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Influence of the heat treatment on the microstructure and properties of austenitic cast steel

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Abstract

Exploitation investigations of a centrifugally cast pipe of austenitic cast steel indicated a significant influence of its microstructure on functional properties. Determination of the possibility of forming the microstructure and properties of the investigated cast steel by heat treatments was the aim of the presented paper. According to the Standard ASTM A 297, material from which the pipe was made is determined as HF type cast steel. The solution heat treatment from a temperature of 1080 °C was performed and followed by the microstructure observations and hardness measurements. It was found, that the solution heat treatment from this temperature will not significantly improve the material strength properties. However, it will visibly influence its fracture toughness. An influence of aging performed after the solution heat treatment on microstructure and hardness was also investigated. Cast steel was aged for 1 hour at 600°C (solution heat treatment from 1080 °C). On the basis of the obtained results it was found, that the solution heat treatment temperature should be the maximum permissible by the Standard i.e. 1150 °C. Heating the supersaturated material (from 1150 °C) even to a temperature of 600 °C should not cause the carbide precipitation in a form of the continuous network in grain boundaries, which would decrease fracture toughness of the investigated cast steel. Due to fracture toughness a service exposure of this material should not exceed 600 °C. The permissible service exposure up to 900 °C, given for this material in the Standard, is correct only on account of heat and high temperature creep resistance but not fracture toughness.

Key words: heat treatment, austenitic cast steel, solution heat treatment, aging, morphology of precipitations

1. Introduction

High temperature creep resisting austenitic cast steels constitute interesting materials due to their technologically friendly production process (casting) and special properties, characteristic for austenitic microstructures (e.g. fracture toughness, corrosion resistance and high temperature creep resistance). Modification of these properties can only be done by the chemical composition modification, changes of crystallisation conditions or an appropriate heat treatment. In the most recent world references there are several papers concerning the role of the chemical composition [1–8], crystallisation conditions [9,10] and heat treatments [11–13] in shaping the microstructure of austenitic cast steel. The influence of microstructure and technological parameters on exploitation properties of austenitic cast steel is also analysed [1,14,15].

The determination of the possibility of designing the microstructure and properties of austenitic cast steel as the heat treatment result – was the aim of the presented paper.

2. Material for investigations

The investigations concerned the centrifugally cast pipe of austenitic cast steel. According to ASTM A 297 Standard the material of which the pipe was made is determined as HF type cast steel. The chemical composition of this cast steel is given in Table 1. This chemical composition is contained within the limits given by the Standard. When analysing the chemical composition it can be noticed that, in the case of the properly performed centrifugal casting due to the thickness of the pipe wall, this cast steel should undergo supersaturation already at the cooling stage after casting. In the other case, in accordance to PN-EN10283 Standard this cast steel should be supersaturated from a temperature of 1080 °C (maximum from a temperature of 1150 °C).

Table 1.
Chemical composition of the investigated cast steel

C	Mn	Si	Cr	Ni	S	P
0.31	0.82	1.21	19.36	9.81	0.003	0.017

3. The obtained results and their discussion

Since the investigated pipe was made by the centrifugal casting method an intensive etching reveals its dendritic structure (Fig. 1a). However such etching makes the phase analysis difficult and can lead to the erroneous conclusion that the material of the cast steel pipe is the monophase. A delicate etching allowed to reveal precipitates of other than austenite phases. The microstructure of the polished section, made from the external pipe surface, is shown in Figure 1b. A network of precipitates in austenite grain boundaries as well as precipitates resembling eutectic in interdentritic places - are seen. An intermediate etching allows to localise those precipitates in relation to the dendritic structure (Fig. 1d). A continuity of carbide precipitates in grain boundaries (in a cutting plane of the investigated pipe: perpendicular to the axis and surface of the pipe) is presented in Figure 1c. It can be noticed, that austenite grains growing dendritically are becoming elongated perpendicularly to the pipe walls. Due to this, phases precipitating on them assume the similar orientation forming the continuous network. A more intensive etching of areas around precipitates indicates that in the vicinity of the observed precipitates there is an impoverishment in alloying additions (carbide-forming) especially chromium. The chromium content decrease facilitates etching of these areas. The mentioned above precipitates in grain boundaries were determined by the authors by means of the EDS analysis (and described in [14]). Those precipitates are the most probably M_6C carbides, however an occurrence of $M_{23}C_6$ carbides is also probable. In addition, the presence of precipitates of the chemical composition suggesting that the intermetallic phase is the most probably the σ phase – was found.

a) b) c) d)

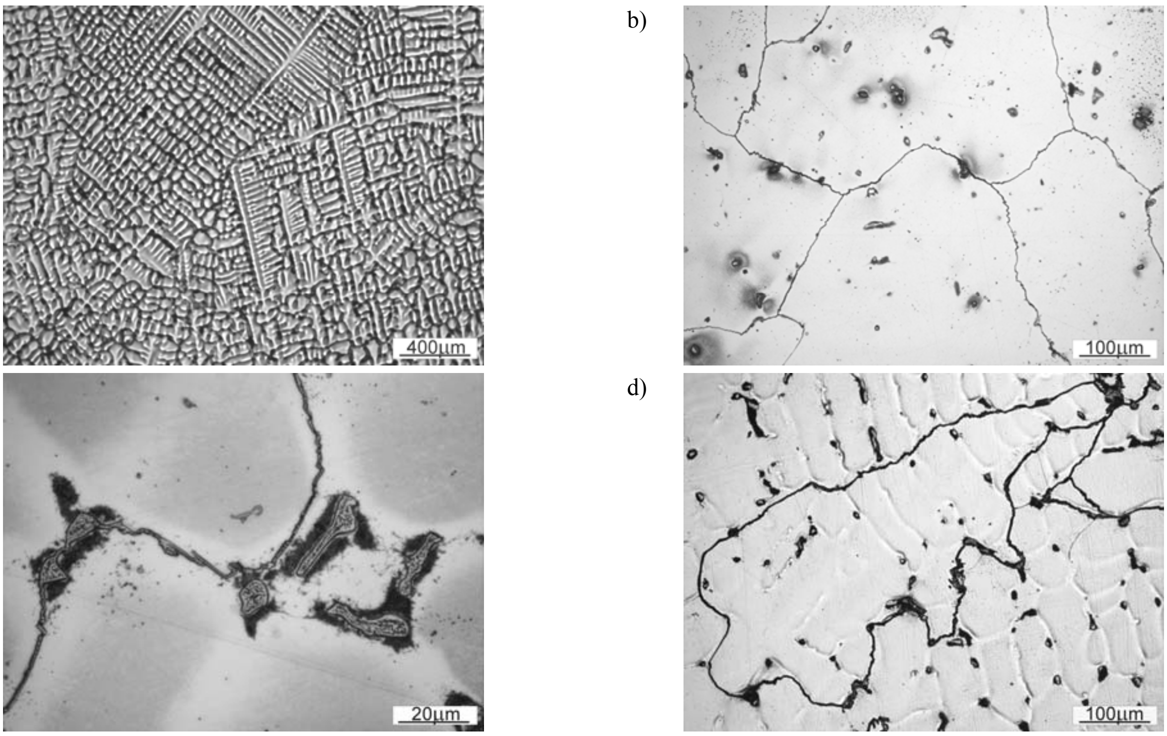

Fig. 1. Microstructures of the investigated cast steel in as-cast condition a) dendritic structure revealed by an intensive etching; b) continuous network of precipitates in grain boundaries; c) precipitates in interdendritic areas; d) localisation of precipitates in relation to dendritic structure. Vickers hardness 204

In order to determine the possibility of designing properties of the investigated cast steel by means of the heat treatment the solution heat treatment from a temperature of 1080 °C was carried out (the recommended solution heat treatment temperature for austenitic cast steels acc. to PN-EN 10283). The microstructure obtained after such operation is shown in Figure 2. Carbide precipitations in austenite grain boundaries became thinner and easily etched areas are not seen near these precipitations. Those changes should be explained by a dissolution of intermetallic phases and a partial dissolution of secondary carbides. Thus, the matrix saturation by alloy elements and carbon does not provide enough strengthening effect to smooth away the decrease of the carbide and intermetallic phases fraction in the matrix. Thus, it should not be expected that the solution heat treatment from this temperature of the investigated cast steel will improve its strength properties and fracture toughness (leaving the continuous network of carbide precipitations).

a)

b)

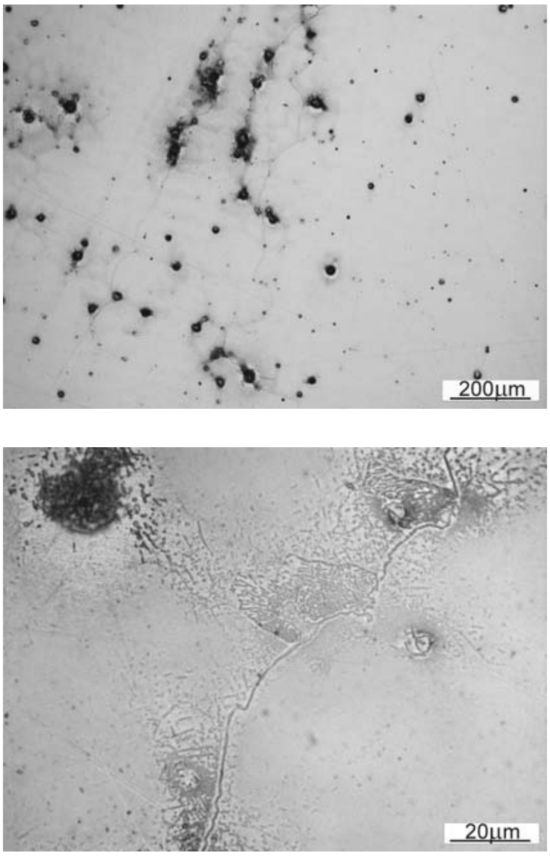

Fig. 2. Microstructure of the investigated cast steel after the solution heat treatment from 1080 °C. Vickers hardness 205. a) morphology of interdendritic precipitates, b) precipitates in grain boundaries

The influence of aging - performed after the solution heat treatment - on the microstructure and hardness of the investigated cast steel was also determined.

The microstructure after aging at 600 °C for 1 hour is shown in Figure 3 (the solution heat treatment from 1080 °C).

a)

b)

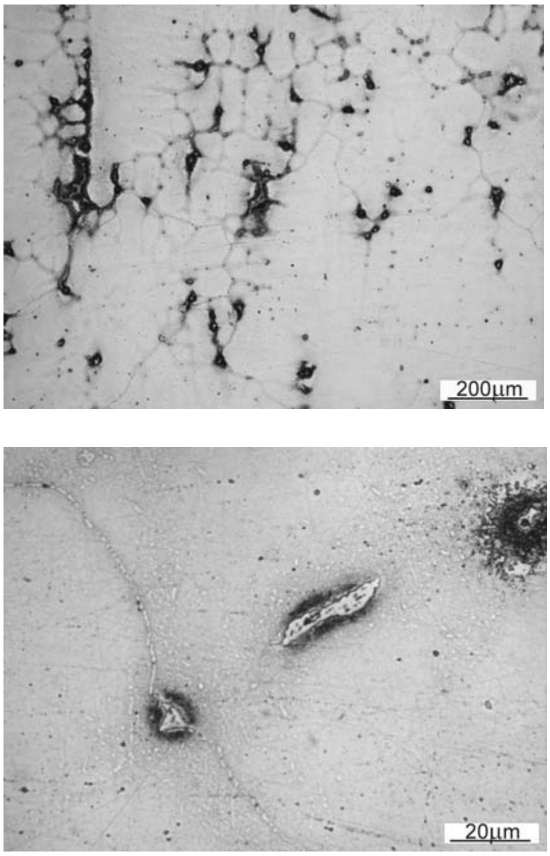

Fig. 3. Microstructure of the investigated cast steel after the solution heat treatment from 1080 °C and aging at 600 °C. Vickers hardness 176. a) morphology of interdendritic precipitates; b) precipitations in grain boundaries

Aging causes the secondary carbide precipitation in grain boundaries (the most probably M_6C or/and $M_{23}C_6$ rich in chromium) and the intermetallic phases precipitation. As the result the matrix saturation with alloying additions decreases and the precipitated carbides fraction (and their morphology) and intermetallic phases does not provide the strengthening effect. Summarising, as the result of both processes the increase of strength properties should not be expected, instead the fracture toughness decrease as compared to the supersaturated state should be feared. On the basis of the presented investigation results it can be stated, that the solution heat treatment temperature should be the maximum permissible one, it means: 1150 °C. The microstructure of the investigated cast steel after the solution heat treatment from this temperature is shown in Figure 4. It can be seen that only fragments of intermetallic and carbide phases, in a form of separated precipitations in interdendritic zones (in zones of eutectic occurrence), were left in the microstructure. Such treatment provides hardness of a similar value as the solution heat treatment from a temperature of 1080 °C. The fracture toughness increase should be expected.

It was found, on the basis of dilatometric examinations [15], that heating the material supersaturated from 1150 °C even to a temperature of 600 °C should not cause the precipitations of carbides in a form of the continuous network in grain boundaries, which would decrease fracture toughness of the investigated cast steel. The temperature operation range above 600 °C should be avoided. The permissible service exposure of this material up to 900 °C - given in the Standard - is only correct on account of its heat and creep resistance but not on account of its fracture toughness.

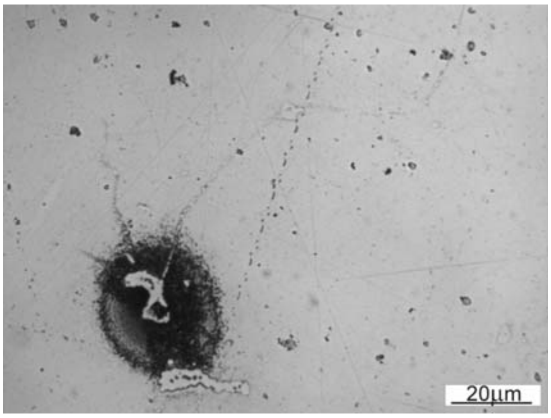

Fig. 4. Microstructure of the investigated cast steel after the solution heat treatment from 1150 °C. Vickers hardness 195

4. Conclusions

It was found that, the solution heat treatment of the investigated cast steel from a temperature of 1080 °C will not significantly improve strength properties while it should influence its fracture toughness.

The solution heat treatment temperature of the investigated cast steel should be the highest permissible by the Standards, i.e. 1150 °C. Heating the material supersaturated from 1150 °C even to a temperature of 600 °C should not cause carbide precipitations in a form of the continuous network in grain boundaries, which could worsen its fracture toughness. Due to the necessity of obtaining the adequate fracture toughness the service exposure to temperatures above 600 °C should be avoided. The permissible service exposure of this material up to 900 °C – given in the Standard - is only correct on account of its heat and high temperature creep resistance, but not on account of its fracture toughness.

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