Microstructural study on retained austenite in advanced high-strength multiphase 3Mn-1.5Al and 5Mn-1.5Al steels

A. Grajcar*, W. Kwaśny

Institute of Engineering Materials and Biomaterials,
Silesian University of Technology, ul. Konarskiego 18a, 44-100 Gliwice, Poland
* Corresponding e-mail address: adam.grajcar@polsl.pl

Materials

ABSTRACT

Purpose: The aim of the paper is to describe crystallographic and morphological features of retained austenite in thermomechanically processed bainite-martensite multiphase steels containing 3 and 5% Mn.

Design/methodology/approach: Two groups of steels were designed and investigated: 3Mn-1.5Al and 5Mn-1.5Al were reference steels, whereas next two steels were microalloyed with niobium. The steels were thermomechanically processed using the Gleeble simulator. The isothermal holding temperature to enrich austenite in carbon was between 350 and 450°C. Metallographic investigations were carried out using light (LM) and scanning electron microscopy (SEM). The retained austenite amount and its carbon concentration was evaluated by X-ray analysis.

Findings: Manganese addition results in the high hardenability of steels leading to bainitic-martensitic microstructures. A high-Al concept and isothermal holding of steel in a bainitic transformation range allow to obtain a high fraction of retained austenite as a result of an incomplete bainitic transformation phenomenon. New complex bainitic morphologies like degenerate upper and lower bainite were identified using SEM. The microstructure and retained austenite characteristics were correlated with the carbon content in γ phase.

Research limitations/implications: Further investigations (TEM, EBSD) to describe in detail the identified structural constituents and the effect of Nb microalloying on microstructure and mechanical properties are needed.

Practical implications: The knowledge of the influence of the isothermal holding temperature on the microstructure and hardness of thermomechanically processed steels are of primary importance for hot rolling of these multiphase high-strength steels.

Originality/value: A problem of the stabilization of retained austenite in advanced high-strength multiphase Nb-free and Nb-microalloyed steels with increased Mn content is discussed.

Keywords: Metallic alloys; TRIP steel; Retained austenite; Nb microalloying; Thermomechanical processing; X-ray quantitative analysis

1. Introduction

Austenite is a very useful structural constituent of advanced high-strength steels (AHSS) used in the automotive industry. Its ability for strengthening in many different ways offers possibilities to obtain a unique range of mechanical and technological properties of steel sheets used for different elements of a car structure. An especially important feature of γ phase is its ability to transform to martensite or to form twinned microstructures during straining. The first one is utilized in low-alloyed multiphase steels containing from 5 to 15% volume amount of retained austenite, which are alloyed with 0.1-0.25% C, 1.5% Mn and up to 2% of Si, Al, Cr and Mo in various combinations [1-5]. In turn, twinning is a main hardening mechanism (often together with strain-induced martensitic transformation) in fully austenitic high-manganese alloys [6-10]. The high ductility level required during technological forming simultaneously at high tensile stresses is obtained owing to a high strain hardening exponent resulting from a lack of localized deformation up to a large range of tensile strain. Recently, there is a trend to increase a fraction of austenitic phase in a multiphase microstructure at simultaneous increase of high-strength constituent amounts (i.e., non-carbide bainite and martensite) at cost just slightly higher compared to conventional DP, TRIP or CP steels. The aim is to improve further the strength level without significant deterioration of ductility. Different approaches include grain refinement, microalloying with Nb, Ti and V, TRIP-aided bainitic steels, ultra-fine bainitic steels and implementation of new processing routes, e.g. quenching and partitioning concept [11-19].

It is well known that the presence of austenite at room temperature requires lowering its martensite start temperature below 20°C. Two-step heat treatment used for TRIP steel sheets is the confirmation that it is possible to obtain some fraction of retained austenite even for steels containing 0.2% C [1, 3-5]. The increase of retained austenite amount by increasing C content is not a good idea due to welding limitations. In this regard, Mn alloying is quite often used at simultaneous decrease in C content [12-20]. Different processing strategies of cold rolling and intercritical annealing as well as the thermomechanical treatment enable to retain from 15 to 40% of austenite in (0.05-0.17)C-(4-7)Mn steels. Sometimes, Si and/or Al are added for hampering of carbide precipitation. An interesting concept is partitioning of Mn to austenite during intercritical annealing of cold rolled initially martensitic steel sheets what decreases M_s temperature effectively. Additional factors fostering the austenite stabilization at room temperature are size stabilization due to grain refinement and mechanical stabilization by increasing the dislocation density in γ phase [16]. It was proved that an intermediate strength-ductility regime between low-Mn TRIP steels and high-Mn TRIP/TWIP steels can be achieved [12-14, 16].

Mechanical and technological properties of steel sheets with increased Mn content are not only dependent on the amount of retained austenite but the mechanical stability of austenitic phase against strain-induced martensitic transformation is equally important [13, 14, 16, 21-23]. Austenite over-stabilized by C or Mn does not contribute to the simultaneous increase in strength and ductility (especially important during drawing, stretching and bending operations) due to a lack of strain-induced martensitic transformation, i.e., TRIP effect. In turn, when the content of C and Mn in retained austenite is too low the martensitic transformation

proceeds to quickly, what manifests by low ductility. In this regard, the most reliable parameter of mechanical stability of γ phase is considered to be a product of retained austenite fraction f_γ and carbon content c_γ in this phase [5,25]. Other factors affecting the mechanical stability of retained austenite are the grain size and arrangement of particles as well as its strength and the stress state [3,5,24]. It is related to the morphology of retained austenite (i.e., blocky, interlath) and other microstructural constituents neighbouring with γ phase particles. Usually, an optimal mechanical stability of retained austenite is achieved when it occurs between bainite of granular morphology [24].

The amount of retained austenite and its carbon content (besides chemical composition) are highly dependent on a processing route, i.e., cold rolling and intercritical annealing or thermomechanical processing. The most important parameters are the temperature and time of isothermal holding in a bainitic transformation range. For example, on the basis of our earlier works [4,25] on 0.24C-1.5Mn-0.9Si-0.4Al-Nb-Ti steel it was established that the maximum fraction of retained austenite can be obtained for the isothermal holding temperatures from 400 to 450°C, while the maximum carbon concentration in the γ phase equal 1.5 wt.% is present at the temperature of 350°C. Below 350°C due to a high $M_{s\gamma}$ temperature, the largest grains of retained austenite located in a ferritic matrix transform to martensite. In a temperature range from 350 to 450°C, the $M_{s\gamma}$ temperature has a negative value, stabilizing retained austenite. However, there are very few reports concerning the dependence of retained austenite content on an isothermal holding temperature in thermomechanically processed TRIP-aided steels with increased Mn content. This problem in undertaken in the present study.

2. Experimental procedure

The paper is a microstructural study on retained austenite in thermomechanically processed multiphase TRIP steels containing increased Mn content. The chemical composition given in Table 1 was designed with the focus on the maximizing of retained austenite amount (increased Mn content) and obtaining carbide-free bainite by low-Si high-Al concept [4,5,11]. Mo and Nb were used to increase strength and for grain refinement during hot-working. Special attention was paid to the effect of Mn and Nb on a final multiphase microstructure. The Mn-Al ingots were produced by vacuum induction melting in the Balzers VSG-50 furnace. Liquid metal was cast in the Ar atmosphere into a cast iron mould. Ingots with a mass of about 25 kg were forged at temperature range from 1200 to 900°C to a thickness of 22 mm. Then, cylindrical samples $\varnothing$ 10x12 mm for hot compression tests were machined. It should be noted that all the steels have comparable concentrations of C, Al, Mo and Si. The difference exists between Mn and Nb contents and it is a basis for steel designation (3Mn-1.5Al, 3Mn-1.5Al-Nb, 5Mn-1.5Al, 5Mn-1.5Al-Nb). Moreover, the steels are characterized by high metallurgical purity (low contents of P and S - Table 1).

In [15] it was found that 3Mn-1.5Al steels are characterized by bainitic microstructure in the initial state (after hot forging) whereas the steels with the higher Mn content have bainitic-martensitic microstructures. All the steels contain also some fraction of retained austenite already after hot forging and air

cooling what confirms the importance of Mn alloying for austenite stabilization. A tendency to macrosegregation of Mn and Al between middle and external zones of the as-cast ingots was also found. The small microsegregation of Al and Mn between individual structural constituents was also detected but it does not seem to be a problem for the homogeneity of obtained structures.

Table 1.
Chemical composition of the investigated steels (mass content, %)

	Steel grade			
	3Mn-1.5Al	3Mn-1.5Al-Nb	5Mn-1.5Al	5Mn-1.5Al-Nb
C	0.17	0.17	0.16	0.17
Mn	3.3	3.1	4.7	5.0
Al	1.7	1.6	1.6	1.5
Si	0.22	0.22	0.20	0.21
Mo	0.23	0.22	0.20	0.20
Nb	-	0.04	-	0.03
P	0.010	0.008	0.008	0.008
S	0.014	0.005	0.004	0.005

The thermomechanical processing of axisymetrical samples was realized by four-step compression followed by multi-step cooling. Parameters of the thermomechanical treatment for 3Mn steels are shown in Fig. 1a and for 5Mn steels in Fig. 1b. The hot compression tests were carried out using the DSI Gleeble 3800 thermomechanical simulator in the Institute for Ferrous Metallurgy in Gliwice, Poland. The specimens were inserted in a vacuum chamber, where they were resistance-heated to a temperature of 1200°C. After austenitizing for 30 s the specimens were cooled to a temperature of first deformation. The deformation temperatures were 1150, 1050, 950 and 850°C. The logarithmic strain was equal to 0.25 at the strain rate of 10 s^{-1} for each deformation step. Results of the four-step hot deformation behavior as well as continuous compression and softening kinetics were presented elsewhere [17,26]. It was found that combined Al and Mn alloying leads to higher flow stresses and critical strains required for initiation of dynamic recrystallization compared to conventional TRIP steels. The effect of Nb results in slightly higher flow stresses compared to Nb-free steels and retardation of dynamic recrystallization as well as softening kinetics of austenite. Dynamic recovery is a main thermally activated mechanism controlling work strengthening during a four-step deformation process.

The present paper focuses on the evolution of multiphase microstructures obtained as a result of multi-step controlled cooling. After the finishing deformation at 850°C the 3Mn steel specimens were cooled to the temperature of 700°C with a rate of 40°C/s (Fig. 1a). An important step of the thermomechanical processing consists in slow cooling of steel for 50 s in a temperature range of 700-650°C. Next, the samples were cooled at the rate of 40°C/s to the holding temperature in the bainite transformation range (T_B). The isothermal holding temperature was 450, 400, 350°C and the holding time (t_B) was constant, 300 s. Finally, the samples were cooled slowly to room temperature.
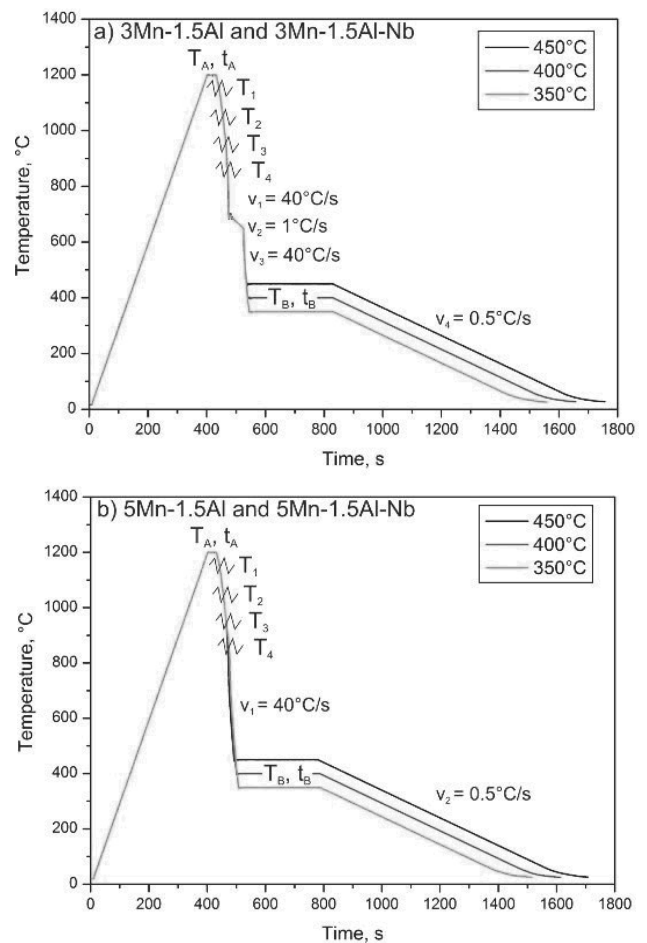

Fig. 1. Schematic drawing of the thermomechanical processing routes for 3Mn-1.5Al (a) and 5Mn-1.5Al (b) steels

Cooling path for 5Mn steels was very similar (Fig. 1b) except slow cooling in a temperature range between 700 and 650°C. Namely, there was a constant cooling rate of 40°C/s between 850°C and an isothermal holding temperature at the bainitic transformation range.

Metallographic specimens were prepared in the plane consistent with the compression direction and corresponding to the plastic flow of steel. For the purpose of retained austenite identification, etching in 10% water solution of sodium pyrosulphate was applied. Metallographic observations were carried out using Leica MEF 4A light microscope applying magnification 1000x. Computer image analysis was used to evaluate metallographically a fraction of retained austenite on the basis of five binary maps in 3Mn steels. Morphological features of microstructure were revealed with the SUPRA 25 scanning electron microscope. Vickers hardness of thermomechanically processed samples was measured with 100N load.

A fraction of retained austenite was assessed more precisely by X-ray quantitative analysis. Samples for XRD were ground using standard metallographic techniques with SiC carbide grinding papers through 1200 grit followed by polishing using

a 3 µm diamond suspension. X-ray diffraction was performed with filtered cobalt radiation and an X-pert PRO diffractometer operating at 40 kV and 30 mA, equipped with an X'Celerator detector. Three alpha and four gamma peaks were quantified by scanning trough a 2-theta range from 40 to 115 deg. To minimize the effect of texture, the volume fraction of retained austenite was quantified on the basis of the integrated intensity of three pairs of diffraction lines using the Averbach-Cohen approach [27]. The positions of maxima of the diffractions lines of austenitic phase (200γ, 220γ and 311γ diffraction peaks) were used to determine the lattice constant of austenite a_γ. The expansion of this parameter is useful for determination of the carbon content in retained austenite [28]:

$$a_\gamma = 3.578 + 0.033C_\gamma \tag{1}$$

where:

a_γ - lattice parameter of the austenite ($\overset{.}{A}$),
C_γ - carbon concentration in the austenite (wt. %).

Martensite start temperature was calculated using the well known Andrews' formula:

$$M_s = 539 - 423C - 30.4Mn - 17.7Ni - 12.1Cr - 7.5Mo \tag{2}$$

where: C, Mn, Ni, Cr, Mo - mass contents of the elements in the austenite.

3. Results and discussion

Force-energetic parameters of the thermomechanical processing of investigated steels were a subject of a detailed analysis of the previous research [17,26]. It was found that the new developed high-Mn high-Al steels have higher flow stresses and critical strains required for initiation of dynamic recrystallization compared to conventional TRIP steels. The values of flow stresses at a final deformation step at 850°C are close to 270 MPa for Nb-free steels and are about 30 MPa higher for Nb-containing steels. Products of the phase transformations from the pancaked austenite taking place during multi-step cooling for 3Mn-1.5Al steels (Fig. 1a) are the microstructures in Fig. 2. The steels are characterized by bainitic-martensitic microstructures containing a small fraction of fine-grained ferrite and some fraction of retained austenite. The realization of phase transformations under conditions of accumulated strain in austenite results in a significant refinement of all the microstructural constituents. A morphology of a bainitic matrix and the amount of retained austenite are dependent on the isothermal holding temperature in a bainitic transformation range. All the microstructures in Fig. 2 are quite homogeneous and comparable both for 3Mn-1.5Al and 3Mn-1.5Al-Nb steels. A small fraction of ferrite, which can be identified in Fig. 2, has a morphology of very fine equiaxial grains but more often it is acicular ferrite. The maximum amount of ferritic phase does not exceed 5%, hence the distribution of carbon into austenite at this transformation temperature range is marginal and the carbon

partitioning between bainitic ferrite and austenite is of primary importance.

Retained austenite is uniformly distributed in a bainitic-martensitic matrix and two morphologies of this phase can be distinguished in Fig. 2 - fine granules and interlath retained austenite. The latter one dominates for the lowest isothermal holding temperature, i.e. 350°C, where the bainitic-martensitic microstructure has a distinct lath character (Fig. 2a, b). Determination of the volume fraction of retained austenite is a very challenging problem due to its very small grain size and high dispersion. The preliminary estimation of the amount of γ phase using a computer image analysis was possible only for 3Mn-1.5Al steel specimens isothermally held at 400 and 450°C. A fraction of retained austenite at 350°C determined using X-ray is the lowest for both steels. It equals to 13.6% for Nb-free steel and is 3% lower for the 3Mn-1.5Al-Nb steel (Table 2).

Increasing the isothermal holding temperature to 400°C results in a change of the morphology of bainitic ferrite into degenerated upper bainite (Fig. 2c, d). It is better visible using SEM (Fig. 3). A characteristic feature of the degenerate upper bainite is a lack of cementite particles, which are replaced by continuous or interrupted layers of retained austenite [29]. Interlath retained austenite has a thickness much below 1 µm. The total amount of retained austenite increased to about 19% what is confirmed by X-ray examination and computer image analysis as well (Table 2). A fraction of γ phase is comparable for both steels and the increase of the retained austenite content with a morphology of fine granules is apparent (Fig. 2c, d, Fig. 3). Some fraction of larger blocky grains of γ phase transformed into fine martensite plates usually surrounded by films of retained austenite (Fig. 3). Fine granules of retained austenite are in majority for the highest isothermal holding temperature applied (Fig. 2e, f). According to the results of computer image analysis the amount of retained austenite is the highest at 450°C (Table 2). However, X-ray examination revealed that a fraction of γ phase at 450°C is very similar to that obtained at 400°C. The amount of austenite determined applying the method of image analysis is probably overestimated. It arises from the fact that binary maps of highly dispersed retained austenite include morphological regions very similar to γ phase areas that have probably undergone into martensite in the last stage of cooling. The etching quality of metallographic specimens strongly influences also final results of distinguishing austenite and martensite. This limitation is not valid for X-ray analysis, which results should be more precise. The majority of fine grains of retained austenite has a diameter of about 1 µm and they are usually located on bainite-ferrite phase boundaries or are a constituent of bainitic-martensitic regions (Fig. 2e, f).

Steels containing a higher manganese content have higher hardenability. Their microstructure consists of bainite and martensite lath packets containing interlath retained austenite (Fig. 4). The identification of the morphology of a matrix and retained austenite using light microscopy is a difficult task due to a very fine-grained microstructure being a result of the thermomechanical processing applied. However, the change of the morphology of bainitic ferrite is proved by the hardness increase with decreasng a holding temperature in the bainitic transformation range (Table 2). Morphologal details of the microstructure can be better identified using SEM (Fig. 5).
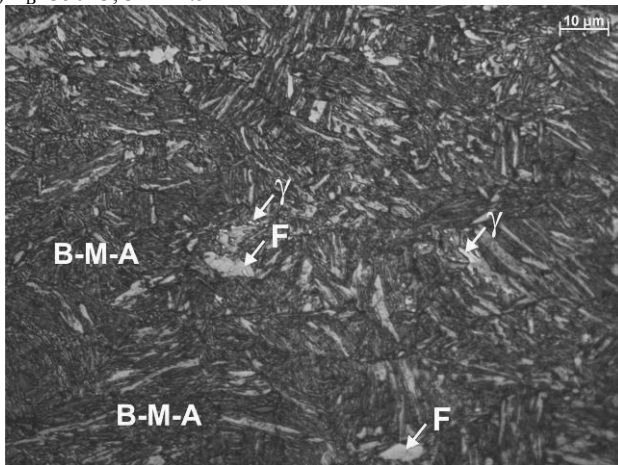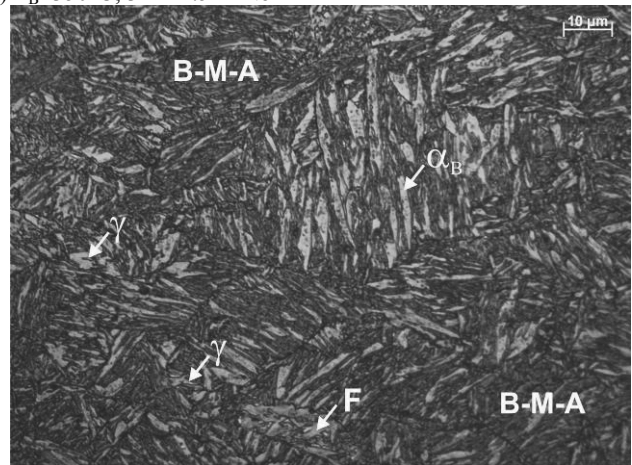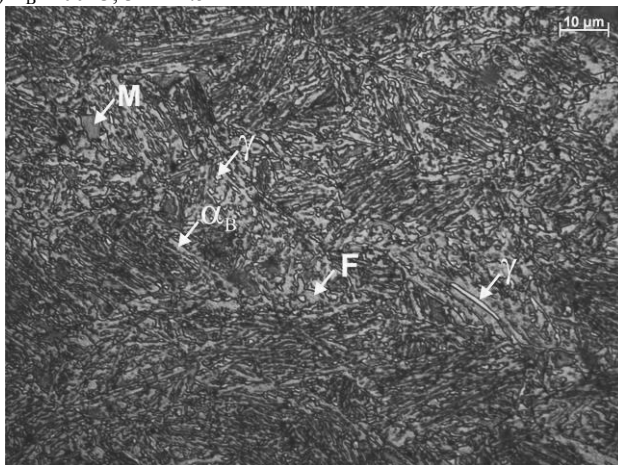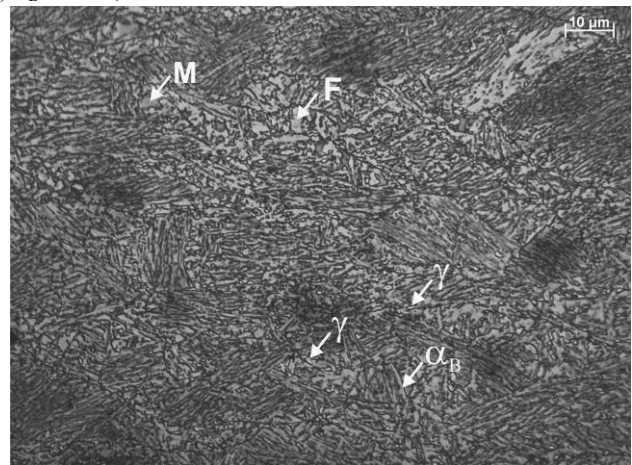
a) T_B=350°C; 3Mn-1.5Al

b) T_B=350°C; 3Mn-1.5Al-Nb

c) T_B=400°C; 3Mn-1.5Al

d) T_B=400°C; 3Mn-1.5Al-Nb

e) T_B=450°C; 3Mn-1.5Al
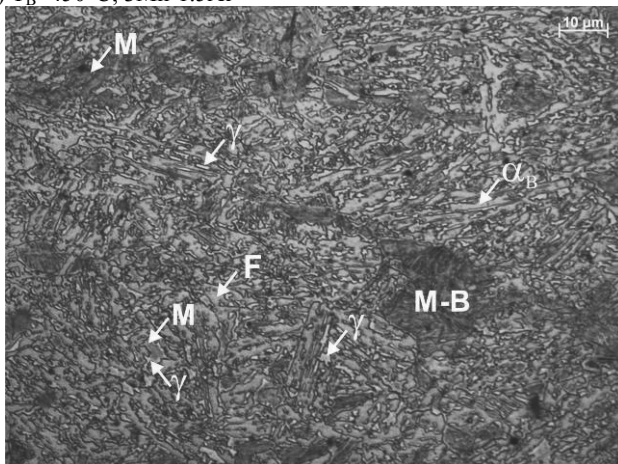
f) T_B=450°C; 3Mn-1.5Al-Nb
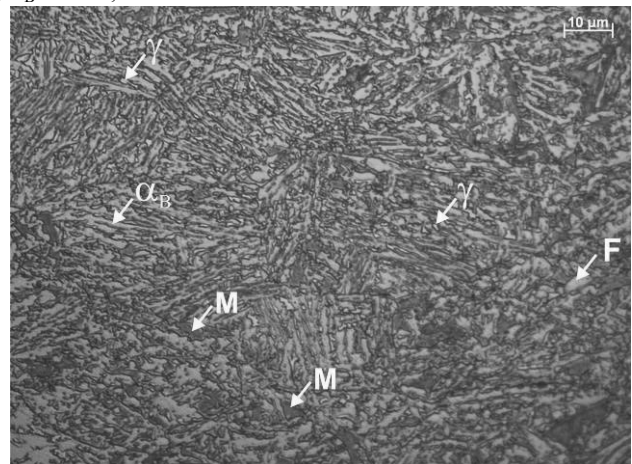
Fig. 2. Bainitic-martensitic microstructures containing fine-grained ferrite and a various fraction of retained austenite for the 3Mn-1.5Al and 3Mn-1.5Al-Nb steel specimens isothermally held for 300 s at the temperature of 350°C (a, b), 400°C (c, d) and 450°C (e, f); B-M-A - bainite-martensite-austenite regions, F - ferrite, α_B - bainitic ferrite, M - martensite, γ - retained austenite

Table 2.
Parameters characterizing retained austenite determined by the X-ray analysis and hardness values for different isothermal holding temperatures

| Steel grade	Isothermal holding temperature, °C	Retained austenite content, %		Lattice parameter of austenite, a_γ, Å	Carbon content in austenite, C_γ, wt.%	Martensite start temperature, $M_{s\gamma}$, °C	HV10
		Image analysis	X-ray				
3Mn-1.5Al	450	19.7±2.1	16.9	3.6182±0.0048	1.22	-78	345±13
	400	18.8±2.2	17.5	3.6188±0.0066	1.24	-85	384±23
	350	nd.	13.6	3.6202±0.0061	1.28	-104	417±11
3Mn-1.5Al-Nb	450	20.4±2.7	15.4	3.6143±0.0061	1.10	-22	376±15
	400	19.5±1.8	18.4	3.6149±0.0045	1.12	-30	435±10
	350	nd.	10.0	3.6167±0.0066	1.17	-53	469±21
5Mn-1.5Al	450	nd.	14.4	3.6079±0.0045	0.91	12	455±30
	400	nd.	13.1	3.6096±0.0065	0.96	-11	460±28
	350	nd.	10.6	3.6108±0.0039	0.99	-26	489±27
5Mn-1.5Al-Nb	450	nd.	10.5	3.6018±0.0041	0.72	80	463±16
	400	nd.	11.2	3.6035±0.0056	0.77	59	509±15
	350	nd.	9.7	3.6049±0.0060	0.82	40	560±29

nd. - not determined
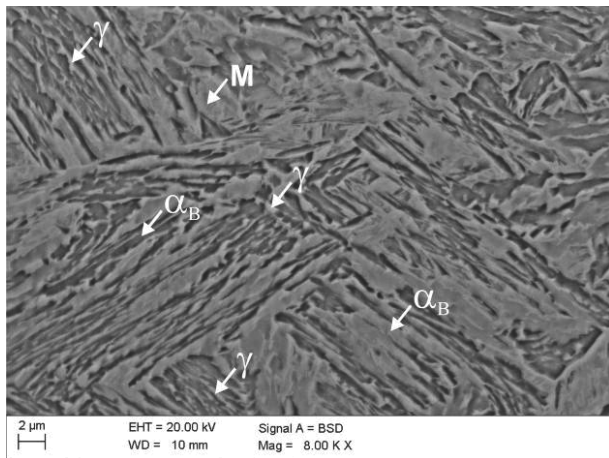
Fig. 3. Degenerate upper bainite containing interlath retained austenite and fine granules of this phase in 3Mn-1.5Al steel isothermally held at 400°C; α_B - bainitic ferrite, M - martensite, γ - retained austenite

Figure 5 is a typical example of the microstructure observed in 5Mn steels. The micrograph obtained for the 5Mn-1.5Al steel held isothermally at 400°C is a mixture of degenerate upper bainite and degenerate lower bainite. This bainite morphology is beyond classification based on conventional concepts due to the incomplete austenite transformation phenomenon, often observed for steels alloyed with silicon and aluminium [29,30]. The bainitic ferrite with interlath retained austenite (degenerate upper bainite) is a main microstructural constituent both for the Nb-free and Nb-microalloyed steels held at 450°C, whereas the degenerate lower bainite dominates for the specimens held at 350°C.

The degenerate lower bainite consists of very fine, fragmented martensite/austenite (MA) particles inside bainitic ferrite laths. They can be apparent in Fig. 5 in larger bainitic ferrite laths. The presence of intra-lath MA particles suggests that the degenerate lower bainite is a product of incomplete transformation of austenite which inhibits the formation of cementite inside bainitic ferrite laths. This problem is more detailed explained in [29]. A bimodal distribution of bainitic ferrite plates (the fine plate and large laths containing intra-lath particles) can be also associated with the coalesced bainite. This process occurs when adjacent platelets of bainite merge to form a single larger plate [19] leaving cementite or MA particles as transformation products depending on a chemical composition and thermal treatment parameters.

A fraction of retained austenite is lower compared to 3Mn steels. It testifies that the hardenability effect of Mn is higher than its efficiency as an austenite former. The amount of γ phase rises with increasing the isothermal holding temperature from 350 to 450°C and it is between 10 and 14% (Table 2). The austenite amount is higher for Nb-free steel. The microstructures of the 5Mn-1.5Al and 5Mn-1.5Al-Nb are very similar (Fig. 4c, d). However, the number of martensite areas is higher for the steel microalloyed with Nb.

The analysis of diffraction patterns allowed not only to assess a volume fraction of an austenitic phase but to determine the lattice constant of austenite and its resulting carbon content (Table 2). Selected X-ray diffraction patterns for 3Mn and 5Mn steels are shown in Figs. 6 and 7. Four γ phase diffraction lines can be identified. Relative intensities of individual diffraction peaks indicate that the specimens are characterized by the preferred crystallographic orientation as a result of the final deformation below the recrystallization temperature of austenite. However, the importance of the texture behavior deserves to be undertaken elsewhere.
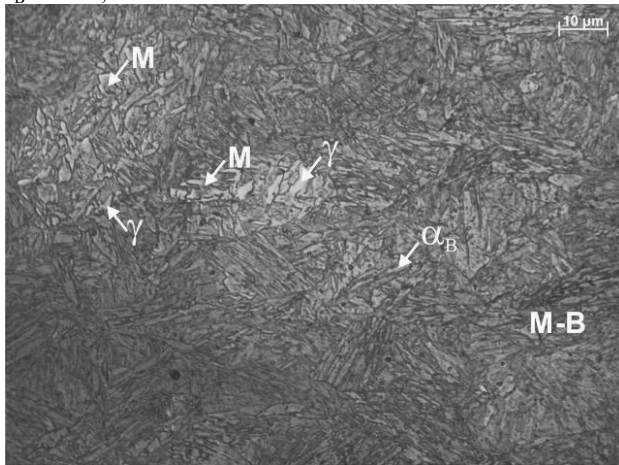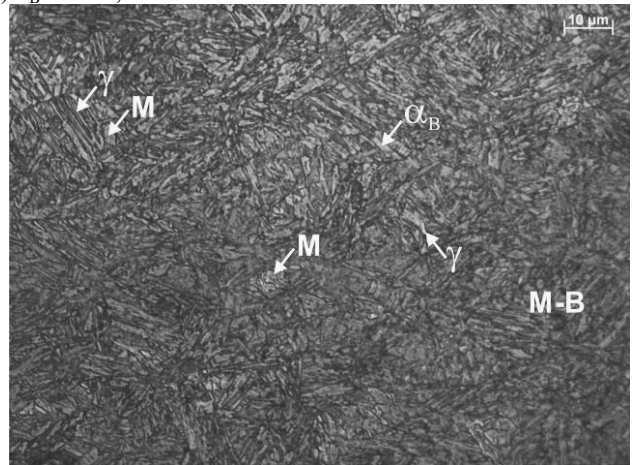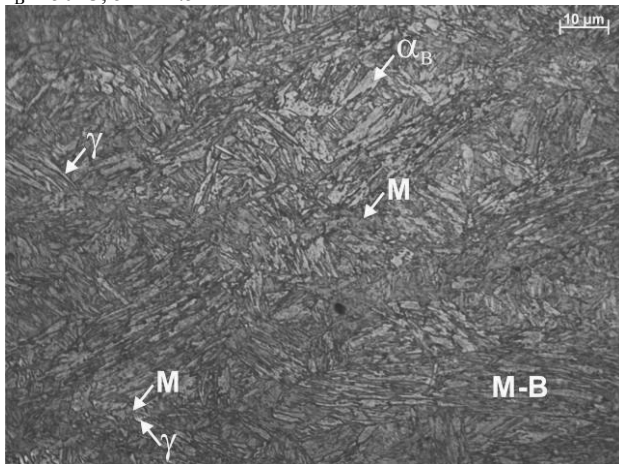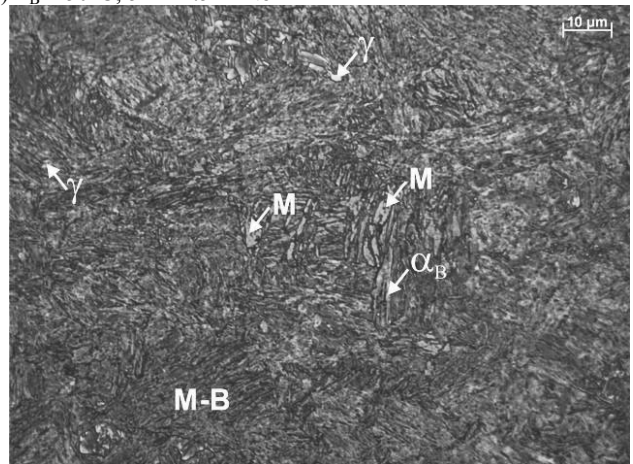
a) T_B=350°C; 5Mn-1.5Al

b) T_B=400°C; 5Mn-1.5Al

c) T_B=450°C; 5Mn-1.5Al

d) T_B=450°C; 5Mn-1.5Al-Nb

Fig. 4. Bainitic-martensitic microstructures containing a various fraction of interlath retained austenite for the 5Mn-1.5Al (a-c) and 5Mn-1.5Al-Nb (d) steel specimens isothermally held for 300 s at the temperature of 350°C (a), 400°C (b) and 450°C (c, d); B-M - bainite-martensite regions, α_B - bainitic ferrite, M - martensite, γ - retained austenite
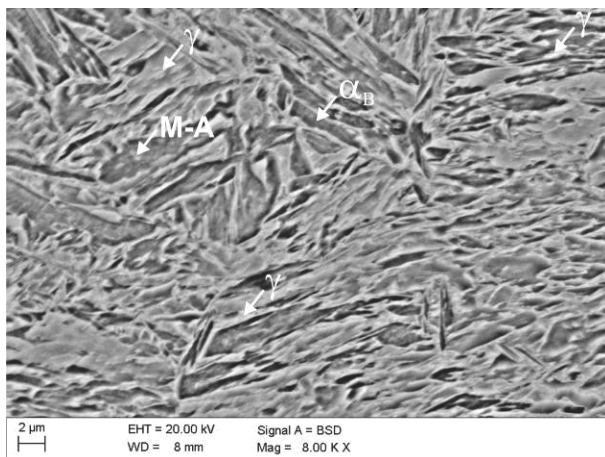

Fig. 5. A mixture of the degenerate upper bainite and degenerate lower bainite in the 5Mn-1.5Al steel isothermally held at 400°C

The value of lattice constant varies in quite a wide range, from 3.6018 to 3.6202Å (Table 2). It is clear that the highest carbon content in austenite for all the steels investigated occurs for the isothermal transformation temperature of 350°C. It is interesting taking into account that the highest amount of retained austenite obtained at the temperature range between 400 and 450°C. However, very similar results were achieved in earlier investigated 0.24C-1.5Mn-0.9Si-0.4Al-Nb-Ti and 0.43-1.5Mn-1Si-1Al-Nb-Ti TRIP steels [4,25]. Below 350°C the carbon diffusion decreases leading usually to rapid lowering an austenite fraction. At 350°C the carbon diffusion conditions are optimal to enrich austenite in carbon but the strong driving force for the bainitic transformation leads to a smaller fraction of retained austenite compared to higher temperatures. With increasing the isothermal holding temperature the diffusion rate of C increases and the enrichment of austenite in carbon is easier. However, the lattice parameter of austenite and its corresponding carbon content decrease slightly what can be a result of the initiation of precipitation processes associated to carbon clustering [31].
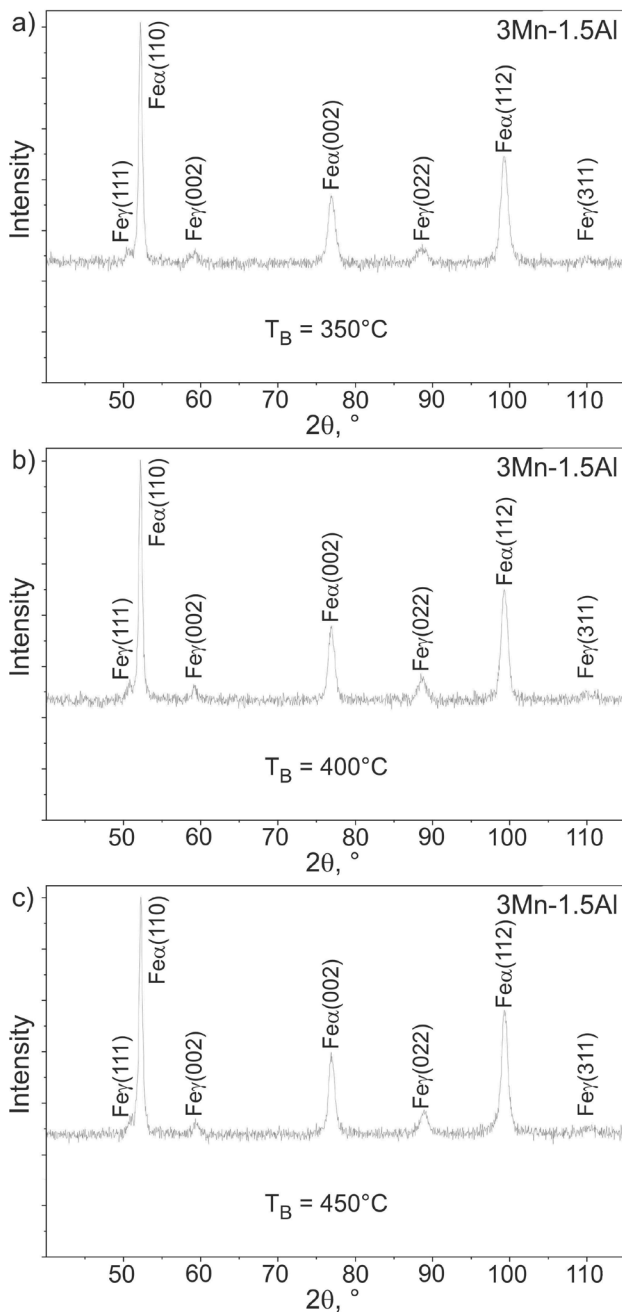
Fig. 6. X-ray diffraction patterns for various isothermal holding temperatures of the thermomechanical processing for the 3Mn-1.5Al steel

At the temperature range between 400 and 450°C the carbon content in austenite is still enough to stabilize a high fraction of retained austenite (Table 2), what is also supported by decreasing the bainitic transformation driving force.

The presence of retained austenite can be also well explained with the help of the calculated temperatures of martensite start for retained austenite quoted in Table 2. They are below room temperature for the steels with a lower Mn content. The martensite start temperature for the steels containing 5% Mn is higher. It is a direct result of the lower carbon concentration in austenite (below 1wt.% - Table 2). The carbon content in retained austenite in carbide-free bainitic steels is determined by T_0 temperature, where austenite and ferrite of the same composition have identical free energies [30,32,33]. According to Takahashi and Bhadeshia [33] manganese lowers the T_0 temperature and a resulting carbon concentration in austenite. Their theory was confirmed by the results of Sugimoto et al. [32]. They reported that the carbon concentration in retained austenite decreases continuously from 1.53 to 1.13 wt.% with increasing the Mn content from 1 to 2.5 wt.%. The higher Mn concentration in the present steels leads to lower carbon contents in retained γ phase (Table 2). The carbon content is enough to lower the martensite start temperature below room temperature and subsequently to stabilize retained austenite in 3Mn-1.5Al and 3Mn-1.5Al-Nb steels. The martensite start is also lower than room temperature for 5Mn-1.5Al steel. However, the opposite is true for 5Mn-1.5Al-Nb steel indicating the destabilizing effect of niobium microalloying on retained austenite. The similar effect can be observed when 3Mn steels are compared (Table 2). It is probably due to niobium carbonitride precipitation but it requires further research with transmission electron microscopy. Some martensite regions in the microstructure of 5Mn-1.5Al-Nb steel can be observed in Fig. 4d. However, they occur locally also in the Nb-free steel (Fig. 4a-c) and in 3Mn steels as well (Figs. 2 and 3), especially as relatively large blocky islands.
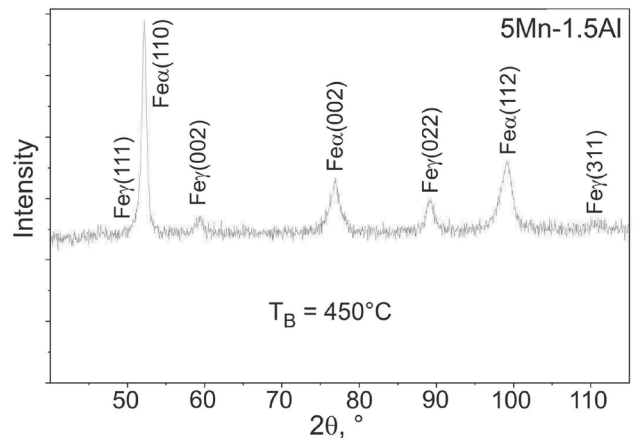

Fig. 7. X-ray diffraction pattern for the isothermal holding temperature of 450°C for the 5Mn-1.5Al steel

Hardness results correspond well with the microstructures obtained for a different isothermal holding temperature in the bainitic transformation range (Table 2). Hardness of the steels containing 5% Mn is about 100 HV higher compared to 3Mn steels. Additionally, Nb-microalloyed steels are characterized by significantly higher hardness compared to Nb-free steels because of the complex effects of higher grain refinement, precipitation of Nb(C,N) and the higher hardenability due to Nb dissolved in austenite. Hardness values decrease with increasing the isothermal holding temperature due to a change of the morhology of a matrix and the increase of the amount of retained austenite.

4. Conclusions

Thermomechanical processing is a very useful method to obtain very fine-grained bainitic or bainitic-martensitic microstructures containing from 10 to 19% of retained austenite in 3Mn-1.5Al and 5Mn-1.5Al steels. The highest fraction of retained austenite is present for all the steels investigated for the isothermal holding temperature in a bainitic transformation range between 400 and 450°C, although the highest carbon content in austenite occurs for a temperature of 350°C. Increasing Mn concentration results in decreasing the carbon content is γ phase, similarly like Nb microalloying. These effects have a detrimental influence on the stabilization of austenite through raising its martensite start temperature. Martensite regions forming at a final cooling step are present especially in 5Mn steels and locally in 3Mn steels as large blocky regions. The increased stability have film-like austenite regions occurring in majority in the investigated steels. Inter-lath retained austenite is a constituent of degenerate upper bainite for an isothermal holding temperature range between 400 and 450°C, whereas degenerate lower bainite with intra-lath martensite-austenite particles is identified in 5Mn steels isothermally held at a temperature range between 350 and 400°C.

Acknowledgements

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