

The influence of kinetics phase transformation during tempering on the mechanical properties of HS6-5-2 steel

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Materials

ABSTRACT

Purpose: The reason for writing this paper was to describe the kinetics of phase transformations during tempering and the evaluation of the influence of pre-tempering on the properties of HS6-5-2 steel after principal tempering. Moreover, the differences in hardness of samples of the investigated steel in relationship to the heating rate were evaluated. The microstructure development in tested samples, reflecting the extend of the phase transformations during tempering, was discussed too.

Design/methodology/approach: Dilatometric investigations were performed using a DT 1000 dilatometer of a French company Adamel. The microstructure of investigations steels were examined by transmission JEM200CX microscope. Due to high brittleness of test steel it was decided to test their strength with static bend test. The tests have been carried out on $\varnothing 5$ samples using INSTRON testing machine. Evaluation of crack resistance of test steels has been made by means of linear elastic fracture mechanics method on the basis of the measurement of stress intensity factor K_{Ic} , in static bend test.

Findings: Change of heating rate during tempering has strong influence on the temperatures of beginnings and the ends of individual transitions as well as on the accompanying dilatation effects. It has been stated that, in test steel in which remains a lot of retained austenite after hardening, the pre-tempering within the range of cementite precipitation and transformation of retained austenite, adversely affects the properties after principal tempering.

Research limitations/implications: Description of the influence of phase transformation on the mechanical properties of HS6-5-2 high speed steel.

Practical implications: It was shown, that advance (by pre-tempering) of selected phase transformations during tempering may affect steel properties after principal tempering.

Originality/value: This results should be of interest to engineers concerned with design new technologies of steel tempering.

Keywords: Tool materials; Tempering; High-speed steel; Retained austenite

1. Introduction

High speed steels owe its name to ability of fast machining and cutting of various materials among others the alloys with iron matrix. They are characterized by high content of carbon and other alloying elements, mainly carbide forming, such as W, Mo, V and Cr. Some types of high speed steels contain also Co [1-3].

One of more important attributes of these steels are their cutting properties dependent on wear resistance, impact resistance and the resistance to heat tempering. Wear resistance depends on the type, content and form of primary carbides (MC , M_6C) and matrix hardness. Whereas the impact resistance is determined by the state of tempered matrix, grain size of former austenite, spatial arrangement and size distribution of primary

carbides. Matrix of high speed steels consists of well tempered martensite and the carbides causing secondary hardness [1, 3].

Microstructure of correctly heat treated high speed steel should consist of a hard and homogeneous matrix with a high volume fraction of fine and uniformly distributed carbides both the undissolved during austenitizing and being formed during tempering.

Among numerous research conducted in recent years on high speed steels one may distinguish the research on modification of chemical composition [4-9], phase transformations during tempering [10-13] and the surface properties of tools made of high speed steels [14, 15].

Main objective of the research presented in this work was the evaluation of the influence of kinetics phase transformation during tempering on the properties of HS6-5-2.

2. Test material

The research was conducted on HS6-5-2 high speed steel with the chemical compositions is given in Table 1.

Table 1.
Chemical composition of the investigated steel

Grade	mass %						
	C	Mn	Si	Cr	Mo	W	V
HS6-5-2	0.85	0.30	0.31	4.14	4.61	6.55	1.94

Prior to testing the samples of investigated steels were soft annealed at 840°C/4hours, and successively cooled at the rate of 6°C/hour to 600°C, and after that to the room temperature together with the furnace.

3. Experimental procedure

The kinetics of phase transformations during continuous heating (tempering) from as-quenched state of investigated steels, was elaborated using a DT 1000 dilatometer of a French company Adamel. Samples with a diameter of 2 mm and a height of 12 mm, after quenching from 1240°C (austenitizing time of 150 s), were heated to 700°C with a heating rate in the range of 0.05 to 35°C/s. Digitally recorded dilatograms (engineering strain elongation $\Delta l/l_0$ in relation to the temperature T) for heated samples were differentiated, what facilitated determination of the start and end temperatures of consecutive transformations.

The microstructure of investigated steels were examined by transmission JEM200CX microscope.

The measurements of hardness were performed with the Vickers HPO250 apparatus.

Samples of test steel, previously hardened from the temperature of 1240°C (austenitizing times were 5 min for sample for testing of flexural strength, 7 min for sample for crack resistance testing, quenched in oil), have been pre-tempered at 20, 250, 450 and 550°C for 1.5 hour and subsequently cooled in air. After that principal tempering has been applied 3x580°C/1h.

Due to high brittleness of test steel it was decided to test their strength with static bend test. The tests have been carried out on $\phi 5$ samples using INSTRON testing machine.

Evaluation of crack resistance of test steels has been made by means of linear elastic fracture mechanics method on the basis of

the measurement of stress intensity factor K_{Ic} , in static bend test. The samples with dimensions of 9x18x90 mm have been used for testing. The samples with a 2 mm deep fatigue cracks on the bottom of mechanically cut notch have been three-point bent on INSTRON testing machine.

Fractographic study has been performed on fractures of samples used for K_{Ic} factor determination. The fracture surfaces have been observed using Hitachi 3500 N type SEM analyzing microscope.

4. Research results and discussion

In as-quenched the structure of HS6-5-2 steel (Fig. 1) according to [14] consists of partially twinned martensite, retained austenite (in amount of about 27.9% vol.), and carbides undissolved during austenitizing, mainly M_6C and occasionally present MC. During heating from as-quenched state up to 700°C the morphology of primary carbides is not changed [12-14].

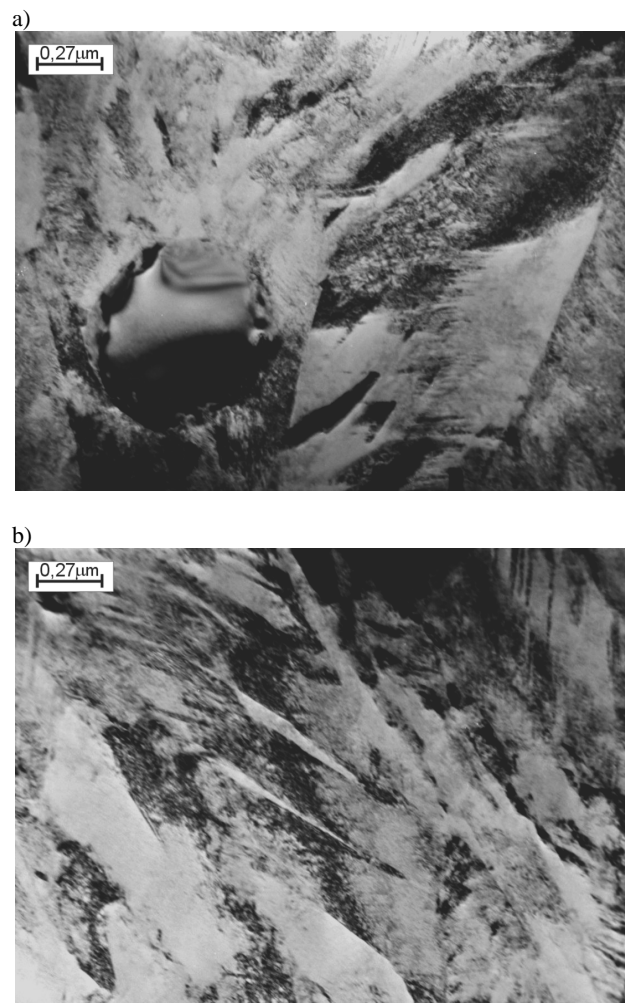


Fig. 1. Microstructures of HS6-5-2 steel samples after quenching from 1240°C. TEM

Fig. 2 presents dilatograms of HS6–5–2 steel heated from quenched state, together with corresponding differential curves. Detailed description the kinetics of phase transformations during tempering of HS6-5-2 steel in a form of CHT (Continuous Heating Transformations) diagram and the changes in structure as a result of continuous heating from quenched state are presented in studies [10, 12, 16].

As it can be noticed, during heating from quenched state in HS6-5-2 steel first there is a contraction related to the precipitation of ϵ carbide. The temperature ϵ_s for heating rate of 0.05°C/s is about 50°C and for 35°C/s is about 140°C , whereas ϵ_f increases from 210°C to 290°C . The temperature ϵ_f approximately equals to the temperature $(M_3C)_s$ at which the cementite begins to precipitate. Within the temperature range of RA_s - RA_f a part of the retained austenite is transformed. It can be noticed, in cooling dilatograms after heating (tempering) to 700°C (Fig. 3), that by far greater part of retained austenite is transformed only by cooling after tempering (heating to 700°C). The decrease of M_s temperature that was found during cooling with the increase of rate of first heating (tempering) to 700°C , from 400°C for $V_{\text{heat.}} = 0.05^\circ\text{C/s}$ to about 350°C for $V_{\text{heat.}} = 35^\circ\text{C/s}$ (i.e. about 50°C) is smaller than for HS18-0-1 steel [11], what also indicates that during heating to 700°C the chemical destabilization of retained austenite in HS6-5-2 steel takes place to a greater degree than in HS18-0-1 steel. It may be connected with higher diffusion coefficient of Mo than W and with higher diffusion coefficients of other elements (including carbon) in the presence of Mo than in the presence of W. This is indicated by so called temperability of matrix standard alloys of quenched high-speed steels with diversified of Mo and W content, described among others in work [10].

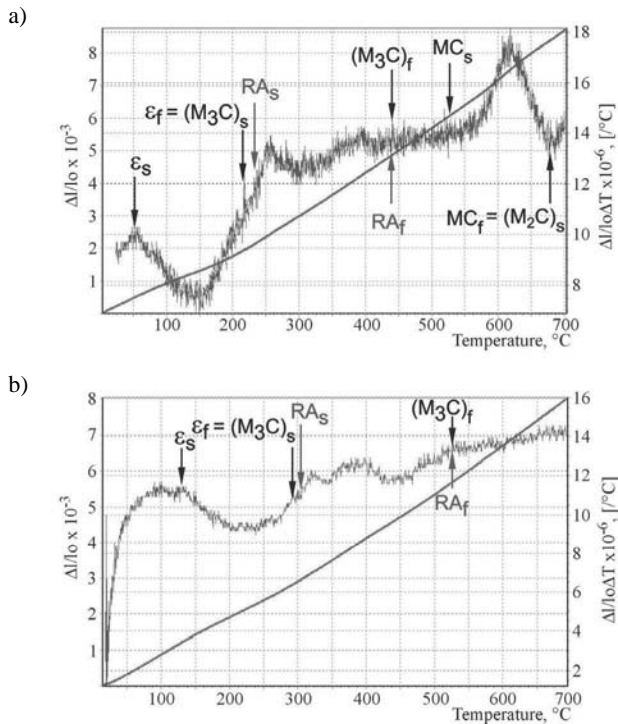


Fig. 2. Dilatograms of HS6-5-2 steel heated from quenched state at the rate of: a) 0.05°C/s , b) 35°C/s

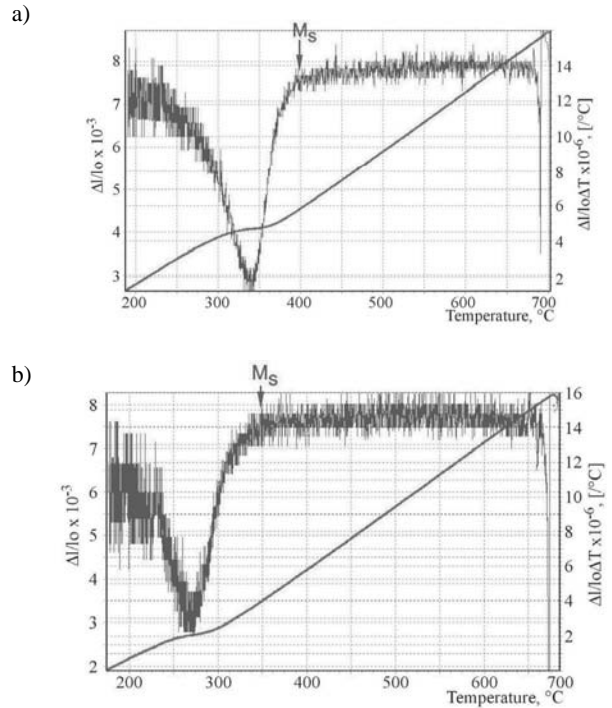


Fig. 3. Dilatograms of cooling at the rate of 1°C/s of HS6-5-2 steel, together with corresponding differential curves. First samples was heated from quenched state to 700°C at the rate of: a) 0.05°C/s , b) 35°C/s

Fig. 4 shows a CHT diagrams for investigated steel. The diagrams contains the ranges of precipitation of ϵ carbide, transformations of retained austenite, precipitation of cementite and alloy carbides of MC and M_2C type [10, 12, 16].

Fig. 5 shows the microstructures of the samples of investigated steel quenched and then heated with a heating rate of 0.05°C/s . Samples were heated up to 210 , 410 , 560 and 790°C , respectively. These are specific temperatures, at which, for a given heating rate (0.05°C/s), the following phenomena were noticed: the end of precipitation of ϵ carbide (before the beginning of transformation of retained austenite), the end of cementite precipitation, the beginning of precipitation of MC alloy carbides and the end of precipitation of carbides of M_2C type. The presented microphotographs indicate a diversified rate of advancement of transformations during tempering, depending on the temperature, that the quenched samples of the tested steel were heated up to.

During heating up to 790°C the morphology of primary carbides is not changed. Heating up to ϵ_f temperature caused the precipitation of ϵ carbide, which could be observed on the pictures taken with the transmission electron microscope. Besides, heating up to this temperature did not cause changes in the structure of high-speed steels. On the contrary, heating up to $(M_3C)_f$ caused some changes in the microstructure of investigated steels. TEM microphotographs show clear martensite strips with cementite precipitations. Heating up to 560°C initiated the dissolution of cementite and nucleation of carbides of MC type. Dispersive precipitations seen after such tempering in the TEM

microstructures are most probably these carbides. In investigated steel, after heating it up to 560°C, numerous precipitations of cementite were observed. The presence of cementite after that heat treatment involving heating the samples up to this temperature seems to be a cause of initiation of the nucleation of MC carbides.

Heating up to 790°C caused the transformation of martensite into ferrite and precipitation of carbides which can be observed on TEM. The carbides revealed on the microphotographs of samples of the tested steel after such tempering, observed in TEM, are of MC, M₂C, M₆C and M₂₃C₆ type [10-12].

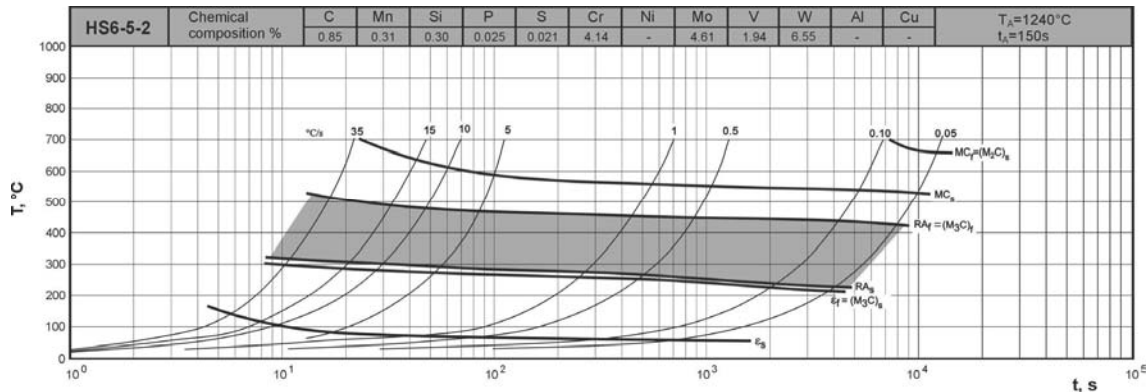


Fig. 4. CHT diagram of HS6-5-2 steel [12]

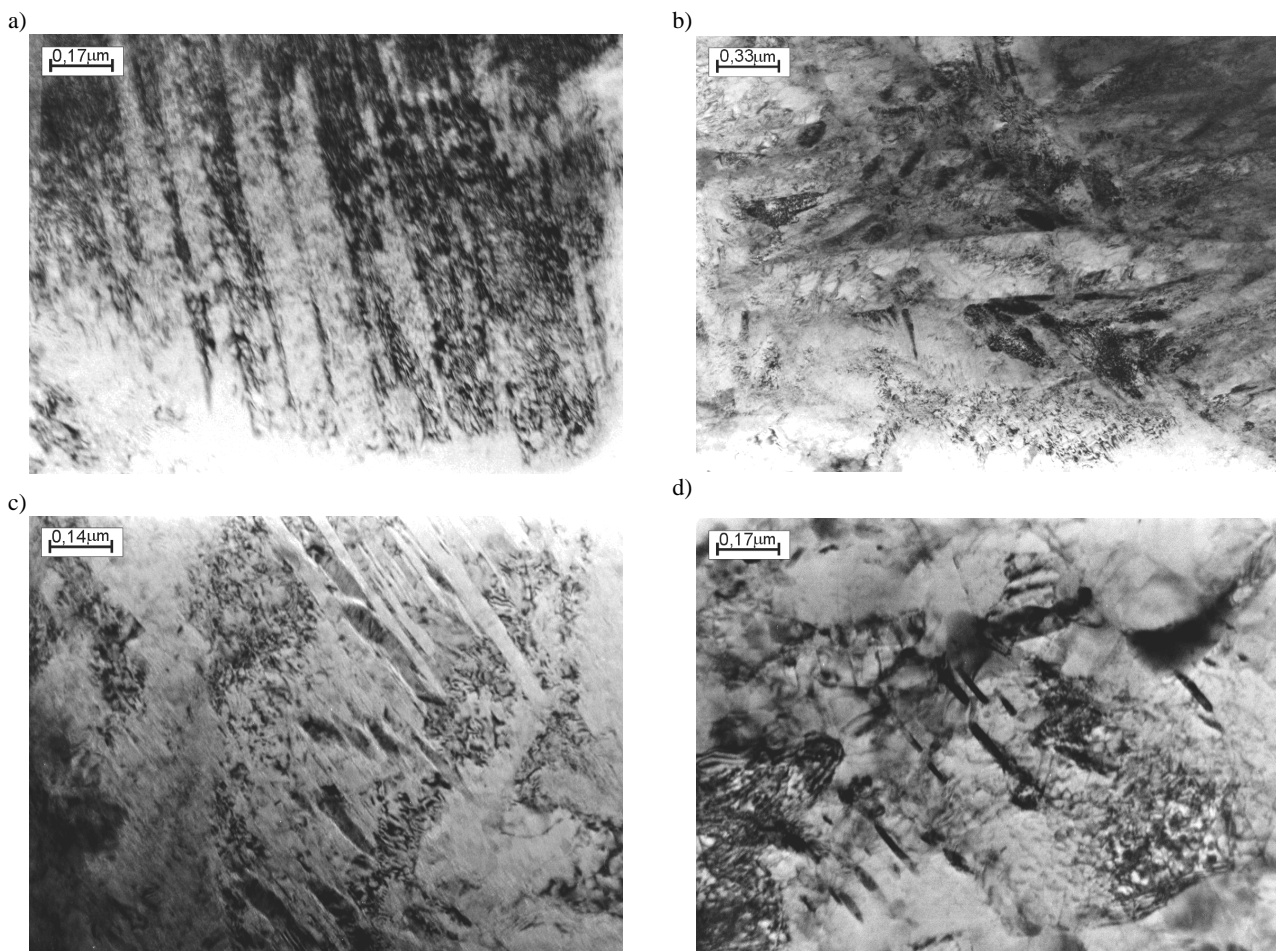


Fig. 5. Microstructures of the HS6-5-2 steel after hardening from 1240°C and heating with the rate 0.05°C/s up to: a) 210°C, b) 410°C, c) 560°C and d) 790°C, TEM

Influence of heating rate from quenched state to 700°C on hardness of investigated steel is shown on Fig. 6. As can be seen, for low heating rates (0.05°C/s and 0.1°C/s) hardness is clearly smaller, what confirms a greater contribution of phase transformations during tempering.

Application of heating rate higher than 0.5°C/s doesn't cause much changes in hardness of investigated steel, though an increase in hardness with increase heating rate can be noticed, what is believed to be a result of smaller degree of phase transformations.

Fig. 7 presents the influence of pre-tempering on flexural strength and hardness of test steel after principal tempering. The temperatures and duration (1.5h) of principal tempering have been selected on the basis of previous research [10, 12]. For test high speed steel a pre-tempering at the following temperatures has been selected: 250°C – i.e. slightly above the temperature of ϵ carbide precipitation end and start of cementite precipitation, 450°C – i.e. after the precipitation of cementite is finished and transformation of part of retained austenite, 550°C – i.e. slightly above the temperature of precipitation start of independently nucleating carbides of MC type.

During pre-tempering at 250°C mainly the ϵ carbide has been advanced. Negligible cementite precipitates on a boundary between martensite and austenite have caused no decrease of carbon content in austenite and, in result of it, during actual tempering the austenite was acting like there had been no pre-tempering, that is why the flexural strength had not changed. However the precipitation of ϵ carbides (mainly) has resulted in more uniform distribution of carbon after their solubilization inhibiting precipitation of cementite and, by it, creation of more favourable conditions for independent precipitation within whole volume of steel. Hence the hardness of samples after pre-tempering at 250°C is higher by about 25HV30. It confirms the above stated suggestion that on the way of advancing of ϵ carbide precipitation (exclusively) one may increase the hardness of steel after actual tempering.

Similarly one may explain obtained flexural strength and hardness after pre-tempering at 450°C. They are almost identical as the ones after pre-tempering at 250°C. Most probably it is a result of higher stabilization of retained austenite and distinct decrease of advance of cementite precipitation in HS6-5-2 steel if compared to HS18-0-1 steel [10, 11, 16].

Pre-tempering at 550°C has started the transformation of retained austenite and has more advanced the precipitation of cementite, also initializing the precipitation of MC carbides. Most probably it has postponed the precipitation of MC carbides during actual tempering resulting in distinctly higher hardness of the steel if compared to standard tempering. Whereas a decrease of strength is most probably connected with transformation of part of retained austenite during pre-tempering.

Fig. 8 presents the influence of pre-tempering on the fracture resistance expressed as stress intensity factor K_{Ic} and on hardness of test steel. Pre-tempering has been conducted at the temperature above the range of cementite precipitation and partial transformation of retained austenite (450°C) and at the temperature slightly higher than start temperature of independently nucleating carbides of MC type (550°C). Compared to the results of hardness measurements performed on samples after bent test the results obtained are insignificantly higher what most probably is a result of a longer austenitizing time during hardening of the samples for K_{Ic} test. Nevertheless the type of hardness changes is compatible for both tests.

During pre-tempering of HS6-5-2 steel at 450°C initially ϵ carbide has started to precipitate of which significant part had solubilized due to cementite precipitation. Besides a part of retained austenite had transformed. It caused significant increase (by 47HV30) of hardness after actual tempering by little decrease of K_{Ic} ($0,7\text{MPa}\cdot\text{m}^{-0,5}$).

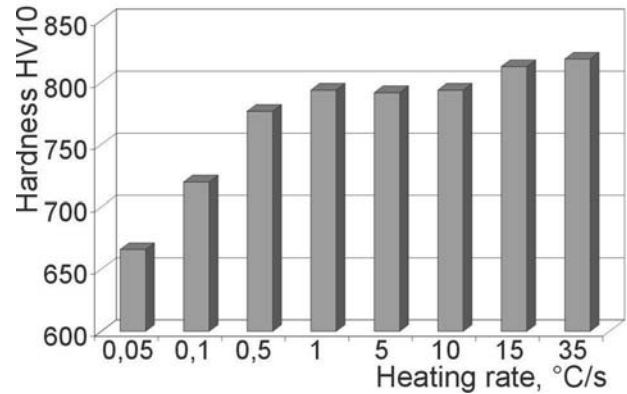


Fig. 6. Influence of heating rate from as-quenched state on hardness of HS6-5-2 steel

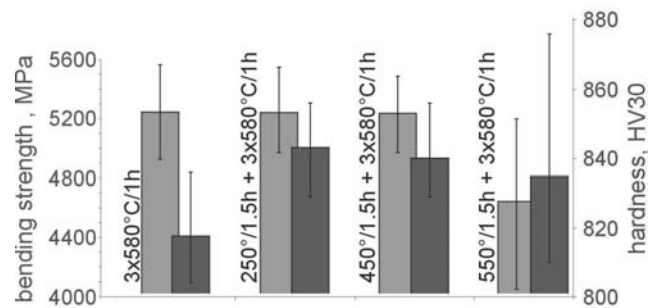


Fig. 7. Influence of pre-tempering on flexural strength and hardness of test steel

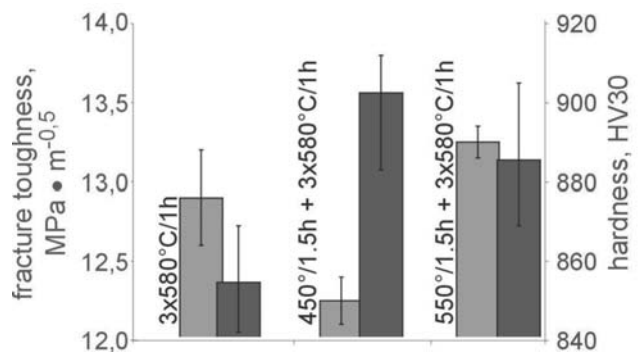


Fig. 8. Influence of pre-tempering on crack resistance expressed as stress intensity factor K_{Ic} and on hardness of test steel

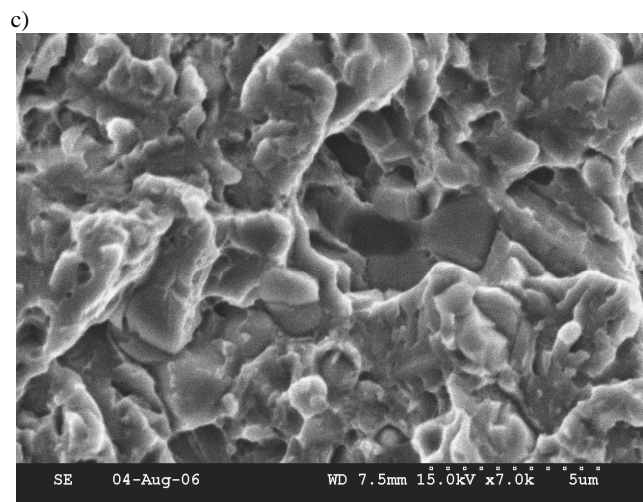
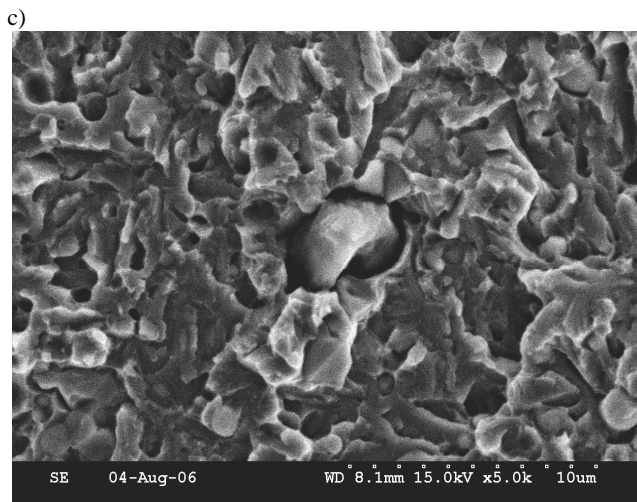
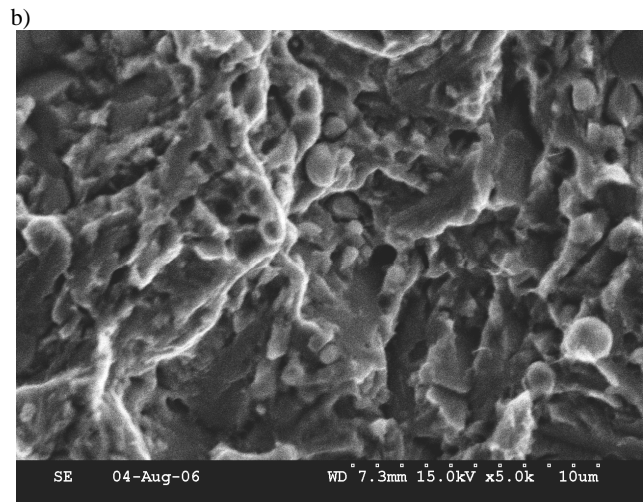
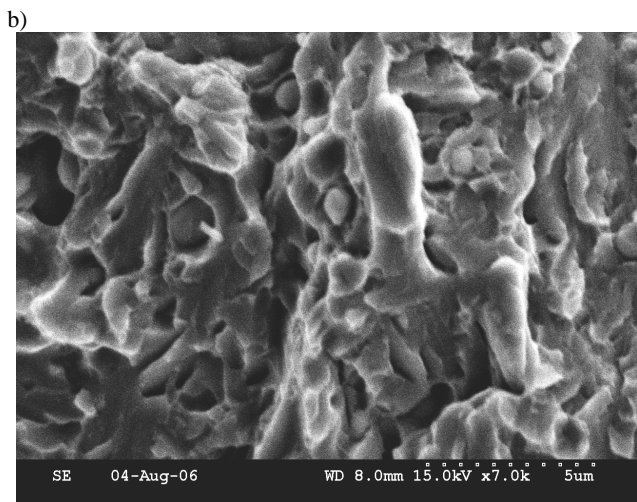
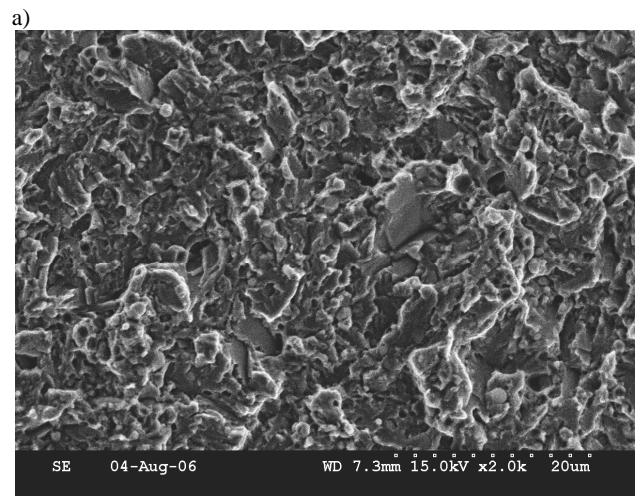
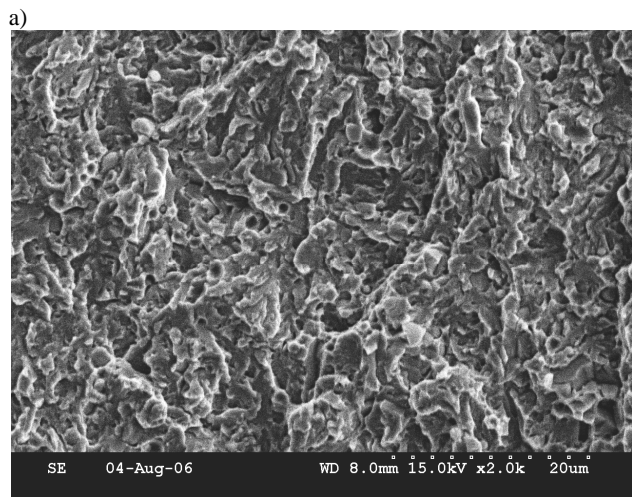


Fig. 9. Fractures of HS6-5-2steel samples after crack resistance test, tempered 3x580°C/1h SEM

Fig. 10. Fractures of HS6-5-2 steel samples after crack resistance test, tempered at 450°C/1.5 + 3x580°C/1h. SEM

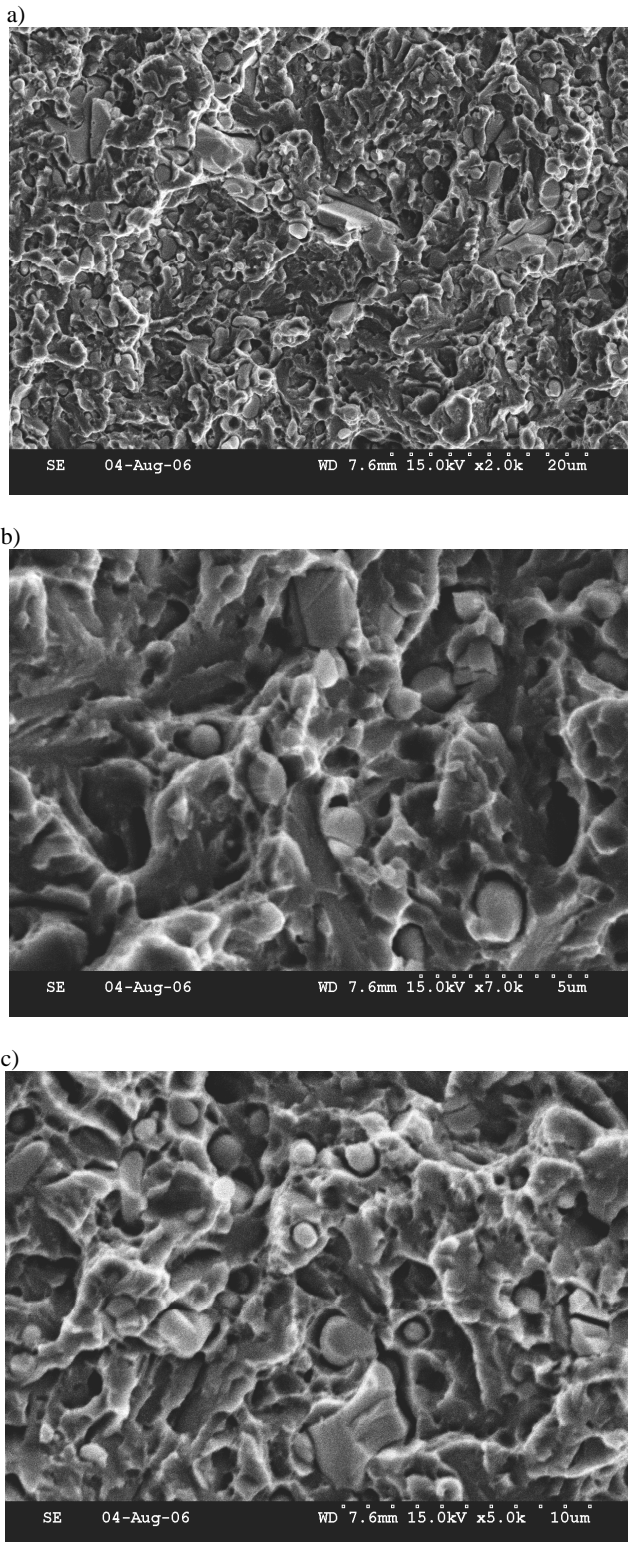


Fig. 11. Fractures of HS6-5-2 steel samples after crack resistance test, tempered at 550°C/1.5 + 3x580°C/1h SEM

Pre-tempering at 550°C has started the transformation of larger amount of retained austenite and increase of advance of cementite precipitation and thus the process of alloy carbides precipitation of MC type during actual tempering had been postponed what resulted in maintaining of high hardness and good fracture toughness.

If one considers a product of K_{Ic} and hardness, any of the variants of pre-tempering applied brings a distinct advantage. Therefore one should consider different approach, by the same flexural strength and K_{Ic} tend to increase of steel hardness. It seems that this goal is achievable by application of pre-tempering within the range of ϵ carbide precipitation (below 200°C for investigated steel) with subsequent principal tempering 3x580°C/1h applying fast heating (eg. in salts) to this temperature.

Fig. 9-11 presents the fractures of test steel after crack resistance test. One may notice that the appearance of the fractures is similar. In all variants the significant influence on flexural strength and K_{Ic} has had a content and distribution of undissolved carbides during hardening process.

5. Conclusions

Heating of the investigated steels from the as-quenched state resulted in the occurrence of four primary transformations: precipitation of ϵ carbide, precipitation of M_3C , retained austenite transformation and precipitation of alloy carbides of MC and M_2C type.

It has been found that in quenched high-speed steels a part of retained austenite has already transformed during heating for tempering, but its significant part transformed only during cooling process after tempering. Change of heating rate during tempering has strong influence on the temperatures of beginnings and the ends of individual transitions as well as on the accompanying dilatation effects.

Advance (by pre-tempering) of selected phase transformations during tempering may affect steel properties after principal tempering. It has been stated that, in test steel in which remains a lot of retained austenite after hardening (~27%), the pre-tempering within the range of cementite precipitation and transformation of retained austenite, adversely affects the properties after principal tempering. Pre-tempering of the investigated steel at the temperature insignificantly higher than start temperature of precipitation of MC type alloy carbides results in decrease of strength properties by insignificant increase of K_{Ic} .

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