

K-CARBIDES PRECIPITATION IN TWO HIGH MANGANESE ALLOYS

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ABSTRACT

Two high Mn alloys marked A (24Mn-1.2C-12Al) and B (28Mn-0.7C-8.6Al) were laboratory hot rolled. After rolling in material A coarser k-carbides were detected, sporadically on the austenite grains boundaries and more frequently on the ferrite ones. Reason was un-balanced Mn a C contents in superposition with the rolling deformations and finishing rolling temperature of 950°C. Hardness corresponded to 432HV30 on average. Using light microscope no carbides on grains boundaries were observed in B both after rolling and followed aging at 500°C/6'. After 6 minutes of aging the hardness of B was approx. by 9 % higher unlike the rolled state. The X-ray analysis nucleation of nano-size k-carbides confirmed. After aging of B at 500°C/30' and/or 60' some coarser k-carbides were detected on ferrite-austenite boundaries and/or austenite-austenite interface. The hardness went up and corresponded to 343 and/or 350HV30. From the X-ray and microstructure analyses follows the 30' and/or 60' dwell supported both very fine and coarser k-carbides precipitation. The calculated stacking fault energies (SFE) of both materials complete all solution. The A showed by 9 mJ.m⁻² higher SFE than the B. In the A alloy lower Mn content and higher Al one resulted to 2.8 times higher ferrite volume fraction in comparison with the B alloy.

Keywords: high manganese alloy, aging, k-carbide, hardness

1. INTRODUCTION

From the group of the high manganese materials, including the Hadfield steel, the TWIP and TRIPLEX alloys have proved to be highly attractive [1, 2]. Higher attention has paid to the TWIP variant and the TRIPLEX one seems to be neglected. The reason can be caused by some difficulties with higher Al content in the TRIPLEX material and its behaviour during own alloy production. The higher Al content makes the density of the high Mn alloy lower what is attractive e.g. for rotate components and makes also the SFE higher, leading to high stability of austenitic matrix [1, 2]. Schumann [3] presented basic diagram of phases stabilities in ternary Fe-Mn-C system. Some information was given about precipitation of the k-carbides in alloys with silicon addition [4, 5], however the variability of chemical composition of the high Mn materials offers existence of different austenite transformation products involving ferrite volume fraction in basis FCC matrix and carbide precipitation [2, 4-6]. For higher strengthening of the TRIPLEX matrix k-carbides of nano-size are prerequisite. Their precipitation can be formed at aging process realised at lower aging temperatures [1, 4, 5]. Among available works is not too concrete information regarding properties after various aging conditions. The aim of work is to show some metallographic parameters and the type of transformed particles in two different high manganese alloys without Si addition. Those alloys were investigated after hot rolling and material B was subsequently aged at 500°C/6' and/or 30' and/or 60'.

2. EXPERIMENTAL APPROACH

Two high Mn alloys marked A (24Mn, 12Al, 1.2C and Fe balac. - in wt. %) and B (28Mn, 8.6Al, 0.7C and Fe balance. - in wt. %) were cast to ingot mould with dimensions of 20x30x170 mm. Both

materials were heated at 1100°C and rolled using laboratory rolling mill Tandem of the VŠB-TU Ostrava. Before each odd double pass from the total 5 the reheating at above mentioned temperature was carried out. The first four reductions corresponded to 5 %, the others reductions to 13 and 20 %. The obtained material was again divided into two pieces and further rolled from the temperature of 1100°C twice with double pass. Final thickness corresponded to 1.9 mm and the finishing rolling temperature was approximately 950°C with air cooling. The measured cooling rate corresponded to 11 °C.s⁻¹. After hot rolling process hardness HV30 (EMCO TEST-MIC 01), microstructure parameters as grain size (by use of linear method) and volume fraction of ferrite in basic austenite matrix (OLYMPUS X70 microscope), microanalyses of Al content in austenite and ferrite (by use of SEM JEOL JSM 6490 LV equipped with energy dispersion analyser (EDA) OXFORD INCA Energy 350) were determined. After rolling and/or followed-up aging (500°C/6' or 500°C/30' and/or 500°C/60') a special attention was paid to carbide particles using the X-ray apparatus DRON with used (Co)_{Kα} radiation. The hardness determination completed the study. The part of solution was SFE calculation of both materials at ambient temperature. The calculation was realised according concept presented previously [7].

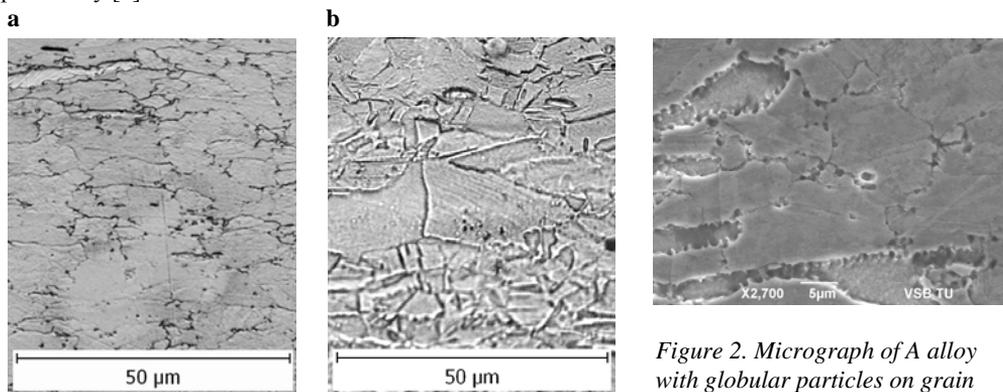


Figure 1. Micrograph of hot rolled alloys a) A, b) B.

Figure 2. Micrograph of A alloy with globular particles on grain boundaries

3. RESULTS AND DISCUSSION

In cast state the both alloys showed fine microstructure. The columnar structure and the presence of micro-porosity require suitable rolling parameters to break the primary structure and re-weld the tiny pits. After hot rolling the microstructures of A and B alloys are demonstrated in Figs. 1a, b and 2. Both micrographs show different features. The basic materials are the FCC structures, partially re-crystallised with different ferrite part. In A and B alloys 7.5 % and 2.7 % (in sequence) of ferrite was observed. Naturally, the higher Al addition contributed higher volume fraction of ferrite formation in basic austenite matrix. Using the SEM and EDA the Mn and Al contents in austenite and in ferrite grains were measured. The ferrite grains showed 14 wt. % of Al on average. This is 1.2 times higher volume fraction of Al than the total chemical analysis of the A alloy (see chapter 2) showed. In austenite was detected declared average chemical composition regarding both Mn and Al. After hot rolling the grain size of the A alloy corresponded to extent of 1-13 μm and to 5 μm on average, while the grain size of the B alloy was lying in extent of 2-20 μm and its average dimension was 7 μm. In the frame of investigation the hardness (HV30) was measured. After hot rolling the HV30 of the A alloy was lying within the extent of 425-441 and the average level showed 432 HV30. In case of the B alloy the minimal and maximal HV30 value was 291-316 and on average 305HV30. The calculated level of the SFE corresponded to 126 mJ.m⁻² (for the A alloy) and to 117 mJ.m⁻² (for the B alloy). Thanks the Al content of the A alloy its SFE was by 9 mJ.m⁻² higher than the SFE of the B alloy. At high temperatures the formation of coarser k-carbides in cooling process from the finishing rolling temperature can be detect in region -a- in the Fig. 3, where the -a- symbolises area of heterogeneous carbide formation in matrix and on boundaries, the -b- discontinuous inter-granular particle formation, the -c- represents formation of periodical structures, the -d- symbolizes area of formation of quasi-periodical structures, the -e- region of homogeneous k-carbide formation and the -γ- basic FCC matrix

[4]. The diagram was plotted for the Fe-28Mn-8.5Al-1C-1.25Si, however it is possible to suppose the Si presence will partially slow down the carbide formation in comparison with the alloy without Si addition. Similarly, the Al content shows the same tendency and at 12 wt. % of Al in the A alloy the influence could be balanced.

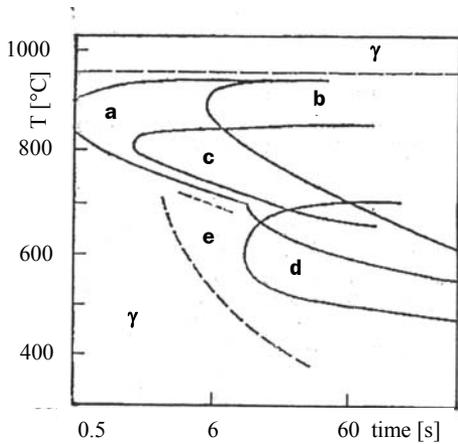


Figure 3. Carbides regions [4]

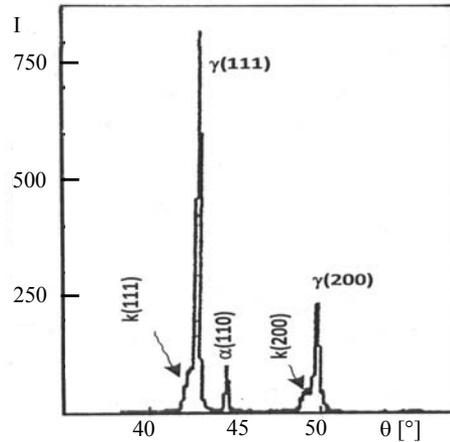


Figure 4. Part of X-ray diffraction

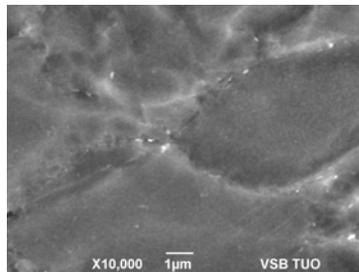


Figure 5. Micrograph of B alloy after aging at 500°C/30'

After hot rolling B alloy was aged 6' or 30' and/or 60' at 500°C. After the shortest aging time none k-carbide precipitation was observed in microstructure by use of light microscopy. Only 9.2 % hardness increase was detected. In given case the average hardness corresponded to 333HV30. It is evidence of very fine regular k-carbides precipitation of nano-size in matrix. It is in agreement with the -e- area in the Fig. 3. In Fig. 4 limited X-ray diffraction profile in θ range of Bragg's angle is presented after 6' of aging. The k-carbide reflections showing a weak intensity are connected with the austenite reflections, what confirms the coexistence of austenite with the k-carbides having the lattice structure of FCC $L1_2$ type. This figure shows characteristic reflections of coexisting tree phases and namely the FCC austenite, ferrite and traces of k-carbides presence. The k-carbides reflexes are linked to the FCC reflections what confirms the k-carbides coexistence with ordered austenitic matrix. The longer aging time the intensive k-carbide precipitation was registered. With longer aging dwell (30', 60') the coarser k-carbides (white particles in Fig. 5) of the micro-size on the ferrite grain boundaries and/or austenite interface were observed. After 30' and/or 60' of aging dwell the average HV30 corresponded to 343HV30 and 350HV30 (in sequence) and also a bit higher k(111) peak the X-ray diffraction profile showed. After longer aging times the hardness increase was 12.5 % and 14.8 % in comparison with alloy after hot rolling. Besides few nano-size k-carbides the quasi periodical carbides were formed and their precipitation was realised in the -d- area from the Fig. 3. Consequently, after 30' and/or 60' dwell of aging further k-carbides precipitation was detected, even when coarser. Using light microscopy the

coarser precipitates were observed on the austenite-ferrite and/or austenite-austenite interface. It is necessary to state, the B alloy showed 1.7 times lower carbon content and 1.2 times higher Mn one than the A alloy. Thanks that fact the austenite matrix of the B alloy was able to keep the C in solid solution during hot rolling and only the aging process lead to carbide precipitates. In the B alloy the number of carbides is much lower than in case of the A alloy as can be seen from the Figs. 2 and 5. With respect to k-carbides specific crystallographic structure for the X-ray detection, e.g. of the intermetallic FeAl or Fe₃Al types, prolonged aging times and temperatures 940-650°C would be necessary to use [8]. The (FeMn)₃AlC formation was connected with higher strengthening of the B alloy. The lower plasticity level and higher strengthening can be awaited with longer aging time as was it presented by Frommeyer [1].

4. CONCLUSIONS

Two high Mn alloys marked A (24Mn-1.2C-12Al) and B (28Mn-0.7C-8.6Al) were laboratory hot rolled. After rolling in the A alloy, numerous coarser k-carbides (up to 1 µm in diameter) were detected, sporadically on the austenite grains boundaries and frequently on the ferrite ones. Reason was un-balanced Mn and C contents in superposition with the rolling deformations and finishing rolling temperature of 950°C. The average HV30 corresponded to 432HV30. Using light microscope carbides on grains boundaries were observed in the B alloy neither after rolling nor after aging at 500°C/6'. After this aging the hardness (HV30) was by 9.2 % higher in comparison with the rolled state. The X-ray analysis confirmed nucleation of nano-size k-carbides. In the B alloy, some coarser k-carbides were detected on ferrite grains boundaries and on austenite-austenite interface after aging at 500°C/30' and/or 60'. With higher dwell time the hardness was higher and corresponded to 343 and /or 350HV30. As it from the X-ray analyses and the Fig. 3 follows, the 30' and 60' dwell supported both very fine and coarser k-carbides precipitation. Thanks the different Al contents of both alloys the calculated SFE showed un-significant difference of 9 mJ.m⁻², whereas the detected ferrite part in basic austenite matrix was 2.8 times higher in case of the A alloy. This was also reason of intensive C push out from ferrite to its boundaries after hot rolling.

5. ACKNOWLEDGEMENT

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