Archives of Materials Science and Engineering

Volume 49
Issue 1
May 2011
Pages 5-14

International Scientific Journal
published monthly by the
World Academy of Materials
and Manufacturing Engineering

Effects of Mn and Nb on the macro- and microsegregation in high-Mn high-Al content TRIP steels

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Received 21.01.2011; published in revised form 01.05.2011

ABSTRACT

Purpose: The aim of the paper is to determine the effects of Mn and Nb on the macro- and microsegregation of new-developed (3-5)Mn 1.5Al TRIP steels in the as-cast state and after hot forging.

Design/methodology/approach: To assess possible macro- and microsegregation the chemical composition was investigated in the macro scale by OES and by the use of EDS in microregions. To reveal the macrostructure the ingots were etched by Oberhoffer's reagent. The chemical composition along a diameter of the ingots was tested as well as in different regions of the plastically deformed flats. Metallographic investigations were carried out using optical (OM) and scanning electron microscopy (SEM).

Findings: The Mn and Nb contents have significant effects on the obtained bainitic-martensitic structures and the morphology of retained austenite. The tendency to macrosegregation of Mn and Al between middle and external parts of the as-cast ingots and hot forged flats was detected. The Nb micoalloying results in the overall refinement of obtained microstructures and reduces the identified segregation of Al and Mn. A little microsegregation of these alloying elements between structural constituents was also detected.

Research limitations/implications: Further investigations to describe in detail the identified structural constituents and the tendency of alloying elements to segregation in the thermomechanically rolled sheets are in progress.

Practical implications: The knowledge of the macro- and microsegregation of Mn and Al in the TRIP steels with a high fraction of retained austenite is of primary importance for the manufacturing paths of these multiphase structure steels.

Originality/value: A problem of segregation of Al and Mn in advanced high strength steels with and without Nb microaddition was described and interesting types of bainitic structures were identified.

Keywords: Metallic alloys; TRIP steel; Retained austenite; Nb microalloying; Macrosegregation; Microsegregation

Reference to this paper should be given in the following way:

A. Grajcar, E. Kalinowska-Ozgowicz, M. Opiela, B. Grzegorczyk, K. Gołombek, Effects of Mn and Nb on the macro- and microsegregation in high-Mn high-Al content TRIP steels, Archives of Materials Science and Engineering 49/1 (2011) 5-14.

MATERIALS

1. Introduction

Conventional TRIP steels contain about 0.2% C, 1.5% Mn and up to 1.5% Si, Al, P (in total) dependent on a chemical composition strategy [1-6]. Silicon, aluminum and phosphorus prevent carbide precipitation during bainite transformation and allow the carbon to diffuse into the retained austenite during isothermal holding of steel at the bainitic transformation range. A high Si content is responsible for a poor surface quality of hot rolled steel and a poor coatability of cold rolled steel. Si causes problems during hot rolling, producing an oxide scale that is difficult to remove in subsequent processing [1-3]. More problematic is the influence of Si concerning hot-dip galvanizing. High-Si steel sheets have poor wettability as a result of a layer of amorphous SiO_2 present on the sheet surface. It has also been shown that Mn and Si can form the mixed oxide Mn_2SiO_4, which was also found to be detrimental to the wettability of the steel [7]. It is well known [1, 2, 6-8] that Al improves reactive wetting during galvanizing. As a result of that, a partial or complete substitution of Si by Al or Al/P is intensively considered [1-4, 6]. Al accelerates the bainite transformation kinetics. As the effect of solid solution hardening by Al is lower, it is often used in higher C content steels. However, there is a trend to decrease C content below 0.2% for reasons of weldability. This discrepancy requires new ideas of chemical composition design. The addition of niobium to Al-based steels has been the object of recent studies [3, 4, 6, 9, 10]. Additions of Mo to Nb-bearing steels are reported to bring about a further improvement of the strength due to its strong solid solution effect and hardenability [9, 10].

Third generation of advanced high strength steels (AHSS) for automotive industry utilize complex interaction of solid solution hardening, precipitation hardening, microalloying, grain refining and TRansformation Induced Plasticity (TRIP) or TWinning Induced Plasticity (TWIP) effects. The new idea of multiphase TRIP steels is to obtain a ferritic or bainitic matrix containing a high volume fraction (up to 40%) of metastable retained austenite of suitable mechanical stability [11, 12]. Higher fractions of γ phase can be stabilized in increased Mn-content steels [11-14]. Increased hardenability of steel due to Mn alloying leads to a considerable decrease in the ferrite fraction as a result of shifting the $\gamma \rightarrow \alpha$ transformation region to the right side on CCT diagram [15]. On the other hand, recently published results [16-18] indicate that a ferrite fraction is not a critical parameter deciding about the mechanical stability of retained austenite instead of the morphology of bainitic ferrite and γ phase and the interactions between all the phases during straining.

Recently, significant levels of austenite (from 18 to 38%) and excellent mechanical properties have been obtained in the experimental low-carbon (5-7)Mn steels [13]. Cold-working followed by annealing in the $\alpha+\gamma$ region resulted in fine-grained ferrite-austenite-martensite microstructures. In specimens in which the austenite partially transformed to martensite during straining the transformation prevents early necking and increases the strain-hardening capacity of the steel. The intermediate strength-ductility regime between low-Mn TRIP steels and high-Mn TRIP/TWIP steels was achieved.

There are also some problems concerning with Mn alloying, e.g. ability to microsegregation, localized deformation, hot-working and corrosion behavior [12-14, 19-23]. A banded microstructure of ferrite / bainite-martensite-austenite as a result of Mn microsegregation during continuous casting is typical for TRIP steels [14, 19]. Pichler et al. [19] noticed that due to the enrichment of Mn and depletion of Al in the centre-segregation zone of sheets, the austenite stabilization is there much more pronounced than in the surroundings. The result of this is a delayed ferrite growth in this region and finally a bainitic-martensitic band. Wietbrock et al. [14] and Kliber et al. [20] reported that local differences in Mn content between dendrites and interdentritic spaces in high-manganese alloys (over 20 wt.%) can reach up to 5-7 wt.% in the as-cast state. After some homogenization strategies the microsegregation of Mn can be much reduced. A global refinement of the microstructure is also efficient to reduce microsegregation [19].

2. Experimental procedure

The present paper addresses a problem of Mn and Al macro- and microsegregation of 0.17C-(3-5)Mn-1.6Al-0.2Si-0.2Mo TRIP type steels with and without Nb microaddition. Presented results are a first part of a project to obtain fine-grained thermo-mechanically processed TRIP steels with ferritic-bainitic-austenitic and bainitic-austenitic structures. The chemical composition given in Table 1 was designed from the point of view of maximization of retained austenite (increased Mn content) and obtaining carbide-free bainite by low-Si high-Al concept (susceptibility to galvanizing) [1, 7-10]. Mo and Nb were used to enhance strength. Special attention was paid to the effect of Nb on a macro- and microsegregation and grain refinement of final microstructures. 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 coding (3Mn, 3MnNb, 5Mn, 5MnNb). Moreover, the steels are characterized by high metallurgical purity connected to low concentration of phosphorus and sulphur (Table 1). Comparable investigations concerning segregation and microstructure were carried out in the as-cast state and after hot deformation by forging.

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

	Steel code			
	3Mn	3MnNb	5Mn	5MnNb
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

Steels were produced by vacuum induction melting in the Balzers VSG-50 furnace at the Institute for ferrous Metallurgy in Gliwice. Liquid metal was cast in the Ar atmosphere into a hot-topped closed-bottom wide-end-up cast iron mould with internal dimensions: bottom – ∅ 122 mm, top – ∅ 145 mm, height – 200 mm (without ingot head). In order to reveal the macrostructure and to evaluate the susceptibility of alloys to macrosegregation the ingots were cut perpendicular to an axis of the ingot within the distance of 30 mm from the bottom. After grinding and polishing, the specimens with a diameter of about 124 mm were etched by Oberhoffer's reagent to reveal a primary structure of the ingots. The chemical composition along a diameter was tested in the macro scale by optical emission spectroscopy (OES). Measurements were carried out in a central zone, within the distance of about 31 mm from the middle and in an external zone of ingots. A result was an average value of three measurements.

Macrosegregation was also assessed at the plane parallel to a major axis of the ingot for the 5MnNb steel. The chemical composition by OES was investigated in the external zone of the ingot (20 mm below its surface) at the height of 30, 100, 135 and 170 mm from the bottom.

The primary structure of as-cast specimens was revealed by nital etching. The specimens were taken from the external zone of the ingots. The microsegregation in different regions of etched specimens was examined by energy dispersive spectrometry (EDS). Comparable linear analyses as well as point analyses of chemical composition of different structural constituents were carried out using EDAX in the SUPRA 25 scanning electron microscope.

Ingots with a mass of 25 kg after austenitizing at 1200°C for 3 hours were forged at a temperature range from 1200 to 900°C to a final width of 160 mm and a thickness of about 22 mm. To compare the microstructures in the as-cast state and after hot forging metallographic investigations were carried out by the use of LEICA MEF4A microscope. The specimens after hot forging were cut near the edge of a flat according to its length. The microstructure of the plastically deformed specimens was revealed by etching in 10% aqueous solution of sodium metabisulfite for better identification of retained austenite. EDS measurements were carried out to assess the differences in chemical composition between surface (about 2 mm below surface) and central regions of a plastically deformed flat. Local microanalysis of chemical composition by EDS in hot forged specimens will be reported elsewhere.

3. Results and discussion

Metallographic investigations revealed that the new-developed steels have slightly different macrostructures. Three distinct zones can be distinguished in Fig. 1a and b for 3Mn and 3MnNb steels. Niobium microalloyed steel possesses in an external region of the ingot a clearly apparent zone of columnar-like crystals. The thickness of the crystals oriented according to direction of heat transfer is about 20 mm. An intermediate zone is consisted of random-oriented elongated crystals (probably dendrites). This zone is especially visible for a Nb-free steel (Fig. 1a), where its fraction is much higher compared to the Nb microalloyed steel (Fig. 1b). A central zone of the ingots comprise fine equiaxial grains. Only two zones can be distinguished in Fig. 1c for the 5MnNb steel. A very similar macrostructure was identified in the 5Mn steel without niobium. A zone of columnar-like crystals can be seen in an external region of the ingot. Its thickness is comparable to the steels containing lower Mn content. A predominant zone of the 5Mn and 5MnNb ingots consists of random-oriented elongated crystals. Due to the higher hardenability of 5Mn steels the dendrites reach an axis of the ingots, as distinct from 3Mn steels, where a central zone contains equiaxial grains.

Results given in Table 2 indicate the good homogeneity of chemical composition in different zones of a transverse section of the ingots. Mo and Si contents are nearly the same like the nominal composition for 3Mn steels, whereas for 5Mn steels they are slightly higher (Table 2). Their concentrations are stable along a diameter of the ingots.

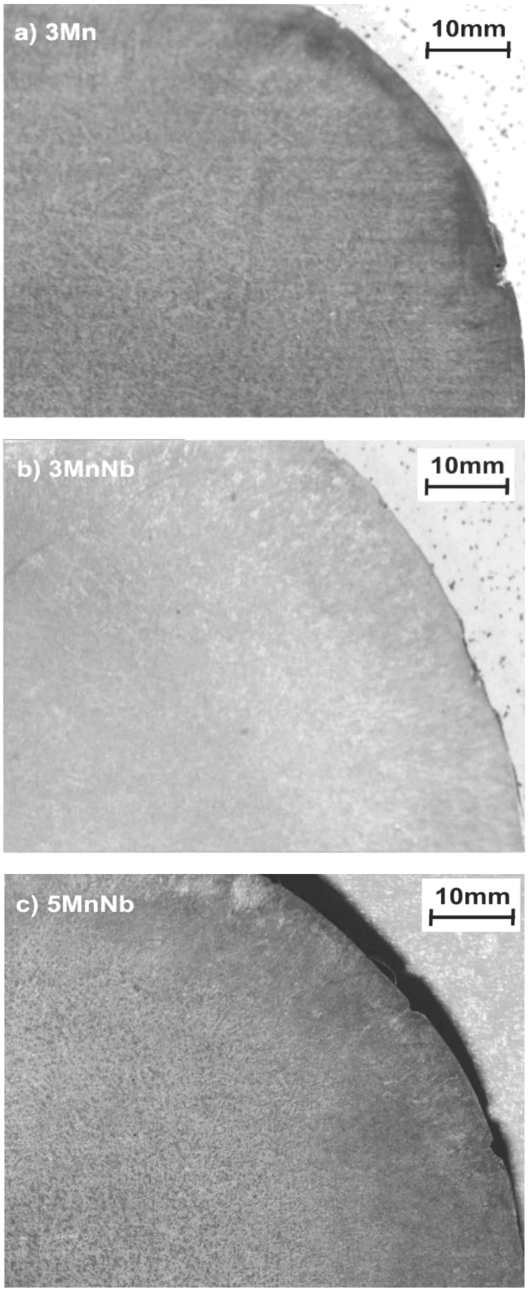


Fig. 1. Macrostructures of the 3Mn (a), 3MnNb (b) and 5MnNb (c) steel ingot (transverse section)

Table 2.
Chemical composition of the investigated steels in different zones of the ingots (wt. %)

Steel	Zone	C	Mn	Al	Si	Mo
3Mn	external	0.21±0.01	2.91±0.02	1.90±0.07	0.22±0.01	0.22±0.01
	intermediate	0.21±0.01	2.90±0.02	1.87±0.07	0.23±0.01	0.22±0.01
	centre	0.20±0.01	2.89±0.02	1.82±0.01	0.22±0.01	0.22±0.01
3MnNb	external	0.20±0.01	2.83±0.02	1.72±0.03	0.23±0.01	0.21±0.01
	intermediate	0.20±0.01	2.82±0.01	1.70±0.03	0.23±0.01	0.21±0.01
	centre	0.17±0.01	2.82±0.01	1.68±0.04	0.23±0.01	0.21±0.01
5Mn	external	0.16±0.01	4.03±0.01	1.73±0.01	0.22±0.01	0.22±0.01
	intermediate	0.16±0.01	4.02±0.01	1.68±0.01	0.22±0.01	0.22±0.01
	centre	0.16±0.01	4.02±0.01	1.69±0.02	0.22±0.01	0.22±0.01
5MnNb	external	0.21±0.01	4.19±0.02	1.63±0.03	0.23±0.01	0.23±0.01
	intermediate	0.21±0.01	4.16±0.01	1.63±0.02	0.23±0.01	0.23±0.01
	centre	0.22±0.01	4.18±0.02	1.62±0.03	0.23±0.01	0.23±0.01

For better comparison of the chemical composition in different zones of the ingots to the bulk content, comparable schemes of the distribution of individual elements along a diameter are shown in Fig. 2a and b. It is characteristic that carbon and aluminium contents are higher compared to the bulk composition (except C in the 5Mn steel). Taking into account that cooling conditions at the bottom are comparable to these along a perimeter of the ingots and lower C concentration in a center zone of the 3MnNb steel (Fig. 2b), it is anticipated that the contents of C and Al should be smaller in the middle of the ingots. The carbon content near the bottom of the ingot is consistent with the bulk content only for 5Mn steel (Table 2). Segregation of carbon in the as-cast samples should not be a problem considering successive hot-working and a high rate of diffusion of interstitial elements at high temperatures. The most interesting results concerning segregation in the as-cast state were observed for Al and Mn and they will be a subject of a further analysis.

The highest difference in alloying element contents between the transverse sections and the nominal composition was observed for manganese. Mn contents are nearly the same along a diameter but much lower compared to the nominal composition (Table 2). This is true for all the ingots but the difference is about 0.4 wt.% for the 3Mn steels and twice higher for steels containing the higher Mn content. The most significant difference of chemical composition in different zones of the ingot can be noticed for aluminium. Its content is higher compared to the bulk content for all the steels and what is more important it decreases successively from the external to center zones of the ingots (Figs. 2a, b). The concentration gradient is relatively small because the maximum difference in Al content between the external and center zones does not exceed 0.08%. The effect is more visible for the steels without Nb both for 3Mn and 5Mn steels (Table 2). The change of Al along a diameter is according with the thermodynamic calculations performed by Pichler et al. [19]. They reported that the first formed solid phases should contain more Al than the nominal composition and hence Al is depleted in the liquid phase. The opposite is true for Mn. With advancing solidification the concentration of Mn increases whereas the Al content decreases. Due to the enrichment of Mn and depletion of Al in the centre zone of the ingots, the austenite stabilization is expected there [19]. Due to a small distance from the bottom the observed changes in Al content are relatively small.

The measurements of the chemical composition at the plane parallel to a major axis of the ingot (Fig. 2c) for the 5MnNb steel are in excellent accordance with the results obtained for the cross section (Table 2). The carbon and aluminium contents are higher compared to the nominal composition whereas the Mn concentration is lower. The contents of all the alloying elements are almost stable independent on the height of the ingot. It testifies a similar chemical composition in lower, central and upper parts of the ingots. However, it is only true for the external zones of the ingots. Taking into account the lower Mn content and higher Al content in the external parts of the ingots and the overall balance of chemical composition in the alloys, it is expected that the opposite situation in Al and Mn concentrations – comparing to the bulk content – should occur in the middle part of the ingots.

Figure 3 reveals primary microstructures of the steels both at the transverse and longitudinal section of the ingots. The dendrites are chaotically arranged and especially visible in 3Mn steels (Figs. 3a-d). Additionally, some places of different morphology also occur. It is connected with a microsegregation of Mn, what will be reported later. In 5Mn steels the dendritic structure is better visible at the longitudinal section especially for the steel without Nb, where it can be seen that the main axes of dendrites are parallel to the axis of the ingot (Fig. 3f). A positive effect of Nb microaddition on a refinement of a primary microstructure can be also seen when comparing Fig. 3e, f with Figs. 3g, h.

At higher magnifications some morphological details of the as-cast microstructures can be seen. In all the steels a growth of austenite transformation products began at clearly visible primary boundaries of γ phase (Fig. 4). Primary austenite grains are relatively huge (also for Nb-microalloyed steels). Lath-type microstructures indicate that the steels show high hardenability. The obtained structures are dependent on Mn content. Bainite is a main structural constituent of steels containing 3% Mn (Figs. 4a, b), whereas the steels with a higher Mn content are characterized by martensitic-bainitic microstructure (Figs. 4c, d). A few ferrite grains can be found in 3Mn steels (Fig. 4a) whereas 5Mn steels consist of distinct martensitic-bainitic laths (Figs. 4c, d). Each primary austenite grain consists of a few bainitic or bainitic-martensitic colonies. A characteristic feature of all the microstructures is the presence of retained austenite. It should be noted that this phase occurs already for air cooled specimens, without a special heat treatment used for TRIP steels [1-6].

It is expected that under conditions of carbon redistribution, a retained austenite fraction should be much higher. For 3Mn steels, this phase occurs mainly as fine granules distributed on primary austenite grains and inside the grains according with a growth of bainite laths (Figs. 4a, b). In 5Mn steels a majority of retained austenite is of an interlath type (Figs. 4c, d). An interesting morphological feature of steels containing 5% Mn is a structural constituent inside large bright laths located usually between individual packets of bainitic-martensitic regions (Figs. 4c, d). This phase does not occur in steels with lower Mn content. The morphology of this structural constituent is clearly visible in SEM (Fig. 5a). These are fine, regularly arranged plate particles inside bainitic ferrite laths. Linear analysis of the chemical composition across the bainitic and martensitic laths did not revealed any significant changes for individual structural components. A little increased Mn content can be seen at boundary regions of bainitic ferrite laths (Figs. 5a, c). The thickened continuous layer of the thickness of about 0.5 μm is probably retained austenite as a result of carbon redistribution during slow cooling. It is usually characteristic that the regions with higher Mn contents contain lower Al content, what is obvious considering the austenitic and ferritic nature of Mn and Al, correspondingly. It can be also observed for 3Mn steels, where some ferrite grains contain higher Al- and lower Mn contents (Figs. 6a, b). Both the linear and point analyses in Figs. 5 and 6 have mainly comparative meaning and the absolute values of chemical composition obtained by EDS should not be recognized as real values. Anyway, the Al content in ferrite grains is higher (Figs. 6a, c – point 2) when comparing to the rest of bainitic-martensitic-austenitic regions (Figs. 6a, c – point 1). The opposite is true for manganese.

The EDS spectrum (Fig. 5c) from the whole region in Fig. 5a did not revealed a significant increase of any alloying element, i.e. C, Nb, Al, N, Mn, Mo (with the exception of slightly increased Mn, Si and Al contents measured during the whole EDS investigations). In this way, the possibilities of the identification of fine particles in Fig. 5a as Nb(C,N) or AlN were excluded. It was not concluded only on a basis of the example in Fig. 5 but after the analysis of many similar regions in 5Mn and 5MnNb steels. The presence of many fine particles in a steel without Nb microaddition (Fig. 4c) confirms the conclusion concerning Nb(C,N). As for AlN, its precipitation at such a scale is low probable when taking into account a low concentration of nitrogen in all the steels (below 50 ppm).

The obtained structure in Fig. 5a is similar to that discovered in MoNiCuCrB steels by Zajac et al. [17]. They classified new, previously unexplored bainitic microstructures, which are beyond classification based on conventional concepts. The structure in Fig. 5a is most similar to degenerate lower bainite. It consists of very fine, fragmented lens-like martensite/austenite (MA) particles inside 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 ferrite laths. It is characteristic that MA particles replaced cementite and have their longest axes inclined at some 60° to the growth direction of the bainitic ferrite laths. Formation of degenerate lower bainite and inhibited cementite formation is connected with Al, Si, Mn and Mo alloying and cooling rate at which the transformation products are formed. This problem is more detailed explained in [17].

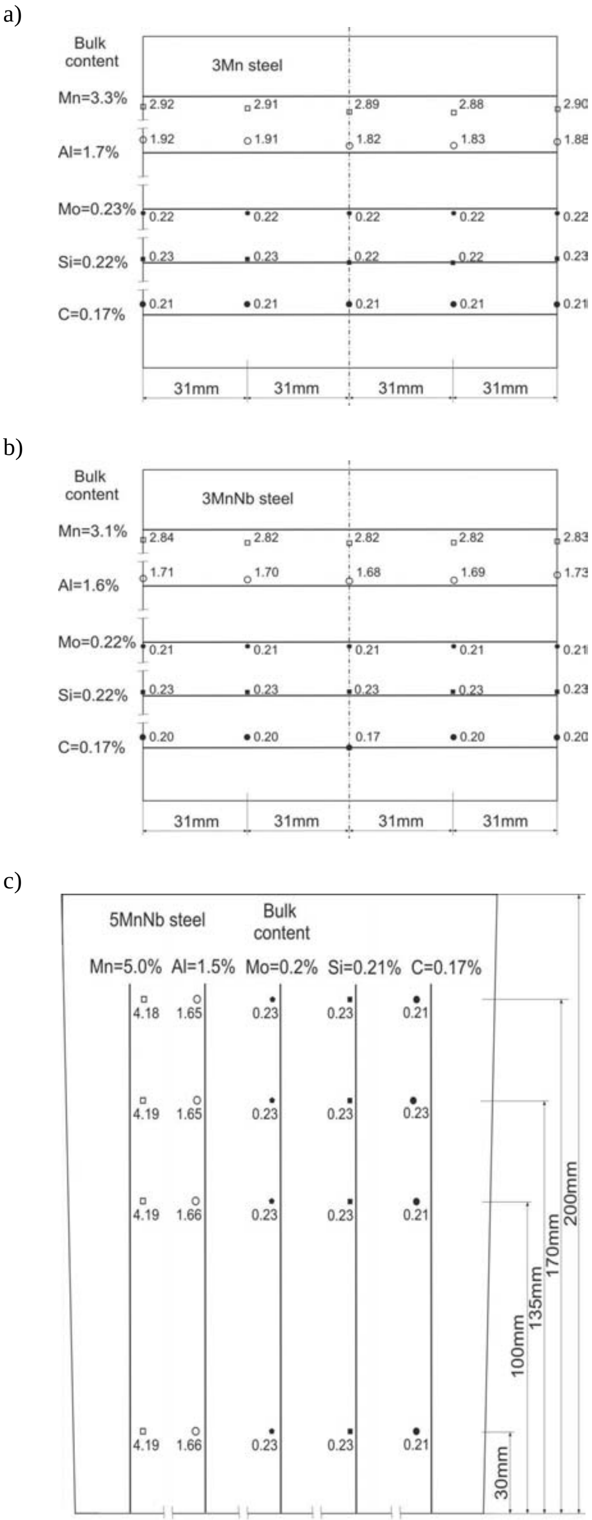


Fig. 2. Chemical composition measured by OES at the transverse sections of the 3Mn (a), 3MnNb steels (b) and at the longitudinal section of the 5MnNb steel ingot (c)

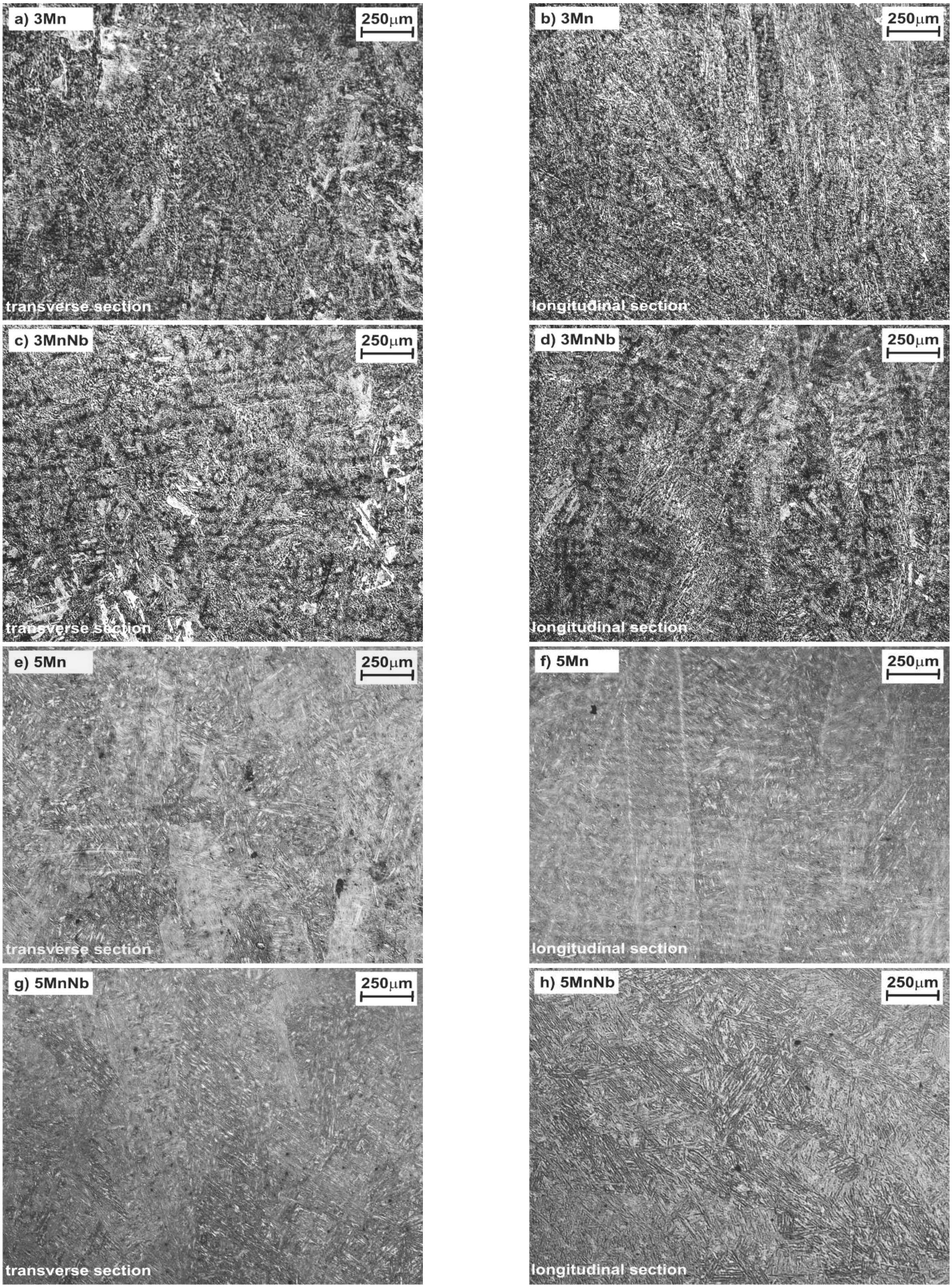


Fig. 3. Primary microstructures revealed at different sections of the 3Mn (a, b), 3MnNb (c, d), 5Mn (e, f) and 5MnNb (g, h) steels

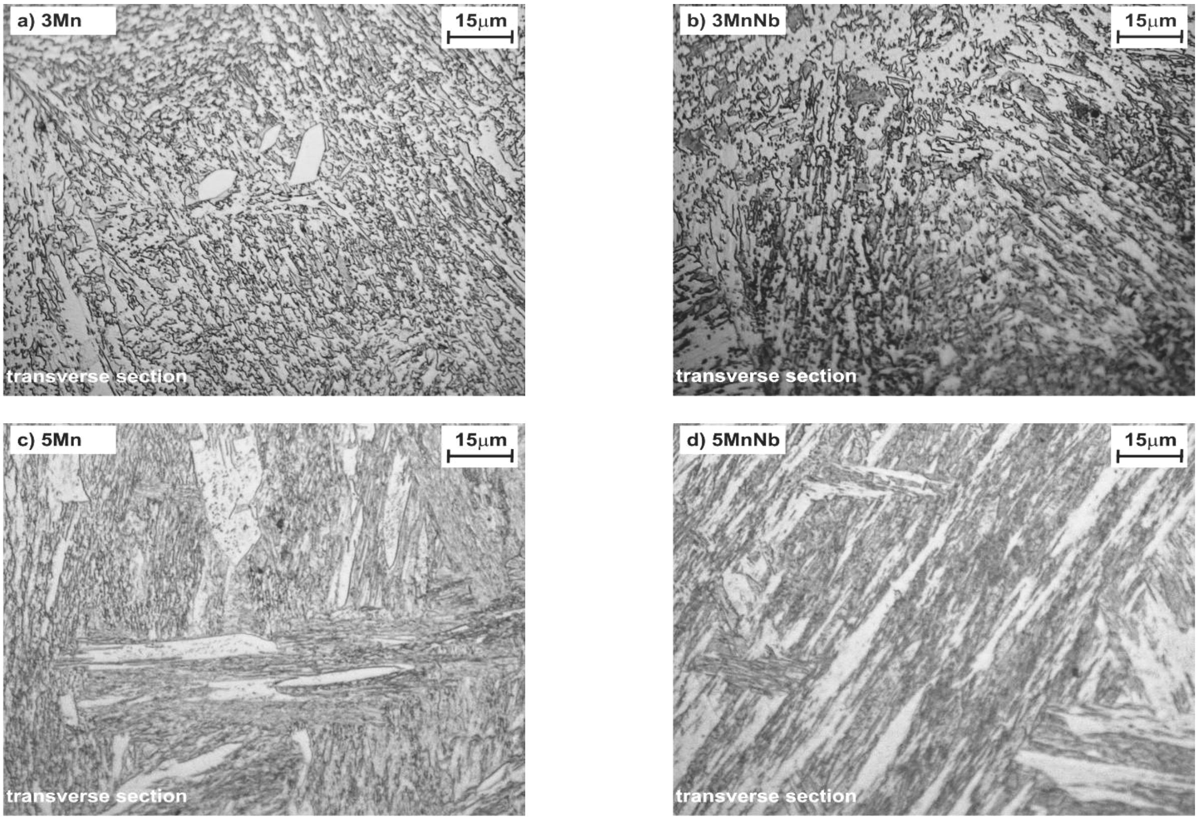


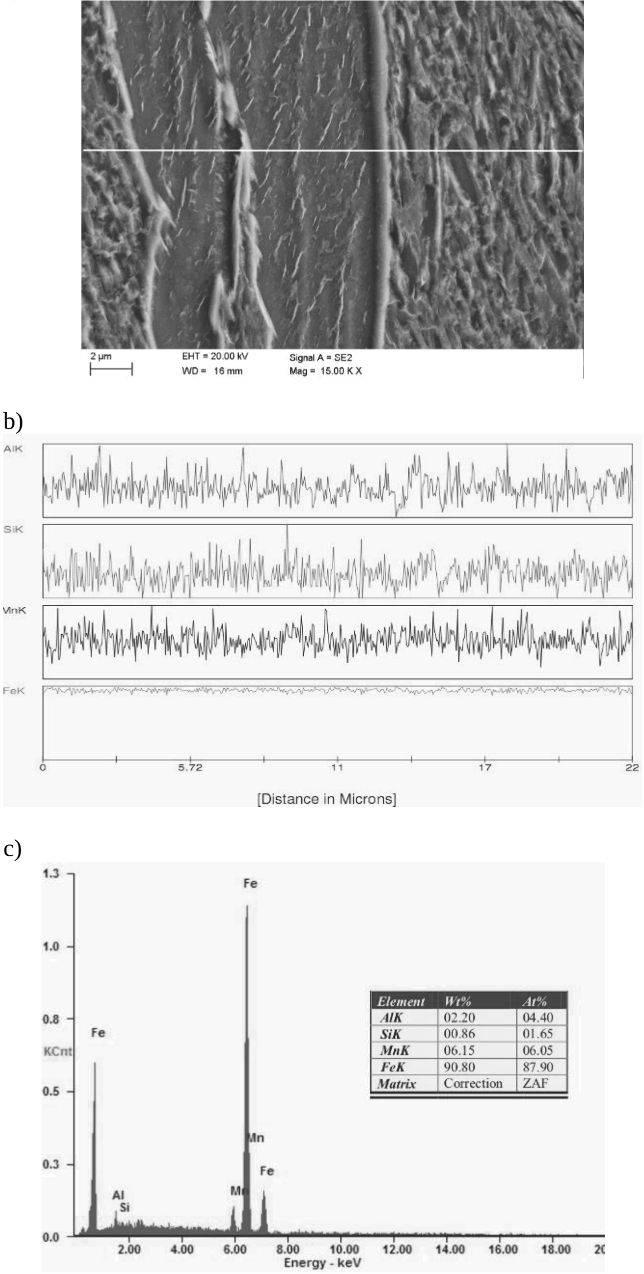
Fig. 4. As-cast microstructures of the 3Mn (a), 3MnNb (b), 5Mn (c) and 5MnNb (d) steels consisting of various fractions of bainite, martensite and retained austenite dependent on the Mn and Nb content

Excluding of $(Fe,Mn)_3C$ precipitation is comprehensible when considering Al-Si concept to inhibit the precipitation of carbides in TRIP steels [1-6, 9, 10]. No increased carbon content in EDS spectra (Figs. 5c, 6c) allow to exclude also the precipitation of molybdenum carbides or auto-tempering of martensite. Taking into account the effect of manganese on decreasing the stacking fault energy of austenite [9, 11-13] the presence of grouping stacking faults or ε martensite can not be excluded. A more detailed analysis of this problem deserves to be undertaken elsewhere.

After hot forging the dendritic structure was disappeared but the steels preserve lath morphology. A significant refinement of the microstructure is visible (Fig. 7). It concerns both the primary austenite grains and bainitic-martensitic regions. The microstructures are homogeneous what indicates that austenitizing of the ingots at 1200°C for 3 hours followed by hot forging is sufficient to remove main microsegregations of as-cast specimens. It is expected that applying the thermomechanical rolling as a next step of manufacturing process should reduce any remaining microsegragations. The fractions of retained austenite are higher when comparing to the as-cast state (Fig. 4). It is mainly due to deformation-induced carbon enrichment of the γ phase and it confirms the importance of the size stabilization effect of retained austenite in TRIP steels. At first sight the fraction of retained austenite is higher for 3Mn steels (Figs. 7a, b) but it could be not true when considering the interlath morphology of retained austenite in 5Mn steels (Figs. 7c, d). It requires the determination using X-ray. The fraction of retained austenite is comparable in 3Mn steels but the granules of the γ phase are smaller for the Nb-microalloyed steel (Fig. 7b). There is also a lack of large, blocky grains of austenite, which transformed into martensite at a final stage of cooling of specimens to room temperature. These structural constituents usually located between individual packets of bainite colonies are characteristic for the 3Mn steel (Fig. 7a). Their martensite start temperature is higher compared to fine granules of retained austenite inside bainite packets.

The martensite/austenite particles inside bright laths of bainitic ferrite are also present for the plastically deformed specimens (Figs. 7c, d). However, both the bainitic-martensitic laths and fine MA particles are highly fragmented compared to the as-cast specimens (Fig. 4b, c). A partial transformation of fine austenite particles into martensite is a result of the relatively high martensite start temperature of the steels because of high aluminum content. Again, the positive effect of C redistribution during the thermomechanical processing is expected to lower M_s of retained austenite below room temperature. The M_s temperature of interlath austenite is much lower, so it can be stable even for air cooling conditions.

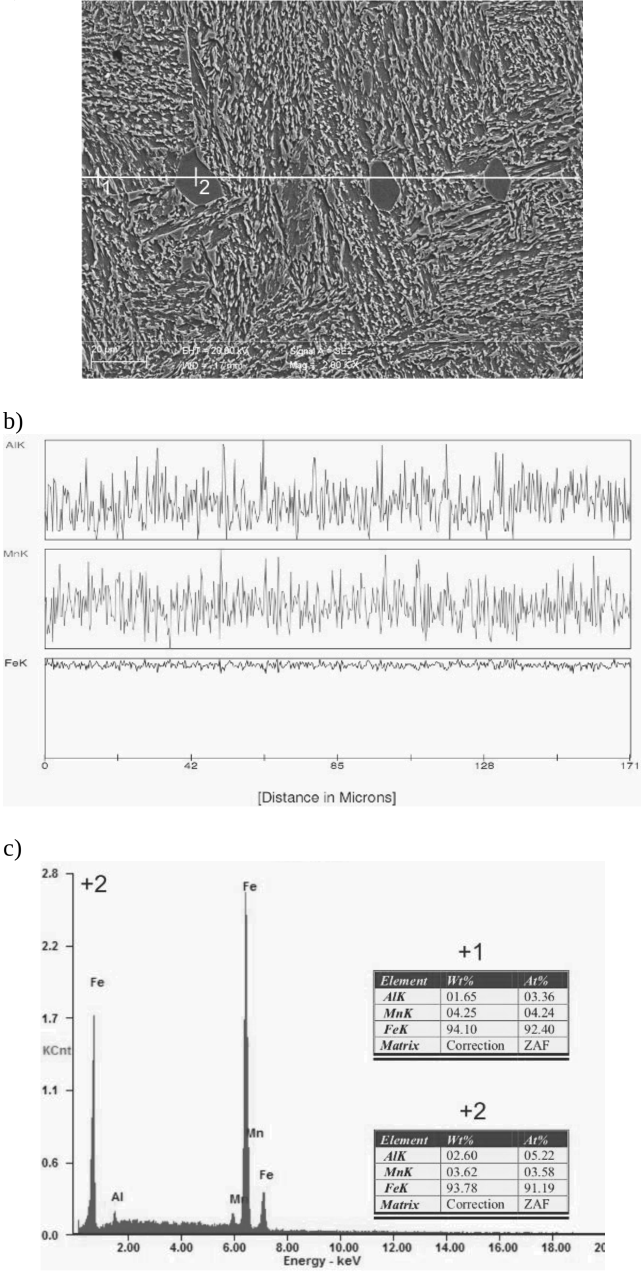
a) 5MnNb steel



Element	*Wt%*	*At%*
AlK	02.20	04.40
SiK	00.86	01.65
MnK	06.15	06.05
FeK	90.80	87.90
Matrix	Correction	ZAF

Fig. 5. Intra-lath martensite/austenite (MA) constituents as the example of the degenerate lower bainite structure in the 5MnNb steel (a), linear distribution of Al, Si, Mn and Fe concentration along the white line (b) and EDS spectrum from the whole region in Fig. a (c)

a) 3MnNb steel



+1

Element	*Wt%*	*At%*
AlK	01.65	03.36
MnK	04.25	04.24
FeK	94.10	92.40
Matrix	Correction	ZAF

+2

Element	*Wt%*	*At%*
AlK	02.60	05.22
MnK	03.62	03.58
FeK	93.78	91.19
Matrix	Correction	ZAF

Fig. 6. Bainitic-martensitic microstructure of the 3MnNb steel containing retained austenite and a few grains of polygonal ferrite (a), linear distribution of Al, Mn and Fe concentration along the white line (b) and EDS spectrum from the point 2 in Fig. a (c)

A problem of banding due to Mn enrichment in a center segregation zone of as-cast ingots and its potential inheriting by the hot-forged specimens required the EDS measurements in the middle and external parts of the plastically deformed flats. The results for all the steels are gathered in Table 3. Neglecting the absolute values of Mn and Al concentrations, it can be seen that the 3Mn and 5Mn steels have higher content of Al and lower content of Mn in the external part of the flat compared to its middle region.

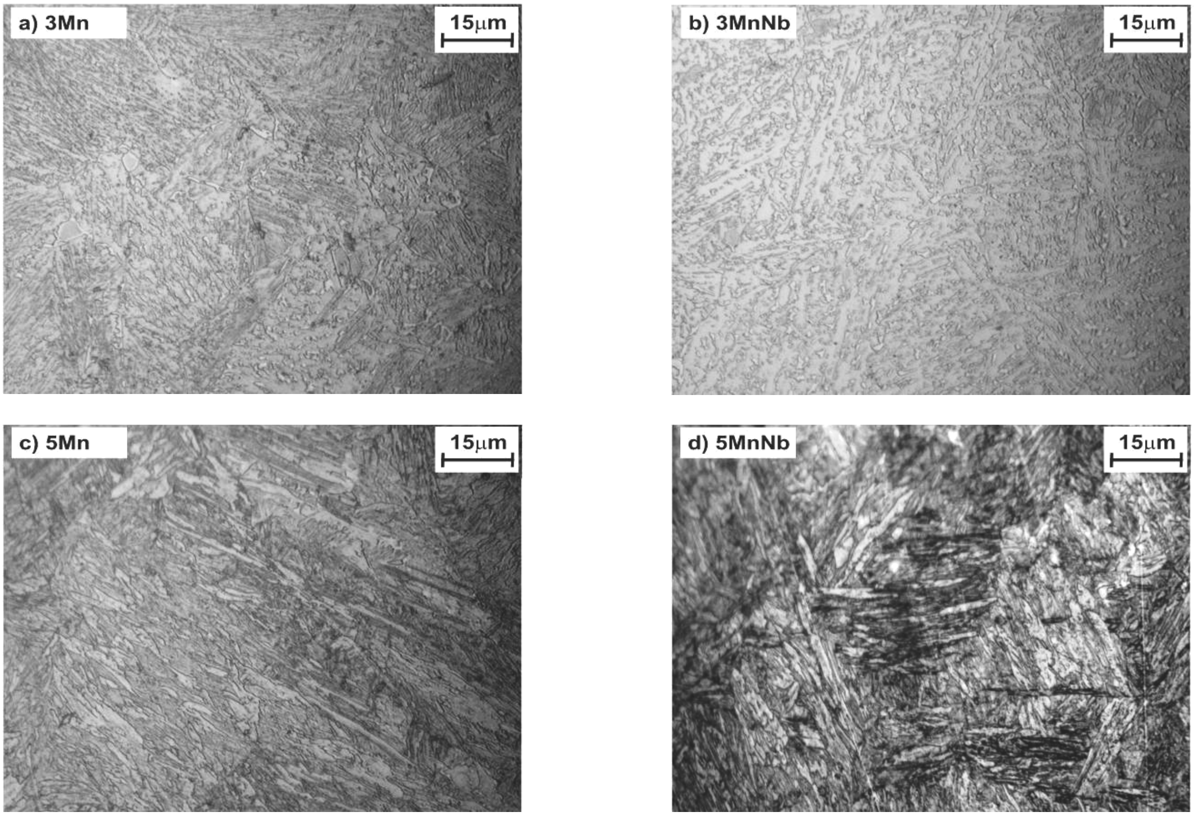


Fig. 7. Bainitic-martensitic microstructures of the 3Mn (a), 3MnNb (b), 5Mn (c) and 5MnNb (d) steels containing retained austenite of a different morphology dependent on the Mn and Nb content (after hot forging)

Table 3.
EDS measurements of the Mn and Al concentration in the middle and external parts of the plastically deformed flats (wt. %)

Zone	3Mn steel		3MnNb steel		5Mn steel		5MnNb steel	
	Mn	Al	Mn	Al	Mn	Al	Mn	Al
middle	3.26	1.34	3.17	1.01	6.03	1.16	5.25	1.55
external	2.59	2.27	3.09	0.96	5.12	1.33	5.42	1.81

The same situation has taken place for the as-cast specimens (Table 2). It indicates that despite a lack of significant microsegregation of Mn and Al in the as-cast state, the tendency to segregation of these alloying elements between middle and external parts of the ingots and the hot-forged flats exists. However, it should be noted that it is not true for the Nb-microalloyed steels (Table 3). The concentrations of Mn and Al are nearly the same at the whole thickness of the flats. A little increased Al content at the external zone of the flat was identified only in the 5MnNb steel. The most probable reason for the reduction of segregation of Mn and Al between the middle and external parts of the flats is the overall refinement of the structure and shorter diffusion paths of all the alloying elements under conditions of hot working. However, it requires a more detailed analysis.

4. Conclusions

The new-developed high-Mn high-Al TRIP steels are characterized by bainitic or bainitic-martensitic microstructures containing a large volume fraction of retained austenite already at the initial state. Retained austenite is present as fine granules for 3Mn steels whereas the interlath morphology of this phase prevails in 5Mn steels. A high fraction of retained austenite was obtained due to Mn alloying and Al-concept was beneficial for inhibition of carbide precipitation. New bainitic structures – similar to degenerate lower bainite – was identified in steels containing higher Mn content. The characteristic feature of this bainitic phase are the intra-lath fine martensite/austenite particles inside bainitic ferrite laths instead of cementite.

The tendency to segregation of Mn and Al between middle and external zones of the as-cast ingots was found. Manganese content is lower whereas Al content is higher in the external zone of the ingots. The opposite is true in the middle. The difference in Mn concentration compared to the bulk content are the highest for 5Mn steels and reaches 0.8 wt%. The segregation between central and external zones remains after hot forging but it is reduced especially for Nb-microalloyed steels. 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.

Niobium microalloying has the positive effects as for the overall refinement of the structure and retained austenite and decreasing the tendency of Mn and Al to segregation both for the as-cast and hot-forged steels.

Acknowledgements

The research work financially supported by the Polish Ministry of Science and Higher Education in a period of 2010-2012 in the framework of project No. N N508 590039.

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