DOI: https://doi.org/10.37434/tpwj2022.09.02

INFLUENCE OF INOCULANTS ON THE FEATURES OF WELD STRUCTURE FORMATION IN LOW-ALLOYED STEELS (REVIEW)

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The paper presents a review of studies on the influence of inoculation of dispersed refractory compounds into the weld pool on formation of weld metal microstructure in low-alloyed high-strength steels. Features of the process of primary structure formation are considered in the presence of refractory nonmetallic inclusions in the liquid metal, as well as on the interface of δ-dendrites and γ-phase. Possibilities of inoculant influence on the temperature range of bainite transformations, formation of microstructural components with higher brittle fracture resistance, and improvement of weld metal toughness values are shown.

KEYWORDS: low-alloyed steels, welding, structure, inoculants, δ-dendrites, austenite, bainite

INTRODUCTION

Low-alloyed steels of higher and high strength (HSLA steels) have become widely accepted in fabrication of welded metal structures. Owing to application of special welding technologies and selection of appropriate welding consumables, welded joints of these steels are capable of providing high values of strength, ductility and toughness. Numerous studies were published in recent years in special literature devoted to the problems of formation of metal microstructure in welded joints of HSLA steels. Particular attention is paid here to the role of bainitic components of weld microstructure. Numerous studies, considering bainite initiation and growth in BCC iron alloys are indicative of the fact that the features of the process of formation of this microstructure are still not quite clear. To date two main mechanisms of γ-> α transformation in iron alloys have been quite well studied and described, namely the diffusion and the shear mechanism. Diffusion mechanism occurs at temperatures close to A_{c3}. It is described by the process of carbon redistribution between γ-(austenite) and α-phase (ferrite) and realized as pearlite transformation. The shear mechanism takes place at temperatures close to A_{c1}, it runs almost instantly and is realized in the form of martensite transformation. The complexity of description of the mechanism of intermediate transformation (bainite) consists in that it involves both these processes. There is a quite large number of factors, which determine the priority of this or that mechanism during bainite structure formation. This is exactly why the bainite structures are characterized by a considerable quantity of morphological forms (upper, lower, globular, acicular, lath, lamellar, intragranular, polyhedral bainite, etc.). Depending on the kind and composition, each of these structures has its values of strength, ductility and toughness. Formation of a particular kind of bainite in the weld metal has a significant influence on the mechanical properties of welded joint as a whole, so that a more profound understanding of the features of the processes of bainite initiation and development requires further investigations in order to expand our knowledge base on this subject.

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The main alloying element, used in HSLA steels to increase the strength values, is carbon. Increase of the content of carbon (precipitating with formation of carbides) promotes steel hardening. Carbides, however, also lower brittle fracture resistance of the metal. To improve the steel strength, it is necessary to increase its carbon content, and, thus, more efforts are required to slow down carbide formation and their refinement. Well-known is the difference in carbide distribution between bainite, which forms at high or at low temperatures, i.e. between the intergranular and intragranular ones, respectively. In the upper region of bainite transformation, when the effectiveness of carbon diffusion in the solid solution is quite high, a considerable quantity of carbon has enough time to go beyond the grains, reach the boundaries, and precipitate in the form of carbides, while the ferrite grains proper remain free from the carbide precipitates. In the region of lower temperatures of bainite transformation, the carbon diffusion rate decreases markedly. Slower carbon diffusion, associated with lower temperature of bainite transformation, enables a certain quantity of carbon to precipitate in the oversaturated bainitic ferrite. In this case, fine dispersion of lamellar carbides in the single-crystal variant forms inside the grains, although more than one variant of carbide precipitation can be observed. Thus, both the size of primary austenite grains and the temperature range of bainite transformation have a significant role during formation of secondary microstructure.

INFLUENCE OF INOCULANTS ON PRIMARY STRUCTURE FORMATION

In general, whole technologies of low-alloyed steel welding consist in forming a fine-grained ferrite structure in the weld metal. It is believed that it should be promoted by formation of fine-grained structure of primary austenite. It should be noted that there are a lot of disputes in publications on the influence of the size of primary austenite grains on bainite transformation. Some researchers believe that the fine size of austenite grains leads to faster growth of bainite, others believe that the small grain size reduces the probability of the real transformation, and some do not notice any changes in bainite transformation at reduction of primary austenite grain size [1–3]. Differences of opinion on this issue are largely determined by the fact that initiation, growth and decomposition of primary structure, on the one hand, are poorly amenable to direct investigation because of the high temperature and speed of the processes, and on the other hand, because of a lack of experimentally confirmed physical values for description of thermodynamics and kinetics of the processes, as a perfect computer model of formation of primary structure of HSLA steels is not available. However, based on the considerations that development of metal microstructure, and of its mechanical properties, respectively, starts from formation and decomposition of the primary structure, the task of expanding our knowledge base on this question is highly relevant.

So far it is not possible to predict the size of austenite grain in the weld metal, as the factors, controlling the grain size, are by far not completely understood. The theory of grain growth envisages that nonmetallic inclusions, contained in the weld metal, should control the grain size through grain boundary blocking (Zener effect). Practice showed, however, that such an analogy is not justified, as austenite grains form as a result of δ-ferrite transformation, while Zener pinning-effect describes blocking of the grain boundaries during their growth from the liquid phase. The driving force of grain growth is usually equal to just several Joules per mole, whereas the activation energy of austenite transformation from δ-ferrite grows unlimitedly at overcooling. In this case blocking of δ/γ interfaces cannot be effective. The mechanism of blocking the boundaries of columnar austenite grains does not agree with the shape of these grains, either, as the movement of δ/γ interfaces along the direction of the maximum temperature gradient has no obvious limitations. In this case, if the pinning process were effective, the austenite grains should be isotropic as a result of their forming. North et al. in work [4] presented a description of such crystallization. However, additional investigations are needed to clarify these issues. The size of columnar austenite grain should in a certain way correlate with grain size in the base metal on the fusion line, as solidification occurs through epitaxial growth of these grains [5]. This connection, cannot be simple, however, as during solidification those grains, where the crystallographic orientation coincides with ⟨100⟩ direction, are located in parallel to the direction of the highest temperature gradient. Such grains grow quickly and inhibit the growth of grains with another crystallographic orientation. An experimental study, illustrating the influence of crystallographic texture on grain size [6], shows that nonmetallic inclusions located in the base metal (for instance, carbonitrides), can limit coarsening of weld metal grains on the fusion line and, thus, eventually result in smaller grain size in the fusion zone.

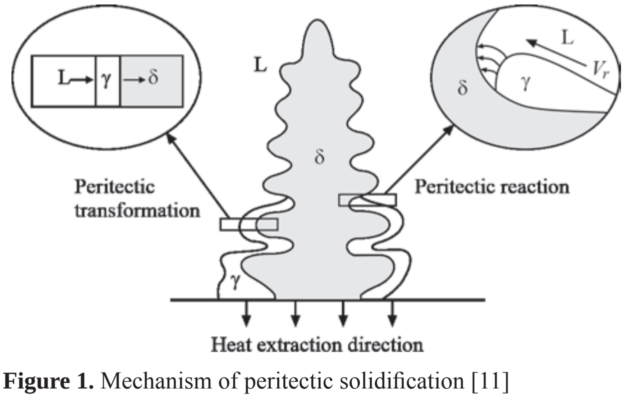

Figure 1. Mechanism of peritectic solidification [11]

When considering the processes of primary structure formation, it is necessary to take into account the fact that they proceed at contact of the three phases and are described by peritectic reactions, respectively. Peritectic crystallization of metal at its cooling occurs in two stages. At the first stage a peritectic reaction proceeds in the point of contact of three phases (L-liquid + δ-ferrite + γ-austenite) in the temperature range a little lower than the peritectic temperature, leading to separation of L-liquid and δ-ferrite with lateral growth of γ-austenite around δ/L interface. At the second stage peritectic transformation begins with thickening of the layer of γ-austenite at the expense of δ-ferrite phase and γ-austenite tip advancing into the liquid L-phase (Figure 1). Development of high-temperature laser scanning together with the confocal microscopy [7] allows observation with a high resolution of phase transformations in the high-temperature region in peritectic steel. Results of these investigations [8, 9] showed that the peritectic transient process is controlled by diffusion of the solutes. It was found that partial remelting of δ-phase also is influenced by diffusion of solutes. It is shown that γ-phase initiates and grows on the interface between δ- and L-phase, while γ-phase quickly separates them in the process of growth. A conclusion was made that the interface enrichment in carbon is increased with increase of the cooling rate that decelerates initiation of

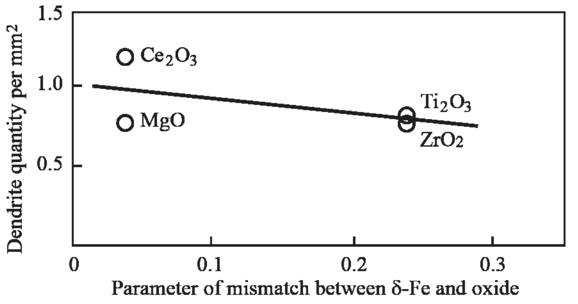

Figure 2. Dependence between the dendrite quantity and parameter of mismatch between δ-Fe and oxide [17]

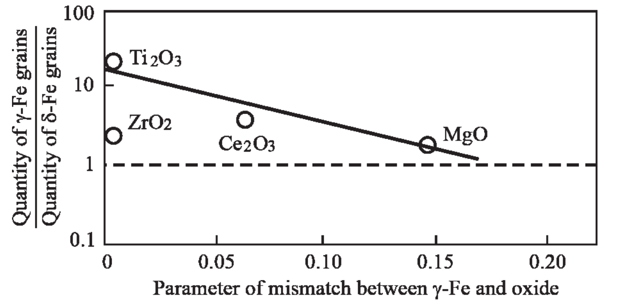

Figure 3. Dependence between mismatch parameter and density of γ-phase nucleation centers on the interface with δ-dendrite [17]

γ-phase. Moreover, it is shown [10] that the interface movement speed is influenced by elastic energy of the interface and coefficient of solute distribution.

Refractory inclusions with melting temperature above that of the metal melt, present in the thin layer on the surface of δ-dendrites, where γ-phase initiates and develops, depending on the wettability index, can be absorbed by the growing phase or can accumulate on the interfacial front and influence the interfacial energy. Experimental results given in publications confirm this conclusion. So, work [12] shows the results of investigation of the influence of inoculation of such refractory oxides as MgO, ZrO_2, Ti_2O_3, Ce_2O_3 into the steel melt. Determination of wettability index between the refractory oxides and liquid iron and δ-Fe showed that the contact angle of wettability changes depending on time and temperature of contact. This is indicative of the possibility of interfacial reactions. From the viewpoint of thermodynamics, reactions with oxygen evolution can be in place at temperatures characteristic for steel pool melts [13, 14, 17]:

$$Al_2O_3 \rightarrow 2Al + 3O, \Delta G^0 = 1225000 - 393.8T \text{ (J/mole)},$$
$$MgO \rightarrow Mg + O, \Delta G^0 = 89960 + 82.0T \text{ (J/mole)},$$
$$2TiO_2 = Ti_2O_3 + O, \Delta G^0 = 379908 - 97.069T \text{ (J/mole)},$$
$$3TiO_2 = Ti_3O_5 + O, \Delta G^0 = 387866 - 112.215T \text{ (J/mole)}.$$

It results in accumulation of decomposition products on δ→γ interface. As a result of the conducted studies, Bhadeshia et al. [15] came to the conclusion that increase of oxygen content in the steel melt does not affect the size of primary austenite grains, while authors of [16] express the idea that formation of γ-phase is influenced by accumulation of alloying elements on the interface, for instance magnesium, as a result of MgO decomposition. The density of distribution of potential centers of new phase initiation depends on the energy on the initial phase boundary. The activation energy is primarily affected by increase of grain boundary energy, as a result of increase of the content of alloying elements on them. Increase of grain boundary energy will lead to increase of the speed of formation of new phase nuclei.

Experimental data given in publications confirm this conclusion. So, work [15] gives the results of studying the influence of inoculation of such refractory oxides as MgO, ZrO_2, Ti_2O_2, Ce_2O_3 into the steel melt. It was found that formation of solidification microstructure is influenced by chemical composition of the inclusions, as well as parameter of mismatch between γ-Fe and oxide and δ-Fe and oxide (Figure 2).

Moreover, it is found that with increase of the parameter of mismatch between δ-Fe and oxide the quantity of γ-Fe grains which formed in the body of one dendrite, becomes greater (Figure 3).

INFLUENCE OF INOCULANTS ON SECONDARY MICROSTRUCTURE FORMATION

Change of primary structure morphology by inoculation of dispersed refractory compounds before welding influences formation of weld metal secondary microstructure. Work [18] gives the results of investigations on adding refractory oxides, carbides and nitrides to the weld pool. It is shown that depending on physico-chemical properties of the compounds, the inoculants influence the size of primary structure grains. Increase of primary austenite size should reduce the effectiveness of carbon diffusion that is confirmed by the change of temperature range of bainite transformation (Figure 4).

Lowering of bainite transformation temperature is accompanied by inhibition of carbon diffusion that causes carbide precipitation in the ferrite grain body, and by development of lower bainite structure in the weld metal (Figure 5) by deceleration of the processes

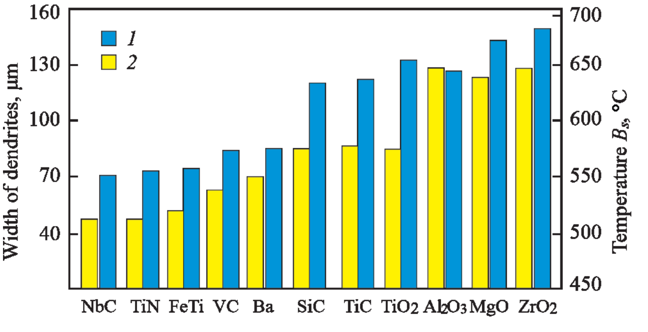

Figure 4. Influence of inoculation of refractory compounds on the change of dendrite size (*1*) and temperature of the start of bainite transformation B_s (*2*)

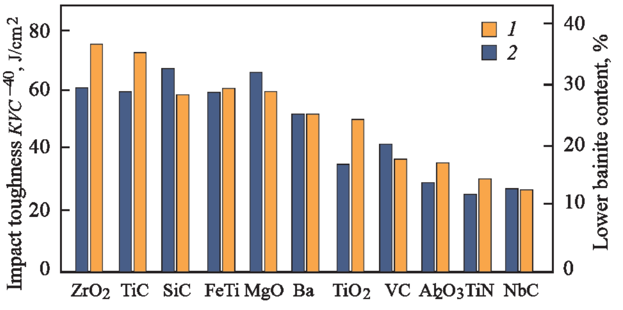

Figure 5. Influence of inoculation by refractory compounds on lower bainite content in the microstructure (*1*) and weld metal impact toughness (*2*)

of upper bainite and Widmanstatten ferrite formation. Change of microstructural composition results in improvement of weld metal toughness values (Figure 5).

Experimental results given in work [18] confirm the possibility of inoculant influence on weld metal structure. It is found that addition of dispersed particles of refractory compounds with the respective physico-chemical properties to the weld pool allows changing the primary structure grain size, and promotes shifting of bainite transformations to lower temperature region. Increase of lower bainite content in the weld microstructure, owing to upper bainite and Widmanstatten ferrite, as a result of development of such processes, allows improvement of their mechanical properties.

CONCLUSION

Formation of weld metal microstructure takes place during a continuous process, which starts from initiation and development of primary structure and ends by formation of a secondary microstructure. Primary structure grain size depends on the energy of interfaces between δ- and γ-phases, and it is determined by the effectiveness of carbon diffusion during $\gamma \rightarrow \alpha$ transformation. Inoculation of refractory compounds to the weld pool liquid metal enables influencing the processes of primary structure formation, temperature range of bainite transformation and formation of a secondary microstructure with higher content of lower bainite in the metal of HSLA steel welds.

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SUGGESTED CITATION

V.V. Holovko (2022) Influence of inoculants on the features of weld structure formation in low-alloyed steels (Review). *The Paton Welding J.*, **9**, 11–14.

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https://pwj.com.ua/en

Received: 28.06.2022

Accepted: 11.11.2022