

COMPARISON OF TRIP STEEL MICROSTRUCTURES AFTER THERMO-MECHANICAL TREATMENT WITH ISOTHERMAL AND ANISOTHERMAL DEFORMATION STEPS

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POROVNÁNÍ MIKROSTRUKTURY TRIP OCELÍ S IZOTERMÁLNÍMI A ANIZOTERMÁLNÍMI DEFORMAČNÍMI KROKY PO TEPELNĚ-MECHANICKÉM ZPRACOVÁNÍ

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Abstract

The processing strategy of thin-walled hollow parts was verified in this work by physical-material modelling using C-Mn-Si transformation induced plasticity (TRIP) steel microalloyed with Niobium. The specimens were processed by a thermo-mechanical simulator to simulate real processing conditions. Different processing methods were applied to the specimens during subsequent cooling. The influence of intensive forming which was carried out during a continual decrease of temperature was compared with the influence of deformation steps applied at chosen temperatures during cooling. The effect of deformation parameters on ferrite development, grain refinement and phase distribution was investigated. Resulting microstructures were characterized with the use of light optical microscopy, transmission electron microscopy of extraction carbon replicas and a new method of confocal laser scanning microscopy.

Key words: TRIP steel, niobium, confocal laser scanning microscopy, thermo-mechanical treatment

Introduction

Advanced low alloyed TRIP steels have mainly been used in the automotive industry to produce parts of vehicle frames. Among the advantages of TRIP steels are their high impact energy absorption and high strength to ductility balance. These good mechanical properties also

enable the application of TRIP steels to other than sheet steel products, for example to thin-walled hollow parts. It is not only necessary to achieve effective production of shaped parts but also to obtain optimal mechanical properties of products. The unique mechanical properties of TRIP steels are affected by a strain induced transformation of retained austenite to martensite during plastic deformation [1]. This is the reason why a convenient microstructure consisting of ferrite, bainite and retained austenite has to be acquired during the processing. However, even a convenient morphology and a higher volume fraction of retained austenite doesn't ensure better mechanical properties unless it is properly stabilized [1,2]. The stabilization is accomplished due to the chemical composition of steel and its thermo-mechanical treatment (TMT). The choice of a convenient TMT strategy is therefore crucial for achieving good mechanical properties. Various TMT strategies with isothermal and anisothermal deformation steps were therefore carried out in this work to analyse their effect on resulting microstructures with regard to the utilization of TRIP effect.

The commercial TRIP steels usually have a chemical composition of 0,2%C-1,5%Mn-1,5%Si and they might be further alloyed with aluminium, niobium or copper [3]. Microaddition of Niobium should improve mechanical properties of TRIP steel. It strengthens solid solution if dissolved in austenite at the beginning of TMT and precipitated particles of Nb(CN) contribute to the refinement of the steel microstructure. Niobium furthermore increases the volume fraction and stability of retained austenite [4].

Experimental Program

Low alloy C-Mn-Si-Nb multiphase steel was used in this work (Tab.1). Specimens were machined from forged bars and then processed by thermo-mechanical processing simulator. The simulator allowed temperature control and application of various numbers of tensile and compression deformations during the processing. Deformation temperature was monitored by thermocouples fixed to the surface of the central part of specimen.

Table 1 Chemical composition of the steel used (in weight %)

C	Mn	Si	Nb
0.21	1.45	1.80	0.059

Various strategies of TMT with different numbers of deformation steps were applied to specimens, where TMT 1-3 contain four isothermal deformation steps (Tab.2) while TMT 4-7 are composed of different numbers of anisothermal deformation steps (Tab.3). Niobium can be present in the microstructure of steel either as precipitated particles (Nb(CN)) or dissolved in solid solution and its effect on microstructure and mechanical properties of TRIP steels is different in each case. High austenitization temperatures of 1200°C and 1230°C ensure dissolution of niobium particles in austenite and consequently strengthening of solid solution. These temperatures were therefore used for all specimens. Determination of the most suitable austenitization temperature is described elsewhere [5]. A holding time of 300s at 420°C and subsequent air cooling to room temperature were also applied to all specimens.

The microstructure of the middle part of specimens after the treatment was characterized by means of light microscopy (LM) and transmission electron microscopy (TEM) of replicas. In addition to the above mentioned methods, laser scanning confocal microscope (LSCM) LEXT OLS3000 was also applied. It has the advantage of magnification of up to 14400x without the need for special equipment or specimen preparation requirements. Plane

fractions of bainite and ferrite were determined by quantitative evaluation and volume fraction of retained austenite was established by X-ray diffraction phase analysis.

Table 2 Thermo-mechanical treatment with isothermal deformation steps

	Austenitization [°C] / [s]	Deformation steps [°C] / [%]				Total def.
		1. st (compression)	2. nd (tension)	3. th (compression)	4. th (tension)	
TMT 1	1230 / 5	1100 / 25	1100 / 25	750 / 50	680 / 50	$\varphi = 2$
TMT 2	1200 / 10	1100 / 25	1100 / 25	950 / 50	850 / 50	$\varphi = 2$
TMT 3	1200 / 10	1100 / 50	1100 / 50	950 / 50	850 / 50	$\varphi = 2,8$

Table 3 Thermo-mechanical treatment with anisothermal deformation steps

	Austenitization [°C] / [s]	Deformation steps [°C] / [%]			Total def.
		1. st (compression)	2. nd (tension)	Incremental deformation (in turns compression and tension)	
TMT 4	1230 / 5	1100 / 25	1100°C / 25	8x def: 850°C-680°C, by cooling rate 3°C/s	$\varphi = 2,1$
TMT 5	1230 / 5	1100 / 25	1100°C / 25	8x def: 850°C-680°C, by cooling rate 10°C/s	$\varphi = 2,1$
TMT 6	1200 / 5	-	-	40x def: 1200-720°C, by cooling rate 10°C/s	$\varphi = 10,4$
TMT 7	1200 / 5	-	-	40x def.: 1200-600°C, by cooling rate 10°C/s	$\varphi = 10,4$

Results and discussion

TMT 1 was designed according to the results of previous experiments [5]. The microstructure after TMT 1 was ferritic-bainitic with 41% of bainite and about 10% of retained austenite (Fig.1). Transmission electron microscopy of carbon extraction replicas revealed the presence of martensite islands and also of islands of M-A constituent. Martensite became more visible when electrolytical polishing was applied (Fig.2) in addition to mechanical polishing. Equiaxial ferritic grains with a length of several μm were observed by TEM. Fine dispersion of Nb(CN) particles was found in all kinds of ferrite and moreover several coarser particles were observed both at the grain boundaries and within ferritic grains. The above mentioned findings were also confirmed by LSCM (Fig. 3).

The deformation steps were redistributed in TMT 2 and TMT 3. Moreover, the amount of deformation was increased in TMT 3 with two 50% deformation steps at 1100°C. The resulting microstructure after TMT 3 was ferritic-bainitic with about 41% of bainite and 11% of retained austenite (Fig.4). Sporadic smaller dark areas were observed by LM and extraction carbon replicas confirmed the presence of fine pearlitic areas (Fig.5) and the groups of small elongated particles growing from the grain boundaries and forming the base of prospective pearlite. Several different kinds of islands were observed at grain boundaries. Fine islands forming the net at the boundaries probably consist of retained austenite and so do some prolonged islands or films (Fig. 6). Some of the later ones had already partially transformed into martensite and correspond to M-A constituent. Larger islands of M-A constituent were also found outside the boundaries. Larger numbers of coarser particles were observed in the microstructures after TMT 2 and TMT 3 than after TMT with an austenitization temperature of 1230°C. The microstructure after TMT 2 was very similar to the one described for TMT 3, with slightly different phase fractions.

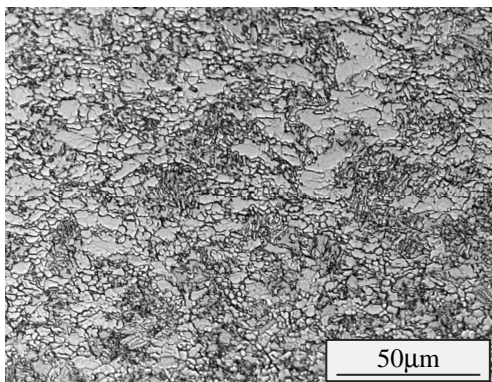


Fig.1 TMT 1, LM

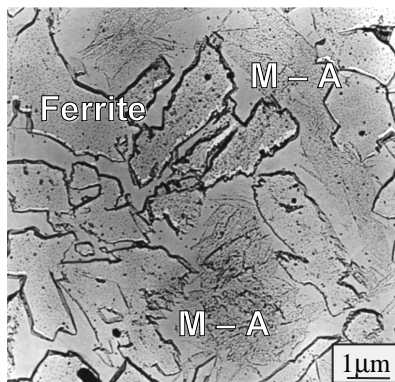


Fig.2 TMT 1, TEM

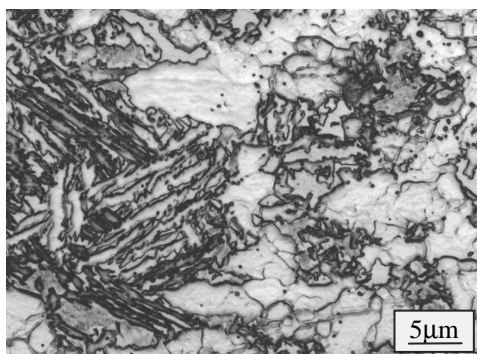


Fig.3 TMT 1, LSCM

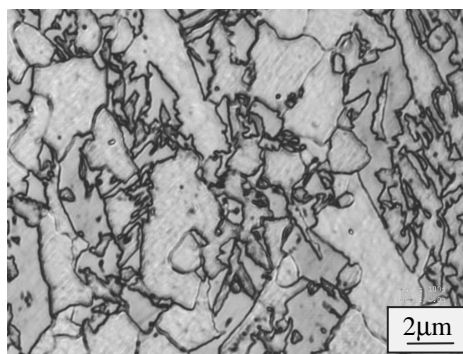


Fig.4 TMT 3, CLSM

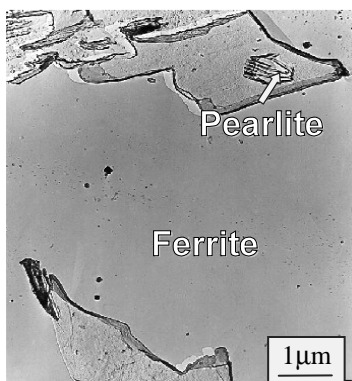


Fig.5 TMT 3, TEM

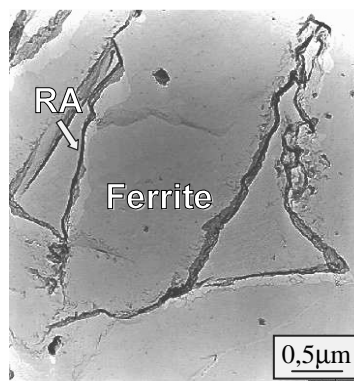


Fig.6 TMT 3, TEM

TMT 4, TMT 5 have nearly the same total deformation as TMT 1 and TMT 3, however their deformation history was very different. This fact is strongly reflected in their coarser microstructures which were significantly different from TMT 1 and TMT 2. TMT 4 and TMT 5 furthermore underwent cooling with different cooling rates among individual deformation steps. Despite this, their microstructures were very similar and consisted of larger bainitic blocks separated by fine ferritic grains (Fig.7). A higher amount of ferrite was achieved

after TMT 4 with a slower cooling rate. Martensite was also observed by TEM either as large islands of complicated shapes at the boundaries of bainitic blocks or as smaller islands formed between the laths of bainitic ferrite or among the grains of proeutectoid ferrite. The fraction of martensite was remarkably higher than in previous cases and bainite was in some areas composed mainly of M-A constituent islands locally interrupted by elongated islands of bainitic ferrite (Fig.8).

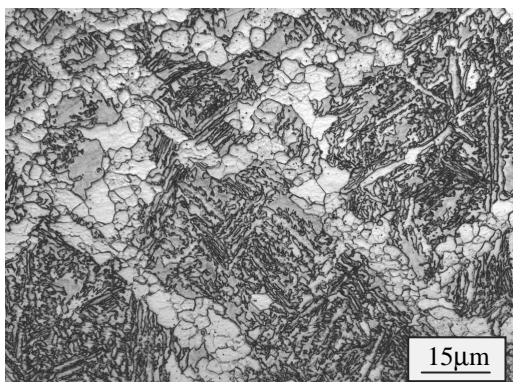


Fig.7 TMT 5, LSCM

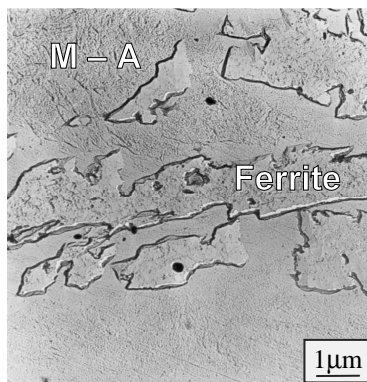


Fig.8 TMT 5, TEM

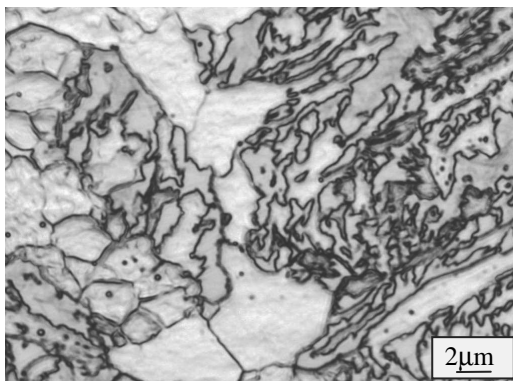


Fig.9 TMT 5, LSCM

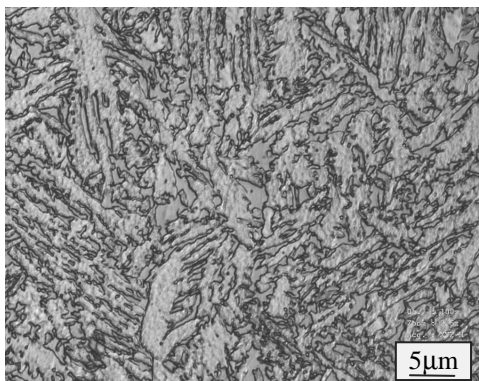


Fig.10 TMT 7, LSCM

The size of proeutectoid ferrite grains was several μm (Fig. 9) and it encompassed Nb(CN) particles. Plane fraction of bainite after TMT 5 and TMT 6 was 56 and 69% and volume fraction of retained austenite was 12 and 11% respectively.

Even more deformation was brought into the steel during TMT 6 and TMT 7 where 40 anisothermal deformation steps were applied. Resulting microstructures were even coarser than in previous cases. The microstructure was still ferritic-bainitic, however very large and compact bainitic blocks were formed with a chain of fine grains of proeutectoid grains lining their boundaries (Fig.10). Fine carbide particles in ferritic grains and several islands of M-A constituent were also found by LSCM. The size of bainitic areas varied between 20 and 50 μm , while the size of ferritic grains was kept down to several μm . Plane fraction of bainite was about 80% which is apparently too much for a TRIP steel. Considering that heating times in TMT 6 and TMT 7 were about one fifth of the heating times in TMT1-3 and austenitization hold was

relatively short, it can be assumed that Nb(CN) particles were not dissolved in solid solution and rather coarse primary carbides persisted in the microstructure. It furthermore implies that niobium could not in these cases influence phase transformations and its beneficial effect on ferrite formation and suppression of bainite formation was not utilized. On the other hand coarsening of microstructure due to the overgrowth of prior austenite grains might be excluded due to the short heating and austenitization times.

Conclusions

Several TMT strategies were applied to the TRIP steel microalloyed by niobium in this work to investigate their influence on final microstructure. Three of them were with isothermal deformation steps and four with incremental anisothermal deformation steps. Resulting microstructures were in all cases ferritic-bainitic with different volume fraction and morphology of phases. The most suitable microstructure for utilization of TRIP effect was achieved after TMT with four isothermal deformation steps which were carried out at temperatures of 1100°C, 950°C and 850°C. This microstructure, consisting of relatively homogeneously distributed bainitic areas, ferrite and about 10% of retained austenite, possesses a high probability of achieving good mechanical properties due to the TRIP effect. TMT with anisothermal deformation steps always resulted in coarsening of bainitic blocks in microstructure while ferritic grains lining the boundaries of prior austenite grains remained fine. The fraction of ferrite was in these cases rather low and thus unsuitable for TRIP steel.

Acknowledgment

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