HEAT TREATMENT, MICROSTRUCTURE AND PROPERTIES OF 75CR1 STEEL, FOR USE IN HEAVY LOADED ELEMENTS

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Abstract: *This study aims to optimize the heat treatment of tool steel 75Cr1 which is used for heavy loaded elements in transmissions. A salt bath was used to quench and temper the steel at different temperatures. Mechanical tests and microstructural characterization were done to define the heat treatment parameters corresponding to the optimal performance of the elements. Optical microscopy, electron back scatter diffraction and x-ray diffraction were used to characterize the microstructure, while tensile tests and toughness tests were employed to determine the mechanical properties after different heat treatments. It was found that the yield strength decreases with increasing annealing temperature and that the toughness decreases with increasing annealing time and temperature. The changes of the mechanical properties are discussed in relation with the thermal treatment and the corresponding microstructures.*

Keywords: HEAT TREATMENT, TOOL STEEL, MICROSTRUCTURE

1. Introduction

Important factors affecting competition in the automotive market are product quality, innovation and development time, price, reliability, safety, fuel economy and emission control [1]. The Continuously Variable Transmission (CVT) is a very efficient system with a high potential for future application in new generation vehicles. This type of automatic transmission provides low fuel consumption, no torque interruption during gear change and optimum power range of the engine [2]. Instead of using fixed gears, the CVT uses two pulleys connected by a pushbelt to transmit torque. The pushbelt is built up of hundreds of steel elements and two maraging steel ring packs. The elements are subjected to high cyclic loads and they should possess a high yield stress and wear resistance. Therefore a tool steel 75Cr1 is used for the production of the elements.

The steel is annealed at varying temperatures and soaking times and subsequently oil quenched and tempered. The steel contains chromium, which participates in the cementite formation. According to [3] the alloying elements partition between the austenitic matrix and the retained (undissolved) carbides during austenitizing. This partitioning fixes the chemistry and volume fraction of the carbides as well as the composition of the austenite. Finer carbides and a larger volume fraction of carbides will control the austenitic grain growth and can influence significantly the mechanical properties. Hence, the goal of the work is to understand better the influence of the heat treatment parameters on the strength, elongation and toughness of the elements made of 75Cr1 steel.

2. Experimental procedures

The nominal composition of the steel is given in Table 1. This high carbon low alloyed steel is classified as tool steel. The as received material has a thickness of 1.8 mm and is in the so called soft annealed (or spheroidization annealed) condition. In this condition the steel microstructure is ferrite with $(Fe,Cr)_3C$ carbides. The steel was subjected to 40 different heat treatments in a salt bath with varying annealing temperatures and times and subsequently oil quenched to form martensite. The different temperatures were 800, 820, 860 and 900 °C, and at each temperature the samples were soaked for 2, 4, 6, 8, 10, 12, 15, 20, 25 or 30 minutes before oil quenching. Afterwards all samples were tempered at 190 °C for 45 min. For every set of heat treatment parameters, 3 tensile samples, 30 samples for the toughness test and several extra samples for the microscopy tests were prepared.

Table 1: *Nominal composition of 75Cr1.*

Element	C	Si	Mn	P	S	Cr	Fe
wt.%	0.74-0.80	0.25-0.40	0.65-0.80	max. 0.025	max. 0.010	0.30-0.45	bal.

XRD analyses were performed with a Siemens Diffractometer D5000 with a molybdenum X-ray tube. The angular range of 25°-40° was scanned with a step size of 0.05° and 5 s/step. Using the formula of Cullity [4], the fraction of retained austenite was calculated for different heat treatments. The electron backscatter diffraction (EBSD) data were acquired in a FEI Nova 600 Nanolab at 20 kV and beam current 2.2 nA. The sample preparation consisted of mechanical and electrolytic polishing with A2 Struers electrolyte, followed by 40 min OP-U polishing with pressure less than 5 N. Tensile specimens were prepared according to the ASTM standard E8M-04 (gage length 50 mm). Tensile tests were carried out on a Zwick Z250 testing machine with a speed of 20 MPa/s until the yield point, and 0.0067 mm/s cross head speed for strain controlled plastic deformation until fracture. The Izod impact tests were executed on a Zwick/Roell HT55P machine.

3. Results

Equilibrium calculations of the Fe-C-Mn-Cr-Si system were performed using the Thermo-Calc software. From the Fe-C isopleth in Fig. 1 one can derive that for 0.75% C the Ae_1 and Ae_{cm} temperatures are 722 and 770 °C respectively and that ferrite, cementite and austenite are in equilibrium in the temperature range between 722 and 734 °C.

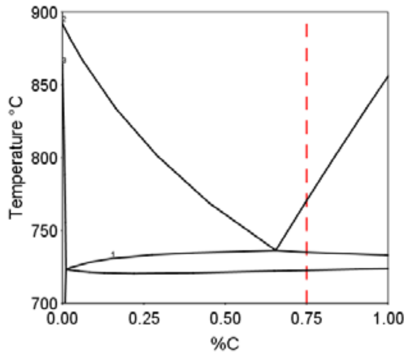

Fig. 1 *Fe-C isopleth of the alloy, the carbon content of the 75Cr1 steel is indicated with a red dashed line.*

Fig. 2 displays the variations in the equilibrium phase fractions for the 75Cr1 steel grade as a function of temperature. According to Thermo-Calc calculations cementite is the stable carbide among the other complex carbides. When the annealing temperature is above 734 °C, the maximum equilibrium cementite fraction is 1.6%.

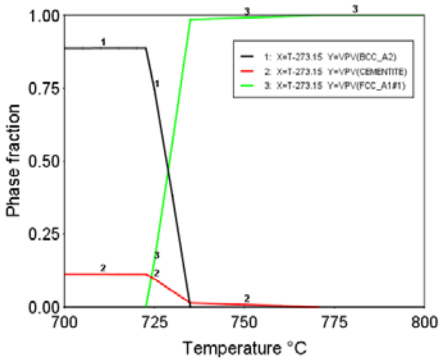

Fig. 2 *Relationship between phase fraction and temperature in 75Cr1.*

The microstructure of the as received material in the soft annealed condition is visualized in Fig. 3. Etching with 4% HNO_3 in ethanol (4% Nital) for 8-11 s reveals clearly the grain boundaries and the spherical carbides (Fig. 3a). The volume fraction of carbides measured by quantitative metallography on multiple samples was about 12-13%. The EBSD maps (Fig. 3b, c) show that around 8% spherical iron carbides are present in the ferritic matrix. This underestimation of the carbide fraction measured with EBSD in comparison to the optical microscopy data is most probably due to the smaller area of the EBSD scan and the incorrectly indexing of some carbide orientations.

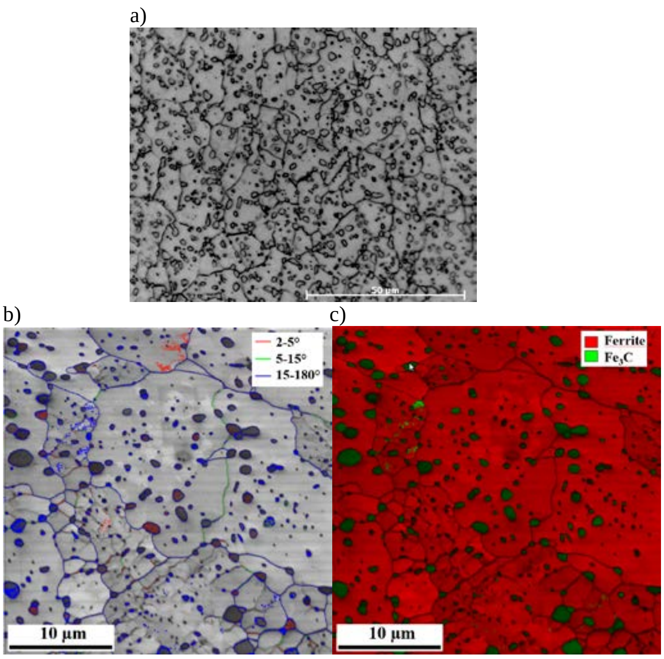

Fig. 3 *Microstructure of the as received material; a) optical microscopy after Nital etching, b) EBSD image quality and grain boundary map, c) EBSD image quality and phase map superimposed.*

Fig. 4 shows the image quality and inverse pole figure maps of a sample annealed at 860 °C for 15 minutes which is oil quenched and afterwards tempered. The martensitic structure typical for high carbon steel is visible. In comparison with the starting material, less and smaller carbides can be observed. Some small regions of austenite in between martensite could be distinguished in Fig. 4c.

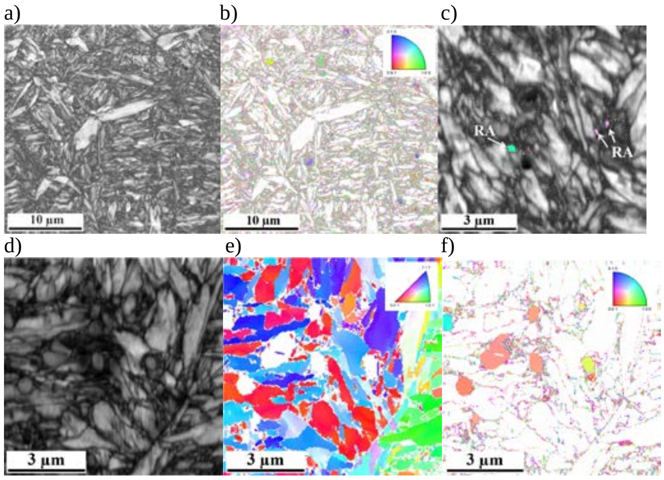

Fig. 4 *Microstructure of 75Cr1 annealed at 860 °C for 15 min, quenched and tempered; a) image quality map, b) same area as in 'a', inverse pole figure map of the carbides, c) image quality map with retained austenite (RA), d) image quality map e) inverse pole figure map of the ferrite, same area as in 'd', f) inverse pole figure map of the carbides, same area as in 'd'.*

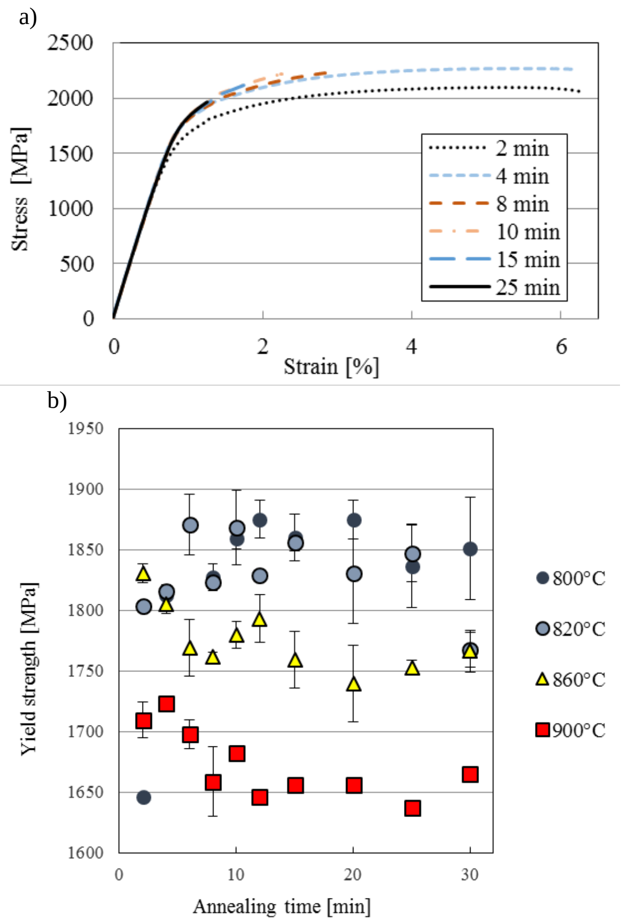

Fig. 5 *a) Tensile stress-strain curve of 75Cr1 annealed at 860 °C for various times, quenched and tempered b) Yield strength versus annealing time for different annealing temperatures and times (also quenched and tempered).*

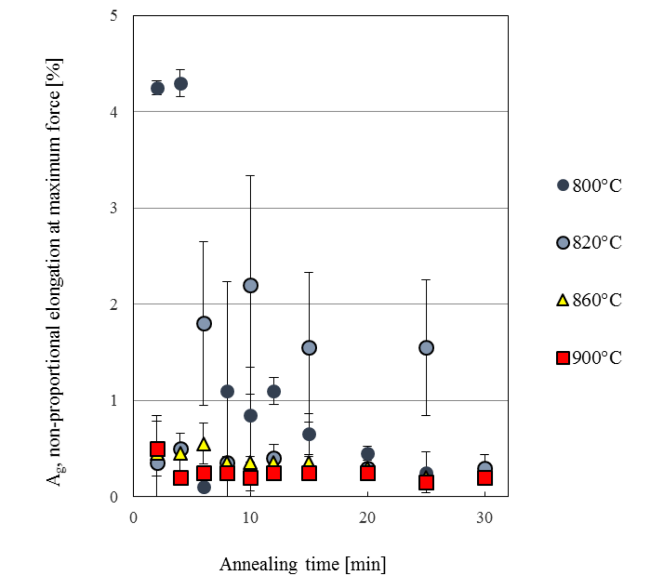

Fig. 6 Non-proportional elongation at maximum force versus annealing time deviated from tensile stress-strain curves of 75Cr1.

The effect of the annealing time on the tensile properties of the tool steel for 800°C is shown in Fig. 5a. The longer the austenitizing process, the larger is the loss of ductility. This effect is less clearly visible at higher annealing temperatures (Fig. 6). A summary of the values for the yield strength for all heat treatments is plotted in Fig. 5b. The increase of the annealing temperatures causes a decrease in the yield strength. On the other hand, if the annealing temperature is 860 °C or 900 °C, the yield strength increases at shorter annealing times. Correspondingly the elongation at fracture decreases with increasing temperature and increasing holding time, as can be seen in Fig. 6. The Izod impact test (Fig. 7) shows clearly that higher austenitizing temperatures and longer soaking times lead to a decrease in toughness.

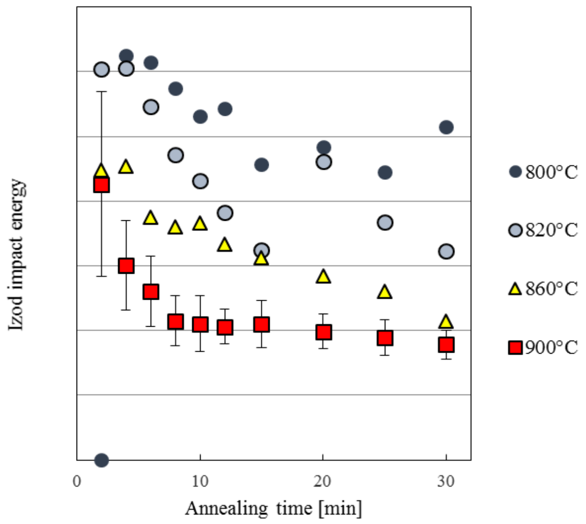

Fig. 7 Izod impact test for the different austenitizing conditions.

Fig. 8 shows that the amount of retained austenite, as measured with XRD and calculated with the formula of Cullity, increases clearly with temperature and also with time. The carbide fraction which was determined with ImageJ on Picral 4%-etched samples, reveals clearly a decrease in carbides with increasing annealing time and temperature.

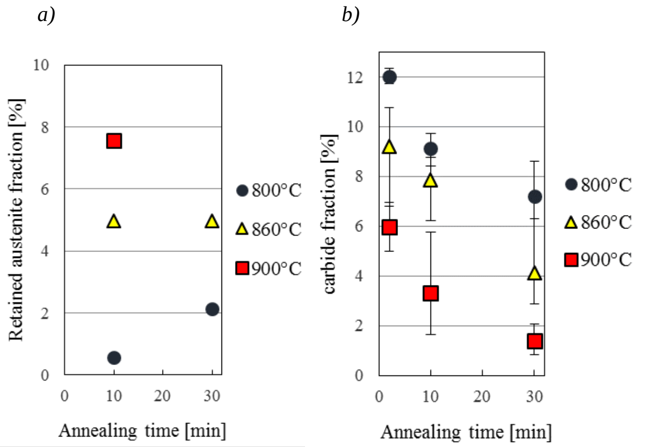

Fig. 8 *a) Retained austenite fraction after different heat treatments, determined with XRD- Cullity's formula, b) Carbide fraction using Image J software on samples etched with Picral 4%.*

4. Discussion

Higher annealing temperatures and times lead to a decrease in the impact toughness, and to a decrease in the yield strength. Analyses of selected samples show that the retained austenite fractions are higher and the carbide fractions are lower when the annealing temperatures and soaking times increase. With increasing austenitizing temperatures and times, more cementite can dissolve in the austenite and correspondingly more carbon and alloying elements can go into solution. Austenite grain growth can be expected at elevated temperatures especially when a critical amount of cementite is dissolved. Liu *et al.* [5] studied an Fe-2.03Cr-3.91C (at%) alloy and found that the undissolved cementite was enriched in Cr during austenizing. They suggest that the main part of the reaction for the studied alloy is controlled by Cr diffusion. Hillert *et al.*[6] concluded that the dissolution rate of cementite is controlled by the rate of diffusion of carbon in the very early stages and of the alloying element in the later stages.

The enrichment of the austenite matrix with carbon and alloying elements will lead to a decrease of the Ms temperature and as a consequence more untransformed austenite will retain at room temperature [4]. Retained austenite, depending on its carbon content and distribution in the microstructure, may contribute differently to the strength and ductility. A lower tensile strength can be observed after quenching and tempering in cases where the fraction of retained austenite is high. This austenite fraction can transform under strain during loading –this is the so called transformation induced plasticity (TRIP) effect. Additionally the very thin film-like morphology of the retained austenite situated between the martensite laths can contribute to the ductile fraction and correspondingly to an increase of the toughness [8,9]. On the other hand formation of the high carbon martensitic structures formed after short soaking times and low temperatures leads to an increase in strength (Fig.5b) and toughness (Fig. 7).

Hence the combination of high strength and high ductility could be obtained via quenching from low temperatures after short soaking times. High temperatures and high soaking times lead to lower strength and lower ductility.

5. Conclusions

The changes in the microstructure and mechanical properties of a tool steel 75Cr1 were studied as a function of the austenitizing temperature and time. It was found that:

- Lowering the annealing temperature and time leads to a higher strength and ductility due to less retained austenite. A lower annealing temperature also leads to less carbon saturation of the austenite and a more ductile material.

- High austenitizing temperatures and high soaking times will lead to a decrease in ductility.

Acknowledgements

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