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Kinetics of phase transformations of undercooled austenite in 18CrNiMo7-6 steel applied for toothed wheels

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Abstract

Toothed wheels constitute essential part of the steel structural elements market. The most often hypereutectoid structural steels for carburizing are used for toothed wheels. The final producers of toothed wheels are not demanding regarding a microstructure, provided that the determined requirements will be fulfilled (e.g. metallurgical purity determined by ultrasounds). Therefore delivered forgings can be in an annealed or quenched state. This results from the situation that the final heat treatment or heat-chemical one is being done at one of the last stages of the toothed wheel production. An essential factor allowing to develop the proper heat treatment is the knowledge of the kinetics of phase transformations of undercooled austenite and its relating to technological conditions, being at the producer disposal, as well as to forging dimensions. Such investigations should be carried out on real melts used for forgings for toothed wheels production together with an analysis of microstructure changes on the forging cross-section. They should be based on calculation methods determining the distribution of cooling rates on its cross-section in dependence of an applied cooling medium.

The mentioned above problems in relation to 18CrNiMo7-6 steel - are elucidated in this paper. The aim of the investigations was the description of the kinetics of phase transformations of undercooled austenite in this steel. The CCT diagram was constructed for the austenitizing temperature determined on the basis of phase transformations temperatures (the so-called critical points).

Keywords: heat treatment, kinetics of phase transformations, steel for toothed wheels, cct diagram, hardenability

1. Introduction

Development of the way of preventing technological problems occurring at toothed wheels production requires an indication of a heat treatment allowing to avoid formation of banded structures [1-3]. Knowledge of the transformation kinetics of undercooled austenite and referring it to technological conditions allows to develop an adequate heat treatment of alloys on ferrous matrix [4-19]. To obtain a technological success the investigations should be carried out on real melts being utilised for forgings for a production of toothed wheels. These tests should be accompanied by an analysis of microstructure changes on a forging cross-section, and based on calculating methods, determining the cooling rate distribution on this cross-section in dependence on an applied cooling medium.

The presented above problems, in relation to 18CrNiMo7-6 steel, belonging to alloy structural steels for carburizing and classified as alloy steel for carburizing [20] – are undertaken in the hereby paper.

2. Material and investigation methods

Investigations were carried out on material taken from a forging of 18CrNiMo7-6 steel (acc. to DIN - 1.6587) of a chemical composition shown in Table 1.

Table 1.
Chemical composition (weight %) of the investigated steel

C	Si	Mn	P	S	Cr	Mo	Ni	Cu	Al
0.16	0.24	0.43	0.008	0.005	1.55	0.26	1.44	0.20	0.033

The investigation material was taken from the middle part of the forging of a circular cross-section and a diameter of 335 mm. In as-delivered condition the microstructure was characterised by intensive banding and consisted of ferritic and pearlitic zones (Fig. 1). Dilatometric samples were cut in parallel to the forging axis.

a)

b)

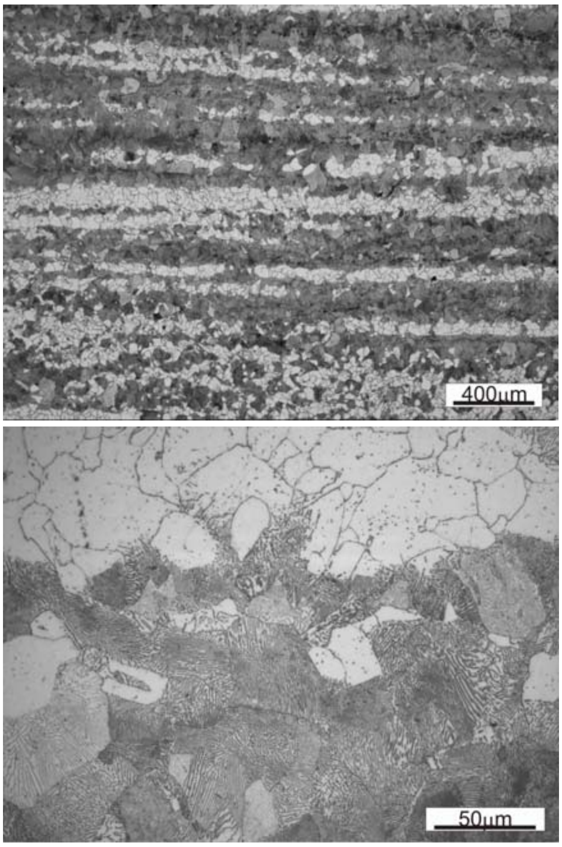

Fig. 1. Microstructure of the investigated steel in the middle part of the forging: a) banding structure, b) microstructure in banding zones. Longer side of the picture is in parallel to the forging axis. Etched by 2 % nital

The heat treatment was performed in the computer controlled laboratory oven of the Carbolite Company, type RHF 1600. Dilatometric tests were done by means of the optical dilatometer LS-4. Samples of dimensions ø4x25 mm were used.

3. The obtained results and their discussion

Temperatures of phase transformations (so-called critical points) were determined on the basis of the dilatometric curve of heating with a heating rate of 0.05 ºC/s. To begin with, the dilatometric curve for as-delivered condition was obtained (Fig. 2).

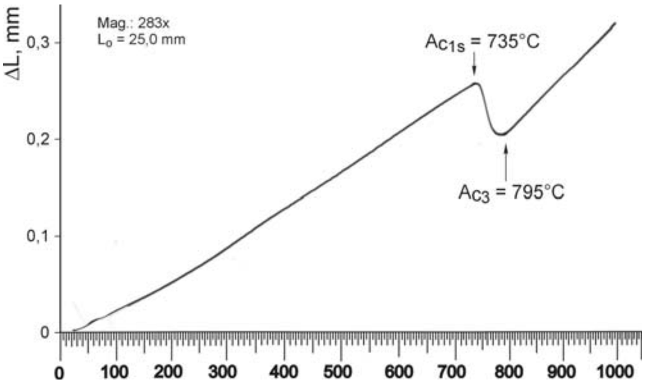

Fig. 2. Dilatometric curve for the investigated steel in as-delivered condition

The dilatometric heating curve, presented in Figure 2, allowed to determine the Ac_{1s} temperature (beginning of austenite nucleation) equal to 735 ºC, and Ac_3 (end of the ferrite into austenite transformation) equal to 795 ºC.

However, the CCT diagram of steel samples of a microstructure characterised (in its as-delivered condition) by a distinct banding does not meet the condition of the microstructure homogeneity required for the proper estimation of the transformation kinetics of undercooled austenite (which should be passably homogeneous). Therefore, on the grounds of the determined characteristic temperatures, the heat treatment aimed to remove the banded structure was designed.

Steel was exposed to the modified full annealing. The modification lied in obtaining – after finishing the annealing - the homogeneous microstructure (without banding), while the occurrence of spheroidal carbide precipitates in the ferritic matrix (characteristic for the equilibrium state) was not necessary.

Those assumptions were realised by austenitizing of the investigated steel for 4 hours at a temperature of 850 ºC (55 ºC above the determined Ac_3 temperature) followed by cooling – with a cooling rate of 3 ºC/min - to a temperature of 600 ºC (below the range of the expected diffusional transformations) and a further cooling together with the oven. The microstructure obtained as the result of such annealing is presented in Figure 3.

It can be seen that due to a high hardenability of the investigated steel, the applied heat treatment did not allow to obtain ferritic matrix with spheroidal carbides precipitations. Nevertheless the microstructure does not exhibit banding, which was the main purpose of annealing. Thus, it was assumed that the annealing task was fulfilled and enabled the proper construction of the CCT diagram.

a)

b)

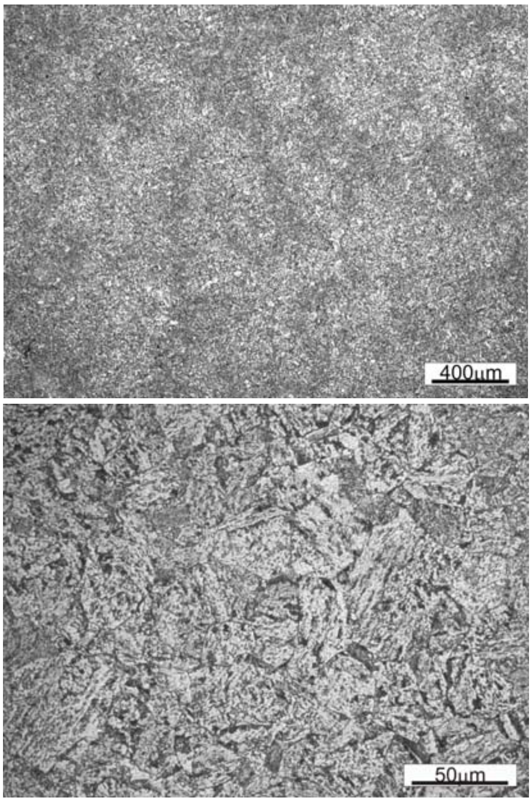

Fig. 3. Microstructure of the investigated steel after the modified full annealing: a) lack of a visible banded structure, b) bainitic microstructure. Etched by 2% nital

Since the change of the initial microstructure can significantly influence characteristic temperatures of the investigated steel, the dilatometric heating curve was performed once more but - this time - after the modified full annealing. This dilatometric curve with marked critical temperatures, is shown in Figure 4.

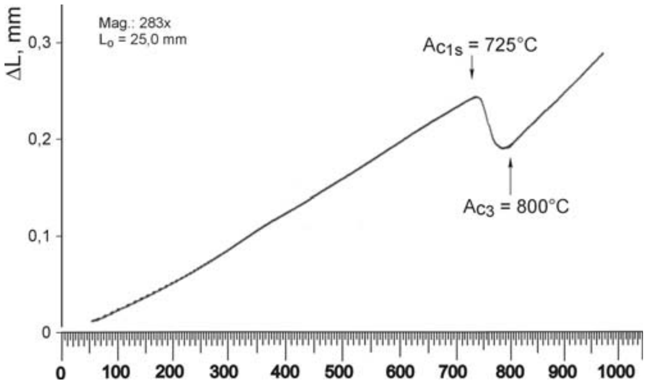

Fig. 4. Dilatometric curve of the investigated steel after the modified full annealing

Since the investigated steel is hypoeutectoid one, the following temperatures should be determined: Ac_{1s} (beginning of the pearlite into austenite transformation), Ac_{1f} (end of the pearlite into austenite transformation) and Ac_3 (end of the ferrite into austenite transformation). However, it was not possible to estimate the Ac_{1f} temperature, since there was no point on the dilatometric curve indicating a distinct change of its character. Therefore (similarly as before), only temperatures: Ac_{1s}=725 ºC and Ac_3=800 ºC were determined.

Thus, it can be stated that the applied heat treatment did not change in practice the Ac_3 temperature (correction by 5 ºC in plus) and slightly decreased the Ac_{1s} temperature (lowered by 10 ºC).

On the basis of the obtained Ac_3 temperature, the temperature of 50 ºC higher (it means: 850 ºC) was assumed as the proper one for creating the CCT diagram. This diagram is presented in Figure 5, while the microstructures of the investigated steel after cooling – in accordance with curves from the CCT diagram – are presented in Figure 6.

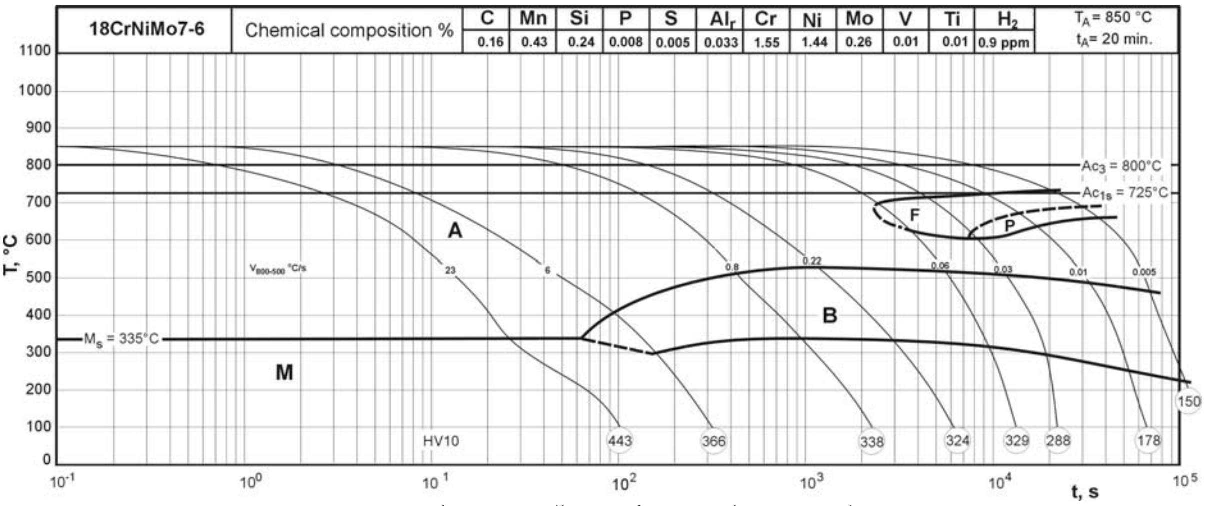

Fig. 5. CCT diagram for 18CrNiMo7-6 steel

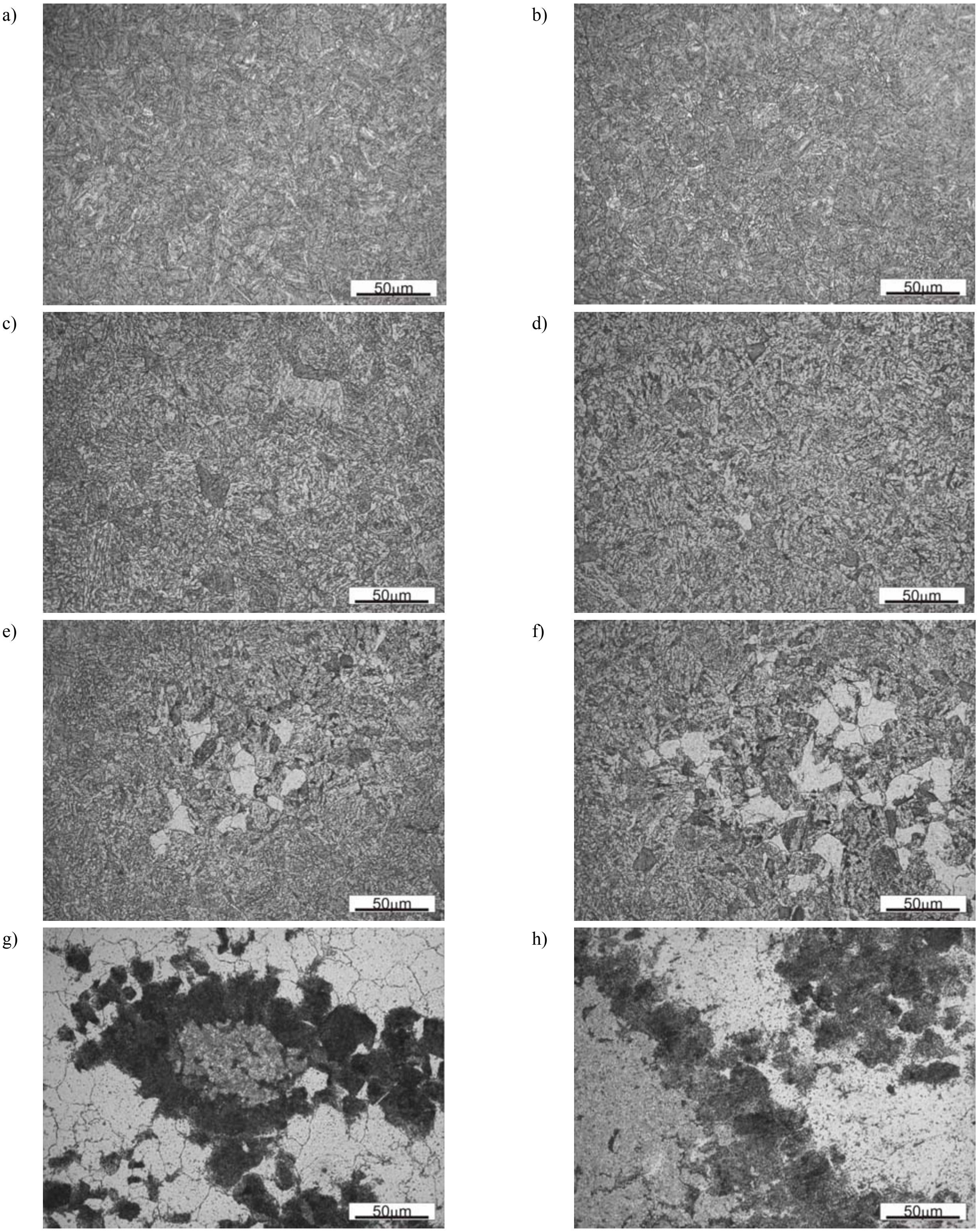

Fig. 6. Microstructure after cooling with various cooling rates $V_{(800\text{-}500\ ^{\circ}C)}$: a) 23 °C/s; b) 6 °C/s; c) 0.8 °C/s; d) 0.22 °C/s; e) 0.06 °C/s; f) 0.03 °C/s; g) 0.01 °C/s; h) 0.005 °C/s

When describing the kinetics of undercooled austenite for 18CrNiMo7-6 steel it is possible to notice that this steel has a very high temperature: M_s = 335 ºC. This temperature is only slightly less than a temperature of 350 ºC considered to be the transition temperature between a lower and upper bainite [21-23]. It should be mentioned that this delimitation is treated as follows: bainite being formed below a temperature of 350 °C is characterized by a high fracture toughness, while bainite being formed above this temperature can be characterized by a low fracture toughness [21] and is so-called structural notch. This is rather rigorous assumption, aimed at ensuring the proper qualities of the produced steel element. Nevertheless, one can find references [24], in which a delimitation temperature between dangerous upper bainite and lower bainite is even higher than 400 ºC.

However, it seems that rather the difference between M_s and the bainite start temperature and not so much as the bainite formation temperature decides whether the forming bainite will be characterized by a high or a low fracture toughness. This depends on the situation whether a continuous network of carbide precipitations will be or will not be formed in boundaries of bainitic ferrite. Admittedly the solution of this problem was not the aim of this work, but it can be assumed – in all likelihood – that the advantageous bainitic microstructure will be formed in the investigated steel, below a temperature of 400 ºC and perhaps even 450 ºC. It can be seen from the CCT diagram that the range of the bainitic structure formation reaches even temperatures higher than 500 ºC, what generates a danger, that at cooling rates lower than 6 ºC/s the investigated steel can partially have an unfavourable microstructure (of upper bainite).

The obtained CCT diagram indicates a broad range of the bainitic structure for cooling rates from 10 ºC/s to approximately 0.08 ºC/s. This elucidates why the bainitic structure was obtained in the investigated steel as the full annealing result. Decreasing the cooling rate to 0.06 ºC/s causes already a precipitation of small amounts of ferrite. Further slowing down of the cooling rate to 0.03 °C/s causes that, due to the formation of ferrite grains in austenite and their growth, carbon diffuses from ferrite into austenitic matrix. The carbon diffusion causes its concentration increase in the austenitic matrix at the ferrite grain boundary, which – in turn – causes starting of the cementite nucleation at this interface boundary. A further ferrite grain growth is related to cementite lamellas growing in these grains from the grain boundary side. The separation of the ferritic range from the pearlitic one can be only symbolically marked on the CCT diagram (broken line). Characteristic for the obtained CCT diagram, is the effect that the beginning of diffusive transformations – especially ferrite precipitation causing an austenitic matrix enrichment in carbon - decreases the bainite start temperature.

The analysis of the CCT diagram in combination with the chemical composition of the investigated steel indicates that, an increased hardenability of this steel (shifting of the diffusive transformations range - formation of ferrite and pearlite - on the CCT diagram to the right) is related to alloy additions especially manganese, molybdenum, nickel and chromium [21]. However, separation of the diffusive transformation range (formation of ferrite and pearlite) from the intermediate transformation (formation of bainite) was caused by additions of chromium and molybdenum assisted by additions of manganese and nickel [21]. A large content of chromium and nickel causes also more significant shifting to the right region of diffusive transformations as compared to bainitic transformation [21].

Thus, it can be stated that the bainite start temperature is a decisive factor concerning the application of the investigated steel. The microstructure characteristic for upper bainite starts to dominate from the cooling range of 0.8 ºC/s. This confirms the previous discussion concerning the temperature, at which ‘dangerous’ upper bainite is formed. Cementite precipitations on bainitic ferrite boundaries (even in the case of structures characteristic for upper bainite occurring in small amounts in a sample cooled with the cooling rate of 6 ºC/s) due to a low carbon content in the investigated steel, not necessarily form a continuous network resulting in a decreased fracture toughness.

As can be noticed, constructing of the CCT diagram was finished at the cooling rate of 0.005 ºC/s, at which the bainitic zone occurs in the microstructure of the investigated steel. Despite the recommendation that the CCT diagram should be constructed, if possible, up to the moment of obtaining diffusive transformations only (such structure has the forging in as-delivered condition), the diagram construction was finished at the above given cooling rate due to the technological reasons (the producer is cooling the forgings in the open air).

5. Conclusions

The performed studies allowed to:

1. determine the kinetics of undercooled austenite of the investigated steel,
2. determine dangers related to wrong microstructure (the so-called structural notches) of 18CrNiMo7-6 steel.

The obtained results enable to formulate the following conclusions:

1. Investigated steel exhibits a tendency of forming the upper bainite microstructure.
2. Low carbon content and a high M_s temperature can – due to a different morphology of upper bainite – decrease a danger of significant worsening of a fracture toughness.
3. Applying accelerated air-cooling or oil quenching it is possible to obtain a bainitic microstructure at the forging cross-section of a diameter of 335 mm.
4. The beginning of pearlite precipitation occurs as a result of cementite lamellas nucleation on migrating ferrite grain boundaries.

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