DOI: https://doi.org/10.37434/tpwj2023.04.02

MATERIALS AND TECHNOLOGICAL APPROACHES TO WELDING OF COMBINED JOINTS BETWEEN MARTENSITIC AND AUSTENITIC STEELS USED IN POWER MACHINE BUILDING (REVIEW)

M.O. Nimko, V.Yu. Skulskyi, T.V. Ivanenko

E.O. Paton Electric Welding Institute of the NASU
11 Kazymyr Malevych Str., 03150, Kyiv, Ukraine

ABSTRACT

A review of recent foreign research literature is given, dedicated to the choice of welding materials and relevant approaches to welding of combined joints between martensitic and austenitic steels, which are widely used at the moment in the global and domestic power machine building. The typical representatives of martensitic and austenitic steels are considered; the problems arising in the joints of dissimilar steels during high-temperature operation, in particular, the process of carbon diffusion from a less alloyed steel to a more alloyed one and arising of thermal stresses due to the difference in the coefficients of thermal expansion of the combined steels are mentioned; technological approaches to reduce the negative impact of the abovementioned factors are described, as well as the promising welding materials used in the manufacture of such joints and their comparative analysis are given. At the end of the review, a table is given in which literary references are classified depending on the grade of combined steels and welding materials used in the manufacture of welded joints in relevant literary references.

KEYWORDS: welding materials, joints of dissimilar steels, martensitic steels, austenitic steels, nickel alloys, carbon diffusion, decarbonized interlayer

INTRODUCTION

The main thermodynamic cycle of thermal power plants used in modern heat power engineering is the Rankine cycle with steam overheating. To implement this heat cycle, different sections of a water-steam circuit at a power plant should have different parameters of temperature and pressure of the working body. In order to make the construction of a power plant cheaper, the sections with lower steam parameters are manufactured of low-alloy bainitic steels with 0.50–2.25 % Cr (tubes of boiler walls, operate at a temperature of 550–580 °C) and martensitic steels with 9–12 % Cr (upper sections of boiler wall tubes, collectors, main steam lines, for temperatures of up to 625–630 °C), sections with higher steam parameters that require enhanced fatigue strength and scale resistance — from more valuable austenitic steels (superheater coils, operating temperatures are up to 660–680 °C) [1]. To manufacture a closed water-steam circuit, these sections are joined between each other by welding, forming a joint of dissimilar steels. For example, in thermal power plants operating at ultra-supercritical steam parameters and maximum temperatures of 600 °C, 4674 joints may be used only between martensitic and austenitic steels [2].

The appearance of new steels, predetermined by the need of increasing the steam parameters to supercritical and ultra-supercritical levels and the corresponding increase in efficiency of power plants, required new studies of the features of welding of combinations of these steels with steels operating on other temperature modes.

The aim of the work was a description of the state-of-the-art of developing problems typical for producing joints between modern martensitic and austenitic steels, as well as analysis of welding materials, in particular, nickel-based materials used in recent decades in research works on creating approaches to welding of dissimilar steels.

BASIC STEEL GRADES FOR TUBE SYSTEMS OF OVERHEATED STEAM

This section considers alloying of the main types of martensitic and austenetic steels, welding of whose combination represents a relevant problem for modern thermal power engineering.

To the class of modern martensitic steels, most common in power machine building, steels of the system with a base alloying 9Cr–1Mo: P91 (X10CrMoVNb9-1 according to EN), T92 (X10CrWMoVNb9-2) belong (Table 1).

P91 steel was developed at the National Laboratory in Oak Ridge in the late 1970s and is a modified version of P9 steel (9Cr–1Mo) with controlled additions of vanadium, niobium and nitrogen to improve the high-temperature mechanical properties of steel due to dispersion strengthening by precipitation of carbonitrides MX (where M is vanadium or niobium, and X is carbon or nitrogen). Typical operation temperatures

Table 1. Nominal chemical composition of the most widespread modern martensitic steels [1]

Steel grade	Mass fraction, %									
	C	Si	Mn	Cr	Mo	W	V	Nb	N	B
P91	0.10	0.35	0.45	9.0	0.95	–	0.22	0.08	0.05	–
P/T92	0.09	0.25	0.50	9.0	0.50	1.8	0.20	0.05	0.05	0.003

of this steel in high pressure vessels and tube systems of thermal power plants amount to 580–600 °C [1].

P92/T92 steel is the next modification of P91 steel. The new steel was developed in the "Nippon Steel Corporation" in the 1990s for operation at temperatures of 600 °C and higher and the steam pressure of about 25 MPa. The further improvement in creep resistance in this steel was achieved by the addition of tungsten, which was partially replaced by molybdenum and boron. The initial purpose of adding W and B was an increase in the fraction of solid solution-strengthening and limitation of coarsening of dispersion precipitates [1].

With regard to austenitic steels, it was found that steels of 08Kh18N10 type (S304H, 316L, etc.), originally developed for using in the corrosive environment, also have excellent high-temperature properties. Austenitic stainless steels used in the power machine building and mentioned in this review, are given in Table 2.

Steels operating in corrosion environment have a low carbon content, usually below 0.03 %, which helps to prevent the formation of intercrystalline corrosion. Such steels are designated by the letter L at the end of the name. High-temperature variants of these steels (with the designation H at the end of the name) have a carbon content of about 0.08 %, which contributes to a slight enhancement in the creep resistance.

To provide a sufficient resistance to corrosion and oxidation at temperatures of ~700 °C, it is necessary to increase the content of chromium in steels to 20–25 %. Two classic high-temperature steels in this range of chromium content are 310 and Alloy 800H. However, both have very low creep resistance at 700 °C, although steel 310 is also used at higher temperatures because of its excellent corrosion resistance. To improve the creep resistance, steel HR3C was additionally developed [1].

Additions of Ti and Nb in austenitic steels help to prevent the development of intercrystalline corrosion and additions of Mo improve the pitting corrosion resistance and provide a slight increment to the creep resistance. Nitrogen helps to increase the strength at elevated temperatures by two mechanisms: due to a solid solution strengthening and through the formation of fine-dispersion niobium nitrides.

PROBLEMS ARISING DURING OPERATION OF DISSIMILAR JOINTS

Welded joints of dissimilar steels are characterized in certain areas by a very sharp gradient in the microstructure, physical properties, chemical potential and, as a consequence, in mechanical properties. The main factors responsible for failures during the creep of joints of dissimilar heat-resistant steels are:

- difference in physical properties;
- decarbonization/carbonization in the area of contact of steels due to the difference in the chemical composition of steels and, accordingly, their chemical potential.

Additional stresses in the joints of dissimilar steels can arise because of mismatch of physical properties: for example, austenitic steels have a coefficient of thermal expansion (CTE) by 30 % greater than the martensitic — typical CTE values are 18 and 13–14 μm/m·K, respectively [3]. A cyclic temperature effect on the operation of power plants plays an important role in premature failure of these joints. The difference in the coefficients of thermal expansion and thermal conductivity between the base metal and the weld metal lead to the formation of thermal stresses in a welded joint during numerous launches and

Table 2. Nominal chemical composition of austenitic steels [1]

Steel grade	Mass fraction, %									
	C	Si	Mn	Cr	Ni	Mo	N	Nb	Ti	Other
S304H	0.10	0.2	0.8	18.0	9.0	–	0.1	0.4	–	3.0 Cu
316L	0.03	0.6	1.6	16.0	12.0	2.5	–	–	–	–
347H	0.08	0.6	1.6	18.0	10.0	–	–	0.8	–	–
310	0.08	0.6	1.6	25.0	20.0	–	–	–	–	–
HR3C	0.08	0.4	1.6	25.0	20.0	–	0.2	0.45	–	–
Alloy 800H	0.08	0.5	1.2	21.0	32.0	–	–	–	0.5	0.4 Al
Tempaloy AA-1	0.10	0.3	1.5	18.0	10.0	–	–	0.3	0.2	0.02 B 3.0 Cu

shutdowns of power plants during operation. These cyclic stresses imposed on residual welding stresses and stresses from external load/internal steam pressure can cause critical failure of such joints in the conditions of fatigue and creep. Typically, such failures occur in the HAZ of martensitic steel directly near the fusion line as a result of the propagation of circumferential cracks [4].

A more significant mechanism that affects the time and place of failure in the joints of dissimilar steels, which are exposed to the creep effect, is a structural heterogeneity caused by residual stresses and a gradient of the chemical composition of steels. A difference in chemical potentials between martensitic and austenitic steels leads to migration (diffusion) of carbon across the fusion line of steel with a lower chromium content in steel with a higher chromium content [3]. As a result, after a while, at a high-temperature exposure in steel with a lower content of chromium (usually martensitic steel), near the fusion line a decarbonized interlayer begins to form and propagate, which has reduced hardness and strength, and can serve as a place of specimen failure in the future. Also it was noted that:

- carbon migration significantly reduces the nanohardness and the yield strength of a decarbonized interlayer, which becomes the weakest area in the whole joint [5];
- when the operation temperature grows, the probability of failure on a decarbonized interlayer during a high-cyclic fatigue increases [6];
- a decarbonized interlayer significantly reduces the fatigue resistance of joints during creep at high loads, however, when the test loads are reduced, this effect becomes less noticeable [7].

The abovementioned mechanisms impose restrictions on the use of dissimilar steels of austenitic materials for welding, which, in addition to thermal stresses due to a sharper CTE gradient and carbon diffusion on the side of the martensitic steel to the austenitic weld, can also lead to the formation of a decarbonized interlayer in the fusion zone. In [8], it was shown that microcracks may originate in such a decarbonized interlayer. Such microdefects are mainly originated in the transition zone and have an intergranular nature. It is noted [9] that during failure in a decarbonized interlayer, the interaction of adjacent carbonized and decarbonized areas is critical, that have extreme physical and chemical heterogeneity in the narrow area: in the carbonized zone, microcracks originate and further propagate to the decarbonzied zone, where increased stresses are located.

To overcome the mentioned problem, it was proposed to use nickel-based welding materials, which both in laboratory studies as well as during operation showed a significant improvement in the service life of the joints compared to the joints, welded by chromium-nickel austenitic materials [4]. This is clear, as far as nickel-based weld metal has an intermediate CTE between martensitic and austenitic steels, which respectively leads to a decrease in the value of cyclic thermal stresses in welded joints. Additionally, the use of nickel materials significantly reduces the degree of carbon migration from martensitic steel into the weld metal due to a low gradient of carbon activity, as well as a low carbon diffusion coefficient in nickel alloys.

Therefore, at present time, nickel-based materials are preferred in the manufacture of welded joints between martensitic and austenitic steels. There is a point of view that the use of nickel materials can five times increase the service life of welded joints compared to the joints that use austenitic iron-based materials [4].

The main disadvantage of nickel-based materials is their worse weldability compared to austenitic materials, which requires the advanced qualification of welders. Also, in the literature information can be found describing a tendency of nickel-based metal to the formation of hot cracks in the joints of dissimilar alloys, but most of the information refers to dissimilar joints between nickel alloys and austenitic steels [10, 11].

TECHNOLOGICAL APPROACHES TO WELDING OF DISSIMILAR STEELS

In addition to the traditional performance of the multipass welded joint, the approach with the preliminary cladding of martensitic steel by nickel welding material, subsequent heat treatment of the cladding edge on the mode of tempering of martensitic steel and, then, final welding with the austenitic side of the joint [2, 12–14] (Figure 1).

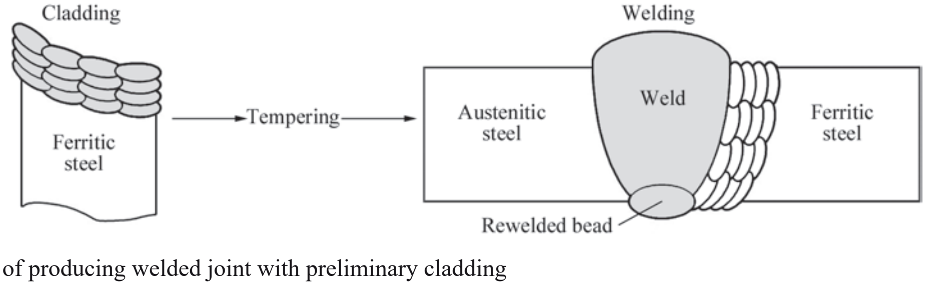

Figure 1. Scheme of producing welded joint with preliminary cladding

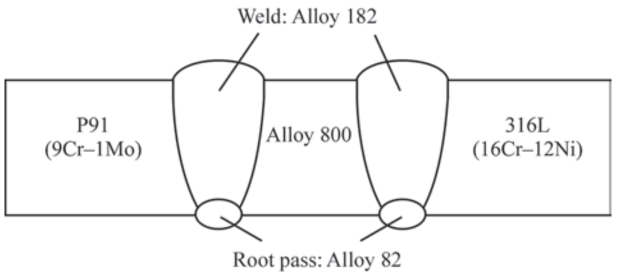

Figure 2. Scheme of joining martensitic and austenitic steel through the insert of the material Alloy 800 [13, 17, 18]

The further technological approaches, aimed at reducing the effect of difference in thermophysical properties between martensitic and austenitic steels, are based on the use of inserts with intermediate CTE. The use of an intermediate insert of Alloy 800 (high-nickel INCOLOY) was proposed, and it is shown that the use of such inserts can lead to a significant decrease in cyclic thermal stresses [4]. Alloy 800 was selected based on the fact that it has an intermediate coefficient of thermal expansion between martensitic and austenitic steels, as well as acceptable creep resistance and oxidation resistance. It is assumed that the use of trimetallic joint with an insert of Alloy 800 can four times increase the duration of service life of welded joint of dissimilar steels compared to conventional bimetallic joint [4].

In the series of works [10, 15–18], Sireesha et al. studied the microstructure and mechanical properties of welded joints P91 + Alloy 800 and Alloy 800 + + 316L as a part of trimetallic joint of P91/Alloy 800/316L: based on the tests on hot cracks formation and results of mechanical and metallographic tests, the comparison of welding materials was carried out. Based on the results of these works, the autors proposed to use the materials Inconel 82 to produce root passes and Inconel 182 — for filling (Figure 2). On the joints produced by the recommended welding materials, the effect of high-temperature aging at 625 °C on the microstructure and mechanical properties and crack formation in thermal cycling in the range of 20–650 °C preriliminary heat treated by the recommended welding materials were studied. The results of this study showed that even in the most unfavorable conditions of thermocyclic testing there is no formation of cracks or oxide undercuts.

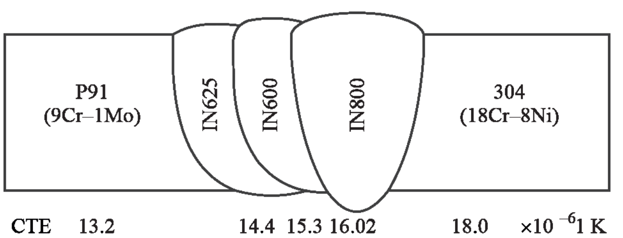

Figure 3. Scheme of producing transition three-metallic weld with a gradient CTE [21, 22]

But in the later work of Lakh et al. [19], the limited prospect of using similar joints with an insert of Alloy 800 is shown. As a result of testing the joints from ferritic-martensitic steels (P9, P91, 2.25Cr–1Mo) with Alloy 800 at 550 °C on creeping, the authors found that the creep resistance of such joints is lower than the creep resistance of the corresponding ferritic steels. It was determined that with a decrease in the load during the creep tests, the place of failure is shifted from the base metal of ferritic steel into the area of intercritical temperatures of HAZ of ferritic steel, which leads to the formation of cracks of type IV. At the lowest loads, the failure occurs in the fusion zone between the ferritic steel and the weld metal and is associated with very low ductility during creeping in this area.

A new approach to welding joints of dissimilar steels was proposed, which was based on the use of transition joints with a filler metal from several steels or alloys [20]. It was assumed that filler materials should have the value of CTE in the range between martensitic and austenitic steels and should be arranged during the weld formation at an increase of CTE from martensitic steel to austenitic.

This approach was partially tested in two works [21, 22], in which two joints of P91 steel with 304 steel were produced by friction stir welding: the first used three-layer transition weld Inconel 625/Inconel 600/Inconel 800 (Figure 3), in the second — a single-layer weld made by the material Inconel 600. Three materials in the first joint were selected based on their value of CTE (Inconel 625 — 14.40·10⁻6 K^{-1}; Inconel 600 —15.30·10⁻6 K^{-1}; Inconel 800 — 16.02·10⁻6 K^{-1}), that provided a gradient variation of CTE in the joint between P91 steel (13.18·10⁻6 K^{-1}) and 304 steel (18.0·10⁻6 K^{-1}). The results of the calculations (finite element method) and the creep tests showed that three-metallic joint has better mechanical properties and creep resistance than a simple joint due to a lower stress distribution in the joint. The disadvantage of the three-metallic joint was the presence of a soft interlayer in HAZ of P91 steel, the formation of which could be associated with carbon diffusion in the adjacent alloy Inconel 625 (the alloy contains 22 % of Cr).

In this connection, the promising research area is the development of heat-resistant welding material based on nickel or iron-nickel, which could have the lowered content of chromium (~ 9% Cr) for prevention of carbon diffusion and CTE close to martensitic steels (~ $14\cdot10^{-6}$ K^{-1}). One of the first steps in this direction was the study of Electric Power Research Institute (EPRI) and the development of welding material HFS6 associated with them developed in the 1980s. HFS6 has never reached a commercial use due

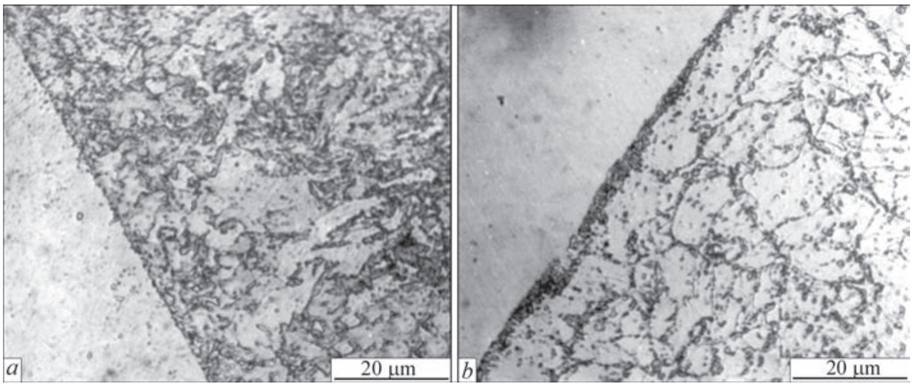

Figure 4. Microstructure of HAZ of P91 steel together with the fusion zone (weld Inconel 182) before and after thermal cycling: *a* — tempering at 760 °C, 50 h; *b* — tempering at 760 °C, 50 h + thermocycling 20–625 °C [16]

to the succeptibility to the liquation cracking in the weld metal, but the data obtained in the research program on its development were of great importance in the future [23]. In 2002, the development of electrodes EPRI P87 was initiated and already in 2009 it was reported about their successful commercial use in the construction of thermal coal power plant in the USA. On the basis of chemical composition of electrodes, a solid wire was later developed for using in TIG welding. A characteristic feature of this nickel-based material is the content of chromium and carbon, close to the content of these elements in P91 steel, which should facilitate the elimination of carbon migration (and formation of carbides of type I) across the fusion zone between the weld, made by P87 and martensitic steel. In addition, P87 has an acceptable CTE value similar to CTE of other nickel-based materials [24].

A broader overview of the types of transition welds, which can be found in the joints of ferritic and austenitic steels in power machine building is given in [4].

CARBON DIFFUSION WHEN USING NICKEL-BASED MATERIALS

Most researchers assume that traditional nickel-based materials (Alloy 82/182/617/625 etc.) can not effectively restrain carbon diffusion from martensitic steel to a weld, since most nickel alloys used as welded materials contain a large number of carbide-forming agents (Cr, Mo, Nb, Ti).

Regarding the joints between martensitic and austenitic steels, in [16] it was shown, that in the joint P91 + Alloy 800, produced by the electrodes Inconel 182, after a long tempering for 20 and 50 h at a temperature of 760 °C and a subsequent thermocycling in the range of 20–625 °C/3025 h at the boundary between P91 steel and nickel weld, carbon migration and formation of a dark-etched area with an increased hardness occur (Figure 4).

In [12], already after tempering at 760 °C/2 h, the formation of a carbonized interlayer in cladding by Inconel 182 and the formation of a soft area in HAZ of P91 steel (Figure 5) was observed. Analysis of the results of mechanical tensile tests of the miniature specimens from different areas of the cladded joint P91 + 316L showed that HAZ of P91 steel had the lowest values of long-term strength and resistance to ductile instability both at a room temperature and at a temperature of 550 °C.

Some of the works give controversial data on the effectiveness of traditional nickel alloys in counteracting carbon migration, possibly predetermined by the lack of a single approach to the evaluation of a decarbonized HAZ area of a low-carbon steel in the combined welded joints.

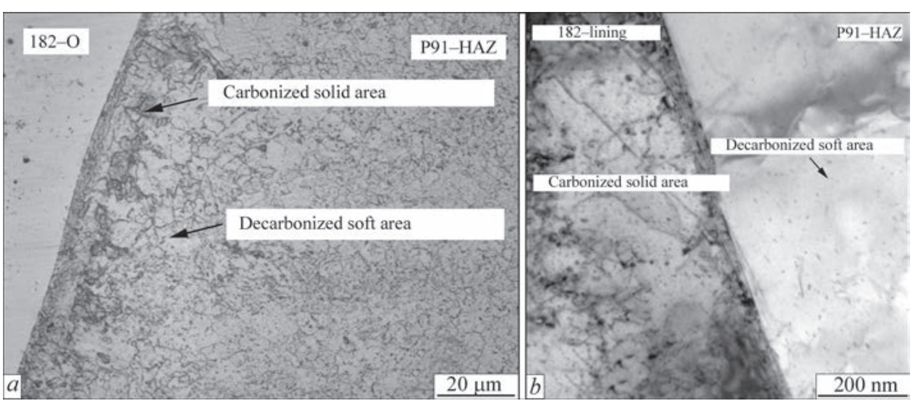

Figure 5. Microstructure of the fusion zone of P91/Inconel 182 after tempering at 760 °C, 2 h: *a* — optical microimage (etched side of P91); *b* — TEM image [12]

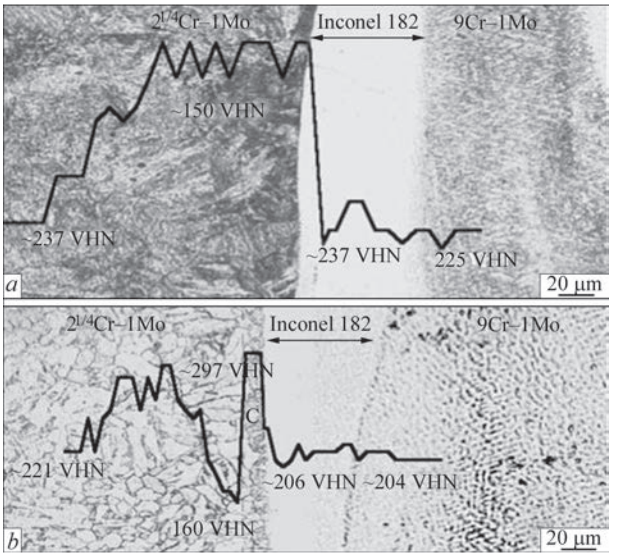

Figure 6. Microstructure and hardness profile in the transition joint, made with a cladding layer of Inconel 182 of 0.1 mm thickness after heat treatment at 750 °C with the duration: *a* — 1 h; *b* — 15 h [25]

For example, in the work of Anand et al. [25] by using of the calculation methods, carbon diffusion was studied in surfacing "substrate of 2.25Cr–1Mo/intermediate interlayer of Inconel 182 (0.1 mm)/layer of 9Cr–1Mo (2 mm)" at a temperature of 750 °C and an exposure of up to 15 h.

The results of the calculation in the triple Fe–Cr–C system on the basis of the finite difference method showed that the efficiency of lining with a nickel material depends on its thickness: in the joints, the lining thickness of which was 40 μm, the formation of strengthened carbonized areas near the fusion zone and their absence in the joints with the lining of 80 μm thickness and higher was predicted, i.e., with an increase in lining thickness, its effectiveness in the removal of carbon migration grows. These results were confirmed experimentally by metallography and measuring hardness of deposits, tempered at 750 °C for 1 and 15 h, which showed the absence of formation of solid areas in a nickel weld or steel 9Cr–1Mo (Figure 6).

Regarding the new material EPRI P87, in [26] its high efficiency in restraining carbon migration from martensitic steel T92 was shown, which significantly exceeds that compared to traditional nickel alloys. It is shown that in the joints made by Inconel 82 after tempering at 760 °C, 30 min, a decarbonized area in some places was 18 times wider, and in the case of Inconel 617 — by 24 times wider than in the joints of EPRI P87 (Figure 7). In [27] on the plate of P91 steel, three deposits were performed: by the material EPRI P87 (9 % Cr), by austenitic stainless 309 steel (22–25 % Cr) and by Inconel 625 (20–23 % Cr). After surfacing, the plate was normalized at a temperature of 1060 °C, 2 h and tempered at a temperature of 760 °C, 2 h. The results of metallographic examinations shown in Figure 8, confirm the benefits of a new material in restraining decarbonization.

SELECTION OF MATERIALS FOR WELDING TYPICAL JOINTS

Analysis of literature data shows that recently for welding of typical combined joints in the power machine

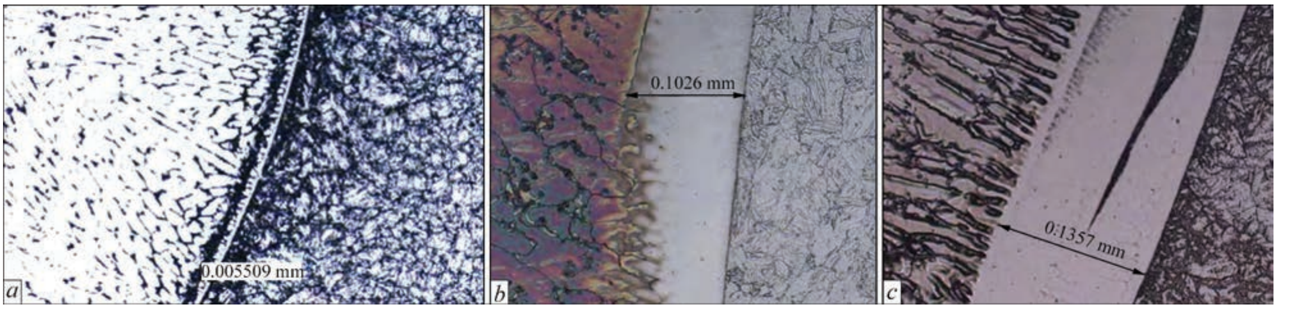

Figure 7. Transition zone between T92 steel and weld, made by the material EPRI P87 (*a*), Inconel 82 (*b*) and Inconel 617 (*c*) after temperting at 760 °C, 30 min [26]

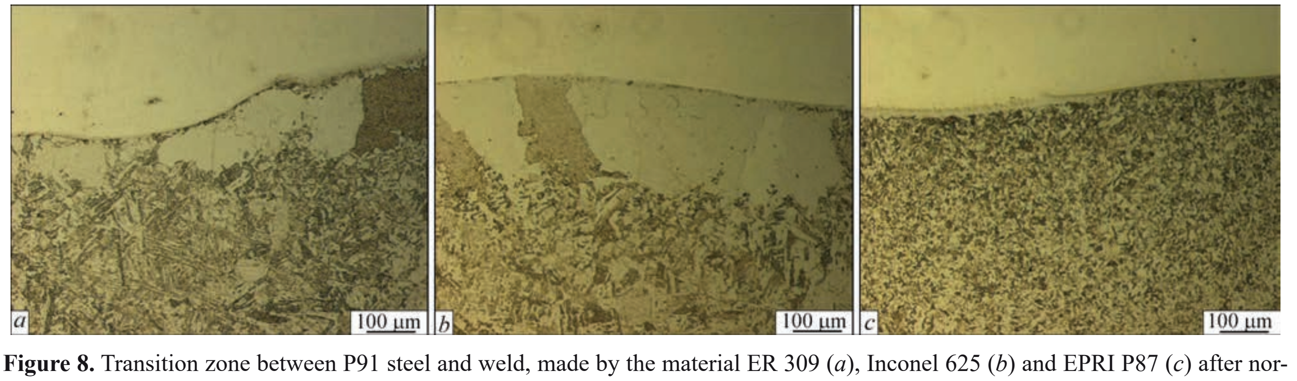

Figure 8. Transition zone between P91 steel and weld, made by the material ER 309 (*a*), Inconel 625 (*b*) and EPRI P87 (*c*) after normalization at 1060 °C, 2 h and tempering at 760 °C, 2 h [27]

building, the materials most commonly used in research works are Alloy 82 and Alloy 182 (Table 3). The typical chemical composition of the most common nickel-based welding materials is given in Table 4.

Some works provide comparative studies of microstructure and mechanical properties of various welding materials regarding a specific combination of welded steels. Thus, in [33], the comparisons of welding materials T-304H, Alloy 617 and Alloy 82 were made based on the criteria of tests on impact toughness and static bending of the joints T92 + 304 (postweld tempering at 630–670 °C, 2 h). It is shown that in the produced compositions of welded joints, the welds produced by the electeodes T-304H had the lowest impact toughness. When comparing the materials Alloy 617 and Alloy 82, the first was preferred, since the welds in the welded joints produced by Alloy 617 had a higher value of ultimate strength than the welds produced by Alloy 82.

Comparative studies of properties of the joints of steels T92 + 304, made using welding materials Alloy 625 and Alloy 82, showed that welded joints with the welds of Alloy 625 had the best microstructural and mechanical properties [35].

Huysmans et al. [2] performed aging at 625 °C, 10000 h of joints T92 + 304, produced by the weld-

Table 3. Typical combined joints found in recent research literature and materials used for their welding

Martensitic steel	Austenitic steel	Welding materials	Literary sources
P91	S304H	Alloy 625, Alloy 600	21, 22, 28
		EPRI P87	23
	S304L	Alloy 182	14
		AWSER90S-B9 (9CrMoV–N)	29
	316L	Alloy 182	10, 12, 13, 15, 16, 17, 18, 19
		Alloy82	10, 15
	310HCbN	EPRI P87	23
	347H	Alloy 82	30, 31, 32
		Rutox (19Cr–20Ni)	30
		ER90S-B3	32
		EPRI P87	23
T92	S304H	Alloy 82	33, 34, 35, 36, 37
		Alloy 617	2, 33, 38
		T-304H	33
		Alloy 625	35
		EPRI P87	2, 23
	316H	Alloy 82	39, 40, 41, 42
	Tempaloy AA-1	EPRI P87	26
		Alloy 82	
		Alloy 617	
	310	Alloy 82	2
		EPRI P87	23
	347HFG	EPRI P87	23
	HR3C	Alloy 82	43
		EPRI P87	2

Table 4. Chemical composition of typical nickel-based welding materials

Alloy	Mass fraction, %									
	C	Si	Mn	Cr	Ni	Mo	Nb	Ti	Fe	Other
Alloy 82	≤ 0.1	≤ 0.5	2.5–3.5	18.0–22.0	≥ 67.0*	–	2.0–3.0**	≤ 0.75	≤ 3.0	Cu ≤ 0.5
Alloy 182	≤ 0.1	≤ 1.0	5.0–9.5	13.0–17.0	≥ 59.0*	–	1.0–2.5**	≤ 1.0	≤ 10.0	Cu ≤ 0.5
Alloy 600	≤ 0.15	≤ 0.5	≤ 1.0	14.0–17.0	≥ 72.0*	–	–	–	6.0–10.0	Cu ≤ 0.5
Alloy 617	0.05–0.15	≤ 1.0	≤ 1.0	20.0–24.0	≥ 44.5	8.0–10.0	–	≤ 0.6	≤ 3.0	Co 10.0–15.0 Al 0.8–1.5 Cu ≤ 0.5 B ≤ 0.006
Alloy 625	≤ 0.1	≤ 0.5	≤ 0.5	20.0–23.0	≥ 58.0	8.0–10.0	3.15–4.15**	≤ 0.4	≤ 5.0	Co ≤ 1.0 Al ≤ 0.4
EPRI P87 [24]	≤ 0.1	≤ 0.3	≤ 1.5	≤ 9.0	Balance	≤ 2.0	≤ 1.0	–	≤ 38.0	Negligible additions of Al, Ti, N, B
Note. *The content of Ni + Co is indicated. **The content of Nb + Ta is indicated.										

ing wires Alloy 617, Alloy 82 and EPRI P87 (TIG). It was determined that in the joints produced by the wire Alloy 617, when aging in HAZ of T92 steel, carbides of type I precipitate, which can lead to preliminary failures in a high-temperature creep. After the same aging conditions, the material P87 showed itself better than other, and it is stated that preference should be given to the technologies of a preliminary cladding of the edge with tempering before conventional welding. Alloy 82 was recognized by the authors of the study as a good alternative to the material P87. However, it was established that at longer high-temperature exposures in the joints, produced by Alloy 82, near the fusion line of T92 steel, more single-oriented carbides are formed, which can subsequently lead to failure.

According to the results of the mentioned works for the joints T92 + 304, based on the improvement in the quality of joints, the following conditional comparative classification of welding materials can be composed: T-304H < Alloy 617 ≈ Alloy 82 < Alloy 625 < EPRI P87.

The materials Alloy 617, Alloy 82 and EPRI P87 were also studied in [26] in the joints T92 + Tempaloy AA-1. In the tests on static tension and impact toughness, the best results were shown by the material Alloy 82; Alloy 617 and EPRI P87 showed similar lower values, which still exceeded the minimum level recommended by the European standards. During tests of all three joints on bending, the angle of bending was maintained at 180° without the formation of microcracks. The measurements of microhardness in the state after tempering at 760 °C, 30 min showed that the joints produced by the material Alloy 82, the strongest softening occurred in the decarbonized area. It is assumed that namely in this area, in the future the failure will be occurred during high-temperature operation under high pressure.

The comparative evaluation of the materials Alloy 82 and Alloy 182 with respect to the joint of steels P91 + 316 is given in [10, 13]. In general, these materials have similar properties, but researchers give preference to Alloy 182. In particular, it is noted that in the conditions of fatigue thermocycling, the welds of type Alloy 182 showed a lower tendency to failure formation than the welds made by Alloy 82.

In [30] the comparisons of mechanical properties and microstructures of joints P91 + 347 were made, produced by the wire Alloy 82 (TIG), electrodes Rutox-Ast, Rutox-B (MAW) and in a combined method «root Alloy 82 (TIG) + filling Rutox-Ast (MAW)». The best results by mechanical properties (ultimate strength, ductility) were demonstrated by the joint produced by the wire Alloy 82, at the second place — Rutox-Ast (MAW), further — combined method and Rutox-B (MAW) at the end.

In recent years, the use of autogenous methods of welding of dissimilar joints of martensitic and austenitic steels, in particular, TIG welding using activated fluxes being in a suspended state in a solution of a carrier, such as methanol, ethanol or acetone (A-TIG). Typical joints of martensitic and austenitic steels were investigated in connection with this method of welding: P91 + 316L [44, 45], P92 + 304H [36, 46], P91 + 304L [47].

The researchers, who compared A-TIG process and conventional TIG welding with filler wire, note [36] that welds made with autogenous A-TIG welding have increased values of tensile strength and lowered values of impact toughness compared to conventional welding with the wire. The reason for this is that the structure of autogenous weld consists of untempered martensite, whereas when performing multi-pass welding by the wire, additional tempering of the already produced weld with the help of subsequent beads is performed. Due to this fact, the joints produced with the help of A-TIG process require much longer exposure in the postweld heat treatment than the standard joints to acquire the acceptable mechanical properties.

CONCLUSIONS

1. One of the major problems during welding of martensitic and high-alloy austenitic steels is a diffusion reduction of carbon concentration in less alloy metal, contacting with a more alloyed metal containing energetic carbide-forming elements — primarily, chromium, as well as molybdenum, titanium, niobium. The areas depleted with carbon are the zones of possible failure of joints during operation.

2. The main technological measure aimed at reducing carbon depletion of martensitic steel remains traditional nickel-based seam welding. The most common are electrode materials of type Alloy 82, Alloy 182, Alloy 600, Alloy 617 and Alloy 625.

3. It was established that the use of nickel-based welding materials does not completely restrain the carbon redistribution due to their high chromium content (13–24 %), as well as other carbide-forming elements.

4. Highly-effective in restricting carbon depletion of martensitic steel is a new nickel-based electrode of grade EPRI P87 with 9 % of chromium, developed by the Electric Power Research Institute EPRI, USA. It was experimentally shown that under the same thermal conditions of producing joints, the width of depleted carbon interlayer when using electrode P87 will be 18–24 times lower than in welding with the electrodes Alloy 82 and Alloy 617.

5. The redistribution of carbon between the less alloyed martensitic steel and the more alloyed weld is affected not only by the difference in the chemical potential of carbon in steels, caused by different content of carbide-forming elements, as well as a stressed state caused by different value of coefficients of thermal expansion of contacting metals. To reduce the stress level in the fusion zone, there are welding methods with the minimization of gradient of CTE in the joint, which involve welding using a transition insert or by layering of several electrode materials with an intermediate (between α- and γ-steels) coefficient of thermal expansion.

6. Based on studying the structure and mechanical characteristics of joints of dissimilar steels, the following most acceptable welding materials for subsequent combinations of steels were identified: T92 + SS304 — EPRI P87; P91 + SS316 — Alloy 182; P91 + SS347 — Alloy 82.

On the example of the joint T92 + SS304, it was found that as to the efficiency of impact on metal quality in the joint zone, the welding materials can be positioned as follows: T-304H < Alloy 617 ≈ Alloy 82 < Alloy 625 < EPRI P87.

REFERENCES

1. Shibli, A. (2014) *Coal power plant materials and life assessment*. Woodhead Publ.
2. Huysmans, S., Vekeman, J., Hautfenne, C. (2017) Dissimilar metal welds between 9Cr creep strength enhanced ferritic steel and advanced stainless steels–creep rupture test results and microstructural investigations. *Weld World*, **61**, 341–350.
3. Dak, G., Pandey, C. (2020) A critical review on dissimilar welds joint between martensitic and austenitic steel for power plant application. *J. of Manufacturing Processes*, **58**, 377–406.
4. Bhaduri, A. K., Venkadesan, S., Rodriguez, P., Mukunda, P. G. (1994) Transition metal joints for steam generators-An overview. *Int. J. of Pressure Vessels and Piping*, 58(**3**) 251–265.
5. Wu, Q., Xua, Q., Jianga, Y., Gonga, J. (2020) Effect of carbon migration on mechanical properties of dissimilar weld joint. *Engineering Failure Analysis*, **117**, 104935, 1–10.
6. Zhang, W.-Ch., Zhu, M.-L., Wang, K., Xuan, F.-Zh. (2018) Failure mechanisms and design of dissimilar welds of 9 % Cr and CrMoV steels up to very high cycle fatigue regime. *Int. J. of Fatigue*, **113**, 367–376.
7. Yong, J., Zuo, Zh., Jianming, G. (2015) Carbon diffusion and its effect on high temperature creep life of Cr5Mo/A302 dissimilar welded joint. *Acta Metallurgica Sinica*, 51(**4**), 393–399.
8. Frei, J., Alexandrov, B. T., Rethmeier, M. (2019) Low heat input gas metal arc welding for dissimilar metal weld overlays. Pt III: Hydrogen-assisted cracking susceptibility. *Welding in the World*, **63**, 591–598.
9. Ul-Hamid, A., Tawancy, H.M., Abbas, N.M. (2005) Failure of weld joints between carbon steel pipe and 304 stainless steel elbows. *Eng. Failure Analysis*, **12**, 181–191.
10. Sireesha, M., Albert, S. K., Shankar, V., Sundaresan, S. (2000) A comparative evaluation of welding consumables for dissimilar welds between 316LN austenitic stainless steel and Alloy 800. *J. of Nuclear Materials*, **279**, 65–76.
11. Singh, S., Singh, A.B., Kumar, M. et al. (2021) Dissimilar metal welds used in AUSC power plant, fabrication and structural integrity issues. *IOP Conf. Series: Materials Sci. and Eng.*, **1017**, 012022.
12. Karthick, K., Malarvizhi, S., Balasubramanian, V., Gourav Rao, A. (2018) Tensile properties variation across the dissimilar metal weld joint between modified 9Cr–1Mo ferritic steel and 316LN stainless steel at RT and 550 °C. *Metallography, Microstructure and Analysis*, **7**, 209–221.
13. Lee, H.-Y., Lee, S.-H., Kim, J.-B., Lee, J.-H. (2007) Creep–fatigue damage for a structure with dissimilar metal welds of modified 9Cr–1Mo steel and 316L stainless steel. *Int. J. of Fatigue*, **29**, 1868–1879.
14. Mahajan, S., Chhibber, R. (2020) Investigations on dissimilar welding of P91/SS304L using nickel-based electrodes. *Materials and Manufacturing Processes*, 35(**9**), 1010–1023.
15. Sireesha, M., Shankar, V., Albert, S. K., Sundaresan, S. (2000) Microstructural features of dissimilar welds between 316LN austenitic stainless steel and alloy 800. *Mater. Sci. and Engin.*, **A292**, 74–82.
16. Sireesha, M., Albert, S. K., Sundaresan, S. (2002) Thermal cycling of transition joints between modified 9Cr–1Mo steel and Alloy 800 for steam generator application. *Int. J. of Pressure Vessels and Piping*, **79**, 819–827.
17. Sireesha, M., Albert, S. K., Sundaresan, S. (2003) Metallurgical changes and mechanical behavior during high temperature aging of welds between Alloy 800 and 316LN austenitic stainless steel. *Mater. Sci. and Technol.*, 19(**10**), 1411–1417.
18. Sireesha, M., Albert, S. K., Sundaresan, S. (2005) Influence of high-temperature exposure on the microstructure and mechanical properties of dissimilar metal welds between modified 9Cr–1Mo steel and Alloy 800. *Metallurg. and Mater. Transact. A*, **36A**, 1495–1506.
19. Laha, K., Chandravathi, K.S., Parameswaran, P. et al. (2012) A comparison of creep rupture strength of ferritic/austenitic dissimilar weld joints of different grades of Cr–Mo ferritic steels. *Metallurg. and Mater. Transact. A*, **43A**, 1174–1186.
20. Brentrup, G.J., Snowden, B.S., DuPont, J.N., Grenestedt, J.L. (2012) Design considerations of graded transition joints for welding dissimilar alloys. *Welding J.*, 91, 252–259.
21. Akram, J., Kalvala, P.R., Misra, M., Charit, I. (2017) Creep behavior of dissimilar metal weld joints between P91 and AISI 304. *Mater. Sci. and Engin. A*, **688**, 396–406.
22. Akram, J., Kalvala, P.R., Chalavadi, P., Misra, M. (2018) Dissimilar metal weld joints of P91/Ni alloy: Microstructural characterization of HAZ of P91 and stress analysis at the weld interfaces. *J. Mater. Engin. and Performance*, 27(**8**), 4115–4128.
23. Siefert, J.A., Tanzosh, J.M., Shingledecker, J.P. (2010) Weldability of EPRI P87. In: *Proc. of the 6th Int. Conf. on Advances in Materials Technology for Fossil Power Plants (August 31–September 3, 2010, Santa Fe, New Mexico, USA*), 995–1013.
24. Siefert, J.A., Tanzosh, J.M., Shingledecker, J.P. et al. (2011) EPRI P87: A promising new filler metal for dissimilar metal welding. *Welding J.*, 90(**3**), 30–34.
25. Anand, R., Sudha, C., Karthikeyan, T. et al. (2009) Effectiveness of Ni-based diffusion barriers in preventing hard zone formation in ferritic steel joints. *J. of Materials Sci.*, **44**, 257–265.
26. Urzynicok, M., Jachym, R., Kwiecinski, K. et al. (2013) Application of EPRI87 in dissimilar welding austenitic-martensitic welded joints of TEMPALOY AA-1 and T92 steel grades. In: *Proc. of 7th Int. Conf. on Advances in Materials Technology for Fossil Power Plants (October 22–25, 2013 Waikoloa, Hawaii, USA*), 992–1005.

27. Coleman, K., Gandy, D. (2007) Alternative filler materials for DMWs involving P91 materials. In: *Proc. of 5th Int. Conf. on Advances in Materials Technology for Fossil Power Plants (October 3–5, 2007, Marco Island, Florida, USA*), 940–967.
28. Sirohi, S., Pandey, C., Goyal, A. (2021) Role of the Ni-based filler (IN625) and heat-treatment on the mechanical performance of the GTA welded dissimilar joint of P91 and SS304H steel. *J. of Manufacturing Processes*, **65**, 174–189.
29. Thakare, J.G., Pandey, C., Mahapatra, M.M., Mulik, R.S. (2019) An assessment for mechanical and microstructure behavior of dissimilar material welded joint between nuclear grade martensitic P91 and austenitic SS304 L steel. *J. of Manufacturing Processes*, **48**, 249–259.
30. Mittal, R., Sidhu, B.S. (2015) Microstructures and mechanical properties of dissimilar T91/347H steel weldments. *J. Materials Proc. Technology*, **220**, 76–86.
31. Senthil, T.S., Siva Kumar, G. (2015) Dissimilar steel welding of super heater coils for power boiler applications. *American J. of Materials Research*, 2(**5**), 44–49.
32. Arunkumar, N., Duraisamy, P., Veeramanikandan, S. (2012) Evaluation of mechanical properties of dissimilar metal tube welded joints using inert gas welding. *Int. J. of Eng. Research and Applications*, 2(**5**), 1709–1717.
33. Liang, Z., Gui, Y., Zhao, Q. (2018) Investigation of microstructures and mechanical properties of T92 martensitic steel/Super304 austenitic steel weld joints made with three welding consumables. *Archives of Metallurgy and Materials*, 63(**3**), 1249–1256.
34. Cao, J., Gong, Y., Zhu, K. et al. (2011) Microstructure and mechanical properties of dissimilar materials joints between T92 martensitic and S304H austenitic steels. *Materials and Design*, **32**, 2763–2770.
35. Chen, G., Zhang, Q., Liu, J. et al. (2013) Microstructures and mechanical properties of T92/Super304H dissimilar steel weld joints after high-temperature ageing. *Materials and Design*, **44**, 469–475.
36. Sharma, P., Kumar Dwivedi, D. (2019) Comparative study of activated flux-GTAW and multipass-GTAW dissimilar P92 steel-304H ASS joints. *Materials and Manufacturing Processes*, 34(**11**), 1195–1204.
37. Xu, L., Wang, Y., Jing, H. et al. (2016) Deformation Mechanism and Microstructure Evolution of T92/S30432 Dissimilar Welded Joint During Creep. *J. Mater. Eng. and Performance*, **25**, 3960–3971.
38. Liang, Z., Zhao, Q., Deng, J., Wang, Y. (2017) Influence of aging treatment on the microstructure and mechanical properties of T92/Super 304H dissimilar metal welds. *Materials at High Temperatures*, 35(**2**), 1–8.
39. Čiripová, L., Falat, L., Ševc, P. et al. (2018) Ageing effects on room-temperature tensile properties and fracture behavior of quenched and tempered T92/TP316H dissimilar welded joints with Ni-based weld metal. *Metals*, **8**, 791–806.
40. Falat, L., Výrostková, A., Svoboda, M., Milkovič, O. (2011) The influence of PWHT regime on microstructure and creep rupture behaviour of dissimilar T92/TP316H ferritic/austenitic welded joints with Ni-based filler metal. *Kovove Materialy*, 49(**6**), 417–426.
41. Falat, L., Čiripová, L., Kepič, J. et al. (2014) Correlation between microstructure and creep performance of martensitic/austenitic transition weldment in dependence of its post-weld heat treatment. *Eng. Failure Analysis,* **40**, 141–152.
42. Falat, L., Kepič, J., Čiripová, L. et al. (2016) The effects of postweld heat treatment and isothermal aging on T92 steel heat-affected zone mechanical properties of T92/TP316H dissimilar weldments. *J. of Materials Research*, 31(**10**), 1532–1543.
43. Cao, J., Gong, Y., Yang, Z.G. (2011) Microstructural analysis on creep properties of dissimilar materials joints between T92 martensitic and HR3C austenitic steels. *Mater. Sci. and Engin. A*, **528**, 6103–6111.
44. Fei, Z., Pan, Z., Cuiuri, D. et al. (2020) Effect of post-weld heat treatment on microstructure and mechanical properties of deep penetration autogenous TIG-welded dissimilar joint between creep strength enhanced ferritic steel and austenitic stainless steel. *The Int. J. of Advanced Manufacturing Technology*, **108**, 3207–3229.
45. Vidyarthy, R.S., Kulkarni, A., Dwivedi, D.K. (2017) Study of microstructure and mechanical property relationships of A-TIG welded P91–316L dissimilar steel joint. *Mater. Sci. and Engin. A*, **695**, 249–257.
46. Sharma, P., Kumar Dwivedi, D. (2019) A-TIG welding of dissimilar P92 steel and 304H austenitic stainless steel: Mechanisms, microstructure and mechanical properties. *J. of Manufacturing Processes*, **44**, 166–178.
47. Thakare, J.G., Pandey, C., Gupta, A. et al. (2021) Role of the heterogeneity in microstructure on the mechanical performance of the Autogenous Gas Tungsten Arc (GTA) welded dissimilar joint of F/M P91 and SS304L steel. *Fusion Eng. and Design*, **168**, 112616, 1–13.

ORCID

M.O. Nimko: 0000-0002-9672-4921,
V.Yu. Skulskyi: 0000-0002-4766-5355

CONFLICT OF INTEREST

The Authors declare no conflict of interest

CORRESPONDING AUTHOR

V.Yu. Skulskyi
E.O. Paton Electric Welding Institute of the NASU
11 Kazymyr Malevych Str., 03150, Kyiv, Ukraine.
E-mail: vsku@paton.kiev.ua

SUGGESTED CITATION

M.O. Nimko, V.Yu. Skulskyi, T.V. Ivanenko (2023) Materials and technological approaches to welding of combined joints between martensitic and austenitic steels used in power machine building (Review). *The Paton Welding J*., **4**, 11–20.

JOURNAL HOME PAGE

https://patonpublishinghouse.com/eng/journals/tpwj

Received: 12.04.2023
Accepted: 25.05.2023