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The kinetics of phase transformations during the tempering of HS18-0-1 high-speed steel

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Materials

ABSTRACT

Purpose: The reasons for write this paper was described the kinetics of phase transformations during tempering of hardened HS18-0-1 high-speed steel. Moreover, the influence of the heating rate on the retained austenite transformation.
Design/methodology/approach: CHT diagram was with dilatometric method determined. The influence of the heating rate on the retained austenite transformation as well as the results of threefold tempering at 560ºC were also determined. The microstructure development of investigated steel was observed using SEM, TEM and optical microscopy.
Findings: During heating of the samples of the quenched HS18-0-1 steel the occurrence of 4 principal transformations was determined. These are: precipitation of ε carbide, M_3C precipitation, transformation of retained austenite and precipitation of alloy carbides of MC and M_2C type. It was shown that in the quenched high-speed steels a part of retained austenite is already transformed during heating for tempering, but its significant part is transformed only during cooling after tempering as well as during consecutive heatings for temperings.
Practical implications: The obtained CHT diagram may be used to design new technologies of tempering of this steel.
Originality/value: The new CHT diagram
Keywords: Tempering; CHT - diagram; High-speed steel; Carbides

1. Introduction

The tempering of unalloyed steels of low and medium carbon content consists of three basic stages, i.e. precipitation of ε carbide, transformations of retained austenite into lower bainite (fresh martensite) and transformation of ε carbide into cementite [1,2]. In steels containing the elements which evoke the secondary hardness effect (V, W, Mo) there is the fourth stage of tempering, i.e. precipitation of MC and M_2C alloy carbides which, nucleat independently [1÷3].

In the first stage of tempering, in the range of temperatures 100÷200°C, the ε ($Fe_{2.4}C$) carbides of compact hexagonal structure are precipitated. Precipitating ε carbide of high dispersion results in decreasing the martensite tetragonality and strong strength erring of steel. Nevertheless, the fall of carbon content in martensite results in its softening and, consequently, the strength of tempered steel in this range decreases slightly [1,2,4,5].

The second stage of tempering in the range of 200÷320°C is dominated by the transformation of retained austenite. This process results in the origin of a heterogeneous mixture consisting of oversaturated ferrite and iron carbide, called lower bainite. It should be observed that this transformation occurs only in steels containing C>0.3% because the amount of austenite which remained in the steel after quenching depends precisely on the carbon content [6,7].

In the third stage of tempering in the range of 200÷420°C cementite is precipitated and the carbon content decreases in the martensitic matrix. Iron transient carbides (ε) are dissolved and it enables the steel matrix to recover whereas the precipitating cementite, according to Blicharski [8], nucleates independently on grain boundaries of the former austenite and on some particles of transient carbides (in situ nucleation). According to Pacyna and Pawłowski [9], nucleation of cementite takes place also on the boundaries of the newly formed cellular structure.

Above 400°C the diffusion of alloy elements becomes possible. Cementite, unstable in these conditions, dissolveds and the carbides of alloy elements nucleate independently or in situ i.e. cementite transforms gradually in a carbide of another type. The carbides of MC and M_2C type precipitated during tempering, which nucleate independently and which are tiny and coherent with the matrix, increase hardness and strength properties [3,10]. High-speed steels are described in literature [10,14,15,16,17,18]

The first diagrams of kinetics of phase transformations at tempering, the so – called CHT diagrams (Continuous Heating Transformations) of HS18-0-1 and HS6-5-2 high-speed steels were published by Pacyna [11]. According to the investigations [11,12,13] the CHT diagrams contribute to interfering in the degree of advancement of successive transformations during tempering (e.g. by means of the change of heating rate, temperature and time of soaking) and , respectively, to achieving advantageous properties, in particular high fracture toughness.

2. Test material

The investigations were performed on the HS18-0-1 high speed steel whose chemical composition is presented in Table 1. The material was in the worm of bars which had been soft annealed and whose crosswise dimensions were 10x20 mm.

Table 1.
Chemical composition of the investigated steel

Grade	mass %						
	C	Mn	Si	Cr	Mo	W	V
HS18-0-1	0,85	0,27	0,31	4,26	0,50	17,0	1,26

3. Experimental procedure and heat treatment

The dilatometric tests were performed with the DT1000 dilatometer manufactured by Adamel in France. The samples of ∅2x12 mm, after prior quenching from 1260°C (austenitizing time 150s), were heated with various rates up to 700°C. The digitally recorded heating dilatograms enabled drawing the CHT diagrams of the tested steel to be executed in the system temperature – time, according to the characteristic points which were read out from differential curves.

Moreover, the heating dilatograms enabled M_S temperature of retained austenite of the tested steel to by read out which were the base to construct the diagrams of dependences of M_S temperature of retained austenite on the heating rate of tempering. Multiple tempering was carried out to illustrate the stability of retained austenite in the investigated steel. It consisted in consecutive heatings with the same rate 0.05°C/s up to 560°C, holding for 1h and then cooling the sample with 1°C/s.

The measurements of hardness were performed with the Vickers HPO250 apparatus. The photographs of samples microstructure of the investigated steel were taken by the optical Axiovert 200MAT microscope, by the scanning 120 Stereoscan microscope and by the transmission JEM200CX microscope.

4. Research results and discussion

Fig. 1 shows an example heating dilatogram of the investigated steel at heating rate of 0.05°C/s with the corresponding differential curve on which the temperatures of beginnings (letter s) and ends (letter f) of respective transformations are presented. This is the method of interpretations of results which were used to make the CHT diagram of this steel. As it can be observed, this steel reveals at first the shrinkage connected with the precipitation of ε carbide. This shrinkage starts at the temperature of ε_s and ends at the temperature of ε_f. The positive dilatation effect, connected with the transformation of retained austenite, is very clear. It is visible in the range of temperatures $RA_s \div RA_f$. Cementite precipitates in the range of temperatures $(M_3C)_s \div (M_3C)_f$, independently nucleating carbides of MC type precipitate from the temperature MC_s to MC_f and the temperature $(M_2C)_s$ process of precipitation of M_2C carbides starts. The alloy carbides which precipitate in the process of tempering of the HS18-0-1 steel comprise mainly the carbides of tungsten, vanadium and molybdenum.

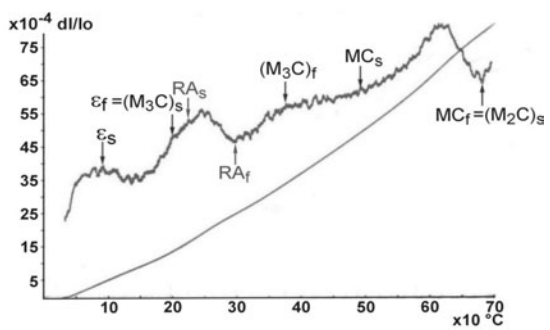


Fig. 1. Dilatograms of heating with the rate 0.05°/s of a sample previously hardened from 1260°C with the corresponding differentiation curve

Fig. 2 presents a new CHT diagram for the HS18-0-1 steel. The diagram contains the ranges of precipitation of ε carbide, transformations of retained austenite, precipitation of cementite and alloy carbides of MC and M_2C type.

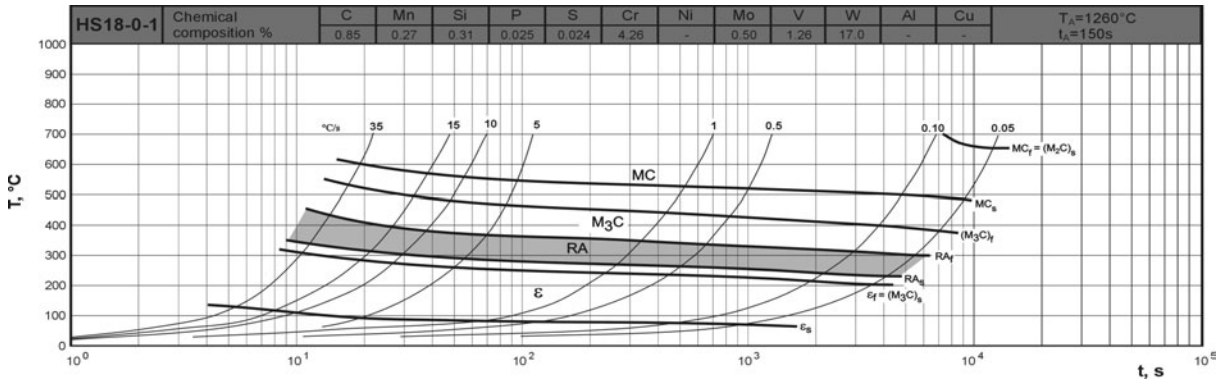


Fig. 2. CHT diagram of the investigated steel

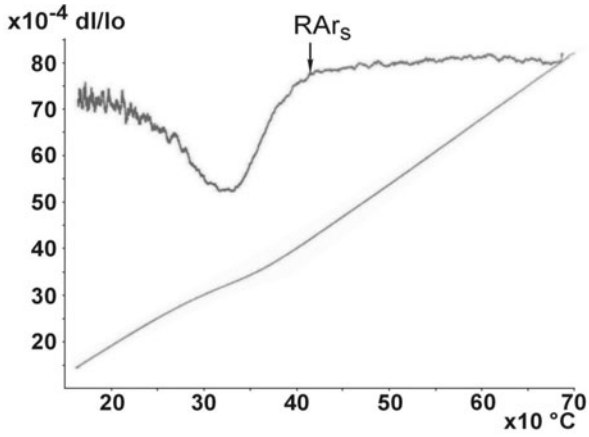


Fig. 3. Dilatogram of cooling of the HS18-0-1 steel sample with 1°C/s cooling rate, previously quenched from 1260°C and heated up to 700°C with 0.05°C/s heating rate

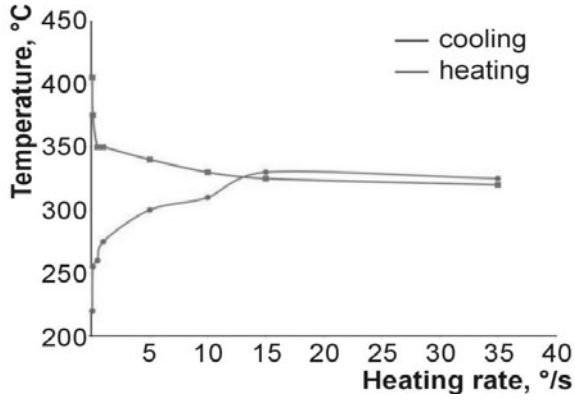


Fig. 4. The effect of heating rate up to 700°C after quenching from 1260°C on the temperature of beginning of transformation of retained austenite at heating (RAc_s) at the next cooling with 1°C/s (RAr_s) for the HS18-0-1steel

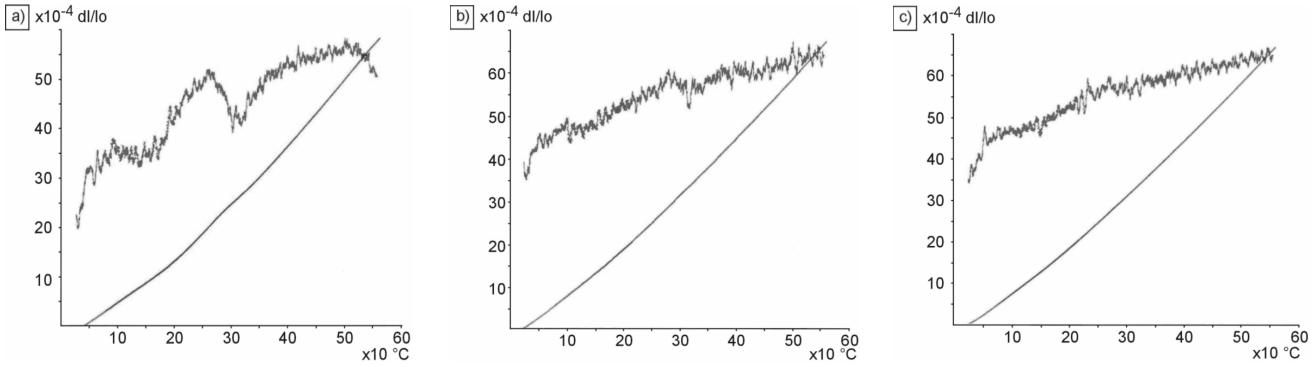


Fig. 5. Dilatograms of heating with the 0.05°C/s velocity up to 560°C for the HS18-0-1 steel: a) from the quenched state, b) second tempering, c) third tempering

Fig. 3 presents the dilatogram of cooling of the sample (with 1°/s rate) after the prior heating from the quenched state (with 0.05°C/s rate, cf. Fig.1) up to 700°C, to gether with the corresponding differential curve with the marked temperature of the beginning of retained austenite transformation during cooling RAr_s. As it can be seen, the dilatation effect from the transformation of retained austenite during cooling (Fig.3) is higher than during heating (cf. Fig.1) which proves that only after tempering a larger amount of austenite is transformed. Fig. 4 presents the diagram of dependences of RAc_s and RAr_s temperatures on the heating to 700°C for the HS18-0-1 steel. As it can be observed, the temperature of the beginning of transformation of retained austenite (RAc_s) increases with the rise of heating rate from 0.05°C/s to 35°C/s. However, at the rates over 15°C/s this parameter does not affect so strongly the RAc_s temperature as it happens for lower heating rates.

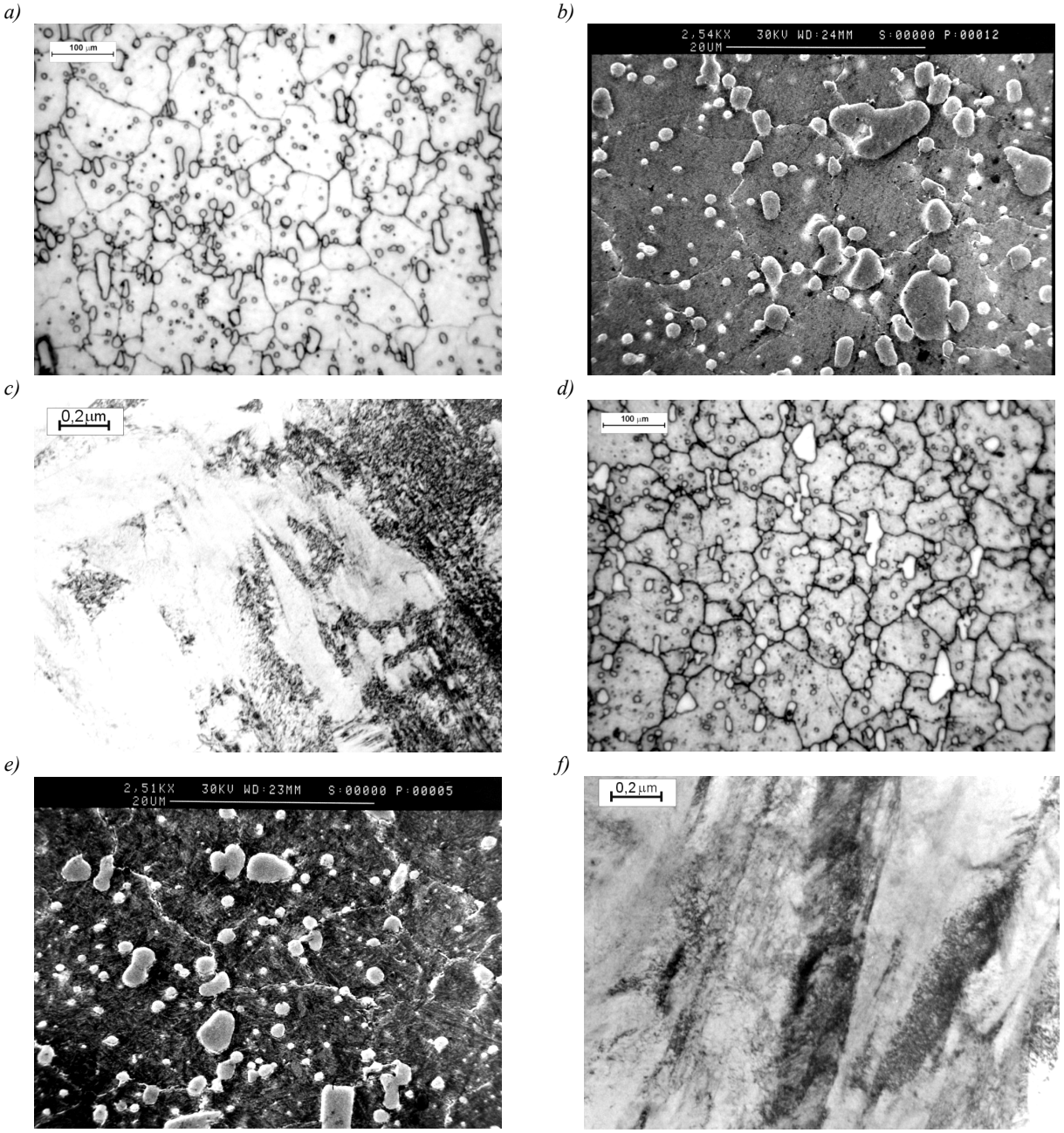


Fig. 6. Microstructures of the investigated steel after hardening from 1260°C and heating with the rate 0,05°C/s up to: a, b, c) 200°C and d, e, f) 370°C – a, d) light microscope, b, e) scanning microscope, after nital etching and c, f) TEM

Multiple tempering was performed to investigate the stability of retained austenite in the tested steel. Fig. 5 shows 3 dilatograms of consecutive temperings of the HS18-0-1 steel. As it can be noticed, the second and third tempering, according to the generally accepted procedure of heat treatment for these steels can be justified from the point of view of transformation of retained austenite because minimum dilatation effects can be observed on dilatation curves, originating most probably from the transformation of the phase.

The microstructures of tested steel which was quenched from 1260°C and following heated with the 0.05°C/s rate (cf. Fig.2) up to 200, 370, 560 and 790°C are shows on Figures 6÷9. These are the characteristic temperatures at which the following phenomena were observed on the tempering dilatograms for the 0.05°C/s heating rate: 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.

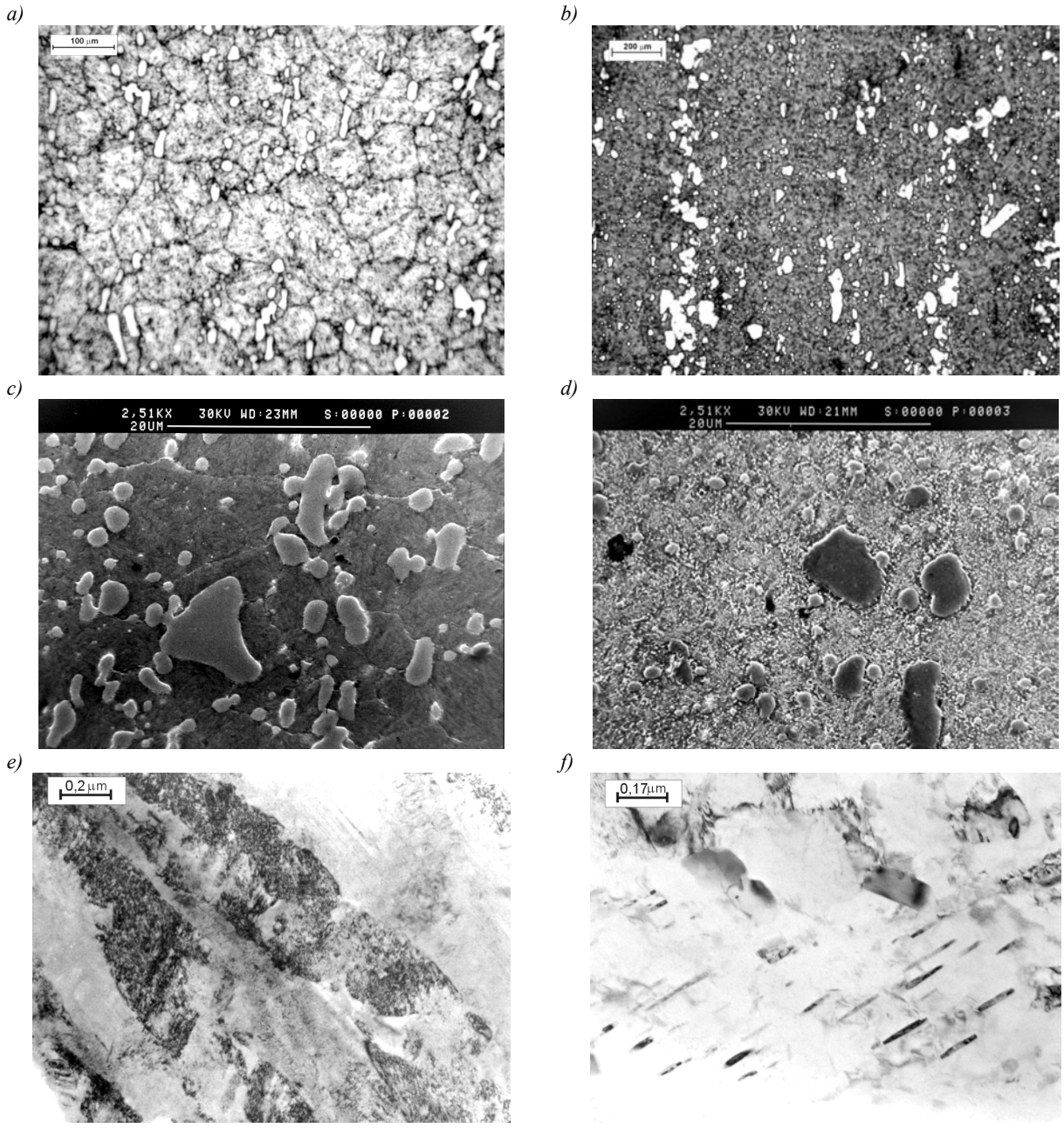


Fig. 7. Microstructures of the investigated steel after hardening from 1260°C and heating with the rate 0,05°C/s up to: a, b, c) 560°C and d, e, f) 790°C – a, d) light microscope, b, e) scanning microscope, after nital etching and c, f) TEM

The presented microphotographs indicate a differentiated rate of advancement of transformations during tempering depending on the temperature up which the quenched samples of the tested steel were heated. During heating up to 790°C the morphology of primary carbides is not changed which is visible on microphotographs made with the optical microscope and the electron scanning one. Heating to 200°C caused the precipitation ε carbide which could be observed on microphotographs made with the transmission electron microscope. Besides, heating to this temperature does not cause changes in the structure high-speed steel. Such changes were caused by heating up to 370°C. These changes are clearly seen, especially on the scanning microscope microphotographs (clear relief coming from martensite). Also the TEM microphotographs show clear martensite strips with cementite precipitations. Heating to 560°C initiated the disintegration and transformation of cementite info

alloy elements of MC type, difficult to be identified. Dispersive precipitations seen after such tempering in the TEM microstructure are most probably these carbides. Heating to 790°C caused the transformation of martensite into ferrite and precipitation of carbides which can be seen well both on TEM microphotographs and scanning microscope ones. The carbides revealed on the microphotographs of samples of the tested steel after such tempering, observed in TEM, are of M_2C type.

Fig. 8 shows the change of hardness of samples of the tested depending on the heating temperature after quenching. As it can be seen, the highest hardness was shown by the sample heated up to 560°C, i.e. to the temperature of precipitation of MC carbides while the lowest belonged to the sample heated up to 790°C, when the coherence of precipitations of MC carbides was broken and the advanced precipitation of M_2C occurred.

Fig. 9. show differences in appearance of fractures of samples the tested steel, heated from the quenched state with the rate 0.05°C/s up to 200, 370 and 560°C. The changes in the structure of samples, caused by tempering, did not cause clear changes in the fracture character which is the result of the high content of primary carbides in the structure of the tested steel and significant strengthening of the matrix.

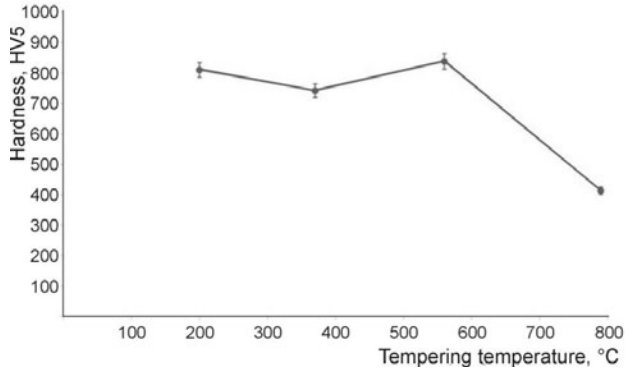


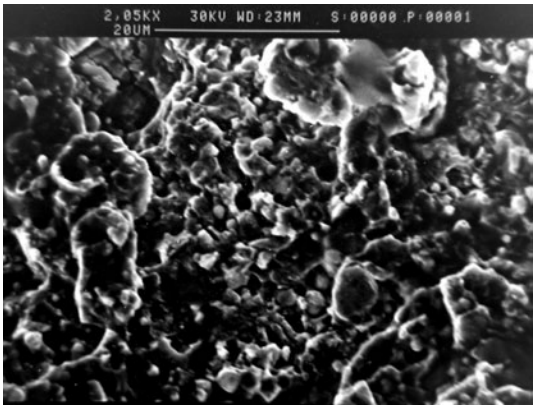
Fig. 8. Dependence of hardness of samples made from the tested steel on heating temperature after quenching

5. Conclusions

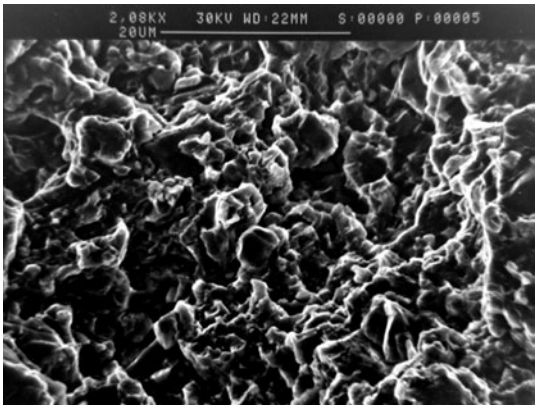
During heating of the samples of the quenched HS18-0-1 steel the occurrence of four principal transformations was determined: 1. precipitation of ε carbide, 2. M_3C precipitation, 3. transformation of retained austenite and 4. precipitation of alloy carbides of MC and M_2C type. The obtained CHT diagram may be used to design new technologies of tempering of this steel.

It was shown that in the quenched high-speed steels a part of retained austenite is already transformed during heating for tempering, but its significant part is transformed only during cooling after tempering as well as during consecutive heatings for temperings. It is worth noting that the change of heating rate during tempering has a strong effect upon the temperatures of the beginning of transformation of retained austenite RA_{Cs} and RAr_s. Because of this, the evaluation of these temperatures must be carried out at strictly determined heating rate.

a)



b)



c)

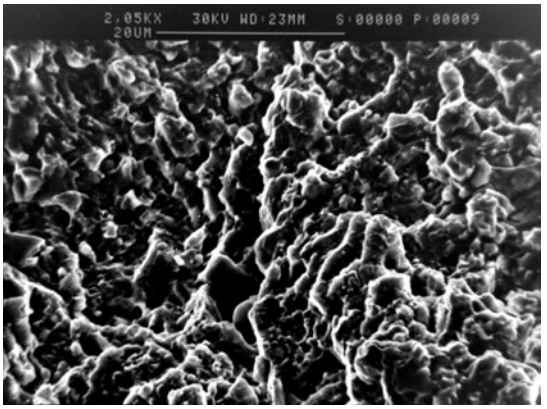


Fig. 9. Differences in the appearance of the fracture of the tested steel, previously quenched from 1260°C and tempered at the heating rate 0.05°C/s up to: a) 200°C, b) 370°C and c) 560°C

The analysis of changes of hardness with heating temperature after quenching revealed that heating up to 560°C resulted in the increase of hardness caused most probably by the precipitation of independently nucleating carbides of MC type whereas heating up to 790°C caused a strong decrease of hardness as a result of loss of coherence of MC precipitation and coagulation of M_2C carbides.

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References

[1] R.W.K. Honeycombe and H.K.D.H. Bhadeshia, Steels. Microstructure and properties. Great Britain 1981.
[2] A.K. Sinha, Physical metallurgy handbook, The McGraw-Hill Companies, Inc. 2003.
[3] W. Rong and G.L. Dunlop, The crystallography of secondary carbide precipitation in high-speed steel, Acta met. 32 (1984) 1591-1599.
[4] S. Murphy and J.A. Whiteman, The precipitation of epsilon-carbide in twinned martensite, Met. Trans. 1 (1970) 843-848.
[5] Y. Ohmori and I. Tamura, Epsilon carbide precipitation during tempering of plain carbon martensite, Met. Trans. 23A (1992) 2737-2751.
[6] J. Pacyna: The effect of retained austenite on the fracture toughness of high speed steels. Steel Research 58 (1987) 87-92.
[7] J. Pacyna, A. Kokosza, Evaluation of retained austenite stability in heat treated cold work tool steel Journal of Materials Processing Technology 162-163 (2005) 327-331.
[8] M. Blicharski, Steel, WNT 2004 (in Polish).
[9] J. Pacyna, B. Pawłowski, Effect of tempering temperature on 30HGSNA steel tougness. Mettallurgy and Casting 10 (1984) 409-421 (in Polish).
[10] L.A. Dobrzański W. Kasprzak, A. Zarychta, M. Ligarski and J. Mazurkiewicz, Structure and properties of W-Mo-V-Co 11-0-2-5 type and W-Mo-V 11-0-2 type high-speed steels, Journal of Materials Processing Technology 64 (1997) 93-99.
[11] J. Pacyna, Physical metallurgy of tool steels cracking. Metallurgy and Casting, No 120, Krakow 1988 (in Polish).
[12] J. Pacyna, A. Jędrzejewska-Strach, M. Strach, The effect of manganese and silicon on the kinetics of phase transformations during tempering – Continuous Heating Transformation (CHT) curves. Journal of Materials Processing Technology 64 (1997) 311-318.
[13] G. Zając, J. Pacyna, The kinetics of phase transformations during tempering in structural steels with nickel. Journal of Materials Processing Technology 162-163 (2005) 442-446.
[14] L.A. Dobrzański, A. Zarychta, The structure and properties of W-Mo-V high-speed steels with increased contents of Si and Nb after heat treatment. Journal of Materials Processing Technology 77 (1998) 180-193.
[15] L.A. Dobrzański, Effects of chemical composition and processing conditions on the structure and properties of high-speed steels, Journal of Materials Processing Technology 48 (1995) 727-737.
[16] J. Richter, M. Hetmańczyk, J. Cwajna, Characterization of carbide phase in nonledeburitic high-speed steel containing Ti and Nb, Journal of Materials Processing Technology 53 (1995) 341-348.
[17] L.A. Dobrzański, A. Zarychta, M. Ligarski, High-speed steels with addition of niobium or titanium, Journal of Materials Processing Technology 63 (1997) 531-541.
[18] L.A. Dobrzański, M. Ligarski, Role of Ti in the W-Mo-V high-speed steels, Journal of Materials Processing Technology 64 (1997) 101-116.