



Hot-working in the $\gamma+\alpha$ region of TRIP-aided microalloyed steel

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ABSTRACT

Purpose: The aim of the paper is to investigate the influence of hot-working in the $\gamma+\alpha$ range and isothermal holding temperature in a bainitic range on a stability of retained austenite in a TRIP-aided microalloyed steel.

Design/methodology/approach: The thermomechanical processing in the $\gamma+\alpha$ range to obtain multiphase structures with the retained austenite in a microalloyed steel was realized. It consisted of plastic deformation of specimens at 750°C or 780°C, oil cooling and isothermal holding in a bainitic region. Degree of deformation was 28 or 50%. To reveal the multiphase structure optical and transmission electron microscopy were used. X-ray diffraction method was employed to determine a volume fraction of retained austenite.

Findings: It was found that hot-working in the two-phase region influences essentially a multiphase structure of investigated steel. The ferrite fraction is comparable for heat-treated and thermo-mechanically processed specimens but the ferrite grain size is twice smaller in a case of plastically-deformed specimens. The optimum isothermal holding temperature in a bainitic range is 300°C, independent on austenitizing temperature. The specimens forged in the $\gamma+\alpha$ range and isothermally held at this temperature made it possible to obtain about 10% of retained austenite.

Research limitations/implications: Investigations concerning the influence of isothermal holding time in a bainitic range on the stability of retained austenite should be carried out.

Practical implications: The established conditions of the thermomechanical processing can be useful in a development of the hot-rolling technology for TRIP-aided microalloyed steels.

Originality/value: The realized thermomechanical processing enabled to obtain about 10% fraction of retained austenite in a steel containing 0.5% Si.

Keywords: Metallic alloys, TRIP-aided steel, Thermomechanical processing, Retained austenite

MATERIALS MANUFACTURING AND PROCESSING

1. Introduction

Lowering the fuel consumption to reduce the amount of gas emission is one of the priorities of the automotive industry. The main methods to lower fuel consumption are the efficiency improvement of engine and driving system, the reduction of running resistance and especially the weight reduction of autobody [1]. The weight reduction of automobile can be

achieved by the optimization of body structure, the reduction of vehicle size and the replacing of conventional materials with materials of the lower density. For this reason, the growing significance in the automotive industry have aluminium, magnesium and titanium alloys as well as composite materials [2,3]. However, one of the main factors deciding about material applying is the manufacturing economics. In this aspect, a lot of modern steel grades have still the highest potential of the

application in the automotive industry. Various demands of the automotive industry concerning mechanical and technological properties of steel sheets used for many elements of vehicles are performed by structural steels with microadditions [4,5], Interstitial Free steels and Bake Hardenable steels [6,7]. Several mechanisms can be used to strengthen the ferritic matrix of these steels, i.e. grain refinement, solid solution hardening, precipitation strengthening, strain hardening, texture optimization. An important solution for manufacturing lighter cars was using the steels with a ferritic – martensitic structure [8,9]. They have advantages of composite materials. When these steels deform, strain is concentrated in the lower strength phase, creating the unique high work hardening rate exhibited by these steels.

Further improvements in strength, ductility and safety of cars can be achieved by TRIP-aided (Transformation Induced Plasticity) steels [10-12]. They have a multiphase ferritic – bainitic structure with the retained austenite of the volume fraction from 5 to 20%. It was found [13-18] that the retained austenite transforms to martensite during a tensile test or technological forming, strengthening the regions which are about to reduce in cross section and fracture. It prevents a localization of plastic deformation and delays necking. Moreover, the retained austenite transforming progressively to martensite with increasing strain, increases the work hardening rate at higher strain levels, providing a slight advantage over DP steels. This behaviour enables to manufacture elements of a complicated shape. When some fraction of the retained austenite is present after forming, it transforms to martensite during subsequent plastic deformation, such as a crash event, providing excellent crash energy absorption.

A significant volume fraction of the retained austenite can be obtained in a final multiphase structure due to C, Mn and Si elements and a properly realized multi-stage heat treatment consisting of intercritical annealing of a cold rolled sheet and isothermal bainitic transformation [10,13,16] or hot rolling combined with a slow cooling during $\gamma \rightarrow \alpha$ transformation and coiling in the bainitic range [9]. The carbon concentration in the TRIP steels is most often close to about 0.2% and manganese and silicon are added at concentrations of about 1 to 2% [11,13,14]. A thermal stabilization of retained austenite depends strongly on a Si concentration, suppressing carbide precipitation during an isothermal holding in a bainitic range [13,14,16]. However, manufacturing the steels of increased silicon content involves technological problems connected with casting, a surface quality due to appearance of red scale defects and hot dip galvanizing, caused by a lack of wetting. For these reasons alternative compositions of steels with a reduced content of silicon are considered [16,17]. Investigations concerning a partial or total silicon substitution by Al and P are also carried out [18,19]. Apostolos et al. reported [20] that a stabilization of retained austenite can be increased by a refinement of the austenite grain. Using the thermomechanical processing during the manufacturing process of multiphase sheets is one of the best methods for the grain refinement. The thermomechanical treatment consists in rolling in the austenite region and cooling from a finishing rolling temperature [21,22]. The further increasing the thermal stability of retained austenite can be obtained by finishing the hot-working of steels in a $\gamma + \alpha$ range [23]. However, there are only few works concerning the manufacturing TRIP-aided steels with a decreased silicon concentration and not containing substitute elements [17,24]. Moreover, there is a lack of enough data concerning low-silicon steels with Nb and Ti microadditions.

2. Experimental procedure

The microalloyed steel with a chemical composition given in Table 1 was investigated. The steel contains a decreased silicon concentration in comparison with usually used TRIP-aided steels containing about 1.5%Si. The steel is characterized by high metallurgical purity connected with low concentrations of impurities, i.e. S and P. Melting was carried out in a Balzers' VSG-50 vacuum induction furnace. In order to modify non-metallic inclusions rare-earth elements were used. It was earlier found [21] that they have a positive influence on an isotropy of mechanical properties due to the prevention of elongation of non-metallic inclusions during hot-working. Moreover, fine non-metallic inclusions can control a grain growth during hot-rolling [25]. It is important for obtaining a fine-grained structure of primary austenite. For this reason, the steel contains microadditions of Nb and Ti forming stable, dispersive nitrides, carbides and carbonitrides of MX-type. They favour precipitation hardening and first of all act as obstacles for an austenite grain growth [4]. The fine-grained austenite cooled from a finishing hot-working temperature under controlled conditions transforms into fine-grained products of supercooled austenite. It is beneficial for plates and sheets with ferritic – pearlitic or martensitic structures [5,26] and can be advantageous in a case of TRIP-aided steels due to two reasons. Firstly, the mechanical properties are improved due to a refinement of ferritic – bainitic structure and secondly, the fine grains of retained austenite can increase its thermal stability during cooling to room temperature and prevent decreasing a volume fraction of stable retained austenite because of decreased silicon concentration.

Table 1.
Chemical composition of the investigated steel

Mass contents in percentage (%)								
C	Mn	Si	Al	Nb	Ti	P	S	N
0.26	1.44	0.50	0.04	0.027	0.012	0.016	0.008	0.0042

Casting was performed in the argon atmosphere into ingot moulds of 25 kg volume. The ingots were forged to obtain the rods of 24x24 mm, subjected to homogenizing at a temperature of 1200°C in the N₂ atmosphere. Machining was used in a final stage of a preparation of the specimens with a diameter and a height of 14 mm.

The routes designed for the heat treatment and thermomechanical processing of the specimens were schematically illustrated in Fig. 1. In order to design properly thermomechanical treatment conditions, the knowledge of critical temperatures A_{c3}, A_{c1} and M_s as well as B_s for the γ phase is needed. The temperatures were calculated on the basis of the chemical composition of the steel using Andrews' relationships [27]:

$$A_{c3} = 910 - 203 \cdot \sqrt{C} - 15.2 \cdot \text{Ni} + 44.7 \cdot \text{Si} + 104 \cdot \text{V} + 31.5 \cdot \text{Mo} \quad (1)$$

$$A_{c1} = 723 - 10.7 \cdot \text{Mn} - 16.9 \cdot \text{Ni} + 29.1 \cdot \text{Si} + 16.9 \cdot \text{Cr} + 290 \cdot \text{As} \quad (2)$$

$$M_s = 539 - 423 \cdot C_\gamma - 30 \cdot \text{Mn} - 17.7 \cdot \text{Ni} - 12.1 \cdot \text{Cr} - 7.5 \cdot \text{Mo} \quad (3)$$

$$B_s = 830 - 270 \cdot C_\gamma - 90 \cdot \text{Mn} - 37 \cdot \text{Ni} - 70 \cdot \text{Cr} - 83 \cdot \text{Mo} \quad (4)$$

where: Mn, Ni, Si, Cr, Mo – mass contents of the elements in the investigated steel, C_γ – carbon concentration in the austenite.

The calculated temperatures are the following: $A_{c3} = 828^\circ\text{C}$, $A_{c1} = 722^\circ\text{C}$, $M_s = 385^\circ\text{C}$ and $B_s = 630^\circ\text{C}$. However, it should be noted that both in case of martensite and bainite start temperatures the carbon enrichment of the γ phase during annealing in the $\gamma+\alpha$ range should be taken into consideration. The $M_{s\gamma}$ and $B_{s\gamma}$ temperatures of austenite are decreased then. Moreover, the martensite start temperature for the retained fraction of austenite further decreases during isothermal holding of specimens in a range of bainitic transformation. The authors of the works [19,24] indicate that a stabilization of retained austenite is possible when the carbon concentration in this phase is over 1%.

The heat treatment and thermomechanical processing consisted of austenitizing of specimens at a temperature of 750°C or 780°C , oil cooling and holding for 600 s in a region of isothermal bainitic transformation. A part of the specimens were upset forged with a degree of deformation of 28% or 50% and a strain rate of about 30s^{-1} . An eccentric press PMS50 with flat dice was used for forging tests. An essential step of the thermomechanical treatment is a selection of a suitable cooling rate between the annealing and holding stages. The specimens were oil cooled with a rate of about 90°C/s , which is sufficiently high to prevent the formation of pearlite.

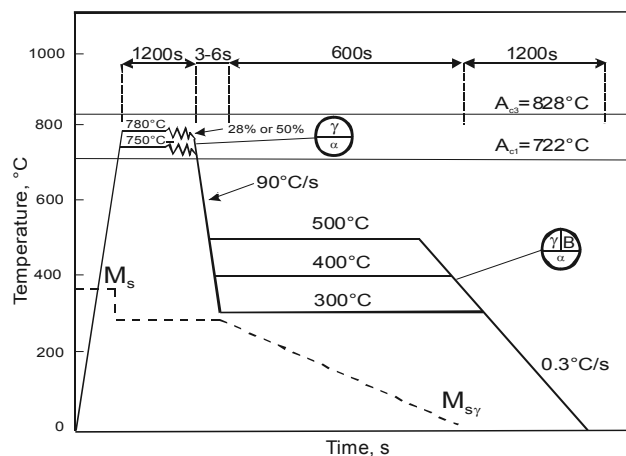


Fig. 1. Schematic representation of the thermomechanical processing of the investigated steel, M_s – martensite start temperature of the steel, $M_{s\gamma}$ – martensite start temperature of the austenite with increased carbon concentration

Optical microscopy was used to a structure identification of processed specimens. Etching was performed using LePera technique [28]. The etchant was a mixture of two solutions, i.e. 1 g $\text{Na}_2\text{S}_2\text{O}_5$ in 100 ml H_2O and 4 g dry picric acid in 100 ml of ethanol. Before the etching, the reagents were mixed and specimens were immersed in the mix for 15 s.

The Leica MEF4a light microscope was used for observations of the multiphase structures and to determine a volume fraction of ferrite. The microstructure was also categorized by using thin-foil transmission electron microscopy and X-ray diffraction. Observations of thin foil structure were carried out in the JEM-200CX transmission electron microscope using an accelerating voltage of 120 kV. To identify a kind of phases forming a multiphase structure of the steel the DRON2.0 X-ray diffractometer was used. The amount of retained austenite was measured by X-ray diffraction from the integrated intensities of diffraction peaks $(200)\gamma$, $(220)\gamma$, $(311)\gamma$, $(200)\alpha$ and $(211)\alpha$ by using $\text{Co K}\alpha$ radiation.

3. Results and discussion

The various conditions of the heat treatment and thermomechanical processing affect strongly a structure of the specimens. The structures shown in Figs. 2-8 contain ferrite, bainite and retained austenite of different volume fraction, grain size and morphology. The structural features depend on the austenitizing and deformation temperature, isothermal holding temperature in a bainitic range and degree of deformation. In a case of the heat-treated specimens, the retained austenite fraction of 8.2% is the largest for an isothermal holding temperature of 300°C (Fig. 2). Increasing the bainitic transformation temperature to 400 and 500°C results in decreasing the volume fraction of the stable γ phase to 4.6% and 3.5% – respectively. The white grains of retained austenite have a various size. Some of them with a grain size up to $2\text{ }\mu\text{m}$ are located on boundaries of ferrite grains. However, most of retained austenite grains have a grain size of about $1\text{ }\mu\text{m}$ and can be observed on interfaces of ferrite and bainite.

Applying the hot-working in the $\gamma+\alpha$ range results in a distinct refinement of the obtained multiphase structure (Figs. 3-5). The grain size of the α phase is about $4\text{ }\mu\text{m}$ and twice smaller in comparison with the heat-treated specimens. They are slightly elongated in a direction of plastic flow. It indicates that after the plastic deformation, the strain hardening can be only slightly removed by static recovery occurring before fast cooling to a range of bainitic transformation temperature. The volume fraction of ferrite is comparable for heat-treated and thermomechanically processed specimens and changes from 63 to 70% (Table 2). A result of the structure refinement in a two-phase region is also a fine-grained structure of ferritic bainite. The largest fraction of the retained austenite obtained after isothermal holding of specimens at a temperature of 300°C (Fig. 3). It is analogous compared to heat-treated steels. At this temperature the bainite transformation kinetics rate is relatively low due to a slow diffusion of the carbon out of the bainite sheaves. However, a fraction of retained austenite averages 10% and is higher in comparison with the specimens heat-treated under comparable conditions. Taking into account comparable volume fractions of ferrite and a resulting carbon content in the austenite, a reason for a higher thermal stability of retained austenite can be found in its smaller grain size. The islands of retained austenite are uniformly distributed in a ferritic – bainitic matrix and have a grain size smaller than $1\text{ }\mu\text{m}$.

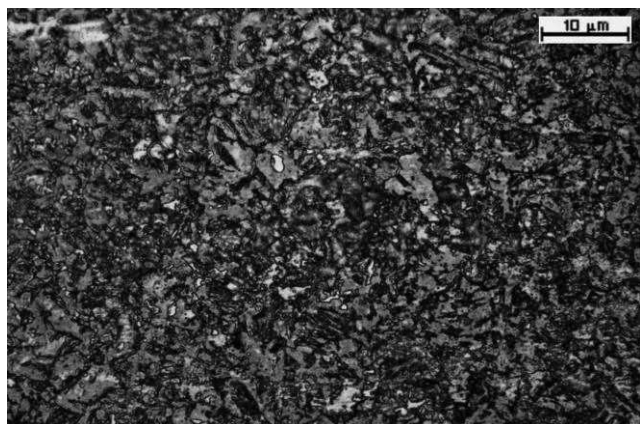


Fig. 2. Ferritic – bainitic structure with the retained austenite of the heat-treated steel isothermally held for 600 s at a temperature of 300°C



Fig. 5. Ferritic – bainitic structure with the retained austenite of the steel 28% deformed at 750°C and isothermally held for 600 s at a temperature of 500°C

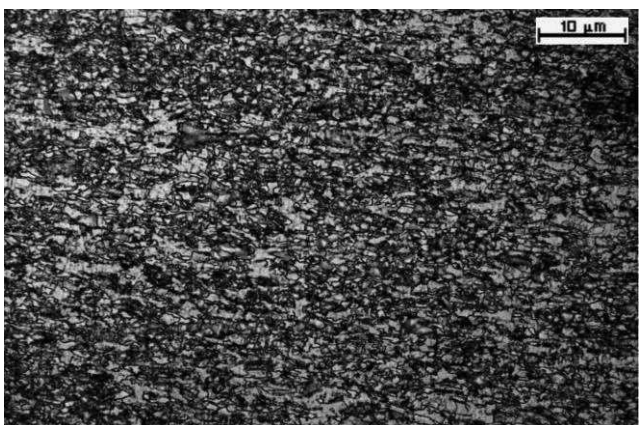


Fig. 3. Ferritic – bainitic structure with the retained austenite of the steel 28% deformed at 750°C and isothermally held for 600 s at a temperature of 300°C



Fig. 6. Ferritic – bainitic structure with the retained austenite of the steel 50% deformed at 750°C and isothermally held for 600 s at a temperature of 300°C

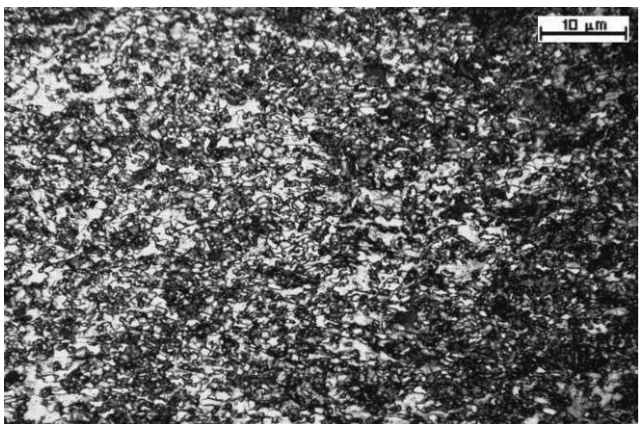


Fig. 4. Ferritic – bainitic structure with the retained austenite of the steel 28% deformed at 750°C and isothermally held for 600 s at a temperature of 400°C



Fig. 7. Ferritic – bainitic structure with the retained austenite of the steel 50% deformed at 750°C and isothermally held for 600 s at a temperature of 400°C



Fig. 8. Ferritic – bainitic structure with the retained austenite of the steel 50% deformed at 780°C and isothermally held for 600 s at a temperature of 300°C

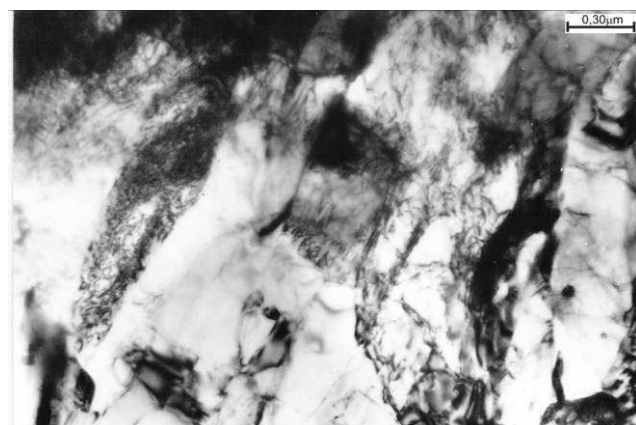


Fig. 9. Fine elongated grains of the retained austenite located in ferritic bainite ($T_D = 750^\circ\text{C}$, $\varepsilon = 28\%$, $T_B = 300^\circ\text{C}$)



Fig. 10. Structure of ferritic bainite with the retained austenite located between its laths and on a boundary of bainite and ferrite with an increased dislocation density ($T_D = 750^\circ\text{C}$, $\varepsilon = 28\%$, $T_B = 300^\circ\text{C}$)

Table 2.

Volume fractions of structural constituents of the heat-treated and thermo-mechanically processed steels

Parameters of the heat treatment and thermomechanical processing				Ferrite fraction %	Bainite fraction %	RA fraction %
$T_A, ^\circ\text{C}$	$T_D, ^\circ\text{C}$	$\varepsilon, \%$	$T_B, ^\circ\text{C}$			
750	-	-	300	69.8	22.0	8.2
750	-	-	400	70.4	25.0	4.6
750	-	-	500	68.9	27.6	3.5
780	-	-	300	67.7	28.0	4.3
780	-	-	400	66.4	30.0	3.6
780	-	-	500	65.4	31.2	3.4
750	750	28	300	64.9	25.0	10.1
750	750	28	400	68.8	25.0	6.2
750	750	28	500	68.6	25.5	5.9
780	780	28	300	63.7	31.0	5.3
780	780	28	400	65.2	30.0	4.8
780	780	28	500	64.3	32.0	3.7
750	750	50	300	62.9	28.0	9.1
750	750	50	400	62.7	31.0	6.3
750	750	50	500	67.8	28.0	4.2
780	780	50	300	63.1	30.5	6.4
780	780	50	400	65.8	30.5	3.7
780	780	50	500	63.1	34.0	2.9

T_A – austenitizing temperature, T_D – deformation temperature, ε – degree of deformation, T_B – isothermal holding temperature in a bainitic range, RA – retained austenite

Increasing the isothermal holding temperature to 400°C did not change a kind of the obtained multiphase structure but a fraction of retained austenite decreased to 6.2% (Fig. 4). It corresponds to a decrease in the amount of fine grains of the γ phase. Further increasing the bainitic transformation temperature to 500°C had a detrimental influence on the volume fraction of retained austenite (Fig. 5). It is due to a faster diffusion rate of carbon and corresponding carbide precipitation in a ferritic bainite.

Hot-working in the $\gamma+\alpha$ range with a higher degree of deformation ($\varepsilon=50\%$) led in obtaining ferritic – bainitic structures with a fraction of retained austenite similarly dependent on an isothermal holding temperature (Table 2). The largest fraction of the stable γ phase exceeding 9% obtained under conditions of the slowest bainite transformation rate (Fig. 6). It can be observed that statically recovered ferrite grains show a considerable elongation in a direction of the plastic flow and have a much higher diversification of their size (Fig. 7). The fraction of ferrite is comparable with heat-treated specimens and thermo-mechanically processed specimens. The retained austenite grains have a similar size of about 1 μm but some of them are slightly elongated, too.

Increasing the hot-working temperature to 780°C resulted in insignificant decreasing the ferrite fraction (Table 2). The grain size of the α phase is comparable, too. However, a volume fraction of retained austenite decreased to about 6.5% for the best conditions of the thermomechanical processing (Fig. 8). The retained austenite grains have a comparable size and a shape compared to specimens deformed at 750°C.

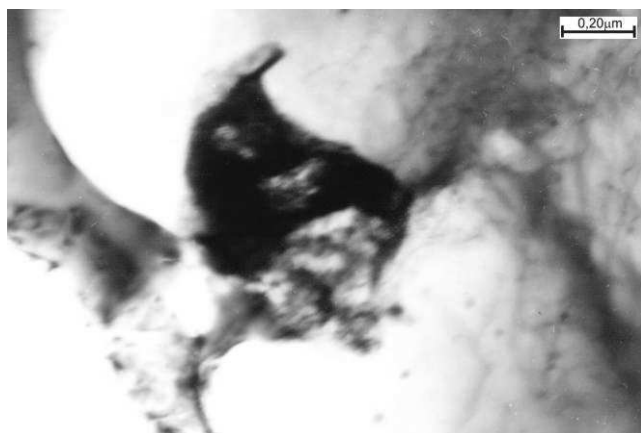


Fig. 11. Retained austenite located on a boundary of three ferrite grains with an increased dislocation density and dispersive precipitations of MX-type phases ($T_D = 750^\circ\text{C}$, $\varepsilon = 28\%$, $T_B = 400^\circ\text{C}$)

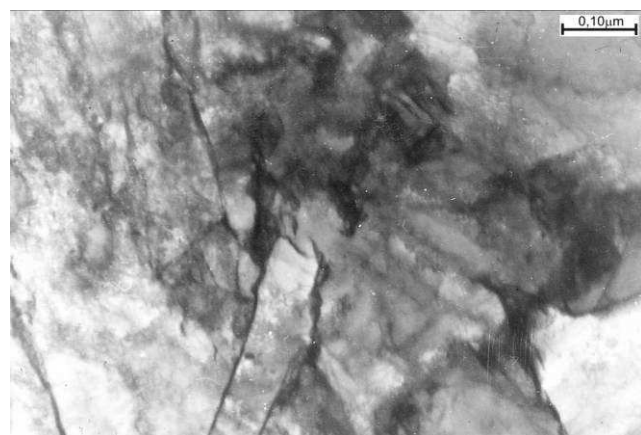


Fig. 12. Fine cementite particles between bainite laths ($T_D = 750^\circ\text{C}$, $\varepsilon = 28\%$, $T_B = 400^\circ\text{C}$)

Figure 9 is a transmission electron microscopy micrograph showing a morphology of retained austenite for the specimen plastically deformed at a temperature of 750°C and isothermally held for 600 s at a lowest bainite transformation temperature of 300°C . It can be observed that the retained austenite has a shape of elongated grains located on the boundaries of ferrite with the increased density of dislocations. The retained austenite can also occur as fine, elongated grains between ferritic bainite laths or on the interfaces of ferrite and bainite (Fig. 10). The ferritic bainite laths do not contain cementite particles. However, a few fine carbides can be seen on the boundary of two ferrite grains. They were only observed locally.

The α phase is characterized by an increased density of dislocations near bainite islands connected with a partially diffusionless character of the bainite transformation (Figs. 9,10). A matrix of the steel is the ferrite with a small density of dislocations generated during hot-working in a two-phase region (Fig. 11).

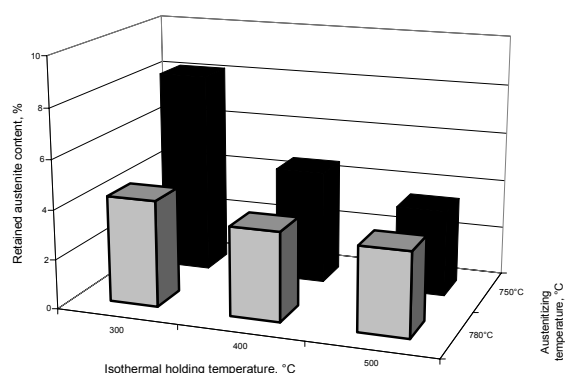


Fig. 13. Influence of the austenitizing temperature and isothermal holding temperature in a bainitic range on a fraction of retained austenite

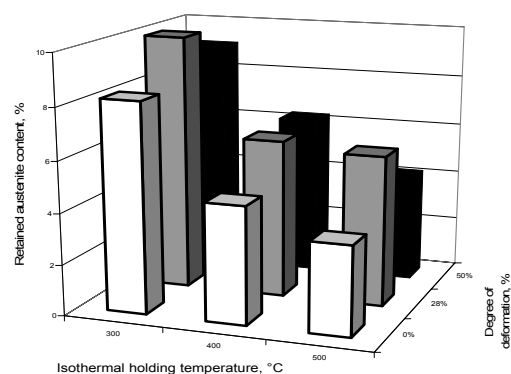


Fig. 14. Influence of the degree of deformation in the $\gamma+\alpha$ region and isothermal holding temperature in a bainitic range on a fraction of retained austenite

Increasing the holding temperature of steel in a bainitic range to 400°C do not change substantially a morphology of obtained structural constituents of the heat-treated and thermo-mechanically processed steels. It was found that the retained austenite can also occur in a form of fine, globular grains with a size of about $0.5\ \mu\text{m}$ located on boundaries of few ferrite grains (Fig. 11). It can be seen that ferrite grains contain also fine, dispersive precipitations – probably interstitial phases with titanium and niobium. At an isothermal holding temperature of 400°C fine cementite particles are identified quite often (Fig. 12). It has a detrimental effect on a stabilization of retained austenite, what was confirmed by the quantitative X-ray phase analysis.

It can be summarized that the higher fractions of retained austenite can be obtained for the specimens austenitized and plastically deformed at a temperature of 750°C (Fig. 13). It is due to a higher concentration of carbon in the austenite (about 0.6%) in a case of the specimens austenitized at a temperature slightly higher than A_{c1} of the steel. The influence of the isothermal holding temperature is evident and connected with a faster diffusion rate of carbon with increasing the bainitic transformation temperature. Hot-working in the $\gamma+\alpha$ range leads to a clear grain refinement of the multiphase structure and a higher stabilization of the γ phase to room temperature (Fig. 14).

4. Conclusions

The investigated steel containing 0.5%Si and microadditions of Nb and Ti can be used for manufacturing the products with a ferritic – bainitic structure and a required fraction of the stable retained austenite. Despite a decreased silicon concentration the used heat treatment conditions made it possible to obtain about 8% of retained austenite to room temperature. The optimum austenitizing temperature was 750°C, for which the carbon concentration in the austenite was about 0.6%. The further enrichment in carbon of the austenite proceeded under conditions of the isothermal holding in a bainitic range. The best temperature of the bainitic transformation in respect of maximum retained austenite volume fraction was 300°C. At this temperature the bainite transformation rate is sufficiently low due to a slow diffusion of the carbon out of the bainite.

The hot-working in the two-phase region resulted in distinct refinement of the multiphase structure. The ferrite fraction was comparable with the heat-treated specimens but a grain size of the α phase was twice smaller. The islands of the ferritic bainite was smaller, too. However, the most important effect observed for the retained austenite stabilization. The increase in the amount of retained austenite fraction up to 10% for an optimum isothermal holding temperature of 300°C was obtained. The thermal stabilization of the higher fraction of retained austenite to room temperature was due to significant decreasing its grain size and probably a higher enrichment of retained austenite in carbon during plastic deformation in the $\gamma+\alpha$ region. The retained austenite has a shape of elongated grains located on the boundaries of ferrite with the increased density of dislocations or occurs as fine, elongated grains between ferritic bainite laths or on the interfaces of ferrite and bainite. The carbides were observed at 300°C only locally.

The established conditions of the thermomechanical processing can be useful in a development of the hot-rolling technology for TRIP-aided microalloyed steels. However, the stabilization of a required, high fraction of retained austenite is possible only under precisely controlled conditions of the hot-working and cooling from a finishing rolling temperature.

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