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J. RYŚ*, W. RATUSZEK*, M. WITKOWSKA*

ROLLING TEXTURE DEVELOPMENT IN DUPLEX TYPE STEEL WITH STRONG INITIAL TEXTURE

ROZWÓJ TEKSTUR WALCOWANIA W STALI TYPU DUPLEX O SILNEJ TEKSTURZE WYJŚCIOWEJ

The present examination concerns a development of deformation textures in stainless steel of duplex type with strong initial texture. The ferritic-austenitic steel was subjected to cold-rolling within the range up to 90% of reduction. Different rolling conditions were applied employing three variants of rolling reductions per pass. The investigations included texture measurements, X-ray phase analysis and microstructure observations by means of optical microscopy.

The occurrence of strong initial textures of both component phases was found after the preliminary thermo-mechanical treatment. Ferrite exhibited the {100}<001> cubic texture and the dominant component of the austenite texture was the {110}<001> Goss orientation. The crystallographic relation between the major texture components of both phases was described by Bain orientation relationship. The experimental results indicate that formation of the ferrite-austenite banded structure in the course of cold-rolling along with the plastically compatible deformation of both phases result in the stability of major texture components up to high reductions and exert a significant effect on the final rolling textures of ferrite and austenite.

Przedstawione badania dotyczą rozwoju tekstur odkształcenia w stali nierdzewnej typu duplex o silnej teksturze wyjściowej. Stal ferrytyczno-austenityczną poddano walcowaniu na zimno w zakresie do 90% deformacji. Zastosowano różne warunki walcowania wykorzystując trzy warianty odkształceń częściowych. Przeprowadzone badania obejmowały pomiary tekstur, rentgenowską analizę fazową oraz obserwacje mikrostruktury za pomocą mikroskopii świetlnej.

Po wstępnej obróbce cieplno-mechanicznej stwierdzono występowanie silnej wyjściowej tekstury obu składowych faz. Ferryt wykazywał teksturę sześcienną {100}<001> a w teksturze austenitu dominującą składową stanowiła orientacja Gossa {110}<001>. Relacje krystalograficzną pomiędzy głównymi składowymi wyjściowej tekstury obu faz opisywała zależność Baina. Uzyskane wyniki wskazują, że tworzenie się pasmowej struktury ferrytu

* WYDZIAŁ METALURGII I INŻYNIERII MATERIAŁOWEJ, AKADEMIA GÓRNICZO-HUTNICZA, 30-059 KRAKÓW, AL. MICKIEWICZA 30

i austenitu w trakcie walcowania na zimno przy równoczesnym plastycznie zgodnym odkształceniu obu faz prowadzi do stabilności głównych składowych tekstury do wysokich stopni odkształcenia i wywiera istotny wpływ na końcowe tekstury walcowania ferrytu i austenitu.

1. Introduction

Plastic deformation of duplex type stainless steels is very complex due to the two-phase character of the structure. The presence of the (α/γ) interfaces considerably affects the deformation behaviour of ferrite and austenite since the phase boundaries effectively hinder the plastic deformation and result in strain incompatibilities between both phases [1÷3].

It is usually assumed that the development of deformation textures in ferritic-austenitic steels depends first of all on the chemical composition, deformation conditions and initial textures of both component phases [2÷6]. It seems however that the morphology of two-phase structure formed upon the steel processing is also very important factor, affecting further deformation behaviour and in consequence texture formation [3, 6].

In duplex type steels both component phases are plastically deformed in the course of rolling. The formation and subsequent reduction in thickness of the ferrite and austenite bands result in so-called "pancake" structure. As a consequence the (α/γ) phase interfaces, which are strong obstacles for dislocation motion, are laying mostly parallel to the rolling plane [1, 2]. The specific band-like morphology of two-phase structure creates different conditions for the plastic deformation and the rolling texture development in comparison to one-phase steels [7÷10]. It is expected in such a case that the occurrence of strong initial textures may additionally affect deformation behaviour of both component phases in the course of rolling, changing a development and a character of the final rolling textures of ferrite and austenite.

The purpose of the present investigations was the examination of texture formation in duplex type ferritic-austenitic steel subjected to cold-rolling within a wide range of deformations. The analysis was performed from the viewpoint of the influence of the strong initial orientations and the applied rolling conditions on the development of austenite and ferrite rolling textures.

2. Material and experimental procedure

2.1. Material preparation

The material examined in the present investigations was the ferritic-austenitic stainless steel of duplex type (00H24N6), with the exact chemical composition given in Table I. Simple chemical composition of the steel under examination, in comparison to super duplex stainless steel grades (i.e. without additions of molybdenum, nitrogen, etc.), was chosen in order to reduce a number of factors affecting a development of deformation texture and microstructure [1÷3].

TABLE 1

Chemical composition of the ferritic-austenitic steel under examination (in wt %)

C	Cr	Ni	Mn	Si	Al.	S	P	N	Fe
0.009	23.7	6.0	1.23	0.34	< 0.020	0.010	< 0.008	0.012	bal.

The steel ingot was industrially homogenized and forged within the temperature range 1100÷900°C. Afterwards the rectangular steel rods were annealed at the temperature 1100°C for 3 hours and quenched in the water. Subsequent to the thermomechanical pre-treatment the steel rods were subjected to reversed rolling at room temperature within the range up to 90% of thickness reduction. Different rolling schedules were applied employing small, medium or high reductions per pass (variants S, M and H respectively). The varying rolling conditions depend on the geometry of the rolling gap, i.e. primarily, on the ratio of the length of contact (l_c) to the mean thickness (h_m) of the sheet ($l_c = \sqrt{r\Delta h}$, where r — radius of rolls and Δh — thickness reduction per pass). The three rolling variants resulted in different stress and strain distributions on the cross section of the sheet [11, 12]. In the case of variant S, i.e. for the ratio (l_c/h_m) < 0.5, the compressive strain does not penetrate the whole thickness of the sheet, hence there is a strain gradient throughout the cross section. In such case the deformation of the centre layer, parallel to the rolling plane, is assumed to be much less than that of the subsurface layers. On the contrary for the ratio (l_c/h_m) > 1.0 (variant H) the compressive strain extends over the entire thickness of the cross section and the deformation of the centre layer of the sheet is much higher [11, 12].

2.2. Experimental procedure

X-ray investigations were conducted by means of Bruker diffractometer D8 Advance, using Co Kα radiation ($\lambda_{K\alpha}$ = 1.79A). 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 component phases, i.e. {110}, {100} and {211} planes for the bcc α-phase and {111}, {100} and {110} planes for the fcc γ-phase. Simulated transformations of the selected ideal orientations and experimental ODFs were carried out additionally to verify the orientation relation between major components of the ferrite and austenite rolling textures.

The phase composition and the initial morphology of two-phase structure after thermo- mechanical pre-treatment as well as the formation of the ferrite-austenite band-like structure in the course of cold-rolling were analysed by means of optical microscopy.

3. Results and discussion

3.1. Initial texture and structure

Texture measurements conducted after the solution treatment revealed the occurrence of strong initial textures in the α- and γ- component phases. The texture of the ferritic α-phase was the {100}<001> cubic texture, whereas the dominant component within the texture of the austenitic γ-phase was the {110}<001> Goss orientation (Fig.1a). A rotation of 45° around the successive <100> axes of the {100}<001>

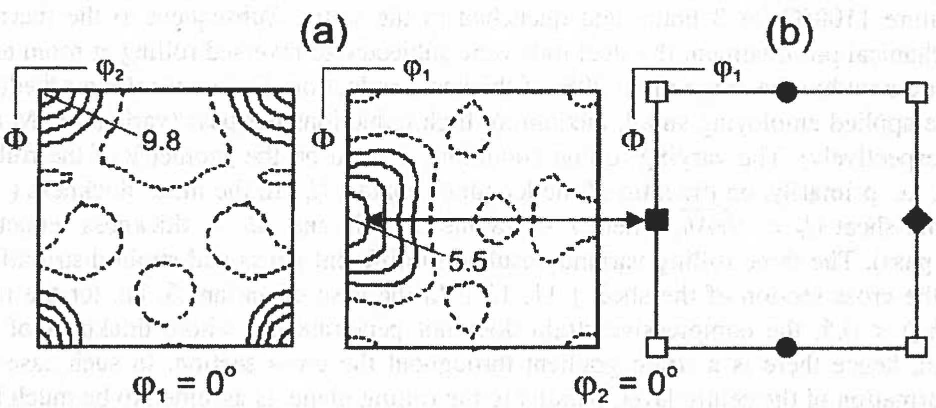


Fig. 1. The initial textures of ferrite and austenite after preliminary treatment in sections $\varphi_1 = 0°$ and $\varphi_2 = 0°$ respectively — (a) and the orientation variant from Bain relation (see Table II) describing crystallographic relationship between the major texture components — (b)

cubic orientation leads directly, according to Bain relation, to rotated cube, Goss or rotated Goss orientation [13] (Table II). Thus the crystallographic relation between the major components of the ferrite and austenite initial textures is described by one of the orientation variants from Bain relationship (Fig. 1b).

TABLE 2

Orientation variants resulting from 45°/<100> rotation of cubic orientation according to Bain relationship (see Fig. 1b)

Orientation			Rotation	Symbol
Cubic	Initial	(100)[001]	———	□
Rotated Cubic	Final	(100)[011]	45°/ [100]	•
Rotated Goss	Final	(101)[$\bar{1}$01]	45°/ [010]	◆
Goss	Final	($\bar{1}$01)[001]	45°/ [001]	■

The specific character of the initial textures in both phases was apparently the result of the applied thermo-mechanical pre-treatment. In comparison to rolling, there is a different strain path in the course of forging with compressive stresses altering between normal and transverse direction. Strong intensities of cube orientation within the ferrite texture after hot deformation of duplex stainless steel have been already reported, for example by Keichel et al. [2], in the case of so-called rod material subjected to forging. On the other hand, the comparison of the austenite textures for rod-materials with those for sheet-materials after hot-rolling indicated at strong intensities of Goss orientation in duplex steels after forging [2]. It is worth to note that the applied temperature of the solution treatment should not result in any substantial phase transformation within the structure of the steel. Hence the phase composition and the orientation relation between both phases were essentially established upon hot deformation.

Based on the metallographic analysis conducted after the solution treatment the volume fraction of ferrite (V_f) was estimated at about 60%. The ferritic α-phase was more continuous and constituted the matrix with islands of the austenitic γ-phase. Hence the two-phase structure under examination was not typical duplex one.

The initial morphology of the ferrite-austenite structure resulted from the applied type of hot- deformation process. Two-phase structures usually exhibit certain degree of anisotropy in the case of forging, however less pronounced in comparison to hot-rolling [1, 2]. On the longitudinal section of the forged rod of duplex steel under examination the austenite islands were elongated parallel to the main axis of the rod, however on the cross-section austenite grains were nearly equiaxed.

3.2. Formation of band-like structure

The anisotropy of two-phase structure produced upon hot deformation was strongly enhanced in the course of cold-rolling. Both component phases were plastically deformed and formed characteristic band-like structure consisting of alternate layers (flat shapes) of ferrite and austenite elongated parallel to the rolling plane. With increasing rolling reduction significant refinement of the microstructure was observed, with the thickness reduction of the austenite layers up to few micrometers. The bands of ferrite were thicker and exhibited more continuous character due to the phase composition of the steel (Fig. 2).

At higher degrees of deformation, starting from about 70÷80% of rolling reduction, the ferrite-austenite banded structure showed the wavy character accompanied locally by the corrugated shape of the (α/γ) interfaces with characteristic offsets (Figs. 2 a÷c). These structural effects may be regarded as a result of the non-homogeneous character of deformation connected with strain localization. It should be noted however that well-defined macroscopic shear bands were not revealed within the deformed ferrite-austenite microstructure. The extension of shear banding is apparently strongly restrained by the (α/γ) interfaces lying mostly parallel to the rolling plane [1, 2]. Based on the examination of slip behaviour in tensile tested ferrite-austenite two-phase bicrystals it was suggested that only slip transfer may occur across the (α/γ) interfaces, since Burgers vectors of slip dislocations in the bcc α-phase and the fcc γ-phase

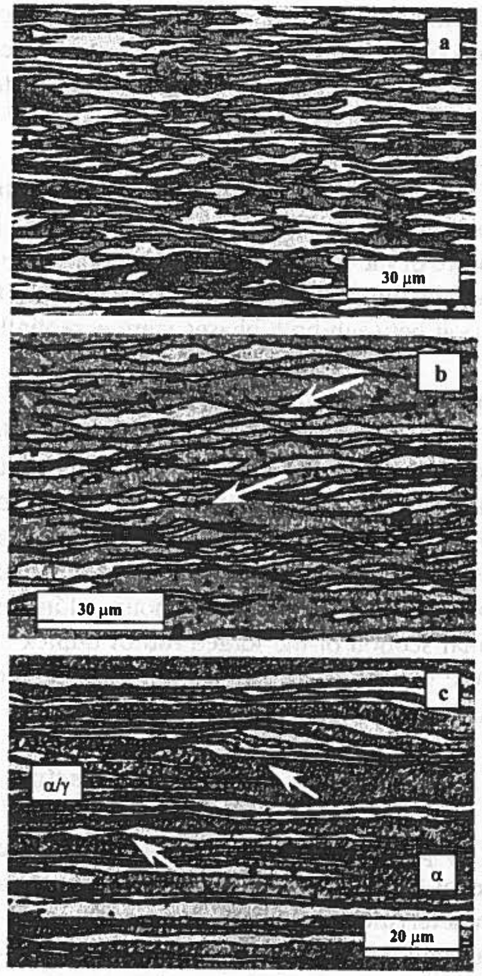


Fig. 2. Ferrite-austenite banded structure in cold-rolled duplex type steel on the longitudinal section (ND-RD) after 70%(a), 80%(b) and 90%(c) of thickness reduction (variants M, H and S respectively)

are different ($\bar{b}_\alpha$ = a/2<111> and $\bar{b}_\gamma$ = a/2<110>) [14]. Similar behaviour may be expected in duplex steel at higher strains in the case of interaction of deformation bands with phase interfaces. This assumption is consistent with the results of microstructure observations of cold-rolled ferrite-austenite steel reported by Blicharski [15]. The author indicates at clearly visible offsets at the (α/γ) interfaces at the points where deformation bands from ferrite areas impinge the interphase boundary. Simultaneously the bands of localized flow appeared in austenite grains running directly from the offsets at the boundary [15]. It seems that the interphase boundaries are strong barriers even for the co-ordinate dislocation motion throughout the band-like two-phase structure [2].

3.3. Rolling texture development

The major components of the ferrite and austenite initial textures, i.e. the cube and Goss orientations respectively, appeared relatively stable orientations within the whole range of deformations (30÷90% reduction) for rolling variant S and up to about 70÷80% of reduction for rolling schedules employing variants M and H (Figs. 3÷5). It should be noted that a significant spread of the initial orientations of both phases and the appearance of other texture components from typical orientation fibres was delayed and in consequence shifted to higher degrees of deformation in comparison to single phase steels [7÷10]. Additionally the intensities of certain characteristic texture components remained very weak up to high rolling reductions.

Although the {100}<001> cubic orientation is not a typical component for the ferrite rolling textures [7, 8], it remained within the texture of the α-phase up to high reductions for all employed rolling schedules, i.e. rolling variants S, M and H (Figs. 3÷5).

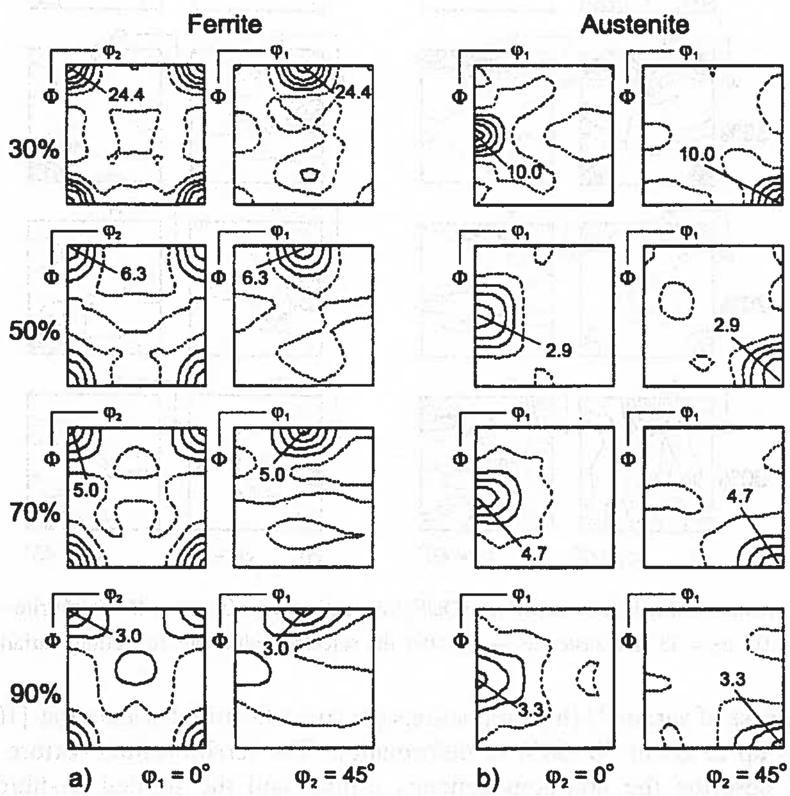


Fig. 3. Orientation distribution functions (ODFs) in section $\varphi_1 = 0°$, $\varphi_2 = 45°$ for ferrite — (a) and $\varphi_2 = 0°$, $\varphi_2 = 45°$ for austenite — (b) after the selected reductions for rolling variant S

For rolling variant S (small reductions per pass) the cubic orientation occurred the strongest texture component within the whole range of deformations. The {100}<001> dominant component of the ferrite texture was only accompanied by relatively week orientations close to {211}<011>. These orientations are laying within the spread of the α_1-fibre (<110>|| RD) and showed higher intensities at very high deformations, i.e. about 90% (Fig. 3a).

Upon rolling with medium reductions per pass (variant M) the {100}<001> orientation is the strongest component of ferrite texture up to 70÷80% of deformation. After 90% of rolling reduction the intensity of the ε-fibre (<001>|| ND) considerably weakens and the non-homogeneous α_1-fibre (<110>|| RD) appears, with the maximum intensity close to the (112)<110> and (111)<110> orientations (Fig. 4a).

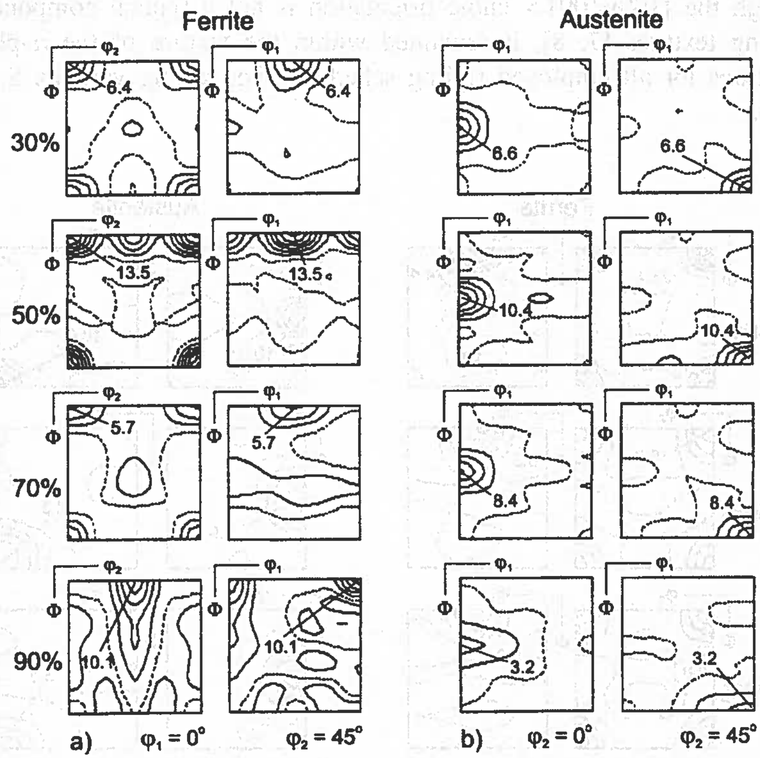


Fig. 4. Orientation distribution functions (ODFs) in section $\varphi_1 = 0°$, $\varphi_2 = 45°$ for ferrite — (a) and $\varphi_2 = 0°$, $\varphi_2 = 45°$ for austenite — (b) after the selected reductions for rolling variant M

In the case of variant H (high reductions per pass) the initial orientation {100}<001> dominates up to about 50÷60% of deformation. The ferrite rolling texture after 70% reduction describe the non-homogeneous ε-fibre and the limited α_1-fibre with the {100}<011> orientation as the strongest component. On the other hand, after 90% of deformation this maximum splits into the two components close to the (100)[0$\bar{1}$2] and (100)[021] orientations (Fig. 5a).

It should be noted therefore that in the case of rolling variants M and H the texture changes connected with the appearance of new orientations proceeded mainly above 70% of reduction, along the $\varepsilon = <001> \parallel$ ND and $\alpha_1 = <110> \parallel$ RD fibres (Figs. 4a and 5a).

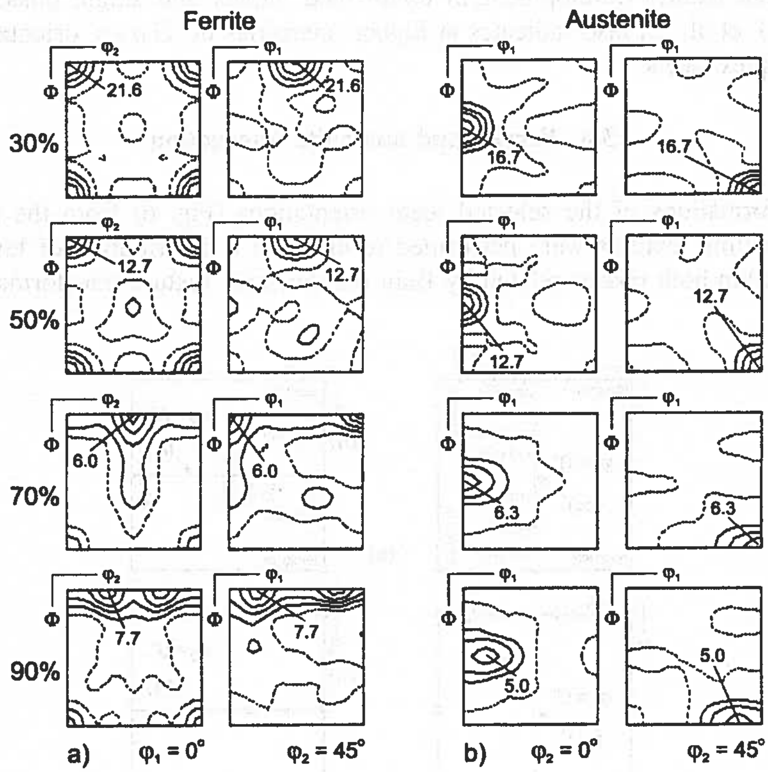


Fig. 5. Orientation distribution functions (ODFs) in section $\varphi_1 = 0°$, $\varphi_2 = 45°$ for ferrite — (a) and $\varphi_2 = 0°$, $\varphi_2 = 45°$ for austenite — (b) after the selected reductions for rolling variant H

In spite of the fact, that the {100}<011> orientation is the characteristic texture component of the cold-rolled single phase ferritic steels [7, 8], its intensity for the case of variants S and M remained very weak up to 90% of reduction (Figs. 4a and 5a). The change from the {100}<001> orientation into {100}<011> one requires 45° rotations of crystal lattice. It seems however that considerable lattice rotations were restrained due to the formation of ferrite-austenite banded structure in the course of rolling.

The texture development of the austenitic γ-phase exhibits comparatively small differences irrespective of the applied rolling schedule and was qualitatively similar to austenitic one-phase steels (ex. [9, 10]) and duplex steel with different initial textures [2÷6]. Relatively small texture changes proceeded along the $\alpha = <110> \parallel$ ND and $\eta = <001> \parallel$ RD fibres (Figs. 3b÷5b). The {110}<001> Goss orientation remained the strongest component of austenite texture within the whole range of deformations for rolling variants S and M, and up to about 70% of deformation for heavy reductions, i.e. variant H (Figs. 3b÷5b). In this case orientations from the limited α- fibre (< 110 >|| ND), including the {110}< 113 > and alloy type {110}< 112 > components, attained relatively high intensities in the range 80÷90% of reduction. Comparison of the austenite texture development in cold-rolled duplex and single phase steels by Keichel et al. [2] also indicates at higher intensities of Goss orientation in the case of duplex steels.

3.4. Ferrite and austenite interaction

Transformations of the selected ideal orientations (Fig. 6) from the ferrite and austenite rolling textures were performed to estimate a contribution of texture components within both phases related by Bain relationship. Texture transformations were

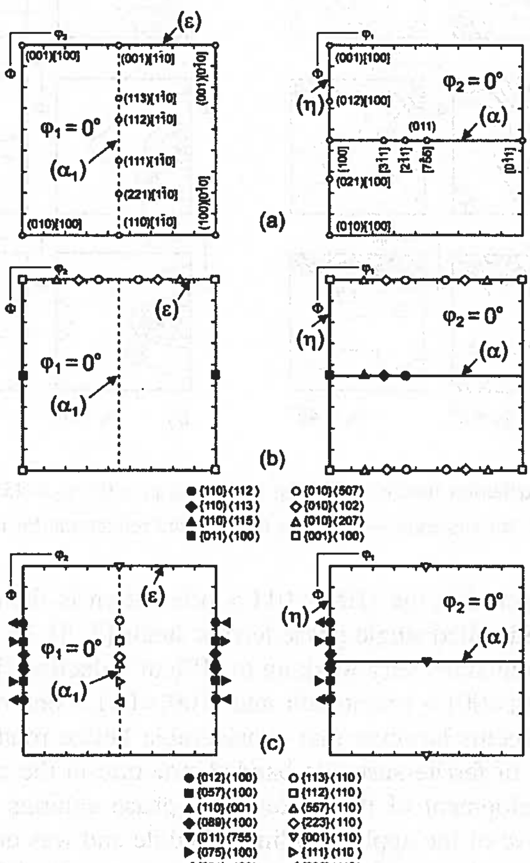


Fig. 6. The orientation fibres from the ferrite and austenite rolling textures — (a) and the crystallographic relations between the selected ideal orientations described by Bain relationship — (b, c)

conducted according to Bain relationship by applying variant selection, i.e. a rotation of 45° about the selected < 100 > axis. Comparison of the experimental ODFs (Figs. 3÷5) and transformed ideal ODFs (Fig. 6) indicates, that even at high deformations, Bain orientation relationship is still giving sufficient description of the crystallographic relation between the ferrite and austenite textures.

Estimation of the most stressed slip systems in both phases was performed to explain deformation behaviour within the strongly textured ferrite austenite band-like structure (Figs. 7 a÷c). Calculations of the ratio m = τ/σ (Fig. 7a) for the initial

(a)

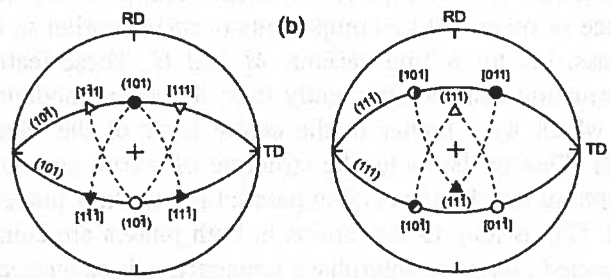
Ferrite - orientation (100)[001]

Type	Slip System	m = τ/σ	Symb
A3	$(011)[\bar{1}\bar{1}1]$	0.408	
A4	$(011)[1\bar{1}1]$	0.408	
B1	$(\bar{1}01)[111]$	0.816+	○ ▽
B4	$(\bar{1}01)[1\bar{1}1]$	0.816+	○ ▷
C1	$(0\bar{1}1)[111]$	0.408	
C2	$(0\bar{1}1)[\bar{1}11]$	0.408	
D2	$(101)[\bar{1}11]$	0.816+	● ▼
D3	$(101)[\bar{1}\bar{1}1]$	0.816+	● ▶
E2	$(110)[\bar{1}11]$	0.408	
E4	$(110)[1\bar{1}1]$	0.408	
F1	$(\bar{1}10)[111]$	0.408	
F3	$(\bar{1}10)[\bar{1}\bar{1}1]$	0.408	

Austenite - orientation (110)[001]

Type	Slip System	m = τ/σ	Symb
A3	$(\bar{1}11)[101]$	0.408	
A4	$(\bar{1}11)[0\bar{1}1]$	0.408	
A5	$(\bar{1}11)[110]$	0.000	
B1	$(111)[\bar{1}01]$	0.816+	△ ⊘
B4	$(111)[0\bar{1}1]$	0.816+	△ ○
B6	$(111)[\bar{1}10]$	0.000	
C1	$(1\bar{1}1)[\bar{1}01]$	0.408	
C2	$(1\bar{1}1)[011]$	0.408	
C5	$(1\bar{1}1)[110]$	0.000	
D2	$(\bar{1}\bar{1}1)[011]$	0.816+	▲ ●
D3	$(\bar{1}\bar{1}1)[101]$	0.816+	▲ ◑
D6	$(\bar{1}\bar{1}1)[\bar{1}10]$	0.000	

(b)



(c)

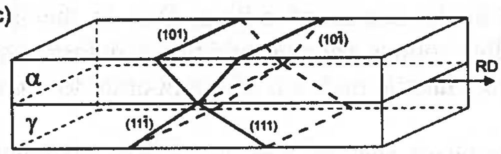


Fig. 7. Values of the relative shear stresses (m = τ/σ) for the (100)[001] and (110)[011] orientations of ferrite and austenite — (a); the stereographic projections and the schematic representation of the most stressed slip systems — (b) and (c) respectively

orientation {100}<001> of ferrite indicate that the highest values of the relative shear stresses (m = 0,816) are in four slip systems. Deformation of ferrite may proceed in two slip planes, the (101) and ($\bar{1}$01), which are symmetrical in relation to <001>|| RD. Similarly, the deformation of austenite with strong {110}< 001 > initial texture may proceed by operation of four equally stressed slip systems, having the same values of the relative shear stresses (m=0,816). Also in this case both active slip planes, the (111) and $(\bar{1}\bar{1}1)$, are symmetrical with respect to <001>|| RD (Figs. 7b, 7c). Taking into account the symmetry of potential slip systems in both phases with respect to RD and the equal values of the relative shear stresses ($m = \tau/\sigma$) in these systems, deformation of the ferrite-austenite banded structure may be regarded as plastically compatible [14]. Such a situation may hold until both major orientations, i.e. the {100}< 001 > cubic orientation and the {110}< 001 > Goss orientation, are dominant components in the rolling textures of the α- and γ- phases respectively.

It is suggested that the operation of the symmetrical slip systems was forced due to the band-like character of the ferrite-austenite structure, which seems strong obstacle for significant orientation changes and resulted in stability of the major texture components of both phases. Thus the texture changes observed upon cold-rolling resulted from relatively small lattice rotations within the whole range of deformations in the case of smaller imposed external strains (variant S) and up to relatively high reductions for rolling variants M and H. It appears therefore that the initial orientations and the morphology of the two-phase structure exert considerable effect on the rolling texture formation of the α- and γ-phases.

When analysing the texture changes in both component phases it appears, that in comparison to austenite, the development of ferrite rolling texture was less continuous and the appearance of other texture components occurred earlier in the case of higher reductions per pass, i.e. for rolling variants M and H. These features of the ferrite rolling texture formation resulted apparently from the accommodation of the imposed external strains, which were higher in the centre layer of the sheet for the case of variants M and H. Due to the band-like structure of ferrite and austenite the largest part of the (α/γ) phase interfaces is laying parallel to the sheet plane and, according to Keichel et al. [2], as long as the strains in both phases are comparable no strong interaction is expected along the interphase boundaries. It suggested however that the observed changes of the ferrite texture were affected by varying strain partitioning between both phases in the course of rolling. Due to the faster hardening rate of austenite and its smaller volume fraction additional deformation mechanisms had to occurred within the more ductile matrix α-phase in order to accommodate the imposed strains.

The results of the phase analysis and the texture examination give no evidence of any significant role of deformation induced ($\gamma \rightarrow \alpha$) phase transformation and its contribution into the texture development. This observation is consistent with the results concerning deformation behaviour and phase transformation in cold-rolled textured polycrystals and single crystals of austenitic steel having the {110}< 001 > Goss orientation [10, 16].

4. Concluding remarks

Texture analysis in cold-rolled duplex type steel with strong initial textures of ferrite and austenite indicates at the significant effect of the initial orientations as well as the band-like morphology of two-phase structure on the formation of the final rolling textures of both component phases. The development of the rolling textures in the α- and γ- phases considerably differs from single phase steels as well as duplex steels with random initial textures after preliminary treatment.

The occurrence of strong cubic texture in ferrite and dominant G o s s orientation in austenite texture, which are related through Bain relationship, resulted apparently from the applied thermo-mechanical pre-treatment including forging. Depending on the subsequent cold-rolling schedule the $\{100\}<001>$ orientation from the initial texture of ferrite and the $\{110\}<001>$ orientation from the austenite texture remained the dominant or major texture components within the whole (variant *S*) or extended range of deformations (variants *M* and *H*). A significant spread of the initial orientations and the appearance of other texture components in the ferrite and austenite rolling textures were delayed and shifted to higher rolling reductions.

It is concluded that the strong initial textures of both phases, with Bain relationship describing orientation relation between the major texture components, enabled plastically compatible deformation of ferrite and austenite. Additionally the specific band-like morphology of two-phase structure formed in the course of rolling restrained considerable lattice rotations upon deformation. Both factors resulted in the stability of the major texture components up to high rolling reductions and considerably changed the texture development of the examined ferritic-austenitic steel.

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