

STRUCTURAL-MECHANICAL PROPERTIES OF V-N MICROALLOYED 34CrMo4 STEEL AFTER CONTROL COOLING PROCESS

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Abstract

34CrMo4 steel is commonly used material for high strength applications such as shafts, forgings and high pressure steel cylinders. Mentioned steel is used in variety of microalloying by niobium, titanium, vanadium and nitrogen, respectively of those elements combinations. In presented paper three different heats of mentioned steel were studied. All heats were microalloyed by nitrogen and vanadium. Three heat treatment modes were applied. The first mode was based on heating at 700 °C, subsequently quenched and tempered at 470 °C. The second mode was the material heated at 890 °C and control cooling process followed. The third mode was similar to the second mode with differences in control cooling process temperatures. Microstructural investigation and mechanical properties testing were part of solution.

Keywords: 34CrMo4 steel, control cooling process, microstructure, mechanical properties

1. INTRODUCTION

The Cr-Mo steel grades are commonly used steels for the variety of applications, especially for the high strength applications. This paper is aimed to the 34CrMo4 steel, which is used for the variety of industrial and constructional applications, such as, shafts, forgings, rolled plates and also for the production of high pressure steel cylinders and vessels [1 - 3]. This steel type can be used in variations of microalloying and its combinations especially by the niobium, vanadium, titanium, boron and nitrogen as is presented in [4, 5]. There are two general ways how to gain high mechanical properties of this type of steel. First way is based on the thermo-mechanical processing and the second way is the control cooling process. This paper is aimed at the experimental control cooling processes, represented by two applied modes of this heat treatment type and one mode of reference experimental mode based on heating, quenching and subsequent tempering, applied to the steel 34CrMo4 after hot reversed extrusion from billet and subsequent reversed hot rolling to the thickness of 10.2 mm. This experimental procedure is one of the possible ways, how to gain high mechanical properties, mainly yield strength (YS) and tensile strength (TS) and homogenous microstructure in transverse and longitudinal testing direction what should reflect in balanced values of notch impact strength (CVN) in both directions of testing.

2. EXPERIMENTAL PROCEDURES

Experimental investigation started by the quantitative chemical analysis of three used heats. This procedure was realized by use of the chemical analyzer Spectrolab 2000. After the chemical composition investigation, 3 heat treatment modes were realized. First applied mode (HT1) was based on the heating at 700 °C and fast cooling in quenching bath with subsequent tempering at 470 °C. Second mode of heat treatment (HT2) was the controlled cooling process based on the heating to the austenitic zone with subsequent fast air cooling to the intercritical zone (IZ), then fast cooling in quenching bath to the temperature just above the martensite start (Ms) temperature followed by air cooling to the temperature 40 °C under the Ms with final step of cooling in the quenching bath to the room temperature. After this controlled cooling process the tempering at 470 °C with final cooling on air followed. Third mode of heat treatment (HT3) was also the controlled cooling process based

on the heating to the austenitic zone with subsequent fast air cooling to the IZ, then fast cooling in quenching bath to the temperature 200 °C above the martensite start (M_s) temperature followed by air cooling to the temperature 90 °C above the M_s with final step of cooling in the quenching bath to the room temperature. After this controlled cooling process the tempering at 470 °C with final cooling on air also followed.

The testing of mechanical properties was realized as the next step of experimental procedures. This procedure was based on the YS, TS, elongation, Brinell hardness (HBW) and CVN in transverse and longitudinal direction testing. Testing of YS, TS and elongation was realized by use of the Zwick/Roell Z 250 machine according to EN ISO 6892-1 standard. Brinell hardness was measured using the M4U750 hardness testing machine according to EN ISO 6506-1. Testing of the CVN was carried out by use of the RKP 450 Charpy Impact Testing Machine according to standard ISO 148-1 at - 50 °C. As the next step of investigation, the metallographic observation by use of Neophot 21 light-optical microscope was realized. This procedure was divided to two blocks. First block was focused on the micropurity evaluation according to ČSN ISO 4967 and the evaluation of gained microstructures on grinded and polished samples in etched state (etching in Nital). Second block was based on the grain size measuring according to ASTM E 112. The final step of the experimental procedure was the fractographic analysis carried out. As samples were used tested CVN samples from the transverse direction of testing. This analysis was carried out to evaluate the morphology and characteristics of fracture samples after the applications of 3 different heat treatment modes. For this observation, the SEM JEOL JSM-6490 equipped with X-ray analyzer EDA was used.

3. RESULTS AND DISCUSSION

Three resulting chemical analysis are presented in **Table 1** and revealed that all 3 used heats were significantly similar with very low content of undesirable P and S elements. Chemical composition of all 3 heats fully corresponds with the standardized chemical composition of 34CrMo4 steel.

Resulting values of all tested mechanical properties are shown in **Table 2** and showed the lowest values of tensile properties in case of HT1 with the low values of CVN in both testing directions and difference of 3 J.cm⁻². The application of HT2 showed that values of tensile properties were between the values of HT1 and HT3 by the tensile properties however CVN values were significantly higher than in cases of HT1 and HT3 and revealed balanced values in both testing directions. Resulting elongation and HBW values fully correspond with the increasing trend of tensile properties as can be seen in **Table 2**.

Table 1 Chemical composition of investigated steel of 3 different heats [wt. %]

Heat	C	Si	Mn	Cr	Mo	P	S	V	N
A	0.37	0.26	0.85	1.15	0.20	0.011	0.002	0.071	0.0102
B	0.36	0.26	0.84	1.12	0.21	0.012	0.002	0.075	0.0133
C	0.37	0.28	0.84	1.15	0.21	0.012	0.005	0.072	0.0105

The non-metallic inclusions evaluation of all 3 heats revealed considerably high cleanness level of all 3 heats. As the **Table 3** shows, first two heats showed only slightly increased values of formable and globular oxides characters impurities. The last heat showed slightly increased amount of inclusions of sulphidic character but on the other hand, the content of formable oxides inclusions was reduced close to 0, see **Table 3**. The grain size evaluation showed in **Table 3** revealed that the finest grain was observed in case of heat B and HT2. The best values of tensile properties vs. CVN in both directions of testing could be ascribed to detected grain size. The grain size of the other 2 heats was similar from value 8 to 10.

Table 2 Resulting mechanical properties of 3 heats after 3 different modes of heat treatment

Heat	Heat treatment mode	YS [MPa]	TS [MPa]	Elongation [%]	CVN transverse [J.cm ⁻²]	CVN longitudinal [J.cm ⁻²]	HBW (2,5/187.5) [-]
A	HT1	676	824	17.6	22	19	248
B	HT2	1120	1257	12.5	34	35	402
C	HT3	1272	1397	9.8	23	22	424

Table 3 Content of non-metallic inclusion in 3 used heats

Heat	Heat treatment mode	Sulphides fine/coarse	Banded oxides fine/coarse	Formable oxides fine/coarse	Globular oxides fine/coarse	Coarse globular oxides fine/coarse	Grain size
A	HT1	0/0	0/0	0.3/0.5	0.8/0.1	0	10
B	HT2	0/0	0/0	0.5/0.3	0.8/0.1	0	11
C	HT3	1.1/0	0/0	0/0.1	0.8/0.1	0	8

Microstructure was evaluated from the areas close to the inner surface of the semi-product of high-pressure cylinders. The reason for this choice of place is that the cylinders are most commonly heat treated in closed state. That means there is no access of quenching medium to the inner surface of cylinders. This causes that trapped heated gases are heating the inner surface and slightly under inner surface area of material for longer time than the outer surface, where the cooling rate is much faster due to the direct contact with the quenching medium. This is resulting in the more noticeable segregation banding in the under inner surface area due to longer times for possible segregation processes. Microstructure after the application of HT1 revealed predominantly low bainite partially mixed with ferrite, accicular ferrite (AF) and pearlite, see **Fig. 1 a and b**. This microstructure is similar to the microstructure after the slow cooling on the air after the hot forming temperatures of reverse rolling which allowed significant segregation of elements such as Cr, Mo, and C to the segregation bands. Segregation bands in this case are moderately thick and of continuous character. The heating at 700 °C was realized approximately 25 °C under the Ac₁ temperature due to the deficient heating temperature, the austenitization process was not achieved. Resulting coarse microstructure contained numerous carbide particles, which precipitated during the tempering process. Significant banded microstructure, which originated during the process of forming and subsequent cooling, was not reduced due to low heating temperature. Due Resulting tensile properties fully correspond with the microstructure and the differences of CVN values, see **Table 2**, in testing directions are consistent with the segregation banding in microstructure.

Figs. 1c and 1d show the microstructure obtained by the HT2 application and reveal considerably finer morphology and segregation bands are significantly reduced. Retained bands are thinner and discontinuous. This corresponds with the highly balanced results of CNV in both testing directions as it can be seen in **Table 2**. Microstructure consists of the tempered martensite, AF and in small portion the low bainite. The phase of AF was obtained by the cooling on air from the fully austenitic zone to the IZ. Subsequent fast cooling in the quenching bath to the temperature just above the M_s temperature ensured that segregation processes were eliminated to the minimum according to the resulting reduced banded microstructure. Cooling on air to the temperature 40 °C ensured the partial low bainitic transformation and by the subsequent quenching in quenching bath was the martensitic microstructure gained.

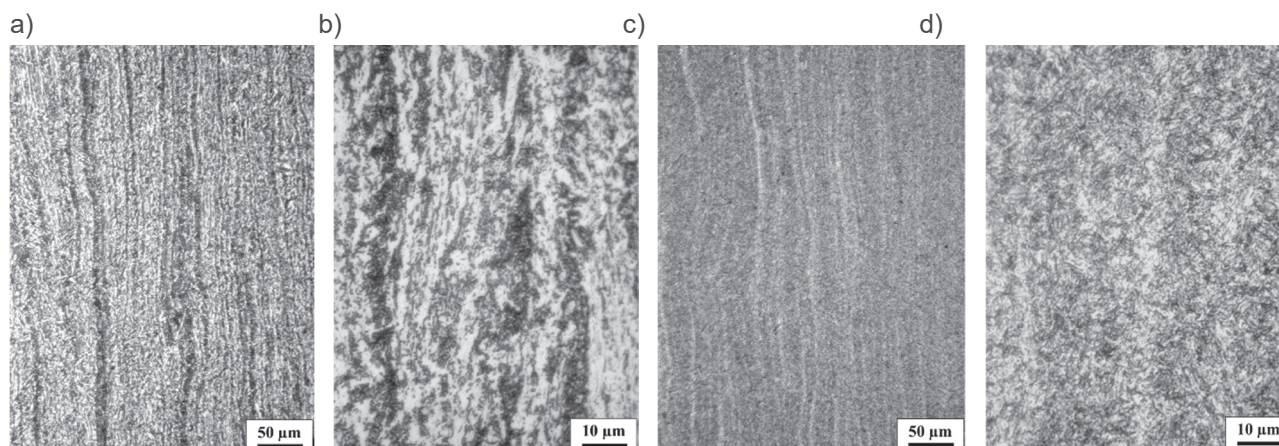


Fig. 1 Microstructure after the application of HT1 a) general view, b) detailed view of microstructure, microstructure after the application of HT2 a) general view, b) detailed view of microstructure

Fig. 2 presents the microstructure after the application of HT3. This microstructure was based on the tempered martensite, AF and polygonal ferrite. Also in this case as it was in case of application of HT2, the AF phase was obtained due to the cooling from fully austenitic zone to the IZ and segregation banding was considerably eliminated, however cooling on air to the 90 °C above the M_s from temperature 200 °C above M_s obtained by the cooling in quenching bath, caused undesirable extended formation of polygonal ferrite which disrupts the homogeneity of microstructure from the mechanical properties, sulphide stress cracking, etc. point of view. This phenomenon appeared due to the long lasting cooling on air from 200 to 90 °C with the contribution of hot trapped gases inside of heat treated cylinder which were continuously heating the inner and under inner surface area of material. Eliminated segregation banding directly reflected in balanced values of CVN in both testing directions as it can be seen in **Table 2**. The martensitic transformation was ensured by the fast cooling in the quenching bath from the temperature 90 °C above M_s to the room temperature. By the subsequent tempering was primarily tempering of martensitic phase and the precipitation of carbides particles in martensite and low bainite realized as it was in case of HT2. The higher YS and TS were ensured by the higher temperature, from which the quenching was realized compared to the HT2. Coarser grain size probably caused the decrease of CVN values and elongation together with higher level of tensile properties.

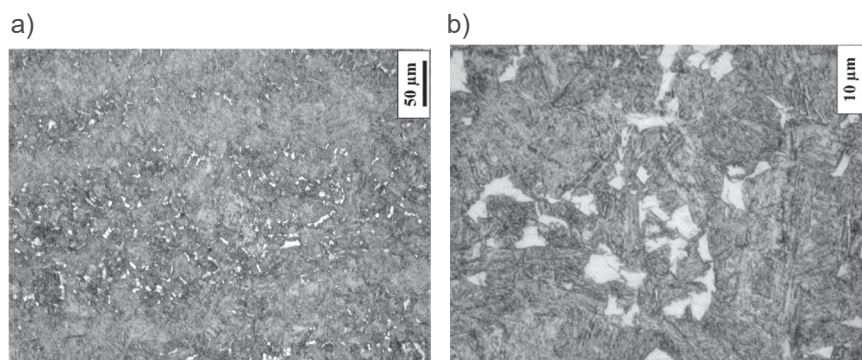


Fig. 2 Microstructure after the application of HT3 a) general view, b) detailed view of microstructure

The micro-fractographic analyses of fractured CVN specimens surfaces showed in all cases of 3 used heat treatment modes trans-crystalline cleavage fracture. As **Fig. 3** shows, the fracture surface revealed combined coarse and fine cleavage facets, which are not decorated by ductile ridges. This could be the reason for the disproportional values of CVN in transverse and longitudinal testing directions. This observed fracture surface correspond with achieved mixed microstructure of low bainite, pearlite and AF and also resulting lowest CVN

values, see **Table 2**, as similar results were achieved and presented in [7]. Fracture surface of specimens after the application of HT2 revealed also trans-crystalline cleavage fracture as it was in case of application of HT1, however cleavage facets were significantly finer and even, see **Fig. 4**. Cleavage facets were partially decorated with thin ductile ridges what was the probable cause of increased and balanced CVN values together with their lower dimensions and therefore increased fracture area. Balanced tensile properties vs. CVN in this case were achieved also due to the finest grain size what directly corresponds with the morphology of fracture surface and elimination of unfavorable segregation banding.

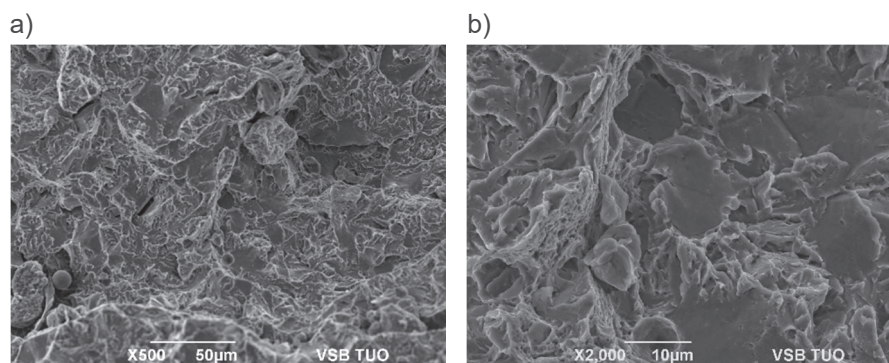


Fig. 3 Fracture surface of CVN specimen after the HT1 a) general view, b) detail of fracture surface

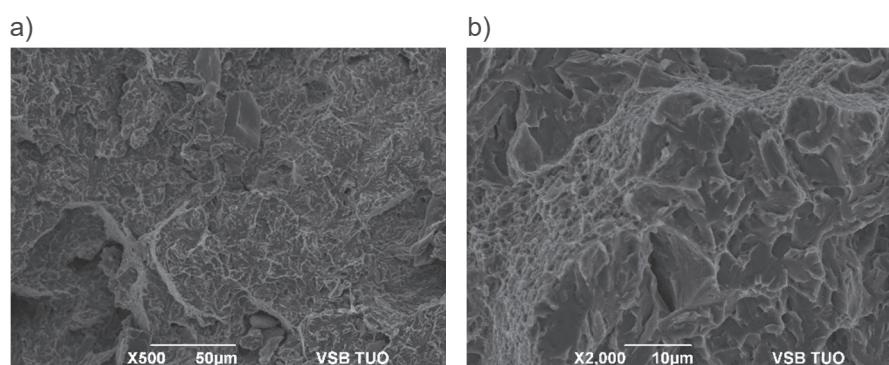


Fig. 4 Fracture surface of CVN specimen after the HT2 a) general view, b) detail of fracture surface

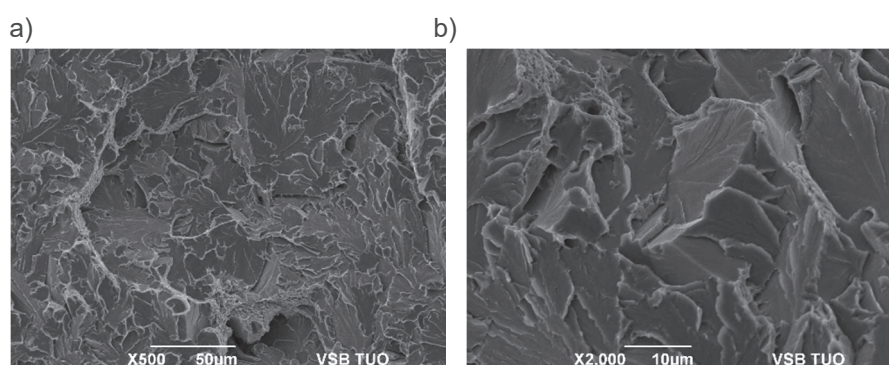


Fig. 5 Fracture surface of CVN specimen after the HT3 a) general view, b) detail of fracture surface

Fig. 5 shows the fracture surface of CVN specimen after the application of HT3. Also in this case the fracture surface revealed the trans-crystalline cleavage fracture with extend cleavage facets with creeks. Ductile ridges were not observed almost in any degree, if so, ductile ridges were short and thin. This morphology fully corresponds with the lowest grain size of used heat for this mode of heat treatment comparing to other 2 used

heats with finer grain size. This fracture also corresponds with the highest tensile properties with the lowest value of elongation.

CONCLUSION

In this paper are presented results of mechanical properties testing, metallographic and fractographic observation results after the application of 3 different heat treatment modes. Best results of achieved mechanical properties from the point of view of tensile properties vs. balanced and high CVN in two testing were achieved in case of HT2, where the segregation banding in microstructure was significantly eliminated compared to the HT1 and HT3 modes. Microstructure obtained by the HT2 mode revealed slightly residual discontinuous segregation banding and its elimination will be possible by increasing the cooling rate from the austenitic to IZ. The subsequent time and tempering temperature should be increased in order to moderately increase values of elongation and CVN at the cost of moderate decrease of TS to approximately 1200 MPa. This experimental procedure proved that the most influencing factor of gaining the high tensile properties vs. as high values of CVN as possible is the grain size as it was achieved and investigated in [6 - 8]. The increase of grain size in input billets will be also the important point of further investigation.

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