Technology and equipment for underwater wet welding and cutting of high strength steel arctic structures using flux-cored wires

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Abstract

The paper describes the experimental findings of underwater wet welding and cutting of API X70 pipe steel and plates with the use of the following flux-cored wires: TiO₂-CaF₂, Fe₂O₃-TiO₂ and Fe₂O₃-TiO₂-SiO₂. For cutting the cored wire with a diameter of 2 mm based on salts of iron and alkali metals was developed. The best weld properties have been achieved during the underwater welding with fluoride flux-cored wires with TiO₂-CaF₂ system, which provides the suitable microstructure of the weld, high solidity of deposited metal and mechanical properties of the weld. The use of flux-cored wire improves a quality of underwater wet cutting of steel plates with thickness up to 8 mm.

1. Introduction

1.1. General information
Currently, research is being actively conducted in the field of underwater welding to support the developing of the Arctic and the Arctic shelf [1]. To produce reliable underwater welded structures, it is necessary to ensure a high level of strength, ductility and impact energy of weld [2, 3]. Underwater wet welding is complicated due to the low stability of the welding arc, porosity, slag inclusions, cracks and hydrogen embrittlement [4-6].

1.2. Purpose of research
The purpose of research is the development of the technology and equipment for underwater wet welding and cutting using flux-cored wires with a high content of alkaline and alkaline-earth metal fluoride compounds for underwater repair welding of pipe steel of API X70 strength.

1.3. Research methodology
For the underwater wet welding of butt and lap joints we have used the samples of pipe steel for gas pipelines with API X70 strength class of size 300×200×21,3 mm. The divers-welders have performed the multilayer flux-cored arc welding (FCAW) at a depth of 14 m with wires of PPS-APL1 (system TiO₂-Fe₂O₃), PPS-APL2 (system TiO₂-CaF₂), PPS-APL3 (system TiO₂-Fe₂O₃-SiO₂) with the diameter of 1.6 mm. The filling factor of the wire was 30 %, each wire contained up to 5 % Na₃AlF₆, 10 % FeMn, 5 % Ni. According to the oscillograms, arc voltage during welding was 38-44 V, current 170-270 A, and wire feed rate 4 m/min.
For mechanical tests according to Russian standard GOST 6996-66 was used the testing machine "Super L60", impact testing machine PH450, hardness tester EMOTEST DuraScan-20, the chemical composition was determined by the optical emission spectrometer Bruker Q4 TASMAN.
The study of microstructure was performed using the microscope Zeiss Axiovert 200 MAT with automatic image analyzer ThioxmetPro. Micro-hardness was measured by micro-hardness tester Buehler Micromet 6040.

2. Reduction of hydrogen in underwater wet welding

The atmosphere of a vapor-gas bubble in underwater welding consists of 62–82\% H$_2$, 11–24\% CO, 4–6\% CO$_2$, metal vapors and salts [7, 8], fig. 1. According to other sources, the hydrogen content in a bubble reaches 85–96\% [9, 10]. Analysis of gases in welding with coated electrodes and flux-cored rutile wire showed that the hydrogen content in the vapor-gas bubble is 93–98\%, carbon monoxide 1,5–6\%, carbon dioxide 0,5–2\% [11–13].

![Fig. 1. Growth of a vapor-gas bubble during underwater wet welding with PPS-APL2 cored wire: 1 - cored wire; 2 - vapor-gas bubble; 3 - steel plate. Shadow video recording with a Fhantom camera with a frequency of 8000 Hz. Pause between frames 1,75 ms](image)

The composition of the gas contained in the pores during underwater wet welding according to [14] has up to 96\% H$_2$ with a small amount of carbon monoxide. According to the data of [15], pores consist of 62–82\% H$_2$, 11–24\% CO, 4–6\% CO$_2$, depending on the electrode coating composition and welding regime. Thermodynamic calculations show a high probability of metallurgical reactions in the vapor-gas bubble [16].

One of the mechanisms for reducing the concentration of hydrogen in weld is to reduce the partial pressure of hydrogen in the atmosphere of a vapor-gas bubble, since the hydrogen solubility in iron corresponds to the known Siverts law. To reduce the porosity in [18, 19], it was proposed to introduce CaCO$_3$, MgCO$_3$ and CaF$_2$. During dissociation, carbonates decompose in a vapor-gas bubble with the formation of CO$_2$, CO, which reduce the partial pressure of hydrogen above the weld pool. With an
increase in CaCO$_3$ from 9 to 12.5%, the volume of porosity decreased from 2.2 to 1% at a depth of 9 m [20].

The second mechanism of hydrogen reduction is the chemical fixing of hydrogen to the HF compound [12, 21-24]. In [12], it was found that with an increase in the concentration of CaF$_2$ from 0 to 86%, the hydrogen concentration in the weld decreases linearly and CaF$_2$ reduces the diffusible [H] in the weld in relation to CaCO$_3$. With 20% CaCO$_3$ in the in electrode coating the diffusible hydrogen concentration is [H] = 54 ml/100 g, and with the introduction of 20% CaF$_2$, the value of [H] decreases by 26%. Similar results were achieved in [9] with an increase CaF$_2$ content in the electrode coating from 0 to 70%, the concentration of diffusible hydrogen decreased from 48 to 30 ml/100 g.

The third metallurgical mechanism for reducing hydrogen is to increase the oxidation of the weld pool and increase the basicity of the slag with the introduction of hematite Fe$_2$O$_3$ with a density 5.3 g/cm$^3$. The content of diffusible hydrogen was reduced from 70 to 30 ml/100 g with the introduction of 35% hematite into the electrode coating [25]. As is known, hematite Fe$_2$O$_3$ transforms into wustite FeO in the slag, which increases the basicity of the slag, besides wustite oxidizes the weld pool and decreases the solubility of the diffusible [H].

The diffusible hydrogen concentration reducing mechanism through saturation of the weld pool with oxygen [26, 27] is also implemented in acid coating electrodes. The increasing in (FeO) in the slag-system leads to the diffusible hydrogen concentration decrease in the weld metal however a growth in the oxygen [O] [28]. However, an increase in the oxidation potential of the slag and the atmosphere of the vapor-gas bubble lead to a growth in the volume of non-metallic inclusions and burnout of alloying elements, which reduces the mechanical properties of the weld [29].

The hydrogen adsorption by weld decreases with an increase in the solubility of water vapor in slags with increasing basicity coefficient [30, 31]. With increasing slag basicity coefficient from 0.4 to 0.5 at the rutile coating, the diffusible [H] in the weld decreases from 41 to 23 ml/100 g [32]. According to [33], with growth of index basicity from 0.52 to 0.75, the diffusible [H] reduces from 48 to 39 ml/100 g. With increasing depth and pressure, iron oxidation is intensified, and the content of wustite and slag basicity increase [6], however, it likewise increases the oxidation of the weld pool, what leads to slag inclusions and a sharp reduction of mechanical properties.

Simultaneous implementation of the three hydrogen reduction mechanisms is possible when creating a low-density slag system based on CaF$_2$ with Na$_3$AlF$_6$. The introduction of CaF$_2$ increases the basicity of the slag and the solubility of H$_2$O vapor in the slag. It is known that water vapor is dissolved in acid slags and in basic slags according to the corresponding reactions [34]:

\[
(Si-O-Si^-) + H_2O = 2 (Si-OH) \quad (1)
\]
\[
(O^-) + H_2O = 2 (OH^-) \quad (2)
\]

In molten fluoride slags, an ionic reaction is possible on hydrogen binding with the formation of OH$^-$ anions [35]:

\[
H_2O + F^-_{\text{slag}} = 2 OH^-_{\text{slag}} + HF^\uparrow \quad (3)
\]

Accordingly, the formation of OH$^-$ anions and the binding of hydrogen to HF can energetically complicate the transition of atomic hydrogen to the weld pool according to the ion equation [35]:

\[
2OH^-_{\text{slag}} + Fe = Fe^{2+}_{\text{slag}} + 2O^{2-}_{\text{slag}} + 2[H] \quad (4)
\]

For modelling the slag system, the ratio of the slag-forming components of the systems Fe$_2$O$_3$-TiO$_2$ and TiO$_2$-CaF$_2$ was chosen for the minimal melting temperature, viscosity and surface tension to improve the drop transfer and the formation of a welds, fig. 2. The TiO$_2$-CaF$_2$ ratio also took into account slag homogeneity and the absence of its separation into two liquid phases [40, 41].
The TiO$_2$-CaF$_2$ slag system consists of TiO$_2$ (density - 4.2 g/cm$^3$), CaF$_2$ (density - 2.5 g/cm$^3$) with surface tension of about 240 mJ/m$^2$ [36] for protection of the weld pool from water penetration. Wetting of the surface of the weld pool is promoted by the low viscosity of the TiO$_2$–CaF$_2$-Na$_3$AlF$_6$ slag-system 0.08 Ns/m$^2$ [36]. The introduction of Na$_3$AlF$_6$ into the wire intensifies the reactions of hydrogen fixing with the formation of hydrogen fluoride HF. When heated in the welding arc complex fluoride Na$_3$AlF$_6$ dissociates according to the reaction:

$$Na_3AlF_6 = 2NaF + NaAlF_4 \ (5)$$

NaAlF$_4$ has a surface tension 86.6 mJ/m$^2$, concentrates in the slag surface layer and reduces the interfacial tension of the molten metal-slag system [42], which improves the drop transfer. The evaporation and dissociation of Na$_3$AlF$_6$ leads to the formation of NaF, AlF$_3$, AlF$_2$, AlF and reduces the partial pressure of H$_2$ in the vapor-gas bubble. Metallurgical reactions of NaF, AlF$_3$, AlF$_2$, AlF compounds with titanium dioxide TiO$_2$ occur with the formation of titanium fluorides TiF$_4$, TiF$_3$, TiF$_2$ with high activity in HF formation reactions at a temperature T = 1000–5000 K, which leads to enrichment of the vapor-gas bubble by fluorides.

3. Experimental research results

Thermodynamic modeling using the Terra program [43] confirms that the introduction of 20% Na$_3$AlF$_6$ vapor into a plasma results in a decrease in H and H$_2$ with formation of HF depending on the pressure of the medium, fig. 3.
Fig. 3. Changing the mass fraction of atomic H, molecular H$_2$ hydrogen and fluoride HF in equilibrium system at dissociation of H$_2$O: 1 - no fluoride; 2 - with the introduction of 20% Na$_3$AlF$_6$ into the system. a - pressure in the system 0,1 MPa, b - 0,6 MPa.

Oscillography of welding currents and voltages in underwater wet welding showed a change in the volume of the vapor-gas bubble, the welding current deviations reach ± 70–90 A, fig. 4.

As the fluoride content increases, the frequency and amplitude of the low-frequency pulsations of the welding current and voltage decrease. The best results in welding-technological properties have been showed for the wire of PPS-APL2 type which had the deviations of the welding current from the arithmetic mean value ±70 A.

Results of visual and radiographic inspection showed that the number of weld slag inclusions and lack of fusion increases with increasing the Fe$_2$O$_3$ content in flux-cored wire. The tensile strength of welded joints was the following during the welding with the wire of PPS-APL1: 439-450 MPa; PPS-APL2: 433–462 MPa; PPS-APL3: 412–421 MPa. Relative elongation of weld metal according to the cylindrical samples reached 9-16 %, the yield strength was up to 322-356 MPa. The study of impact strength at Charpy specimens has showed that the maximum value of impact energy of the weld is achieved by using the flux-cored wire of PPS-APL2 with system TiO$_2$-CaF$_2$ without hematite Fe$_2$O$_3$: $KCV_{+20}$ value reaches up to 52-60 J and $KCV_{-20}$ was up to 37-44 J, that meets the requirements of ANSI/AWS D3.6M:2010, Fig. 5.
The hardness value of the deposited metal of the weld performed by wire PPS-APL1 was 131-160 HV$_{10}$; PPS-APL2: 162-200 HV$_{10}$; PPS-APL3: 159-175 HV$_{10}$. The hardness value of the deposited metal of the weld of the lap joints performed by the wire: PPS-APL1 was: 142-183 HV; PPS-APL2: 138-182 HV$_{10}$; PPS-APL3: 149-170 HV$_{10}$. The width of the heat affected zone was up to 2-3 mm. Thus, the average hardness of butt welds was less than for 10-15 HV$_{10}$ in comparison with the hardness of fillet welds. The average hardness of heat affected zone of the butt welds was more for 15-40 HV$_{10}$ in comparison with the hardness of heat affected zone of fillet joints.

The results of chemical analysis of the weld have shown that in multi-layer welding of butt joints the alloying elements content decreases from the root layer to the cap bead. The greatest change is in the content of manganese, silicon and carbon. When welding with different flux-cored wires the manganese content from root weld to the cap bead is 2 to 3.6 times reduced. The minimum 2 time reduction of manganese from the root weld to the cap bead is achieved during welding by the wire of PPS-APL2. At the same time, the average content of alloying elements in the weld metal when welding with the wire of PPS-APL2 type was, %: C – 0.1; Si – 0.05; Mn – 0.4; Ni – 1.1; P – 0.01; S – 0.008.

The basic microstructure of the samples consisted of ferrite with a secondary phase, primary, polygonal and a small amount of acicular ferrite with micro-hardness of 150-190 HV$_{10}$. Figs. 6 and 7. The microstructure contains the nonmetallic inclusions which have small size and are uniformly distributed over the cross section. Microstructure of heat affected zone consists of bainite when the value of micro-hardness is 280-330 HV$_{10}$.
Fig. 6. Lap-welded joint (a), macrostructure of the weld (b) and microstructure of the weld interface (c) of X70 API pipe steel with a thickness of 21.3 mm after wet underwater welding with PPS-APL2 cored wire.

![Fig. 6](image)

Fig. 7. The microstructure of the primary ferrite (a) and the microstructure of the ferrite with the secondary phase (b) with nonmetallic inclusions during wet welding with PPS-APL2 wire.

Modeling of fluoride systems allowed us to develop a flux-cored wire PPR-APL1 with a diameter of 2.0 mm, type PS D-ARN according to Russian standard GOST 26271-84 (LLC "Scientific and Technical Center" Svarka", St. Petersburg) for underwater wet cutting of carbon and alloyed steels with a thickness of up to 20 mm at depths up to 60 m. The slag system of this wire contains up to 20% fluoride, which stabilizes the welding arc in cutting cycles in water with high cutting quality.

![Fig. 8](image)

Fig. 8. A typical oscillogram of current and voltage during underwater cutting with flux-cored wire PPR-APL1 with a diameter 2 mm.
Fig. 9. Face (a) and bottom (b) sides of the plate with a thickness of 8 mm after cutting with PPR-APL1 wire at wire feed rates 6; 7; 8; 9 m/min. Arc voltage 31-35 V, current 370-430 A

Fig. 10. Complex KOPS-M for underwater wet mechanized welding and cutting

Conclusions

Increasing the metallurgical activity and basicity of slag to reduce the concentration of hydrogen can be achieved without the use of hematite Fe₂O₃ due to the introduction of CaF₂ together with Na₃AlF₆ into rutile cored wire in an amount of at least 5-12% weight of the powder. The saturation of the vapor-bubble atmosphere with fluorides and an increase in the basicity of the slag improves the stability of the arc, reduces the porosity and the volume of nonmetallic inclusions. The best quality, microstructure of the welds and mechanical properties of the welds were achieved in the underwater wet welding and cutting using of wires TiO₂-CaF₂-Na₃AlF₆ slag system with high metallurgical activity in relation to hydrogen.

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