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INFLUENCE OF HEAT TREATMENT ON IMPROVEMENT OF MECHANICAL PROPERTIES OF WELDED JOINTS OF SPARSELY-DOPED TITANIUM ALLOY Ti–2.8Al–5.1Mo–4.9Fe

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ABSTRACT

The most important advantage of pseudo-β-titanium alloys is their high strength, and the disadvantages include the high cost of alloying elements. Sparsely-doped alloys, such as LCB, Timetal 125, etc. were developed to lower the cost of titanium alloys based on β-phase. This class of titanium alloys is promising for application in inexpensive structures. Development of welding technology and modes of heat treatment of such alloy joints is an important task. In this work investigations were performed of the surface of fractures in welded joints of titanium alloy of Ti–2.8Al–5.1Mo–4.9Fe system, obtained after impact toughness testing. It was found that local heat treatment in the vacuum chamber of specimens of welded joints of test titanium alloy Ti–2.8Al–5.1Mo–4.9Fe produced by EBW by mode 4 (LHT in a vacuum chamber at the temperature of 750 °C for 5 min) allows producing higher ductility properties of welded joints, and preventing post-weld cold cracking. Such heat treatment leads to a more uniform arrangement of tough fracture areas on fracture surfaces.

KEYWORDS: titanium, pseudo-β-titanium alloys, sparsely-doped titanium alloys, welded joints, welding, electron beam welding, local heat treatment, tungsten electrode, heat treatment, mechanical properties, fracture surfaces, quality criterion

INTRODUCTION

Owing to its unique properties, sparsely-doped Ti–2.8Al–5.1Mo–4.9Fe titanium alloy is promising for application in different industries. Sparsely-doped titanium alloys are those which do not contain any expensive or deficit elements, such as niobium, vanadium, etc., and their alloying systems are based on such relatively inexpensive components as aluminium, iron, silicon, oxygen, etc. [1–4]. PWI performed work on producing ingots of Ti–2.8Al–5.1Mo–4.9Fe pseudo-β-alloy, using UE-208M electron beam unit with a cold hearth and portioned feed of liquid metal. Note that the produced ingots of Ti–2.8Al–5.1Mo–4.9Fe pseudo-β-alloy are close by their chemical composition to an alloy developed by TIMET Company, USA, which was named LCB (low cost beta) [5, 6].

High-strength sparsely-doped titanium alloys have been widely accepted in the aerospace sector (critical and highly-loaded components and assemblies); defense industry (armour protection elements of combat vehicles and personnel); transport engineering (engine components, wheel disks, springs, load-carrying structures of sports cars); in manufacture of smart goods for sports, medicine and consumer goods [7].

Improvement of mechanical properties of welded joints of sparsely-doped titanium alloys requires application of different methods of heat treatment or thermomechanical treatment (HT, TMT) [8].

In connection with the fact that application of full heat treatment of the whole welded product is not economically viable, local heat treatment (LHT) is as a rule used in the region of the titanium alloy joint. The main LHT objective consists in creation of the conditions to prevent cracking and other defects in the welded joint region, and in improvement of the mechanical properties in welding through application of local preheating up to certain temperatures and subsequent annealing, leading to relaxation of thermal stresses in the welding zone, as a result of the change of the metal structural and phase composition and producing the specified properties [9].

Electron beam and argon-arc welding are traditionally used for joining high-strength titanium alloys.

ARGON-ARC WELDING (AAW)

Tungsten electrode inert gas argon-arc welding (AAW) has been the most widely accepted for joining titanium alloys due to the fact that this welding process is the most inexpensive and versatile. It allows making joints in different positions in space, under the conditions of limited space and does not require any complicated readjustment of the equipment at the change of the welded product thickness or joint type. Welding can be performed both with application of filler metal and without it. Titanium alloy wires or rods are used as filler metal. Inert gas argon practically does not enter into chemical interaction with molten metal and other gases in the arc burning zone. Argon is heavier than air, so that it drives air out of the welding zone and reliably protects the liquid weld pool and near-weld zone of the welded joint from contact with the atmosphere.

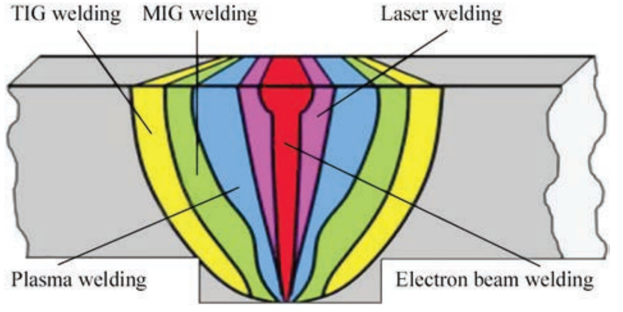

Figure 1. Comparison of weld dimensions for different kinds of welding

ELECTRON BEAM WELDING (EBW)

is welding, at which heating and melting of the contacting surfaces is performed by high-speed electron flows, moving in vacuum under the impact of the electric field. Welding is performed in the chamber in a vacuum, obtained by pumping down the air to a pressure of the order of 10^{-3} to 10^{-4} Pa. The electron beam is formed in a special assembly of the electron beam gun.

EBW features a high power density, being inferior only to laser welding as to this value, but it is greatly superior to electric arc welding. More over, EBW is characterized by minimal area of the heating spot. Figure 1 schematically shows the welding zone dimensions for different kinds of welding.

A significant advantage of EBW is complete degassing of the working area, which results in achievement of high-quality joining of chemically active metals. Absence of the influence of atmospheric oxygen and hydrogen on the weld metal allows achieving its more homogeneous and dense structure, as well as avoiding further corrosion.

The main disadvantage of EBW is the high cost of creating vacuum for welding highly active metals and alloys. That is why this welding method works in a narrowly specialized range of tasks for welding titanium alloys, high-technology expensive parts from these alloys, while meeting high requirements on tolerance and quality of the surface.

The objective of the work is determination of optimal modes of heat treatment of welded joints of sparsely-doped pseudo-β-titanium alloy of Ti–2.8Al–5.1Mo–4.9Fe system, made by electron beam and tungsten electrode argon-arc welding.

MATERIAL AND INVESTIGATION PROCEDURE

Welded joints of test sparsely-doped Ti–2.8Al–5.1Mo–4.9Fe titanium alloy were studied in the work. Multipurpose laboratory UE-208M electron beam unit was used for making the titanium alloy ingots [8]. Ingots of a round cross-section of 110 mm diameter were produced by the technology of cold-hearth electron beam melting (CHEBM) with portioned feed of liquid metal into a water-cooled crucible. Investigations of chemical composition of the produced ingots showed a uniform distribution of alloying elements along the ingot length. Plastic deformation of the billets was performed in a rolling mill of Skoda 355/500 model to the thickness of 10 and 6 mm by a standard procedure and they were annealed at the temperature of 750 °C. Table 1 shows the chemical composition of the studied titanium alloy.

Table 1. Chemical composition of sparsely-doped pseudo-β-titanium alloy [8]

Chemical composition, wt.%								
Al	Fe	Mo	Cr	Ni	Si	O	N	Ti (base)
2.78	4.87	5.13	0.03	0.02	≤0.003	0.08	0.02	89.48

Welding of this alloy was performed by two methods — EBW and AAW.

EBW of Ti–2.8Al–5.1Mo–4.9Fe alloy specimens 10 mm thick was performed in ELA 60/60 unit. Single-pass welding with through-thickness penetration was conducted in the following mode: accelerating voltage U_{ac} = 60 kV, beam current I_b = 80 mA, welding speed V_w = 7 mm/s. Temperature was controlled using thermocouples, fastened from the weld root side. To perform preheating before welding and local post-weld heat treatment directly in the chamber, the welded joints were heated by the electron beam, expanded into a rectangular scan. The width of the region heated at LHT, was determined so that it overlapped the weld and HAZ. In the work, the region width was 30 mm. Electron beam power during preheating and LHT was close to 3 kW that allowed ensuring the temperature on the level of 750 °C in the treated zone. The mode of welding with preheating and LHT is optimal from the viewpoint of cold cracking prevention and better weld formation.

Mode of AAW of Ti–2.8Al–5.1Mo–4.9Fe alloy specimens 6 mm thick was as follows: arc voltage U = 12 V, RPDC welding was performed by an automatic welding machine, welding current I = 350 A, welding speed v_w = 10 m/h, filler wire feed rate V_f = 30 m/h, arc voltage during welding was 12 V, shielding gas (argon) flow rate was 18 l/min in the nozzle and 22 l/min into the protective device for shielding the cooling weld metal. The weld reverse side was also shielded by argon from oxidation, using a copper forming backing. Figure 2 shows the general schematic of EBW (*a*) and AAW (*b*).

The filler metal used at AAW was 2 mm unalloyed titanium welding wire VT1-00zv. It allows varying the degree of weld metal alloying in a narrow range. The relative amount of filler metal in the weld metal was found by determining the penetration area of the joint metal in the transverse sections of the welds. During welding, the filler wire is fed into the pool head part. An oscillator device is connected in parallel to the power source for striking the arc. The os-

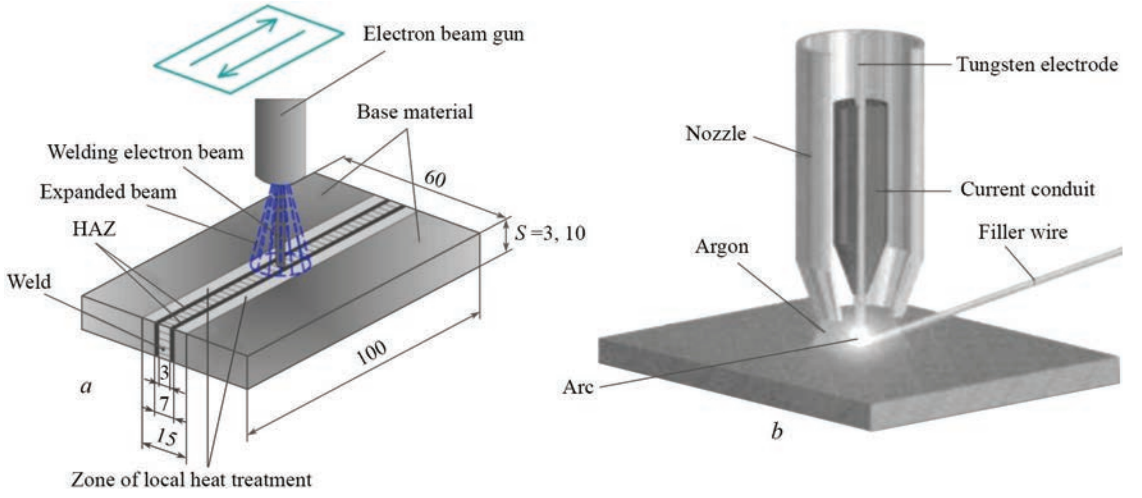

Figure 2. Schematic of welding sparsely-doped Ti–2.8Al–5.1Mo–4.9Fe titanium alloy: *a* — EBW; *b* — AAW

cillator applies high-frequency high-voltage pulses to the electrode, which ionize the arc gap and ensure arc striking after switching on the power source.

Application of the above welding methods is due to the features of the structures being welded and production capabilities.

In this work the fracture mechanism of welded joints of test sparsely-doped Ti–2.8Al–5.1Mo–4.9Fe pseudo-β-titanium alloy, produced by EBW and AAW was studied, also after the impact of different heat treatment modes (Table 2).

Mechanical properties of the studied specimens of base metal and welded joints are given in Table 3 [10].

INVESTIGATION RESULTS

In this work the fracture surfaces of welded joints of a titanium alloy of Ti–2.8Al–5.1Mo–4.9Fe system after impact toughness testing were investigated.

Figure 3 shows the test specimen fracture surfaces. Specimen analysis showed that for all the studied specimens fracture takes place by a mixed mechanism. The fracture surface is clearly defined and has a coarse-crystalline structure. The main crack propagated from the introduced stress raiser strictly in the direction of the applied load. Macroscopic analysis of the specimens showed the absence of shrinkage or “lips” of the cut that is indicative of small macroscopic plastic deformation at all the destruction stages [11]. At the same time, macroscopic analysis of the fracture mode showed that the plastic deformation fraction on the fracture surface essentially depends both on the welding type and on the heat treatment mode (Figure 3, *a*, *f*, *g*).

Microscopic study of fracture surface of specimens Nos 1–3 showed that it is heterogeneous and is characterized by a mixed pattern (Figure 4, *a*, *b*, *d*) [12]. Studies showed that the fracture surface is light-coloured,

Table 2. Modes of heat treatment of sparsely-doped titanium alloy of Ti–2.8Al–5.1Mo–4.9Fe system

Specimen number	Specimen type	Heat treatment
1	Base metal	–
2	EB welded joint (non-heat-treated)	–
3	EB welded joint with preheating up to 400 °C	–
4	EB welded joint with preheating up to 400 °C	LHT at 750 °C for 5 min
5	EB welded joint	Heating up to 760 °C, slow cooling at the rate of 1 °C/min
6	EB welded joint	Heating up to 760 °C, quenching into water, aging at 400 °C for 10 h
7	AAW joint made with VT1-00zv filler wire	Heating up to 760 °C, slow cooling at the rate of 1 °C/min

Table 3. Mechanical properties of welded joints of sparsely-doped titanium alloy of Ti–2.8Al–5.1Mo–4.9Fe system

Specimen number	Ultimate strength σ_t, MPa	Yield limit σ_y	Relative elongation δ, %	Reduction in area ψ, %	Impact toughness *KCV*, J/cm^2
1	1015	939	1.9	–	3.6
2	960	921	3.8	–	6.4
3	992	959	5.1	–	3.6
4	997	964	6.5	–	5.4
5	964	905	4.7	12.6	7.1
6	1204	1199	8.6	1.7	4.2
7	958	958	1.3	2.4	6.5

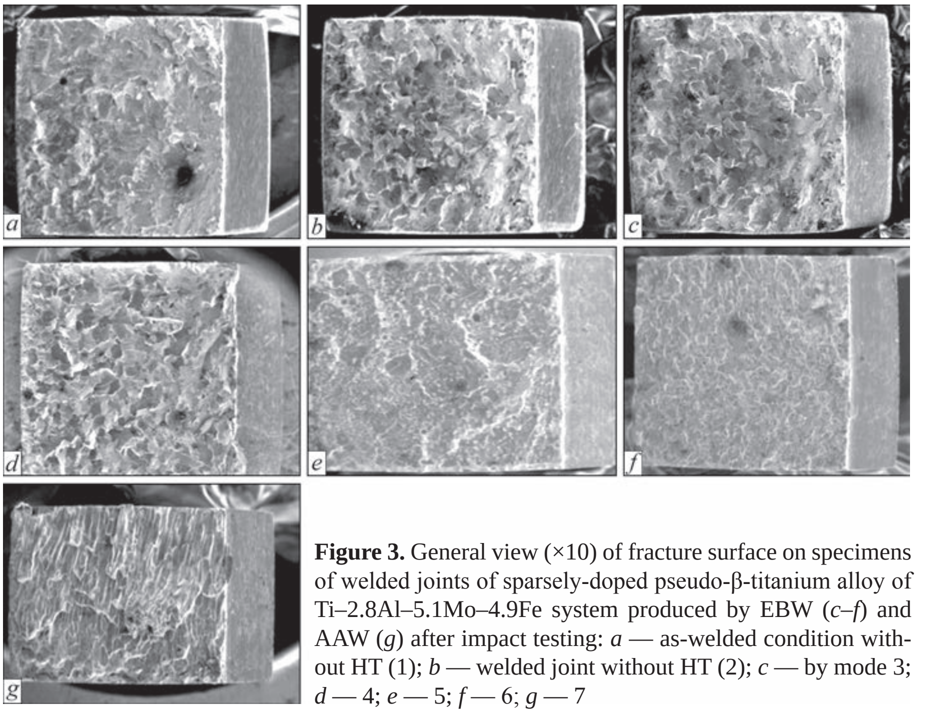

Figure 3. General view (×10) of fracture surface on specimens of welded joints of sparsely-doped pseudo-β-titanium alloy of Ti–2.8Al–5.1Mo–4.9Fe system produced by EBW (*c*–*f*) and AAW (*g*) after impact testing: *a* — as-welded condition without HT (1); *b* — welded joint without HT (2); *c* — by mode 3; *d* — 4; *e* — 5; *f* — 6; *g* — 7

coarse-crystalline, and developed. The fracture surface contains cleavage facets, which alternate with regions of intergranular and tough fracture. Cleavage facets separated by tearing regions, are observed. Specimens differ by the size of destruction facets (from 50 to 200 μm) and fraction of the tough component in the fracture (from 5 to 30 %). In specimens Nos 4–6 (Figure 3, *b*, *d*, *e*, *f*) the fracture surface is flat, and weakly-pronounced, that is indicative of a low rate of crack propagation which forms under the conditions of plane strain state. The fracture mode is predominantly brittle, tough component fracture is not higher than 20–25 %. The fracture surface of specimens Nos 6, 7 contains a series of parallel crystallographic surfaces in the form of steps (Figure 4, *f*, *g*).

A detailed analysis of fracture surface on specimens of welded joints of sparsely-doped Ti–2.8Al–5.1Mo–4.9Fe pseudo-β-titanium alloy showed that the details of fracture surface relief simultaneously contain both the cleavage facets and pits, which formed by the micropore coagulation mechanism (Figure 4, *b*, *f*, *g*). Tear ridges and river patterns are visible on the cleavage facets. Tough fracture elements were also found on fracture surfaces of specimens produced by modes 4, 5, and 6. This is indicative of the fact that after the appropriate heat treatment the material becomes more ductile (Figure 4, *d*, *f*).

Comparison of fracture mode of specimens after application of the appropriate heat treatment modes, performed by EBW and AAW showed (Figure 4, *f*, *g*) that

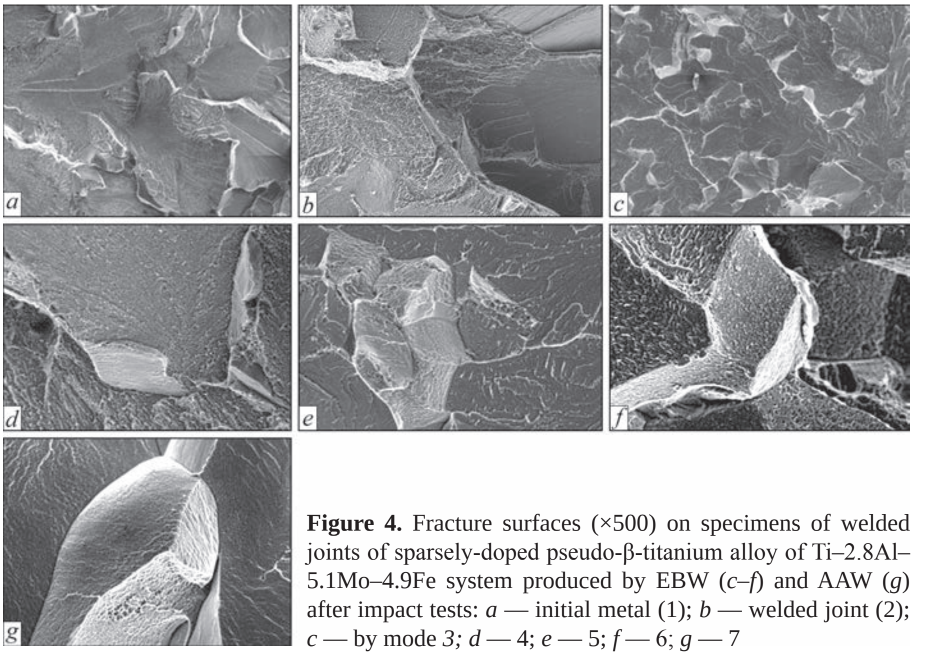

Figure 4. Fracture surfaces (×500) on specimens of welded joints of sparsely-doped pseudo-β-titanium alloy of Ti–2.8Al–5.1Mo–4.9Fe system produced by EBW (*c*–*f*) and AAW (*g*) after impact tests: *a* — initial metal (1); *b* — welded joint (2); *c* — by mode 3; *d* — 4; *e* — 5; *f* — 6; *g* — 7

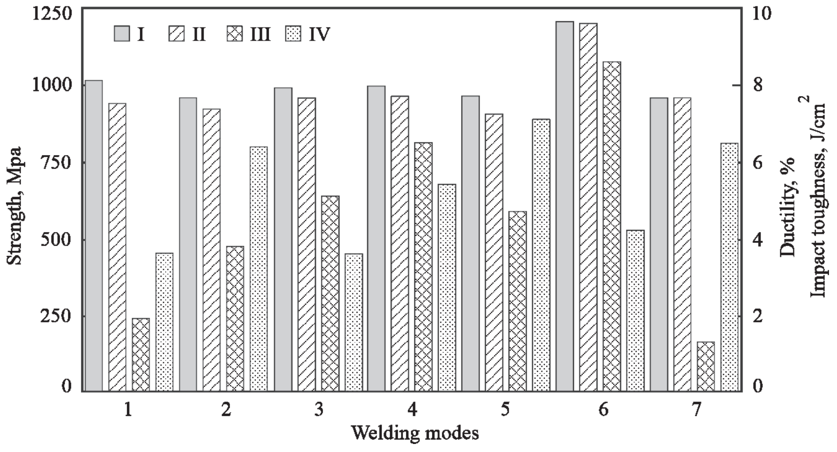

Figure 5. Influence of heat treatment modes on strength and ductility values of Ti–2.8Al–5.1Mo–4.9Fe titanium alloy (I — ultimate strength; II — yield point; III — reduction in area; IV — impact toughness)

specimens after EBW are more ductile and have a greater fraction of the tough component in the fracture.

Analysis of the results of mechanical testing of the studied sparsely-doped Ti–2.8Al–5.1Mo–4.9Fe titanium alloy after heat treatment by different modes (Table 3) showed that heat treatment of this alloy allows changing its mechanical properties in a rather broad parameter range. So, for EB welded joints in the heat-treated condition the highest ultimate strength of test Ti–2.8Al–5.1Mo–4.9Fe titanium alloy was achieved in mode 6 (quenching at the temperature of 760 °C + cooling into water + aging at 400 °C) — 1204 MPa, and the lowest value of 964 MPa was produced in mode 5 (slow cooling from the temperature of 760 °C at the rate of 1 °C/min). That is, appropriate heat treatment allows achieving properties both higher (mode 6) and lower (mode 5) that those of a non-heat-treated welded joint (mode 2, 960 MPa). Respective results were obtained also for the yield limit: the highest values were produced in mode 6 — 1199 MPa, and the lowest values were obtained in mode 5 — 905 MPa.

The ductility values (relative elongation and reduction in area) in the test Ti–2.8Al–5.1Mo–4.9Fe titanium alloy remains low after heat treatment. Maximum relative elongation was achieved in specimens after heat treatment by mode 6 — 8.6 % (126 % higher), and the lowest value was obtained in mode 7 — just 1.3 % (66 % lower). The specimen heat treated by mode 5, had the greatest reduction in area of 12.6 % (92 % greater) at respective elongation (4.7 %) (Figure 5).

Impact toughness of the studied specimens was not high, varying from 4.2 (mode 6) to 7.1 J/cm² (mode 5), the studied heat treatment modes changing it only slightly. The cause for the low values of impact toughness and ductility can be iron content in the test alloy — almost 5 % (see Table 1) that leads to formation of particles of titanium-iron intermetallic during production, which have a negative impact on the level of impact toughness of the test Ti–2.8Al–5.1Mo–4.9Fe titanium alloy [11].

In order to select the optimal mode of heat treatment of test Ti–2.8Al–5.1Mo–4.9Fe titanium alloy, a criterion of heat treatment "quality" in conditional units was proposed. It consists of the contribution of heat treatment modes into the comprehensive increase (or lowering) of total values of strength, ductility and impact toughness of the new titanium alloy relative to a non-heat-treated specimen (specimen 2) (Figure 6). That is, increase of some mechanical properties, for instance, strength, usually causes the respective lowering of ductility. In certain cases, however, it is disproportionate. In this connection, the "quality" criterion was proposed:

$$K_{ht} = \Sigma \frac{(P_{in} - P_{ht})}{P_{in}},$$

where K_{ht} is the "quality" criterion; P_{in} is the initial parameter (ultimate strength, yield limit, relative elongation, impact toughness) of a non-heat-treated specimen; P_{ht} is the heat-treated specimen parameter.

Positive values of K_{ht} parameter determine a comprehensive increase of the properties, and negative ones determine their lowering.

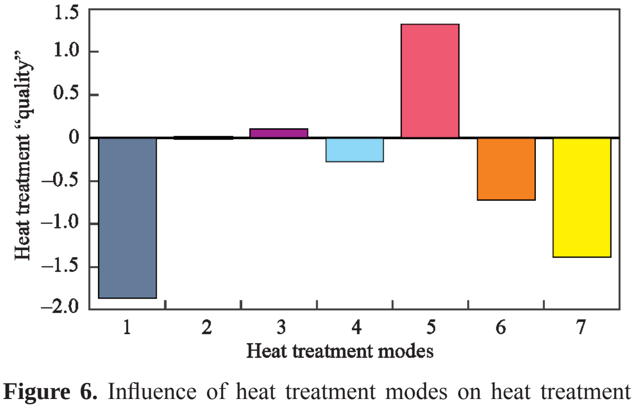

Figure 6. Influence of heat treatment modes on heat treatment "quality"

Analysis of the produced results shows that a comprehensive increase of welded joint properties at EBW can be achieved using mode 5. At application of heat treatment at AAW (mode 7) it is impossible to achieve an optimal combination of strength, ductility and impact toughness.

Use of the same heat treatment modes: heating up to 760 °C and subsequent slow cooling at the rate of 1 °C/min (modes 5 and 7) for welded joints made by two different kinds of welding, namely EBW and AAW, showed that the mechanical properties differ slightly: ultimate strength was 964 MPa (EBW) against 958 MPa (AAW), respectively.

Impact toughness values of specimens with a sharp notch for the two heat-treated specimens also changes only slightly from 7.1 J/cm^2 (EBW) against 6.5 J/cm^2 (AAW). The main difference after heat treatment is found in the ductility properties (relative elongation and reduction in area). Heat treatment of EBW specimens results in much better ductility properties (Table 3). This is attributable to the fact that the ductility properties in the heat-treated EBW specimens are higher due to increase of the tough component on the fracture surface (Figure 3, *f*, *g*). In EBW specimens the fracture contains 28 % of the tough component, and in AAW specimens it contains less than 15 %.

CONCLUSIONS

1. It is found that local heat treatment in a vacuum chamber of specimens of EB welded joints of test Ti–2.8Al–5.1Mo–4.9Fe titanium alloy by mode 4 (LHT in the vacuum chamber at the temperature of 750 °C for 5 min) allows producing higher ductility properties of welded joints and preventing cold cracking after welding. Such heat treatment leads to a more uniform arrangement of tough fracture regions on the fracture surface.

2. It is shown that application of heat treatment of specimens of Ti–2.8Al–5.1Mo–4.9Fe titanium alloy by mode 5 (heating up to 760 °C, slow cooling at the rate of 1 °C/min) allows producing an optimal complex of ductility (δ = 4.7 %, ψ = 12.6%) and impact (KCV = 7.1 J/cm^2) properties.

3. It is found that slow cooling after AAW also ensures an increase of ductility properties of welded joints of Ti–2.8Al–5.1Mo–4.9Fe titanium alloy, but they are lower than those at EBW with slow cooling, that is related to a higher cooling rate at AAW.

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CONFLICT OF INTEREST

The Authors declare no conflict of interest

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