FLASH-BUTT WELDING OF HIGH-TEMPERATURE NICKEL ALLOY USING NANO-STRUCTURED FOILS

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The process of resistance flash-butt welding of high-temperature nickel alloy Rene 80 using nano-structured foils of Ti–Al and Ag–Cu system is considered. Features of welded joint formation and their microstructure are studied. Microhardness distribution in welded joints is shown.

Keywords: *flash-butt welding, high-temperature nickel alloy, lack-of-penetration, microcrack, nano-structured foil, base metal, heat-affected zone, microhardness distribution, microstructure*

Progress of science and technology is making ever higher requirements to high-temperature strength of materials in combination with ductility, thermal and low-cycle fatigue life, resistance to gas environment, and endurance. This stimulates activities on development and introduction of new alloys, in particular, high-temperature casting nickel-based alloys.

High-temperature casting nickel-based alloys are effectively used in industry as material for parts of gas turbine engines [1, 2]. Taking into account one of the methods to improve performance (alloying optimization) an experimental high-temperature casting nickel-based alloy is proposed, which has the following composition, wt.%: 0.17 C; 18 Cr; 8.5 Co; 1.8 Mo; 2.6 W; 0.9 Nb; 3.4 Ti; 3.5 Al; 1.75 Ta; 0.5 ≤ Fe; Ni being the balance. A feature of this alloy is an increased content of aluminium, titanium (3.5 and 3.4 wt.%, respectively) and other alloying elements.

Metallographic investigations showed that the alloy structure is typical for cast metal (Figure 1). It is based on γ-phase dendrites, which are complex-alloyed nickel-based solid solution. Two kinds of phases are located in interdendritic space. These are, probably, γ'-phase based on $(Ni, Cr)_3(Ti, Al)$ compound and MeC type carbides, capable of creating such elements as titanium, tantalum and niobium. Location of these phases indicates that their precipitation occurred at solidification of interdendritic melt from γ'-phase and carbides. Isolated inclusions of these phases were also found in the dendrite volume.

Dimensions of γ'-phase and carbides are larger that those at dispersion precipitation from the solid solution. This, as well as the predominant location along the grain boundaries, changes their role in the alloy strengthening. Level of high-temperature strength determines the retardation of grain-boundary slipping.

Introduction of new alloys is complicated by the problem of producing their permanent joints with each other and other materials. Complex alloying by reactive elements and thermal instability at high temperatures cause certain difficulties in welding high-temperature alloys.

Resistance flash-butt welding (RFBW) provides local high-speed heat input into the joint zone [3]. Considering the experience of previous developments on RFBW of such difficult-to-weld materials [4, 5], it is proposed to perform welding of high-temperature nickel alloy using nano-structured foils.

Nano-structured foils of Ti–Al and Ag–Cu systems were used in this work. Production of such nano-structured foils based on vapor-phase technology has been mastered at PWI [6].

Ti–Al system foil is a multilayer composition of alternating layers of titanium and aluminium, corresponding to γ'-Ti–Al stoichiometric composition. Heating of such foil up to the temperature of about 300 °C, results in titanium and aluminium interaction with formation of intemetallic. Interaction reaction is

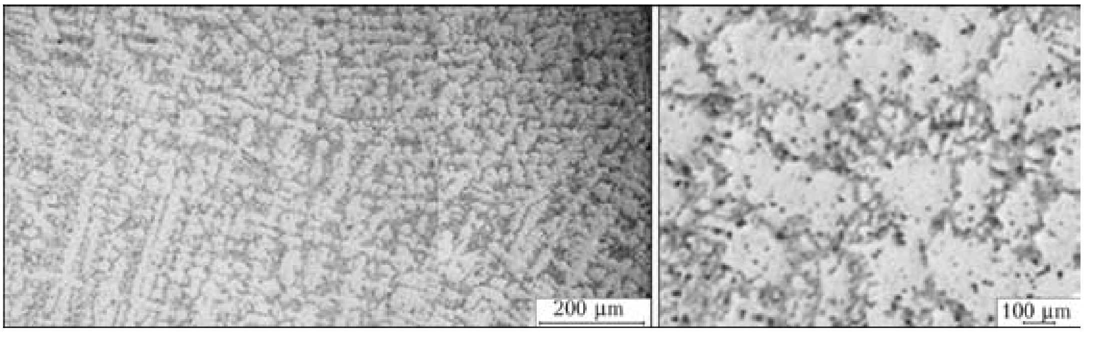

Figure 1. Microstructures of cast high-temperature nickel alloy Rene 80

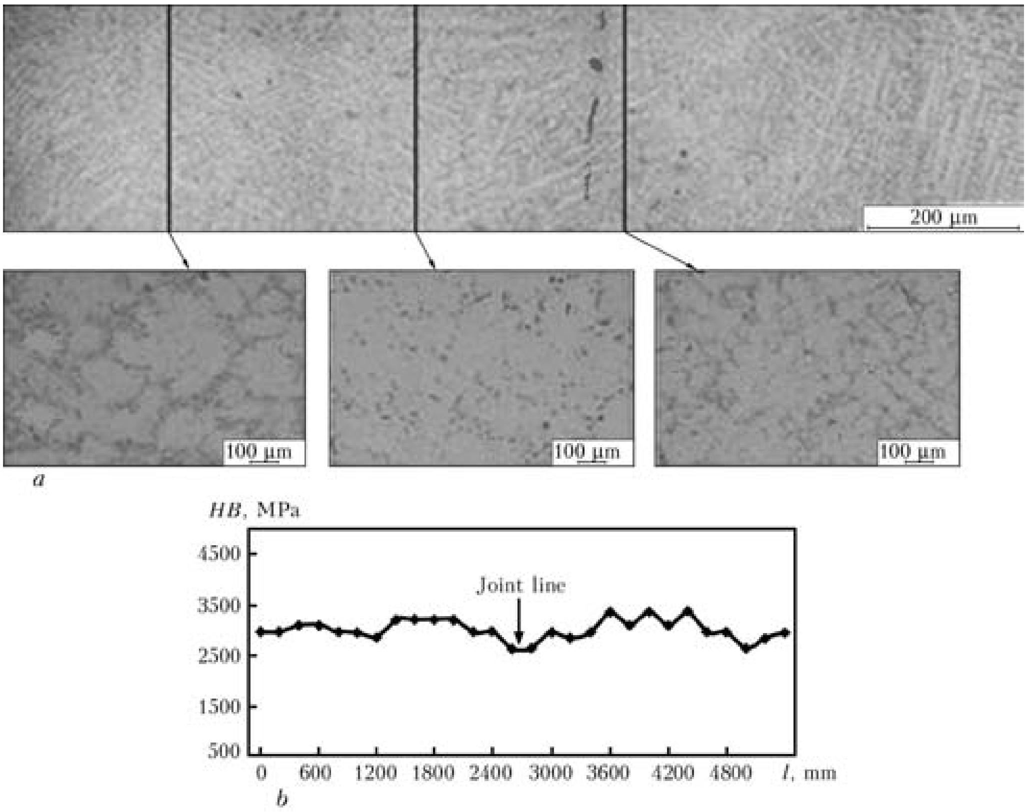

Figure 2. Microstructure (*a*) and microhardness distribution in the joint of a high-temperature alloy produced using nano-structured foils (*b*)

developing at a high rate and is accompanied by evolution of addition heat.

Foil of Ag–Cu system in its composition is close to eutectic condition of Ag–Cu system alloys. Eutectic melting temperature is equal to 779 °C [7]. Melt formation at the contact boundary at the temperature lower than the alloy melting temperature, is favourable for joint formation.

RFBW was performed in an upgraded «Schlatter» machine. Maximum machine power was 25 kW·A and upsetting force was 800 MPa. Welding of samples of 10 × 10 × 100 mm size was performed in air at the pressure of 2 MPa, welding current of 4–5 kA and welding time of 3–5 s.

Analysis of microstructure and chemical inhomogeneity of the joints was conducted in optical microscope «Neophot 32» and scanning electron microscope ISM-840 with Link system microanalyzer.

To assess mechanical properties of welded joint microhardness distribution was determined on microprobe computerized «Micron-gamma» system.

At RFBW of the alloy in a similar joint without using foils such defects as lacks-of-penetration and microcracks were recorded in the welded joint zone (Figure 2). A probable cause for unsatisfactory formation of the joints is non-uniform heating of samples.

The following was noted in the HAZ of the produced joint (Figure 2). A region of equiaxed grains of γ'-phase is adjacent to base metal with arborescent dendrites. The quantity of carbides and γ'-phase is preserved here. In the region adjacent to the weld, the quantity of carbides and γ'-phase decreases considerably. Isolated precipitates of these phases are found along the grain boundaries. Furtheron the dendritic structure and quantity of carbides and γ'-phase are restored.

Microhardness distribution was analyzed to assess the change of strength characteristics in the HAZ.

As is seen, microhardness lowering in the base metal from 3000 to 2500 MPa is recorded in the region with refined grains of the solid solution that forms along the joint line (see Figure 2). Microhardness increase above 3000 MPa occurs in the section of partial dissolution and, probably, further disperse precipitation of strengthening γ'-phase in the solid solution (see Figure 2).

At the next stage, welded joints of the alloy will be produced using nano-structured foils of Ti–Al, Ag–Cu system.

In the joint HAZ decomposition of oversaturated solid solution in interdendritic volumes of the metal and formation of strengthening γ'-phase were found (Figures 3 and 4). In the middle part the thermomechanical impact leads to structure refinement.

In the joint produced using foils of Ti–Al system arborescent dendrites are preserved and no foil fragments were found in the weld (see Figure 3), and in the joints produced using foils of Ag–Cu system equiaxed grains were observed (see Figure 4). In this case, microstructure is the result of solid-liquid interaction of the eutectic melt of Ag–Cu system with the alloy base metal.

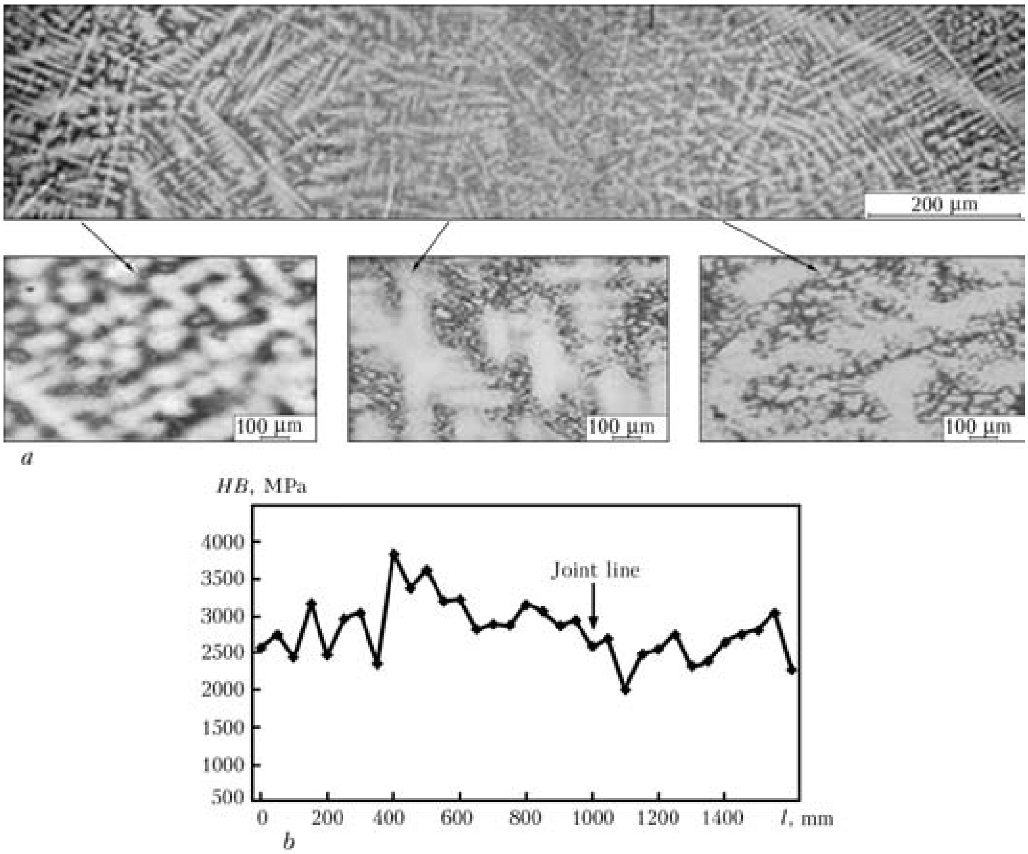

Figure 3. Mictrostructure (*a*) and microhardness distribution in the joint of high-temperature alloy produced using nano-structured foil of Ti–Al system (*b*)

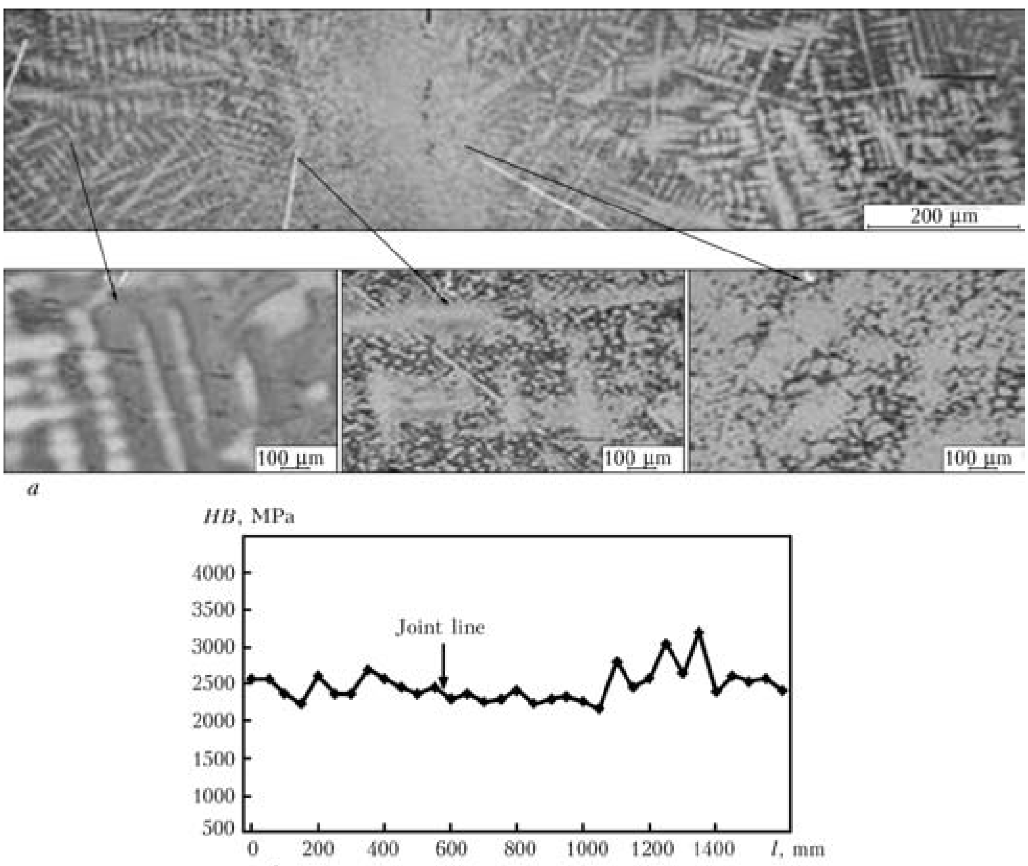

Figure 4. Microstructure (*a*) and microhardness distribution in the joint of high-temperature alloy produced using nano-structured foil of Ag–Cu system (*b*)

Comparative analysis of microhardness distribution showed that in the middle part of the joint produced using foil of Ti–Al system microhardness rises relative to base metal and is equal to about 3000 MPa. On the other hand, in the middle part of the joint made with application of foil of Ag–Cu system microhardness is lower than in the base metal and is equal to 2250 MPa.

Therefore, application of nano-structured foil at RFBW of high-temperature nickel alloy will allow ensuring a uniform highly concentrated heating of the joint zone and lowering the process temperature, thus preventing base metal softening.

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DEVELOPMENT OF FLUX-CORED WIRE FOR ARC WELDING OF HIGH-STRENGTH STEEL OF BAINITE CLASS

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Given are the results of investigation of the effect of alloying on formation of structure and mechanical properties of the weld metal in gas-shielded flux-cored wire welding, as well as of development of composition of a core of the wire providing the yield strength value of not less than 590 MPa and impact energy of more than 50 J at –50 °C. Optimal additional microalloying with zirconium was determined for the basic C–Si–Mn–Ni–Mo alloying system. This microalloying allows decreasing the volume fraction and size of non-metallic inclusions, as well as increasing the share of dispersed components in metal structure, and provides the required level of strength of the weld metal and its low-temperature tough-ductile properties.

Keywords: *arc welding, low-alloy steel, flux-cored wires, properties of weld metal, structure, non-metallic inclusions, microalloying*

Development of new welding consumables, meeting the high requirements made to mechanical property indices, in particular strength and impact toughness, was necessitated by expansion of production and application of low-alloy steels in building and industry.

The aim of the present paper is development of composition of a wire core providing obtaining of a weld metal with yield strength value of not less than 590 MPa and required values of low-temperature impact energy (more than 50 J at –50 °C) [1]. It is a complex task to achieve such a level of indices using traditional alloying systems.

Experience of development of low-alloy consumables, in particular, flux-cored wires, indicates an appropriateness of application of the alloying systems, close on composition to alloying system of steel to be welded, taking into account different conditions for formation of a metal structure at rolling and welding. As a rule, C–Si–Mn–Ni–Mo(Cr–Cu) system makes an alloying basis. Alloying of the metal in C–Si–Mn–Ni–Mo system due to solid-solution hardening [2, 3] provides necessary indices of strength. Regulation of the tough-ductile properties requires selection of alloying and microalloying system, providing formation of the dispersed structural components which have high resistance to brittle fracture [4, 5].

The investigations were carried out on the pilot batches of flux-cored wire of 1.2 mm diameter with slag-forming system of rutile-fluorite type during downhand welding of AB-1 steel grade plates (400 ×

Table 1. Content of alloying elements and additions in the weld metal, wt.%

Weld number	C	Si	Mn	Ni	Mo	Ti	Al	Zr	S	P
1	0.07–0.09	0.2–0.4	1.0–1.4	2.0–2.4	0.15–0.25	0.01–0.015	0.025–0.035	–	0.011–0.016	0.016–0.020
2								0.007–0.009	0.012–0.016	0.016–0.020
3								0.010–0.015	0.011–0.016	0.016–0.020