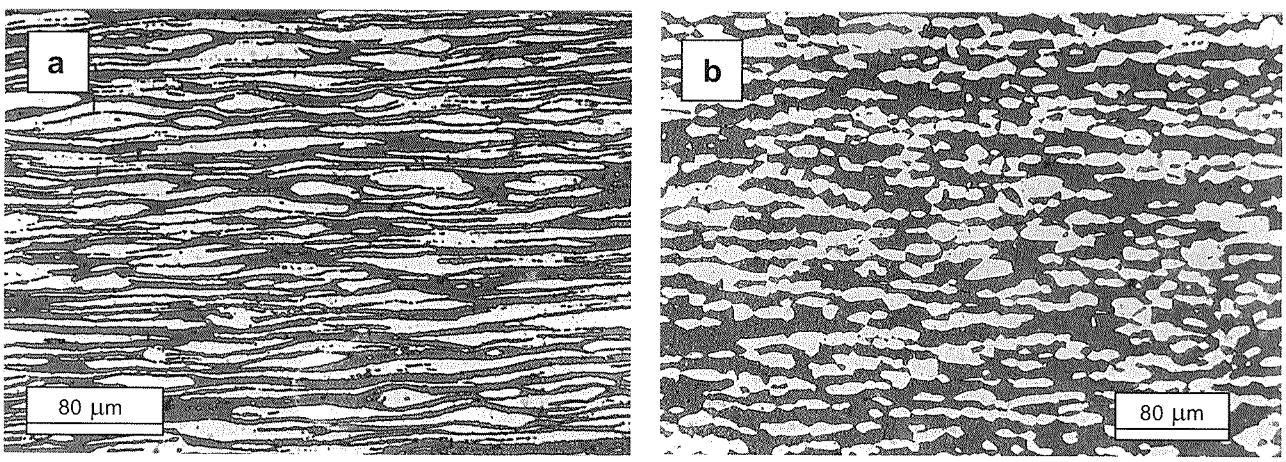

Fig. 2. Morphology of the ferrite-austenite two-phase structure after hot rolling (a) and solution annealing at 1150°C (b) on the longitudinal section (ND_H-RD_H) of the sheet

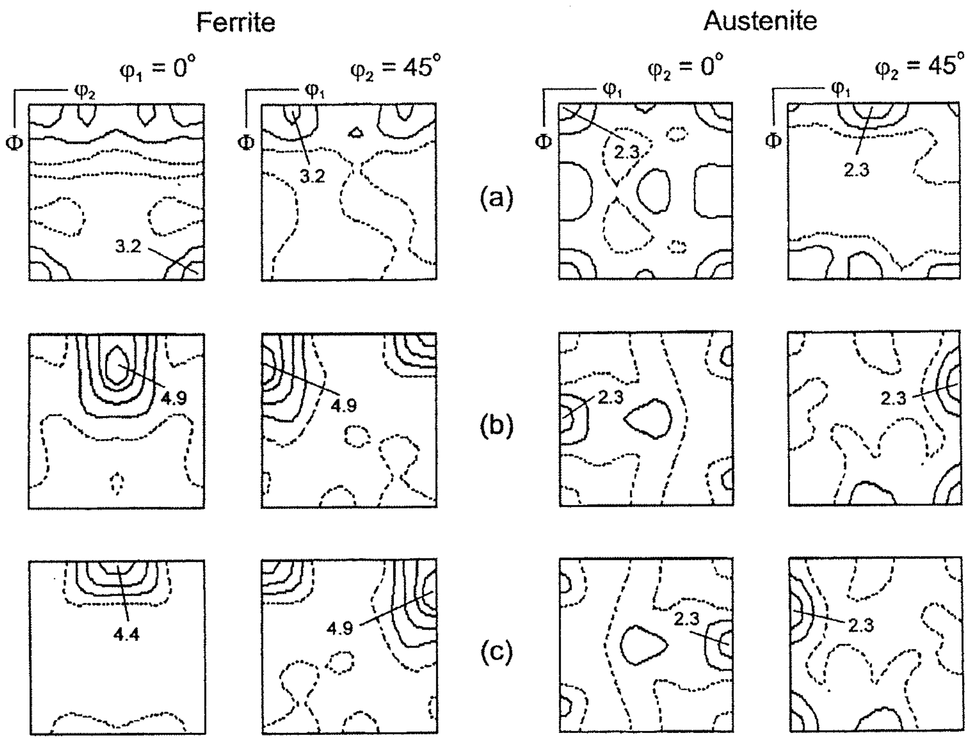

Fig. 3. Orientation distribution functions (ODFs) in sections φ_1 = 0°, φ_2 = 45° for ferrite and φ_2 = 0°, φ_2 = 45° for austenite after hot-rolling (a), solution annealing (b) and the 90°/ND rotation (c) from the centre layer of the sheet

After the solution treatment at 1150°C the austenite texture remained very week. The main texture component, which is the Goss orientation {110}<001> with the intensity f(g) = 2.3, is observed against the background of nearly random texture. The texture of ferrite after solution annealing is stronger with the maximum intensity f(g) = 4.9 corresponding to the ~{114}<110> orientation, which is deviated about 20° from the rotated cubic {001}<110> component. The ferrite texture may be essentially described by the limited fibres; α_1 = <110>||RD and ε = <001>||ND (Fig. 3b).

3.2. Deformation texture and structure

After the preliminary treatment the samples cut of the hot-rolled sheet were subjected to cold-rolling, which was conducted along the perpendicular direction, i.e. $RD_C \perp RD_H$ (Fig. 1). That is why the initial orientation distributions of both constituent phases with respect to the external co-ordinate system were changed due to the rotation of the samples 90° around the direction normal to the sheet ND. After rotation, the strongest component of the initial texture of ferrite was the rotated cubic orientation {001}<110> and the maximum intensity of the austenite starting texture corresponded with the {110}<110> orientation (Fig. 3c). The schematic illustration of the strongest initial orientations in both phases

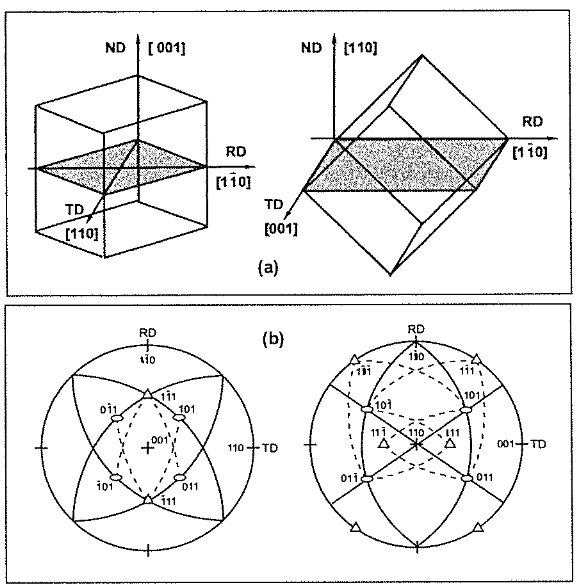

Fig. 4. Schematic illustration of the strongest texture components in ferrite and austenite after 90°/ND rotation (a) and the stereographic projections of the potential slip systems in both phases (b)

is shown in figure 4a. Estimation of the most stressed slip/twin systems within the textured ferrite-austenite banded structure was performed by calculation of the relative shear stresses, i.e. the ratios m = (τ/σ), for the strongest texture components in both phases. It should be noted that the potential slip/twin systems are symmetrical with respect to the common rolling direction <110>||RD_C (Fig. 4b). For the case of ferrite initial orientation {001}<110> the highest values of the relative shear stresses are in two {112}<111> slip systems (m = 0,942) and four {110}<111> slip systems (m = 0,816). In austenite with strong {110}<001> texture component there are four {111}<112> equally stressed twin systems (m = 0,471) and eight {111}<110> slip systems (m = 0,408). Obviously not all of these systems are active simultaneously. It is assumed that the specific morphology of two-phase structure forces the operation of the selected slip systems, symmetrical with respect to RD_C, which ensure compatible deformation of the ferrite and austenite bands (layers) [7, 8].

Changes in the ferrite-austenite microstructure, as observed on the longitudinal sections (ND_C-RD_C) of the cold-rolled samples, are shown in figure 5a, b. It should be noted, that the initial two-phase morphology corresponded with that observed on the cross-section (ND_H-TD_H) of the hot-rolled sheet, i.e. before 90°/ND rotation. In the course of cold-rolling both component phases were plastically deformed and formed characteristic banded structure (so-called pancake structure) consisting of alternate bands (flat shapes) of ferrite and austenite elongated parallel to the rolling plane. With the increasing rolling reduction a significant refinement of the microstructure was observed, with the thickness reduction of the austenite bands to about few micrometers. The layers of ferrite were thicker and exhibited more continuous character due to the phase composition of the steel (Fig. 5a, b).

The measurements of the ferrite and austenite deformation textures carried out in the course of cold-rolling revealed quite different behaviour of both constituent phases. Texture changes after the successive rolling reductions were examined in the selected one-dimensional sections through the ODFs, along the orientations fibres characteristic for both phases (Figs. 6a and 6b). Additional examination concerned the selected experimental and calculated pole figures, {110} for ferrite and {111} for austenite (Figs. 7 and 8).

In the case of ferrite the dominant component of the initial texture was the {001}<110> orientation, which occurred the stable orientation within the whole range of deformations and remained the major texture component up to 85% of rolling reduction (ε = 2.0). The rotated cubic orientation {001}<110> from the orientation fibres ε = <001>||ND and α_1 = <110>||RD is one of the typical components of ferrite rolling textures. The only changes in ferrite rolling texture, as observed in ODF sections φ_1 = 0° and φ_2 = 45° (Fig. 6a), concerned essentially the orientation components of weaker intensities from the α_1- and ε-fibres. In general the inhomogeneous and limited ε-fibre was observed within the whole range of deformations and only at medium strains (about 40%) the full scope of orientations {001}<100÷110> was detected.

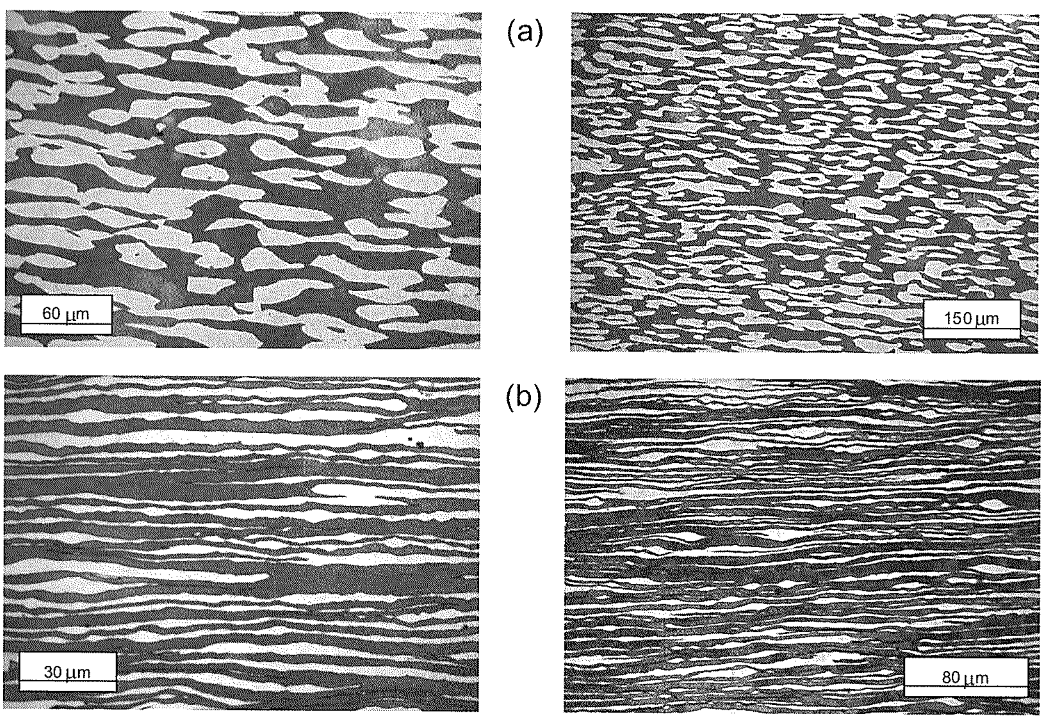

Fig. 5. The morphology of the ferrite-austenite microstructure after 30% (a) and 85% (b) of rolling reduction on the longitudinal section (ND_C-RD_C)

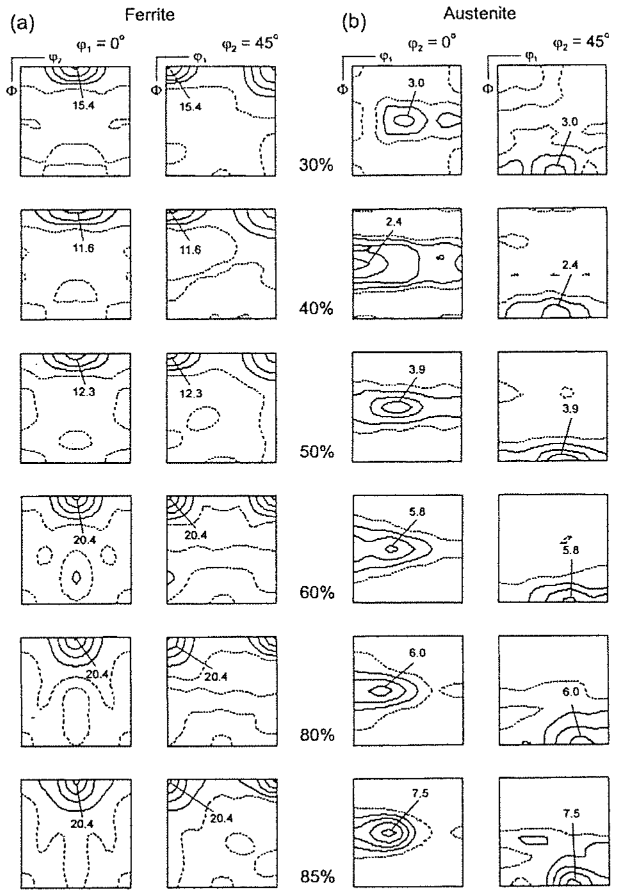

Fig. 6. Orientation distribution functions (ODFs) for ferrite in sections $\varphi_1 = 0°$, $\varphi_2 = 45°$ (a) and for austenite in sections $\varphi_2 = 0°$, $\varphi_2 = 45°$ (b) after the selected thickness reductions in cold-rolled sheets

On the other hand the limited α_1-fibre appeared only at higher deformations, starting from about 60% reduction. This fibre shows relatively small extension along the Φ direction in comparison to single phase ferritic steels and to some degree reflects the slip behaviour within the bands of ferrite. It is usually assumed that in single phase bcc materials the full orientation range in the α_1-fibre is the result of so-called pencil glide, i.e. slip in {110}, {112} and {123} planes [9]. In the case of two-phase (α/γ) structure, with specific band-like morphology, the limited α_1-fibre with strong {001}<110> texture component may be the effect of major contribution of the {110}<111> slip in ferrite texture formation [3, 5]. Stability of the dominant component of the ferrite rolling texture is very well visible on the experimental {110} pole figures, measured after the successive rolling reductions (Fig. 7). The characteristic rearrangement of texture spread, which initially was parallel and finally perpendicular to the rolling direction RD_C (Fig. 7), is the result of only slight orientation changes due to the 90° rotation of the samples around the ND, i.e. the change of rolling direction, $RD_C \perp RD_H$.

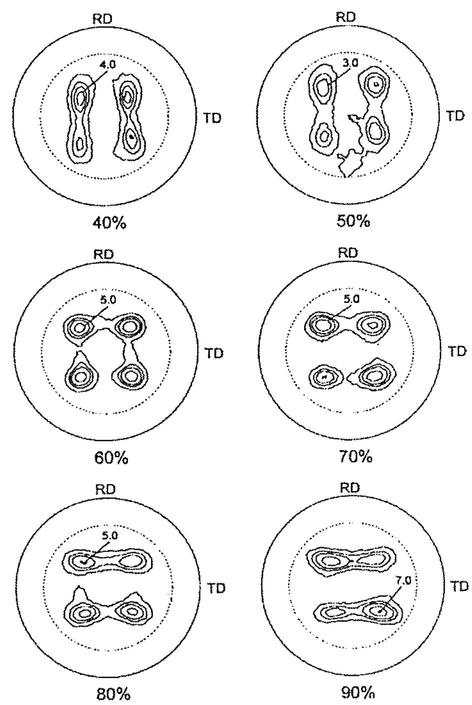

Fig. 7. Changes of the ferrite rolling texture on the experimental (incomplete) pole figures recorded of {110} planes, after the selected rolling reductions

In contrast to ferrite, the gradual development of the final rolling texture was observed in the case of austenite (Fig. 6b). Due to 90°/ND rotation of the samples the major component of austenite initial texture was the {110}<110> orientation. After 30% of reduction the limited {110}<uvw> fibre was observed, with maximum intensity f(g) = 3.0 corresponding to ~{110}<223> and ~{110}<441> orientations. At medium rolling reductions (40% and 50%) the inhomogeneous and relatively wide orientation fibre α = <110>||ND was formed with maximum intensities shifting from the {110}<001> Goss orientation to the {110}<112> alloy type component. In both cases the texture intensities were still very low, i.e. f(g) = 2.5-3.9. Starting from 60% reduction up to 85% the austenite rolling texture is relatively stable and may be described by the limited α-fibre with the maximum shifting within the range {110}<113-112>. At higher strains, i.e. 60-80%, a certain contribution of the components from the η = <001>||RD fibre is observed and after 85% of deformation the {111}<112> orientation from the γ-fibre appeared additionally. Finally, the rolling texture of austenite is an alloy-type texture, which is typical for low SFE materials [10]. It should be noted, that addition of nitrogen strongly decreases the SFE value of austenite [14]. Due to the elemental partitioning upon high temperature annealing the N contents in austenite is estimated to be 7-8 times higher than in ferrite [15]. Taking into account the 0,19% addition of nitrogen to the examined steel, its contents in the γ-phase may reach up to 0.35%.

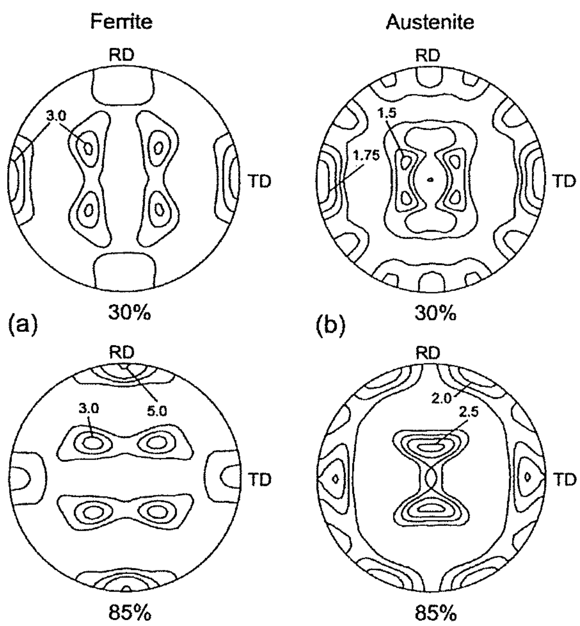

Fig. 8. Comparison of the ferrite (a) and austenite (b) rolling textures after early (ε = 30%) and final (ε = 85%) deformation stages on {110} and {111} pole figures respectively

The differences between texture development of both phases, i.e. stability of major texture component in ferrite and gradual formation of the final texture in austenite, are quite well reflected when comparing the {110} and {111} complete (calculated) pole figures from the early (30%) and final (85%) deformation stages (Fig. 8a, b). Another difference between the rolling textures of constituent phases is texture intensity, i.e. orientation density f(g). Although at higher deformations the texture intensities of austenite increased to maximum f(g) = 7.5, nevertheless remained relatively low in comparison to those of ferrite texture. Within the whole deformation range the intensities of the ferrite rolling textures were found to be at least three times higher in comparison to the orientation density of austenite. This difference results not only from the phase composition of the examined steel (V_V^F = 60% and V_A^V = 40%). The fact, that austenite was the second phase in the form of thin bands (layers), apparently reduced the rotation rate of grains and hence the lower texture intensity of the γ-phase. It should be noted additionally, that cold deformation was performed perpendicularly to the austenite bands elongated upon hot-rolling along the RD_H (Fig. 1). After 90°/ND rotation the initial texture of austenite was gradually changing in the course of cold-rolling and reached its maximum intensities not before the final deformation stage, i.e. at 85% reduction.

4. Concluding remarks

The rotation of the sheet 90 around the normal direction (ND) does not essentially change the ferrite initial texture, which is the very specific orientation distribution in this case. However in austenite there is considerable orientation change with respect to the new coordinate system ($RD_C \perp RD_H$). In general it is concluded that the initial orientation distribution together with the specific band-like morphology of the ferrite-austenite microstructure after thermo-mechanical treatment exert significant influence on the formation of the final rolling textures in the examined ferritic-austenitic steel of duplex type.

Analysis of the ferrite and austenite rolling texture formation indicates that both constituent phases show tendency to develop the textures close to those in single phase steels. However in both α- and γ-phases the process proceeds in different way and the final textures display certain important differences in comparison to the typical ferrite and austenite textures.

In the case of ferrite the initial major texture component {001}<110> remains stable up to high deformations and the final texture essentially does not exhibit the fibrous character. The limited α_1-fibre with strong {001}<110> texture component may be the effect of a dominant contribution of the {110}<111> slip in ferrite texture formation, resulting from the specific band-like morphology of ferrite-austenite structure. The only changes within the ferrite rolling texture in the course of cold-rolling concern essentially some orientations from the texture spread, that is components of weaker intensities.

On the other hand in austenite a gradual formation of the final texture, close to the {110}<112> alloy-type texture, is observed. The fact that austenite is the second phase (V_A^V = 40%) in the form of thin bands within the ferrite matrix significantly restrains lattice rotations. That is why the resulting orientation changes are extended over a wide deformation range and the intensity of the final rolling texture is relatively low. However, the addition of nitrogen and its concentration mainly within the γ-phase considerably decreases stacking fault energy and as the result austenite finally reaches the rolling texture close to that in low SFE alloys.

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