DIFFUSION WELDING OF TITANIUM ALLOYS OF DIFFERENT TYPES

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I. Introduction to the Problem

The application scope of titanium welded constructions in various fields of techniques is steadily widening [1]. A high level and a variety of requirements of such constructions make it efficient and economically more effective to apply chemically heterogeneous alloys in the above constructions. The problem is to obtain permanent welded joints from titanium alloys of different types. Presently there are more than 30 types of titanium alloys in the USSR. Traditionally different types of fusion welding were used because of a good scientific base and adequate application experience [2]. The diffusion welding application is more perspective in cases of welding of long and sandwich constructions of different types [3].

The problems of diffusion welding of titanium alloys in similar combinations (SC) are described in many publications [4] and others. Publications on the problems of diffusion welding of different combinations (DC) are extremely limited [5]. The available information on this problem is referred to the research of weldability of some different combinations of titanium alloys and does not resolve the complex problem. At the same time titanium alloys differ not only by a system but also by a degree of alloying.

Therefore different DC of titanium alloys will differ in character and degree of chemical heterogenuity in a contact area. That will result in substantial differences in their weldability. The lack of information on these problems does not permit to predict the weldability of the variety of DC of titanium alloys. As a rule, the attempts of applying usual technological schemes do not lead to positive results, that is illustrated by the following experiment.

Cylindrical specimens (in diameter 16x30 mm) with polished (R_Z = 0,05) contact surfaces were welded in vacuum 10^{-2} mm of a mercury column at T_W = 900°C. The joint area was screened by getter, the welding pressure was applied at the room temperature and ensured the same for SC and DC deformation rate $\varepsilon = 2{,}0 \cdot 10^{-5} s^{-1}$.

The experimental data point to the fact that while time of welding t_W is increased the KCV_{SC} values are gradually rising up with obtaining KCV_{SC} = 1,0 at ε = 8,0-10,0%. At first there is some ascent and later on a descent of KCV_{DC} (t_W) curve (fig. 1). However it was impossible to obtain a combination adequate in strength to the base metal even at ε = 16,0%. The metallographic analysis showed that the SC joint area did not practically differ from the base metal (fig. 2a) and that the DCs had a strongly developed porosity (fig. 2b).

II. Theoretical Analysis of Weldability

In case of diffusion welding the possibility of obtaining qualitative joints of chemically heterogeneous metals firstly depends on characteristics of the welding final phase - the volumetric interaction phase [3,4].

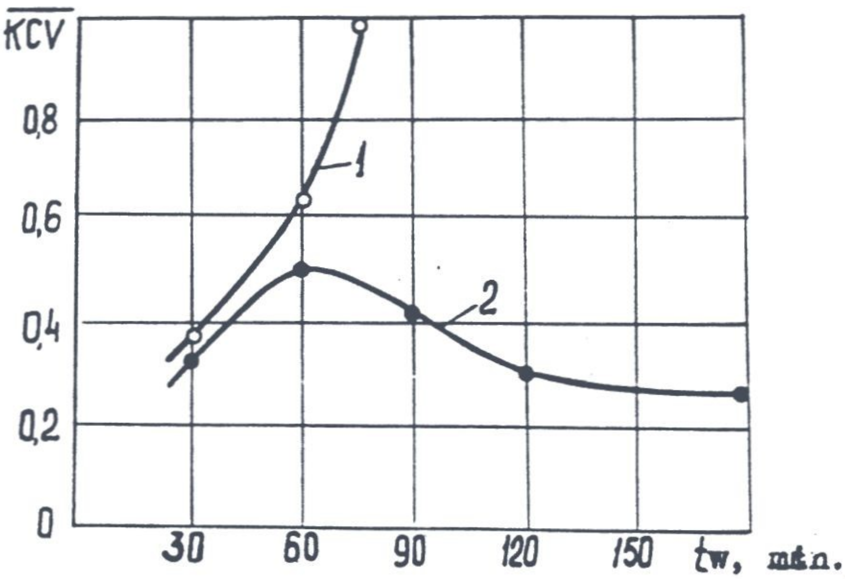

Fig. 1

Dependence KCV(t_W) for combinations OT4-1 + OT4-1 (1) and OT4-1 + BT3-1 (2)

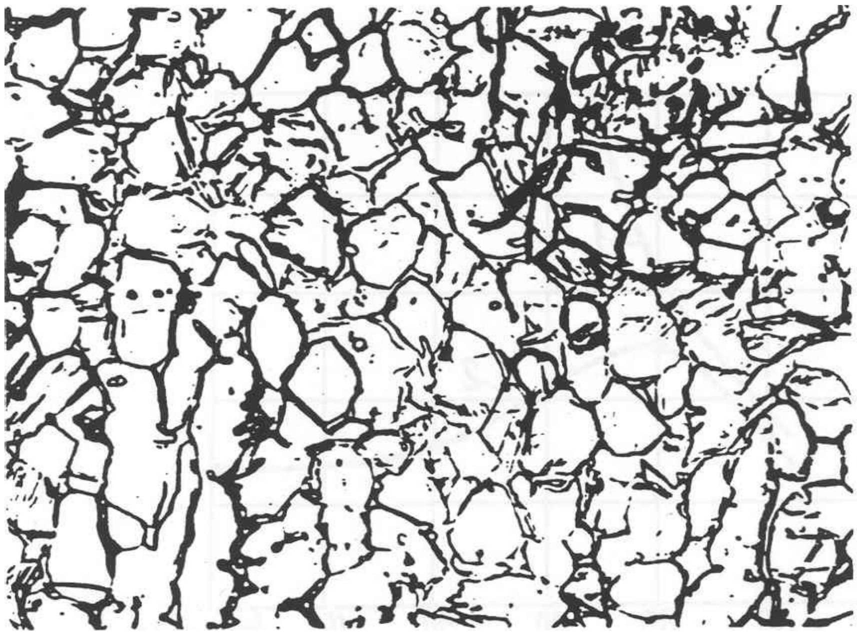

Fig. 2a

Structure of welded joint area of SC (a) and DC (b) after welding at T_W = 900 °C, t_W = 75 min.

Fig. 2b

Structure of welded joint area of SC (a) and DC (b) after welding at T_W = 900 °C, t_W = 75 min.

Therefore in this publication the evaluation of DC titanium alloy weldability was made due to special features of realization of just this phase. A well-known technological scheme [6] more fully gains the above aim. Due to the scheme a physical contact was previously formed at not very high temperature. It was followed by further isothermal annealing at standard welding temperature rate. In order to prevent polymorphism influence the research was carried out at OT4 and BT14 alloys having the same $\alpha+\beta$ transformation temperature. Specimens-immitators of rought surface 1 microprojectors (fig. 3) were previously scalded by hermetic welds 2 in titanium capsules 3 in a pure argon atmosphere in order to prevent oxidation process.

A physical contact was made at T_{ph} = 625...650°C. afterwards the specimens were held without pressure for different time t_{an} at different annealing temperatures T_{an} in $\alpha+\beta$ - area. The welding operation was conducted at radiant heating installation [4]. The quality of welded joints was determined by comparative impact toughness

$$KCV = \frac{KCV_W}{KCV}$$

where: KCV - impact toughness of a tougher welded joint,

KCV_W - welded joint impact toughness.

At fig. 4 dependencies KCV (t_{an}) show that there is a monotonous increase of KCV values at all T_{an}. There is quite a different dependence for DC: while t_{an} is increased, an interval of intensive KCV increase is transforming into perceptible decrease. An interval of KCV increase is more intensive for DC than SC, that is $KCV_{DC} > KCV_{SC}$, one can see an opposite regularity during a decrease interval. precisely $KCV_{SC} > KCV_{DC}$. Consequently at first one can observe an effect of DC improved weldability in comparison with SC, later on it is changed by a loss of strength effect.

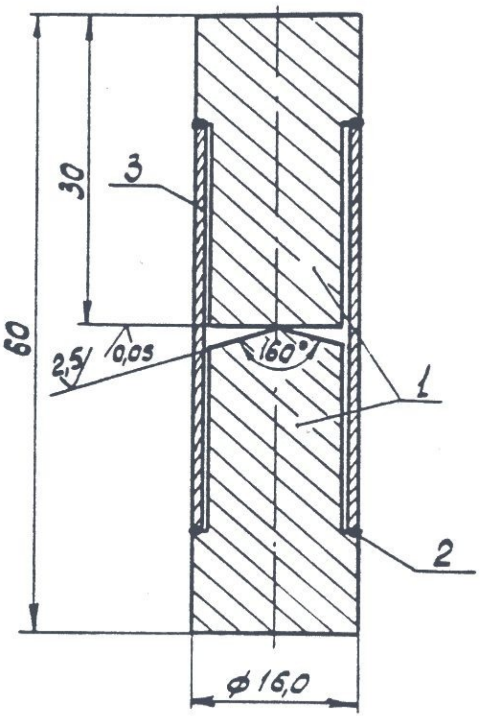

Fig. 3

Welded specimens

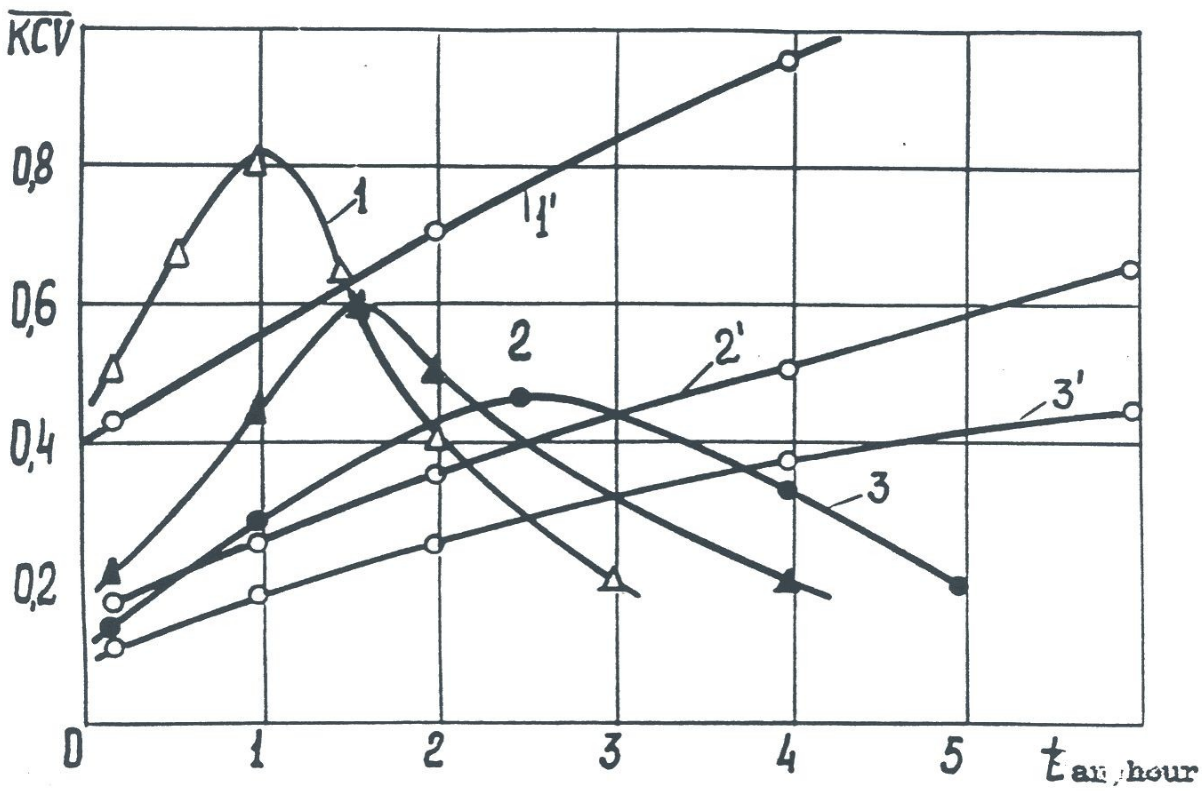

Fig. 4

Dependence KCV (t_{an}) for combinations OT4+OT4 (1', 2', 3') and OT4+BT4 (1, 2, 3) at T_{an} °C: 3,3' - 850, 2,2' - 900, 1,1' - 940.

Thermoactivation analysis of KCV (t_{an}) was carried out by plotting functions on 1g coordinates KCV_{SC} - 1/T (where KCV_{SC} is a comparative impact toughness rate). The analysis has determined an effective energy of $E_{ef.}$ activation process controlling the impact toughness growth of SC welded joints (fig. 5). The determine value $E_{ef.}$ = 102 kJ/mol makes ~0.4 from known [7] value of volumetric self-diffusion activation energy (E_s = 251 kJ/mol) in α-titanium, that corresponds to energy of vacancies migration activation [8].

A thermoactivation analysis or a period of an intensive increase of KCV_{DC} (t_{an}) dependencies determined that $E_{ef.}$ = 96 kJ/mol values were similar to the corresponding $E_{ef.}$ values for SC. Consequently despite the differences in kinetic curves character an impact toughness growth is controlled in both cases by the same process connected with vacancies migration in the are near the contact. $E_{ef.}$ values prove it. However, in studied cases the vacancies flows dynamics is not identical, this process is connected with titanium and i ts alloys diffusion anomality. One aspect of the anomality problem is connected with the fact that the diffusion coefficients of many alloyed elements considerably exceed the coefficient of titanium volumetric self-diffusion [9]. In this case parameters of self- and heterodiffusion of titanium alloys essentially depend on a system and a degree of their alloying [9]. Therefore diffusion mobility of atoms of DC titanium alloys will greatly differ from SC titanium alloys and when putting them into contact there'll be an uncompensated vacancies flow to the direction opposite to the masstransference. Probably this difference results in intensive development of DC volumetric interaction at the initial stages of welding. Furthermore it is known that the diffusion process in metals is followed by the dislocations formation in the diffusion zone[10].

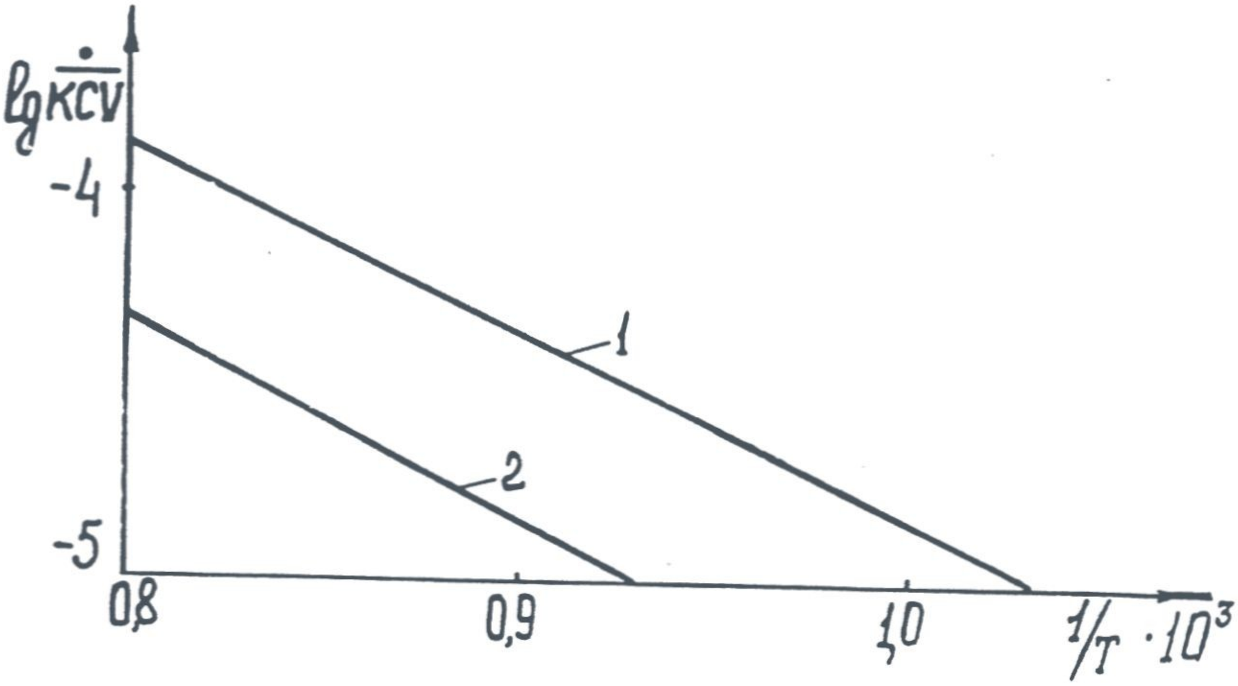

Fig. 5

Dependence KCV - 1/T for DC (1) and SC (2)

Therefore with low t_{an} when maximum gradient of concentraions is in the welded joint surface, dislocations will take place directly near the contact surfaces activizing them additionally and making adhesion more intensive. Consequently the difference chemical composition of DC of titanium alloys with low t_{an} may be considered as an activizing factor which increases the density of interaction zones on the contact surface and intensifies the development of all phases of volumetric interaction stage. Probably this factor causes more intensive KCV_{DC} increase at the initial stages of annealing (fig. 4).

The fact of deterioration of properties of heterogeneous metals welded joints under long thermal influence is usually connected with formation of brittle intermetallides [11]. However as a rule, systems of alloying of titanium elements prevent from intermetallides formation on heating more than temperature of eutectoid decay [12]. Measuring of microhardness of welded joint area of specimens (fig. 1) did not also reveal any brittle inclusions. Probably the KCV_{DC} decrease (fig. 4) was caused by further development of the vacancies flows dynamics. It is known [13] that the existence of uncompensated vacancies flow according to Frenkel effect results in the formation of pores in the diffusion zone. Therefore one can consider that during the process of isothermal holding surplus vacancies condense on micro-areas of a welded joint with an unformed physical contact and promote their growth.

Consequently experimental data (fig. 4) may be interpreted in the following way; due to the low reaction activity depending on the terms of experiment [14], the improved weldability effect is not completely realized, and the loss of strength effect becomes predominant.

To one or another degree determined features of the volumetric interaction stage are realized in different terms of welding of the most DC. In this case essential differences are made during the realization of other stages of the process including a stage of a physical contact formation. This influences both the special features of a welded joint formation on the whole and its mechanical properties. Therefore to resolve the problem of DC titanium alloys weldability it is necessary to study the heteroalloying influence on characteristics of a physical contact formation and quality of welded joints. This will help to optimize a technological scheme of welding of different DC of titanium alloys.

III. Special Features of Formation of a Physical Contact

In case of diffusion welding formation of a physical contact is a necessary condition of obtaining qualitative welded joints. As far as DC of titanium alloys are concerned formation of a physical contact was studied at alloys mentioned in table I.

Table 1

Nos.	Type of alloy	Middle Chemical Composition	Total content of alloyed elements C, %	β-stabilisators content, C_β, %	T°C $\alpha+\beta-\beta$ transformation	Middle size of grain, d mcm	KCV, J/cm
1	2	3	4	5	6	7	8
1.	BT1-0	unalloyed	-	-	890	5,0	150,
2.	OT4-1	1,5Al-1,0 Mn	2,5	1,0	910	6,0...8,0	76,
3.	OT4	3,5Al-1,5 Mn	5,0	1,5	950	4,0...6,0	70,

1	2	3	4	5	6	7	8
4.	BT5	5,0 Al	5,0	-	960	4,0...6,0	45
5.	BT14	4,5Al-3,0Mo-1,0V	8,5	4,0	950	3,0...5,0	60
6.	BT3-1	6,0Al-2,5Mo-2,0Cr-0,3Si-0,5 Fe	11,3	4,5	970	3,0...5,0	50
7.	BT16	2,5Al-5,0Mo-5,0 V	12,6	10,0	870	5,0...8,0	40

Welding operation of cylindrical specimens 16x30 mm in diameter in SC and DC was conducted at radiation heating installation [4].

One of the welded surfaces was polished (R_Z = 0,05 mcm), the other was turned (R_Z = 10,0 - 500 mcm) or finished (R_Z = 1,5 mcm). The specimens were welded in vacuum 10^{-2} mc of a mercury column for different time t_W, and the butt was additionally screened by a getter. Relative area of a physical contact

$$F = \frac{F_c}{F_n}$$

(where F_c - a physical contact area; F_n - nominal area of contacting) was valued by factogramms of previously polished surface and methods of profilographing and optical metallography were used. Profilograph-profilometer of 252 model and MBS-7, MIM-8, "Neofot" optical microscopes were used for that purpose. Relative macrodeformation ε_H was valued by total shortening of both welded specimens. When SC and DC were tested, one of the welded in SC specimen was previously annealed in $\alpha+\beta$ -area, its initial microstructure was purposefully changed. In this case the equality of rates of high-temperature creep of compared combinations was ensured.

Near the top limit of $\alpha+\beta$ -area temperature is optimal in case of welding of many titanium alloys [4]. Therefore welding of

OT4-1 and VT3-1 (R_Z 0,05 + R_Z 10,0) alloys was conducted at T_W = 900°C.

Experimental data show (see fig. 6a) that a physical contact of DC with all t_W and ε' is formed more slowly than of SC, that is the following inequality takes place:

$$F'_{DC} < F'_{SC} \qquad (1)$$

where F'_{DC} and F'_{SC} -middle rates of physical contact formation of DC and SC correspondingly.

The initial size of grain of welded alloys (table 1) approximately ensures their equal rate of high temperature deformation, and thus excludes a formation of mechanical heterogeneity in the contact area. Therefore it' s impossible to explain inequality (1) by difference in microplastic deformation character in the joint area.

In this case the influence of other processes may become important. The above processes are responsible for a formation of a physical contact in case of titanium alloys welding. One of the most important of the above processes is connected with a formation of high-temperature surface microrelief [4].

Actually in α+β -area the intensity of this process depends on chemical and phase composition of an alloy, the process is more active near the point A_{c3} [15]. That's why in the analyzed case (fig. 6) formation of microrelief of BT3-1 alloy occurs slower than of OT4-1, because point A_{c3} of the first alloy is essentially lower than welding temperature. In case of the above alloys welding total contribution of formed microrelieves of both welded surfaces to the process of disappearance of macro-discontinuities will be lower than in case of SC welding.

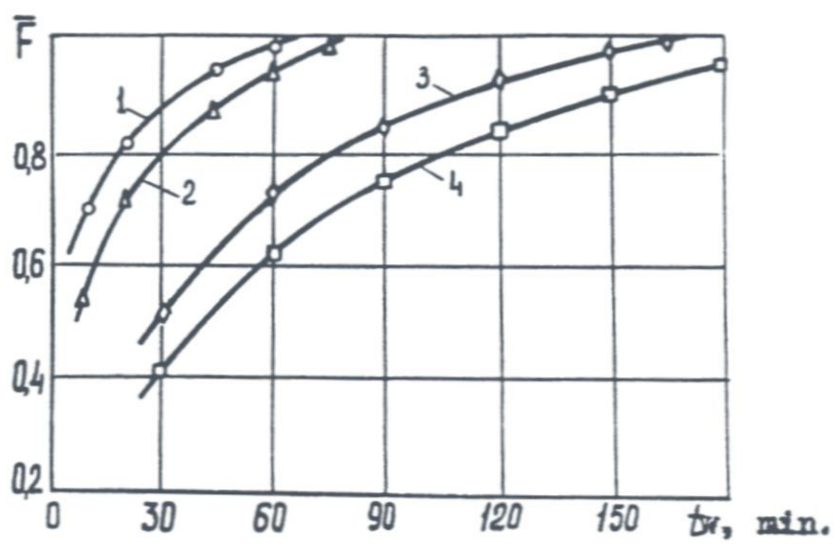

Fig. 6a

Dependence $F(t_W)$ (a) and $F(\varepsilon)$ (b) in case of welding of combinations (R_z 10,0 + R_z 0,05) OT4-1 + OT4-1 (1,3) and OT4-1 + BT3-1 (2,4) with different deformation rate ε', C^{-1}; $0,82\cdot10^{-5}$ (1,2)

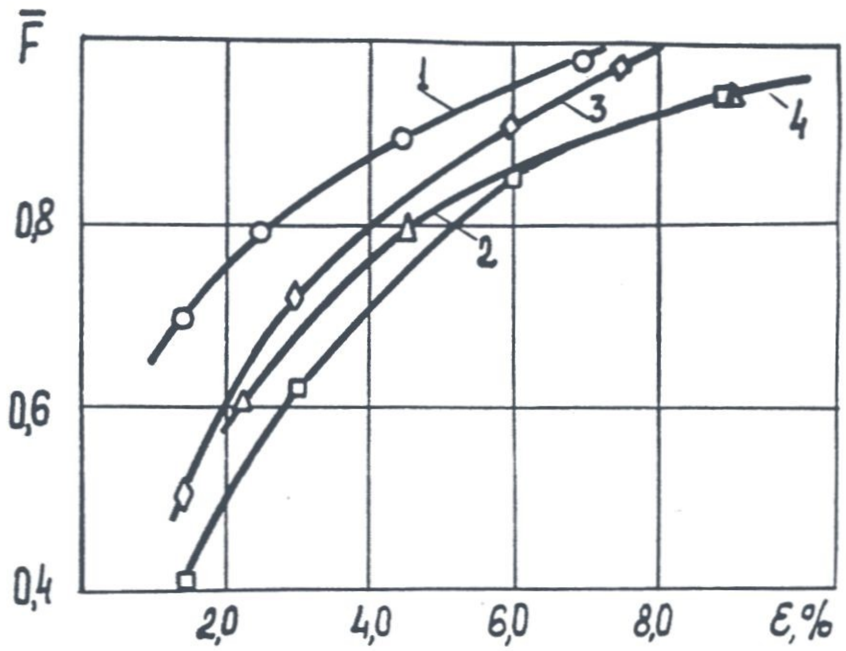

Fig. 6b

Dependence $F(t_W)$ (a) and $F(\boldsymbol{\varepsilon})$ (b) in case of welding of combinations (R_z 10,0 + R_z 0,05) OT4-1 + OT4-1 (1,3) and OT4-1 + BT3-1 (2,4) with different deformation rate $\boldsymbol{\varepsilon}'$, C^{-1}; $0,82\cdot10^{-5}$ (1,2)

Different geometry of defects in contact proves it: SC have flat long marcodiscontinuities (fig. 7a), and DC have thicker macrodiscontinuities (fig. 7b). One can judge of contribution value of formed microrelieves to the formation of a physical contact by results of welded joints annealing [4]. Thus welded joints anneling in vacuum (R_Z 0,05 + R_Z 70,0) at T_{an} a 900°C results in F_{SC} rise up to 0,95 value after preliminary contact formation F = 0,75, the annealing does not practically change F_{DC}.

The influence of special features of microrelieves formation on characteristics of $F(t_W)$ dependencies will be minimum if welded alloys have similar values of point A_{c3}. However the inequality (1) is correct for similar combinations. The welding of OT4 and BT3-1 alloys proves it (fig.8). Consequently special features of microrelieves formation can't be the single factor slowing down the growth of DC physical contact. At the same time it is proved that dependencies $F_{DC}(t_W)$ and F_{SC} (t_W) differ little in case of welding of DC alloys having lesser gradient of concentrations of alloyed elements in contact $\Delta C = C_2 - C_1$ (where: C_1, C_2 -content of alloyed elements in less or more alloyed composition of the given DC correspondingly).

The above phenomenon takes place in case of welding of OT4 + OT4-1 and BT14 + BT3-1 alloys. Consequently, the inequality (1) is connected with diffusion processes in the joint area. These processes are caused by a gradient of alloyed elements concentration. The inequality may be determined by realization of Frenkel effect for analysed DC group.

The realization is simplified if there are prepared areas for vacancies flows in the diffusion zone (1.3). In case of diffusion welding such may be microareas of welded joint with an unformed physical contact. Extra vacancies will promote the growth of microareas condensating on them and slowing down the formation of

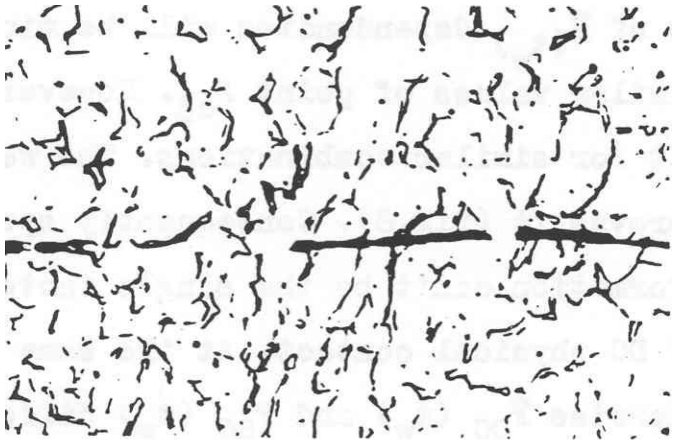

Fig. 7a

Welded joint area structure OT4-1 + OT4-1 (a,b) and OT4-1 + BT3-1 at t_w = 30 min. (a,b) and t_w = 150 min. (c,d), x1000.

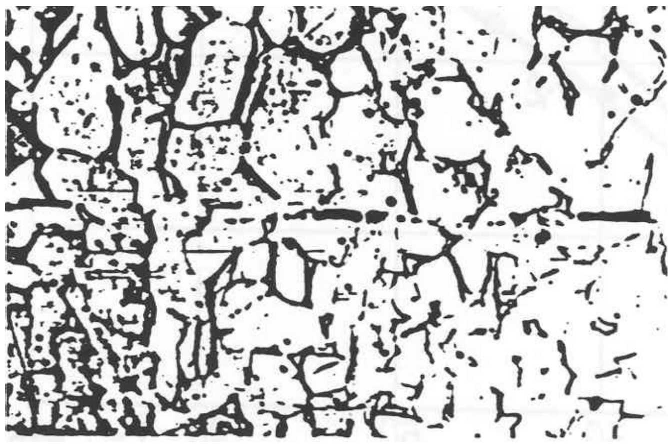

Fig. 7b

Welded joint area structure OT4-1 + OT4-1 (a,b) and OT4-1 + BT3-1 at t_w = 30 min. (a,b) and t_w = 150 min. (c,d), x1000.

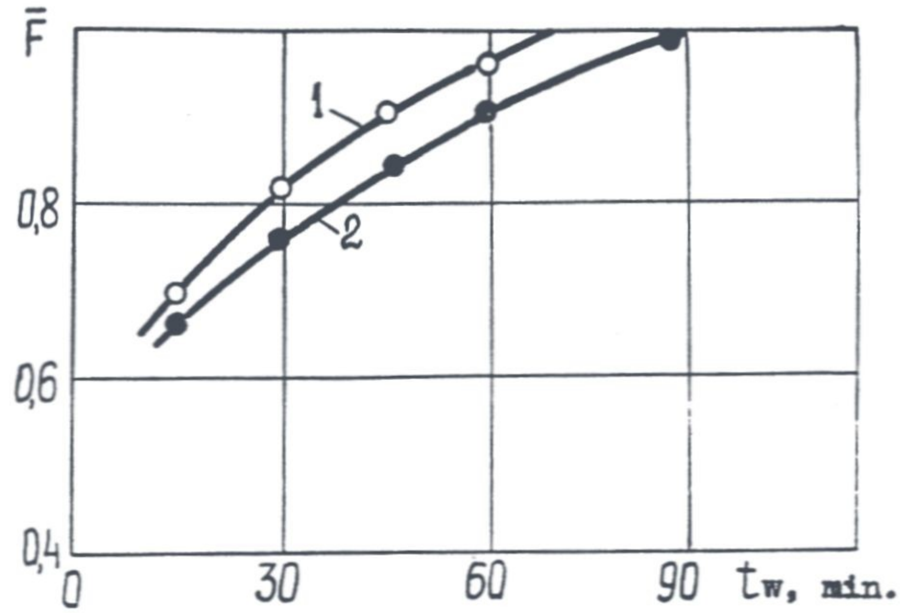

Fig. 8

Dependence $F(t_W)$ in case of welding (T_W = 940°C, $\varepsilon = 2{,}0\cdot10^{-5}$ c^{-1}) of combinations (R_Z 10,0 + R_Z 0,05) OT4 + OT4 (1) and OT4 + BT3-1 (2) a physical contact in such a way. The results of a metallographic analysis show this (fig. 7). With ε_H and t_W growth the SC defects become less in both dimensions of the surface of a section metallographic specimen (fig. 7c). There appear interrupted stitch defects transforming into pores which are removed later on accord to the regularities of the caking process. There is quite an opposite phenomenon in case of DC welding with high values of ε_H and t_W in the contact area and the existence of a great number of small pores there remain discontinuities of large sizes. Their thickness does not differ greatly from the initial one, though they become much shorter (fig. 7d). Experimental data on welding of BT3-1 and OT4 alloys show that F_{DC}(tw) and F_{SC}(tw) dependences are similar when the contact surfaces have low R_Z values (fig. 9). At final stages there is an inequality: $F_{DC} > F_{SC}$.

Probably extra vacancies promote the growth of only rather large discontinuities and intensify the removal of small ones. This fact makes quite reasonable the technological requirement to increase a class of purity of welded surfaces of DC alloys.

Alloyed elements in titanium alloys may promote both lowering (β = stabilizators) and rise (α = stabilizators) of point A_{c3}.

The development of diffusion processes of DC results in redistribution of alloyed elements in the joint area and may change point A_{c3} in a local zone of a contact surface. In case of welding in $\alpha+\beta$ = area the realization of this possibility is determined in the first term by the correlation of approaching diffusion flows of β-stabilizators in the contact area. The flows of the above group of combinations are balanced. Therefore local changes of point A_{c3} do not occur and the structural state of the surface layers differs little from the base material (fig. 7). This occurs when a value of gradient of β-stabilizators

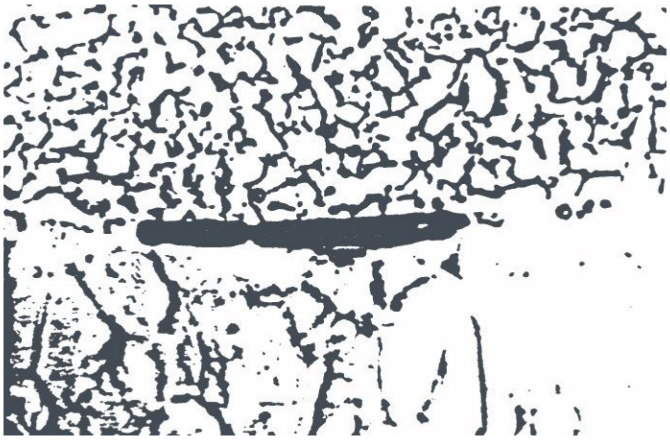

Fig. 7c

Welded joint area structure OT4-1 + OT4-1 (a,b) and OT4-1 + BT3-1 at t_w = 30 min. (a,b) and t_w = 150 min. (c,d), x1000.

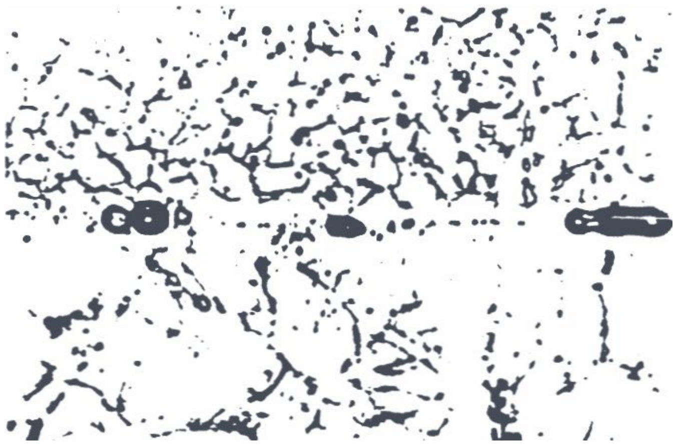

Fig. 7d

Welded joint area structure OT4-1 + OT4-1 (a,b) and OT4-1 + BT3-1 at t_w = 30 min. (a,b) and t_w = 150 min. (c,d), x1000.

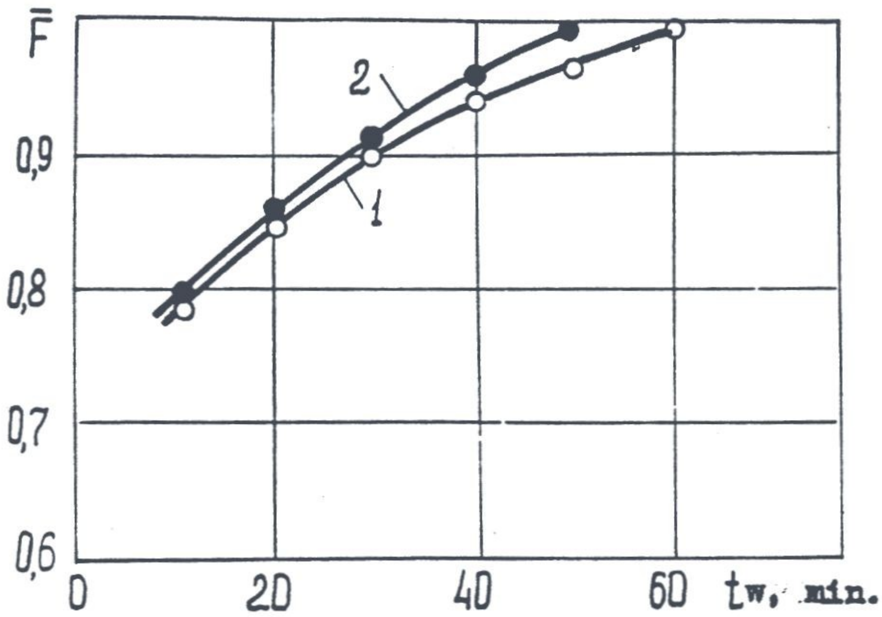

Fig. 9

Dependence $F(t_W)$ in case of welding (T_W = 940°C, $\varepsilon = 1,0 \cdot 10^{-5}$ c^{-1}) of combinations (R_z 1,5 + R_z 0,05) OT4 + OT4 (1) and OT4 + BT3-1 (2) concentrations ΔC_{β} in the contact area is not high or when one of the welded alloys has the surplus of them, and that is compensated by greater diffusion mobility of the other alloy. The situation changes with ΔC_{β} increase, that is confirmed by the following example.

Welding of BT1-0 alloy w1th other alloys in α-area is followed by essential structural changes in its surface layer. The in-flow of β-stabilizators elements brings point A_{c3} down and causes polimorphous $\alpha+\beta$ =transformation, followed by the grain growth in the contact area on the side nearest BT1-0. While ΔC_{β} is increased in a succession of DC (BT1-0 + OT4-1), (BT1-0 + BT3-1) (and BT1-0 + BT16) a thickness of a large grain layer and a size of grain grow (fig. 10).

The heavy integration of grain in the contact area of one of DC alloys increases its resistance to high temperature plastic deformation and reduces the rate of the physical contact formation. In particular this phenomenon is also revealed in case of welding of heterostructural combinations (fig. 11) of alloys having the same chemical composition*. For analyzed DC group F_{DC} decrease is not of great importance (fig. 12). This fact may be explained by the influence of the correlation of initial values of welded alloys

$(\varepsilon_{BT3-1} \sim 2\cdot10^{-5}\ c^{-1},\ \varepsilon_{BT1-0} \sim 1\cdot10^{-6}\ c^{-1})$

in the terms of experiment.

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*Experimental data were obtained together with ph.D. of Technology Rodionov V.N.

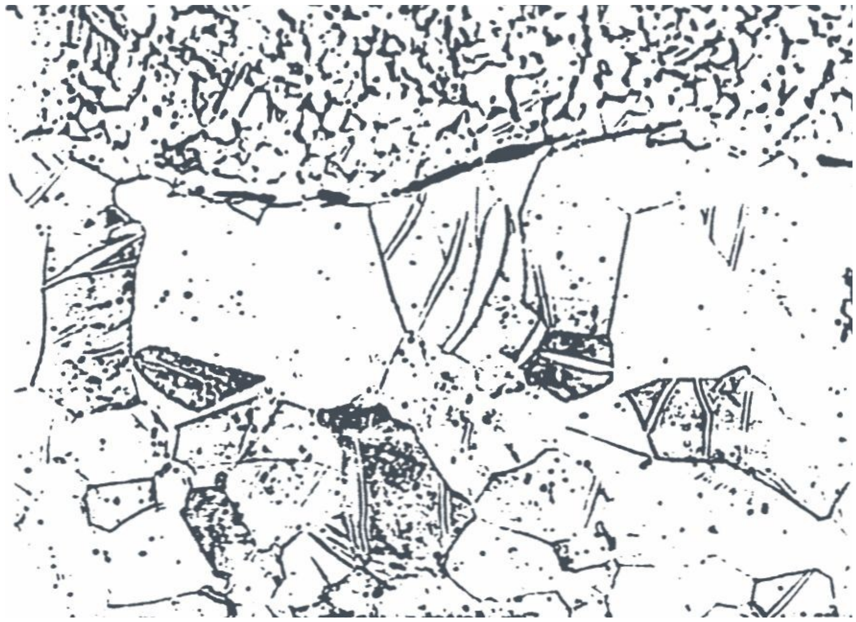


Fig. 10a

Welded joint areas structure (T_W = 857°C, t_W = 30 min.) BT1-0 + OT4-1 (a), BT1-0 + BT3-1 (b) and BT1-0 + BT16 (c) x500

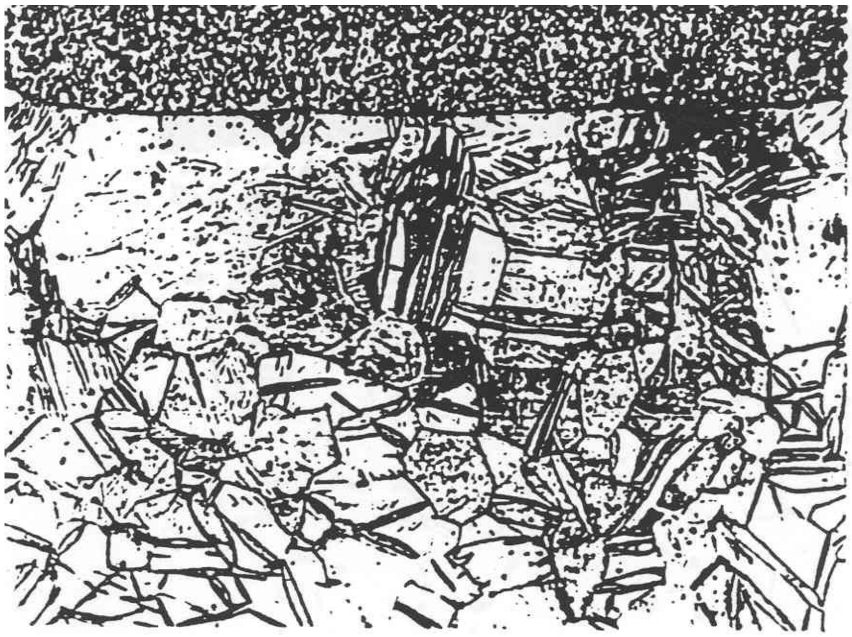


Fig. 10b

Welded joint areas structure (T_W = 857°C, t_W = 30 min.) BT1-0 + OT4-1 (a), BT1-0 + BT3-1 (b) and BT1-0 + BT16 (c) x500

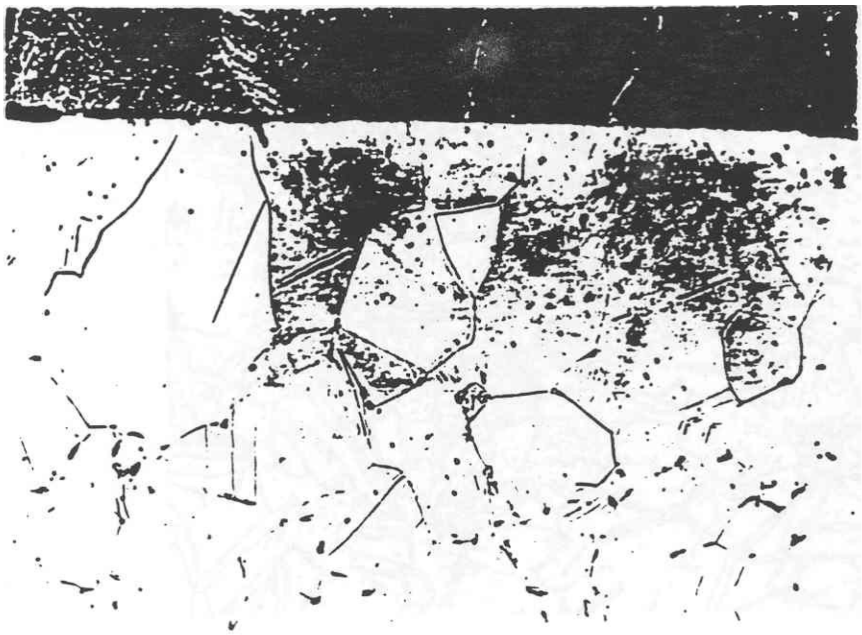


Fig. 10c

Welded joint areas structure (T_W = 857°C, t_W = 30 min.) BT1-0 + OT4-1 (a), BT1-0 + BT3-1 (b) and BT1-0 + BT16 (c) x500

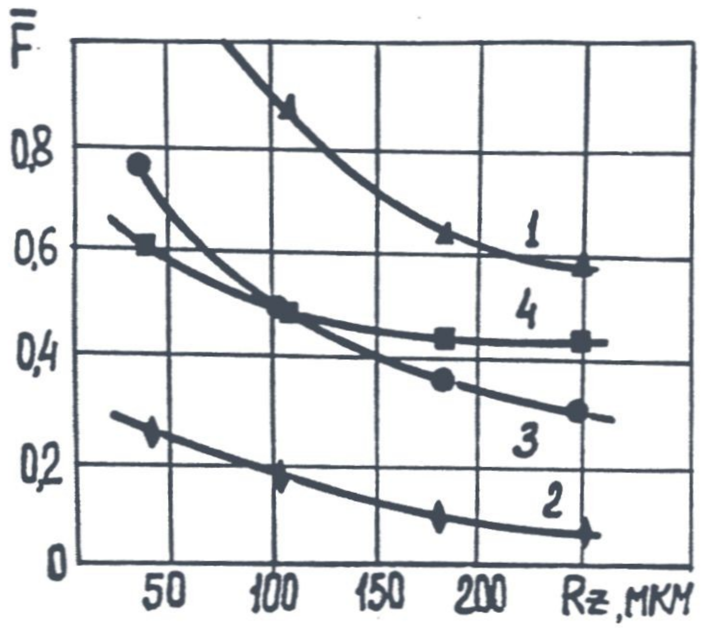


Fig. 11

Dependence $F(R_Z)$ (where R_Z – middle height of microprojections of turned surface) in case of welding T_W = 940 °C, t_W = 40 min., P = 2,0 MPa) of OT4 alloy with different combinations of contacting of small grain (S) and large grain (L) structures:

1- ST+SP; 2- LP+LT; 3- SP+LT; 4- ST+LP

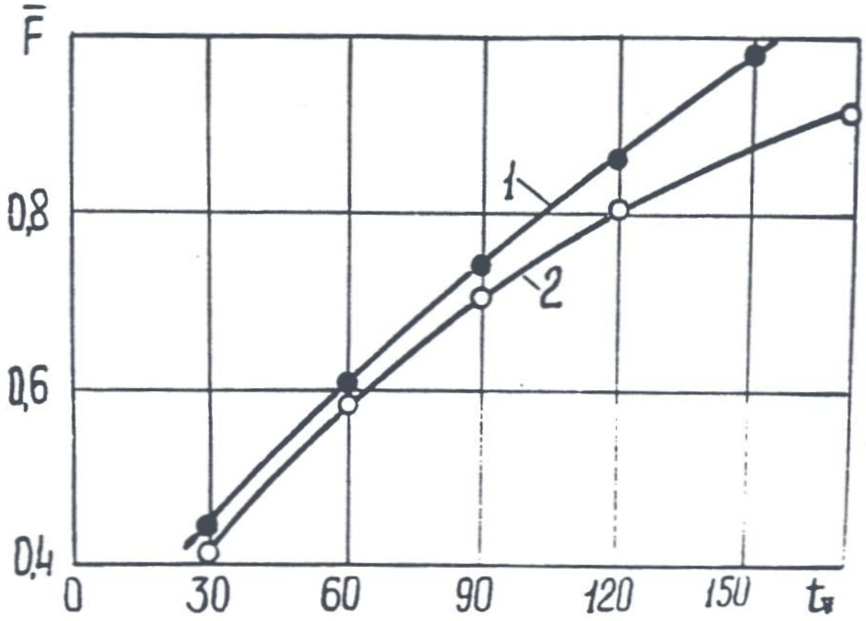


Fig. 12

Dependence $F(t_W)$ in case of welding (T_W = 875°C, $\varepsilon = 1{,}0 \cdot 10^{-5}\ s^{-1}$) of combinations ($R_Z$ 10,0 + R_Z 0,05) BT3-1 + BT3-1 (1) and BT3-1 + BT1-0 (2)

In case of mechanical heterogenuity a physical contact is formed to a great extent due to microplastic deformation of the material having greater ε. Therefore the structural hardening of less plastical material does not seriously influence the characteristic of $F(t_W)$ dependencies (fig. 12). On the contrary, the hardening of a more plastic material results in an intensive F_{DC} lowering in comparison with F_{SC}. Specifically it occurs in case of welding of low α = alloys and pseudoalloyed α = compositions having high values of ε with high β = alloys and pseudo-β = alloys. In this case F_{DC} lowering is provided by the fact that β-stabilizators supersaturation in the area near the contact of a low alloy is followed by decreasing of their concentration in a high alloy. It locally raises point A_{c3} of a high alloy thus reducing the rate of a high temperature microplastic deformation. Besides high alloys are determined by low intensity of microrelief formation [15]. All this causes the most intensive F_{DC} decrease in comparison with F_{SC} in all the above analyzed cases, an example of welding of OT4-1 and BT16 alloys proves it (fig. 13).

It is common knowledge that due to effect of deformation hardening in case of titanium alloys welding an inequality [4] takes place:

$$\left|\frac{dF}{d\varepsilon}\right| \cdot \varepsilon_i > \left|\frac{dF}{d\varepsilon}\right| \cdot \varepsilon_{i+1} \qquad (2)$$

In case of DC welding with adequate R_Z due to realization of Frenkel effect and the structural hardening in the contact area the inequality (2) does not take place. It is correct for the combinations characterized by low values of ΔC and ΔC_β. (OT4 + OT4 -1, BT14 + BT3-1) or in the case when structural hardening takes place in less plastic material (BT1-0 + OT4-1 t BT1-0 + BT3-1). In case of ΔC increasing (see fig. 6b) an approximate equality

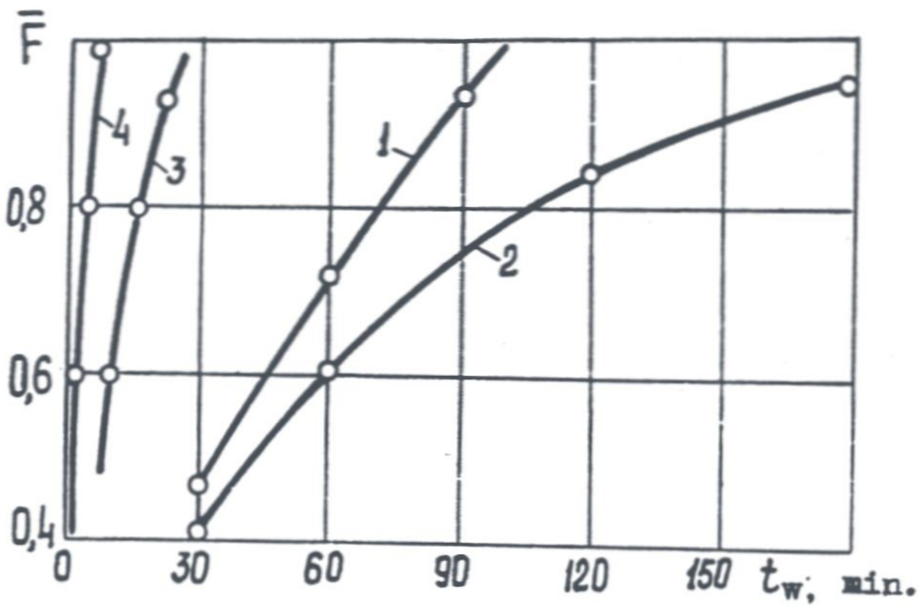


Fig. 13a

Dependence $F(t_W)$ (a) and $F(\varepsilon)$ in case of welding (T_W = 900°C) of combinations (R_z 10,0 + R_z 0,05) OT4-1 + OT4-1 (1) and OT4-1 + BT16 (2,3,4) with different deformation rate ε, s^{-1}: $1,22 \cdot 10^{-5}$ (3), $27,0 \cdot 10^{-5}$ (4)

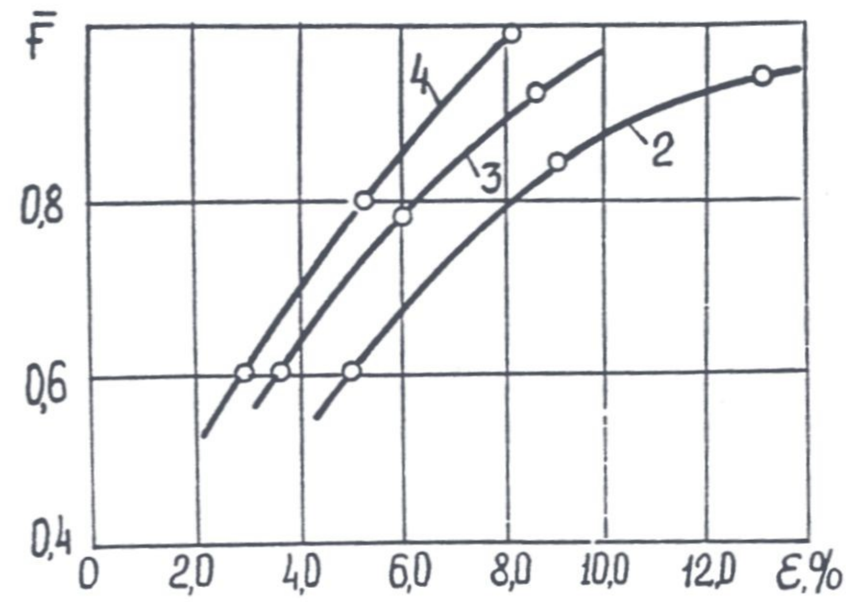


Fig. 13b

Dependence $F(t_W)$ (a) and $F(\varepsilon)$ in case of welding (T_W = 900°C) of combinations (R_z 10,0 + R_z 0,05) OT4-1 + OT4-1 (1) and OT4-1 + BT16 (2,3,4) with different deformation rate ε, s^{-1}: 1,22·10^{-5} (3), 27,0·10^{-5} (4)

takes place:

$$\left|\frac{dF}{d\varepsilon}\right|\cdot\varepsilon_i \sim \left|\frac{dF}{d\varepsilon}\right|\cdot\varepsilon_{i+1} \qquad (3)$$

In case of unfavourable correlation of ε values of welded alloys further ΔC and ΔC_β increase results in the opposite regularity:

$$\left|\frac{dF}{d\varepsilon}\right|\cdot\varepsilon_i > \left|\frac{dF}{d\varepsilon}\right|\cdot\varepsilon_{i+1} \qquad (4)$$

In the process of welding in $\alpha+\beta$ -area structural hardening is intensified by T_W and t_W increase. Therefore together with the inequality (4) the following correlation is correct:

$$\left|\frac{dF}{d\varepsilon}\right|t_{Wi}, T_{Wi} > \left|\frac{dF}{d\varepsilon}\right|t_{Wi+1}, T_{Wi+1} \qquad (5)$$

Patterns (4) and (5) show that in order to diminish an effect of structural hardening and to obtain joints with less residual deformation it is necessary to reduce temperature and duration of welding process by increasing deformation rate. It is evident that optimal T_W, t_W and ε must ensure the formation of a full physical contact before the formation of low plastic layers. Diminishing of roughness of welded surfaces will also provide this aim. It is very important from technological point of view that in case of mechanical heterogenuity a physical contact is formed more rapidly if less plastic material has higher R_z values in DC. Moreover this tendency strengthens while R_z values become less (fig. 11). Probably forcing a hard microprojection into an opposite soft surface is followed by less deformation hardening in comparison with an apposite variant. Therefore the requirement to increase a class of purity of contact surfaces should be first of all applied to more plastic material of DC.

The determined regularities of a physical contact formation connected with structural changes of surface layers take place in one or another degree in case of welding in other temperature ranges. Thus if more high point A_{c3} corresponds to more alloyed composition of DC then it is sometimes more efficient to conduct welding at temperature between points A_{c3} of welded alloys. In addition we can also observe some F_{DC} values decrease in comparison with F_{SC}, especially for DC with low ΔC_{β} values and rather high ΔC values (for example, BT3-1 + OT4-1). In this case a limiting factor is a diffusion removal of α-stabilizators from more alloyed composition. This removal causes a local lowering of point A_{c3} end an increase of a size of grain in the area near the contact (fig. 14).

IV. Influence of Heteroalloying on the Quality of a Welded Joint

Titanium alloys were used for research (table 1), the alloys were welded in SC and DC.

Welded samples (fig. 15) were prepared in a pure argon atmosphere in such a way as having been used before (fig. 1). Then they were heated to the predetermined welding temperature T_W, held parted for t_p= 10 min. and pressed to ε= 1,0...1,5% for t_p= 5 min. According to the above technological scheme welding ensures high reactivity of contact surfaces in connection with formation of surface relief of micro- and substructural character [16].
After welding some specimens were annealed without pressure (in a single thermal cycle) at different T_{an} and t_{an}. The quality of welded joints was valued by KCV values. The welding alloys are characterized by correlation KCV_1/KCV_2, where KCV_1 and KCV_2 - an impact toughness of more or less tough of the welded alloys.

Fig. 14

Welded joint structure BT3-1 + OT4-1 at T_W = 940 °C, t_W = 30 min., x 1000

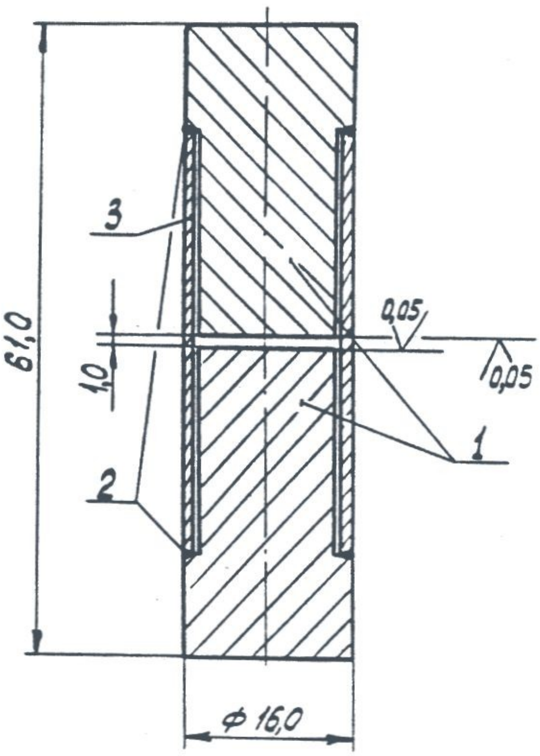


Fig. 15

Welded specimens: 1- specimens; 2- welded seams (welds); 3- titanium capsules

Therefore KCV_{DC} values are also dependent on welded materials properties. According to this fact when analyzing KCV values they took into consideration the character of destruction which may occur along the contact surface (1-type) or in some distance from an incision on tougher material (III-type). An intermediate II-type of destruction including the first two elements is also possible.

The results of impact bend tests of welded joints after welding cycle are given in table 2. It is obvious that KCV_{DC} values are mostly determined by the difference in the degree of alloying of the welded ΔC alloys and also by total content of alloyed C_1 elements in tougher (less alloyed) composition.

With minimum ΔC a welded joint is destroyed according to the first type, and KCV_{DC} values are similar to corresponding KCV_{SC} values (see table 2 pos. 6,14). With maximum ΔC and minimum C_1 KCV_{DC} values exceed corresponding KCV_{SC} values of both welded alloys, that is an effect of improved weldability is realized. This effect is maximum for combinations Nos. 3,4,7,8 and is decreasing with C_1 growth (No. 10,11) and ΔC lowering (No. 2). The realization of the effect is followed by the appearance of tough fractures of III and II types.

The determined regularities are more observable at the first group of DC (table 2). All those tendencies are preserved at the second group, but DC weldability is getting somewhat worse.

Thus it follows that in inoxidable terms under short thermal influence heteroalloying intensifies all stages of the process of point formation. Probably this is connected with characteristic features of diffusion in the contact area resulting in formation of uncompensated flow of vacancies. The latter causes

Table 2

Nos.	Combination of alloys	KCV at T_w, °C 800	825	850	875	900	940	980	$\frac{KCV_1}{KCV_2}$	$\frac{\Delta C}{C_1}$ %
1	2	3	4	5	6	7	8	9	10	11
Group 1. Combinations when more alloyed composition has greater $T_{\alpha+\beta\to\beta}$ - transformation										
1	BT1-0 + BT1-0	0,4(I)**	0,46(I)	0,55(I)	0,65(I)	0,7(I)	0,55(I)	0,4(I)	1,0	0
2	BT1-0 + OT4-1	0,45(I)	0,5(I)	0,7(II)	0,8(III)	0,8(III)	0,4(I)	0,4(I)	1,98	2,5
3	BT1-0 + BT14	0,5(I)	0,6(II)	0,7(III)	0,75(III)	0,75(III)	0,7(III)	0,5(I)	2,5	8,5
4	BT1-0 + BT3-1	0,80 (III)	0,82(III)	0,85(III)	0,87(III)	0,80(III)	0,78(III)	0,5(I)	3,0	1,3
5	OT4-1 +OT4-1	0,45(I)	0,6(I)	0,7(I)	0,75(I)	0,8(I)	0,65 (I)	0,6(I)	1,0	0
6	OT4-1 + OT4	0,4(I)	0,5(I)	0,6(I)	0,7(I)	0,7(I)	0,6(I)	0,5(I)	1,07	2,5
7	OT4-1 + BT14	0,50(I)	0,6(II)	0,9(III)	0,95(III)	1,05(III)	0,75(III)	0,6(I)	1,25	6,0
8	OT4-1 + BT3-1	0,72 (II)	1,0(III)	1,1(III)	1,1(III)	1,05(III)	0,7(III)	0,6(I)	1,5	8,8
9	OT4 + OT4	0,37(I)	0,44(I)	0,5(I)	0,61(I)	0,75(I)	0,65(I)	0,45(I)	1,0	0
10	OT4 + BT14	0,45 (I)	0,6 (II)	0,76(II)	0,78(II)	0,8(II)	1,0(III)	0,45(I)	1,17	3,5
11	OT4 + BT3-1	0,5(I)	0,7 (II)	0,78(II)	0,8(II)	0,85(II)	1,0(III)	0,5(I)	1,4	6,3
12	BT14 + BT14	0,1(I)	0,15(I)	0,22(I)	0,32(I)	0,52(I)	0,58(I)	0,4(I)	1,0	0
13	BT3-1 + BT3-1	0,1(I)	0,14(I)	0,25(I)	0,32(I)	0,42(I)	0,51(I)	0,4(I)	1,0	0
14	BT14 + BT3-1	0,1(I)	0,15(I)	0,25(I)	0,33(I)	0,5(I)	0,55(I)	0,4(I)	1,2	2,8

1	2	3	4	5	6	7	8	9	10	11
	Group 2. Combinations when more alloyed composition has less $T_{\alpha+\beta\to\beta}$ = transformation									
15	BT16 + BT16	0,03(I)	0,04(I)	0,05(I)	0,07(I)	0,1(I)	0,1(I)	0,8(I)	1,0	0
16	BT16 + BTI-0	0,05(I)	0,4(II)	0,55(III)	0,5(III)	0,45(III)	0,4(I)	0,4(I)	3,75	12,5
17	BT16 + OT4-1	0,05(I)	0,3(I)	0,36(II)	0,5(II)	0,8(III)	0,4(I)	0,4(I)	1,9	10,[illegible]
18	BT16+ OT4	-	0,1(I)	0,15(I)	0,2(I)	0,25(I)	0,33(I)	0,2(I)	1,75	7,5
19	BT16 + BT14	-	-	0,1(I)	0,17(I)	0,32(I)	0,3(I)	0,3(I)	1,5	4,[illegible]
20	BT16 + BT3-1	-	-	0,05(I)	0,11(I)	0,20(I)	0,27(I)	0,2(I)	1,25	1,[illegible]

*C_1 and C_2 - composition of alloyed elements in welded alloys more or less alloyed correspondingly (table 1)

** - character of destruction is stated in brackets the dislocations formation in diffusion zone [10] and probably results in displacing of the initial surface of the contact according to Kirkendall effect [13].

In $\alpha+\beta$ -area isothermal annealing of welded joints results in decreasing of KCV_{DC} values (fig. 16-23). Consequently due to duration of thermal influence a heteroalloying factor may provide both improving of titanium alloys weldability and its deterioration. Besides the degree of KCV_{DC} lowering depends on the value of chemical heterogenuity in the contact area and the origin of diffusion elements.

The publication [17] states that Kirkendall effect exists to a low degree for systems (Ti)-(Ti-4,OAl) and (Ti)-(Ti-8,OAl) with annealing in α-area due to similar coefficients of titanium and alluminium diffusion (D_{al}~1,1D_{Ti}), but Frenkel pores-formation does not take place all. Therefore in case of BT1-0 + BT5 alloys welding an effect of improved weldability does not occur (fig. 16) and in case of annealing there is an increase of KCV_{DC} values, fractures of II and III types appear. The KCV_{DC} values increase transforms into slight decrease only at t_{an} > 2,0 h.

The process of softening is improved if along with Al there are some other elements in the composition of welded alloys which diffusion mobility is more than titanium.

Sufficient softening occurs (fig. 17,18) if with adequate ΔC diffusion flows of elements of β-stabllizators over contacting surface are approximately the same. It is possible with a low gradient of concentrations of β-stabilizators (ΔC_β) or when β-stabilizators surplus of one of the welded alloys is compensated by their greater diffusion mobility of another (OT4-4 + BT3-1; OT4-1 + BT14; OT4 + BT14; OT4 + BT3-1). In this case volumetric interaction happens according to common [11] kinetics.

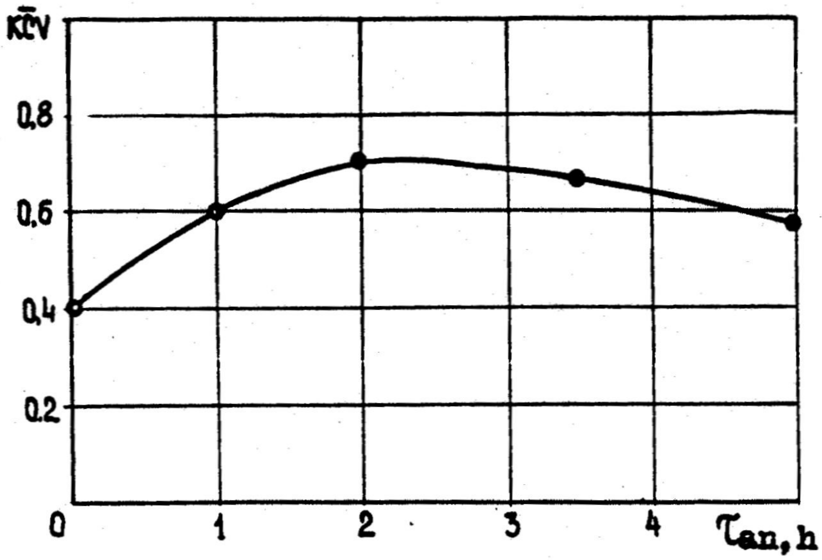


Fig. 16

Dependence KCV(t_{an}) for BT1-0 + BT5 combination after welding and annealing at T_W = T_{an} = 875 °C

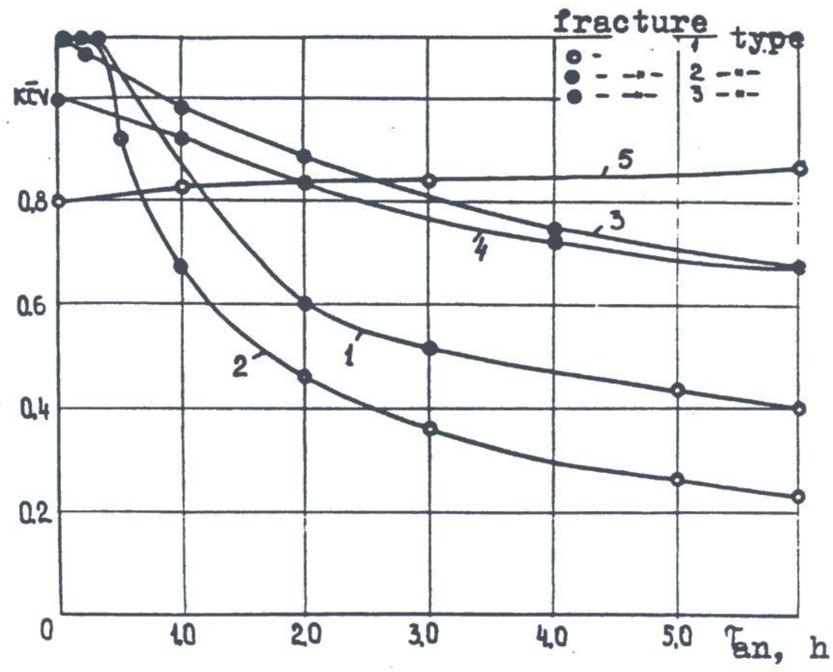


Fig. 17

Dependence KCV(t_{an}) for combinations OT4-1 + BT3-1 (1-4) and OT4-1 + OT4-1 (5) after welding (T_W = 900 °C) and annealing at T_{an} °C: 1 – 850; 2,4,5 – 900; 3 – 940; BT3-1 -initial small grain structure; OT4-1 -small grain (1-3,5) and large grain (4) structure

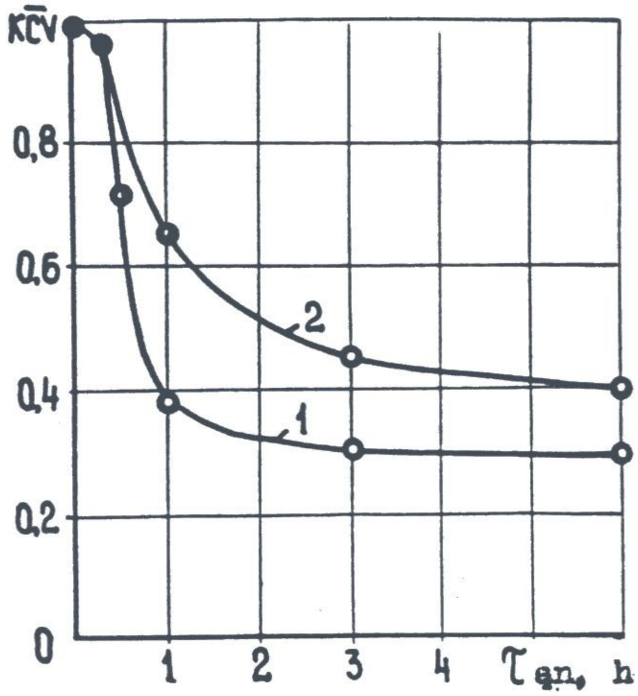


Fig. 18

Dependence KCV(t_{an}) for combinations OT4 + BT14 (1) and OT4 + BT3-1 (2) after welding and annealing at $T_W = T_{an} = 940$ °C

The boundary orientated along the contact surface (fig. 19a) is modified with t_{an} growth, the grains common for both welded blanks appear on it maintaining the initial globular structure. However a chain of small pores appear (fig. 19b) over the initial surface of contacting. The appearance of the chain is probably connected with Frenkel effect realization. Probably this is the main cause of softening that is followed by changing of fracture character from the III type to the II and the I.

The loss of strength becomes less with T_{an} decrease (fig. 17, curves 1,2) and with approaching DC (OT4 + OT4-1; BT14 + BT3-1) with less ΔC (fig. 20).

The structure of welded alloys greatly influences the degree of the loss of strength. As the coefficient of diffusion along the grains boaders is higher than in the grains volume the existence of the small-grains-structure in the contact area promotes the increase of approaching flows. This results in the growth of uncompensated flow of vacancies and the increase of porosity. In case of large-grain-structure a resulting diffusion flow becomes less and decreases softening. Consequently all technological methods leading to the integration of grains of even one of the welded alloys promote decreasing of softening. Welding in β-area is among such methods (fig. 17, curve 3). Specifically, the previous annealing of OT4-1 in β-area (1000 °C, 30 min.) resulting in the growth of initial grain to 300 mcm., excludes fractures of the I type (fig. 17. curve 4). It should be noted that KCV_{DC} lowering is also observed when there ara no pores in the contact area. In such a case this is determined by simple equalizing of concentrations of alloyed elements in the joint area. Besides with t_{an} growth KCV_{DC} values approach the corresponding index for more alloyed composition with maintaining tough fracture (fig. 17, curves 3.4).

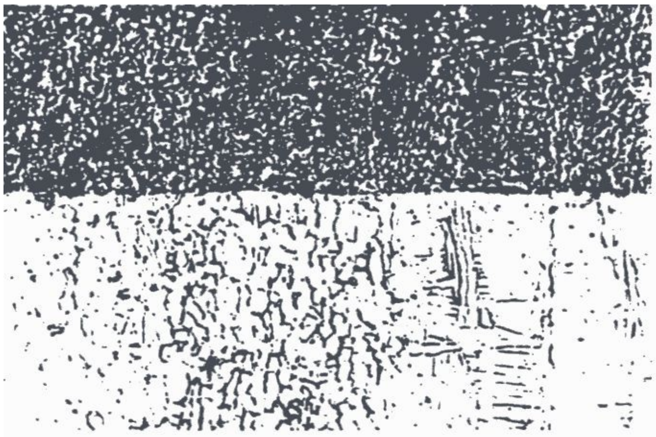


Fig. 19a

Welded joint structure of OT4-1 + BT14 combination after welding (T_W = 900 °C, x 500 – a) and annealing T_{an} = 900 °C, t_{an} = 64, x 1000 – b)

Fig. 19b

Welded joint structure of OT4-1 + BT14 combination after welding (T_W = 900 °C, x 500 – a) and annealing T_{an} = 900 °C, t_{an} = 64, x 1000 – b)

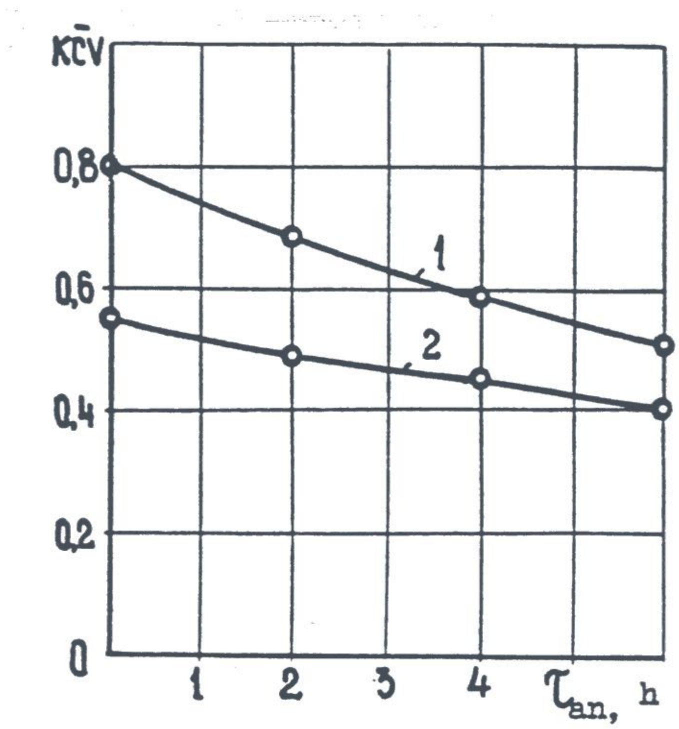


Fig. 20

Dependence KCV(t_{an}) for OT4 + OT4 (1) and BT14 + BT3-1 (2) after welding and annealing at 900 °C (1) and 940 °C (2)

The loss of strength becomes less if diffusion flow of β-stabilizators elements dominates in one of directions of the contact area. Such a situation takes place when one of DC alloys unlike the other does not comprise β-stabilizators or when ΔC_{β} values are high enough. In this case the influx of β-stabilizators to the surface layer of less alloyed composition results in local decreasing of $\alpha+\beta$ -transformation temperature. Specifically it happens in case of annealing of welded joints BT1-0 + BT14, BT1-0 + BT3-1 (fig. 21). In this case a surface layer with large-grain-structure is formed (fig. 22a), that cardinally decreases a resulting flow of vacancies. Therefore even after long electropolishing porosity was not revealed in the joint area (fig. 22b), this fact excludes the destruction of the 1st type.

For DC BT1-0 + OT4-1 ΔC_{β} is lower, therefore the grain in the surface layer BT1-0 is integrated less (fig. 22c). As a result in case of electropolishing there appears a chain of very small pores in the joint area (fig. 22d). In this case KCV_{DC} dependence is identical in quality to the previous one (fig. 21). However after long annealing at 850 °C the destruction happens according to the I and II types.

Intensive softening happens with DC, having maximum ΔC and ΔC_{β} values. High ΔC_{β} value results in a large-grain-layer formation in the contact area and high ΔC value causes porosity. Such a situation occurs in case of annealing of welded joints of BT16 alloy with other alloys (fig. 23), especially with low alloyed.

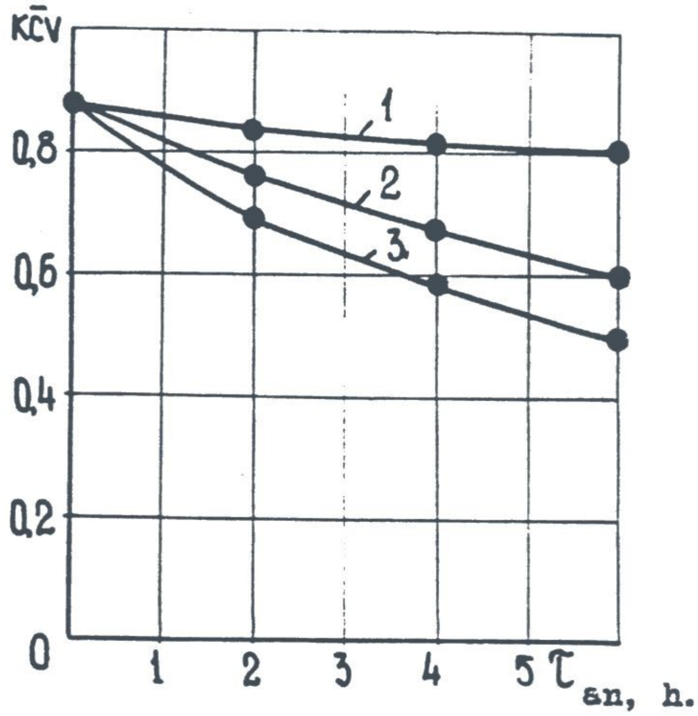


Fig. 21

Dependence KCV(t_{an}) for BT1-0 + BT3-1 combination after welding (T_W = 875 °C) and annealing at T_{an} °C: 1- 850; 2- 875; 3- 900.

Fig. 22a

Welded joint zone structure of BT1-0 + BT14 (a,b) and BT1-0 + OT4-1 (c,d) combinations after welding and annealing $T_W = T_{an} = 850$ °C, t_{an} = 6h; a,c (x100) - etching: b, (x 1000), d (x 500) - electropolishing

Fig. 22b
Welded joint zone structure of BT1-0 + BT14 (a,b) and BT1-0 + OT4-1 (c,d) combinations after welding and annealing $T_W = T_{an} = 850$ °C, $t_{an} = 6h$; a,c (x100) – etching: b, (x 1000), d (x 500) – electropolishing

Fig. 22c

Welded joint zone structure of BT1-0 + BT14 (a,b) and BT1-0 + OT4-1 (c,d) combinations after welding and annealing $T_W = T_{an} = 850$ °C, $t_{an} = 6h$; a,c (x100) – etching: b, (x 1000), d (x 500) – electropolishing

Fig. 22d

Welded joint zone structure of BT1-0 + BT14 (a,b) and BT1-0 + OT4-1 (c,d) combinations after welding and annealing $T_W = T_{an} = 850$ °C, $t_{an} = 6h$; a,c (x100) - etching: b, (x 1000), d (x 500) - electropolishing

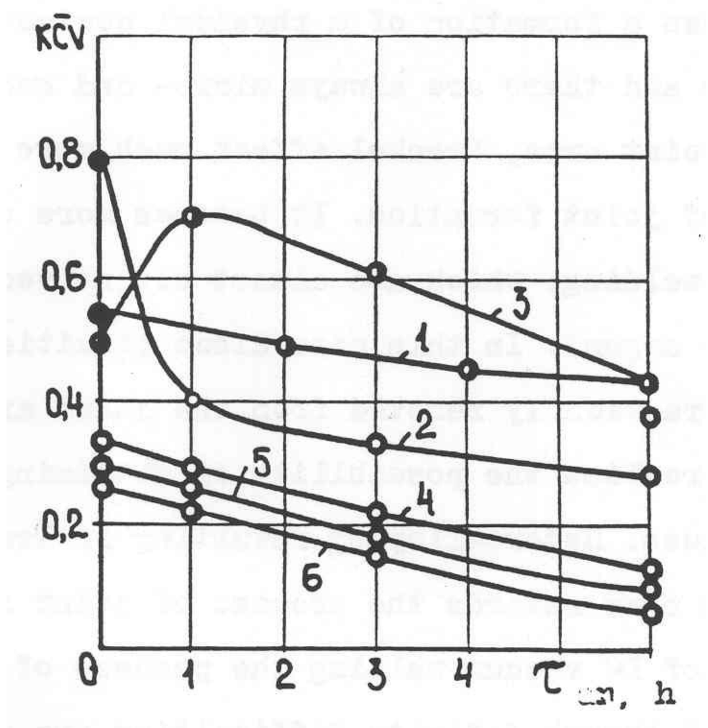


Fig. 23

Dependence KCV(t_{an}) for combinations BT1-0 + BT16 (1), OT4-1 + BT16 (2,3), OT4 + BT16 (4), BT15 + BT16 (6). Welding and annealing at T °C: 850 (3), 875 (1), 900 (2,4,5,6).

V. Technological Aspects of Weldability.
The Choice of Optimal Technological Schemes

Kirkendall and Frenkel effects are exposed simultaniously and they are competitive, but if there are prepared areas for uncompensated vacancies flows in the diffusion zone, then Frenkel effect becomes predominant [13]. Therefore in real terms of welding when a formation of a physical contact occurs comparatively slow and there are always micro- and macrodiscontinuities in the joint area, Frenkel effect much more slows down the process of joint formation. It becomes more obvious in case of inert gas welding, which ere almost not solved in titanium (for example, argon). In this case discontinuities filled by gas are not practically removed from the joint area and it become difficult to realize the possibility of obtaining a joint with KCV=1,0 values. Heteroalloying resulting in Frenkel effect realization much more retards the process of joint formation.

In case of DC vacuum welding the process of joint formation is facilitated though definite difficulties are also preserved in this case. The results of welding of specimens with different initial geometry of welded surfaces prove it. Specimens (in 16x30 mm diameter) were welded in vacuum 10^{-2} mm of a mercury column with additional screening of joint area by getter.

The most difficult problem is to obtain qualitative DC joints characterized by maximum ΔC and ΔC_{β} with absolutely different resistance of welded alloys to high temperature plastic deformation (for example, 0T4-1 + DT16).

The formation of a large grain layer on the side of less alloyed composition slows down the physical contact formation. But even after a full macrocontact formation in the joint area a great number of micropores mainta1ns. They are situated both over the initial surface of contacting and along a new structural border of the joint. Moreover the joint microareas last coming into contact are the most defective ones (fig. 24). Such a structure of joint zone results in low KCV_{DC} values. Found defects are stable enough, therefore even in case of substantial deformation there is no perceptible increasing of KCV_{DC} (fig. 25).
Consequently in a studied case volumetric interaction is the stage limiting the formation of a qualitative joint. The reduction of duration of thermodeformation cycle of welding due to increase of the deformation rate provides the porosity decreasing and physical contact formation with lesser accumulated deformation (fig. 13). However even in this case KCV_{DC} values remain not high (fig. 25) because an area near the contact of high alloy is not practically deformed and its contact surface remains non-active. In this case the activation stage becomes limitary and to obtain high KCV_{DC} values it's necessary to take special measures for increase of the reaction activity of a high alloy surface.

In case of DC welding characterized by high enough ΔC_{β} with moderate ΔC (BT-1-0 + BT3-1 and so on) the tendency to pores format1on is minimum, therefore KCV_{DC}=1,0 joint is formed practically almost after achieving a full physical contact. Consequently this stage of the process is limitary, however its realization is somehow slowed down in comparison with SC due to formation of low plastic layers. Therefore for similar DC a technological scheme providing a physical contact formation at low temperatures with further heating to a high boarder of $\alpha+\beta$ -

Fig. 24

Welded joint zone structure OT4-1 + BT16 (R_Z 0,05 + R_Z 10,0) at T_W = 900 °C, t_W = 2h; ε = 10%; x1000.

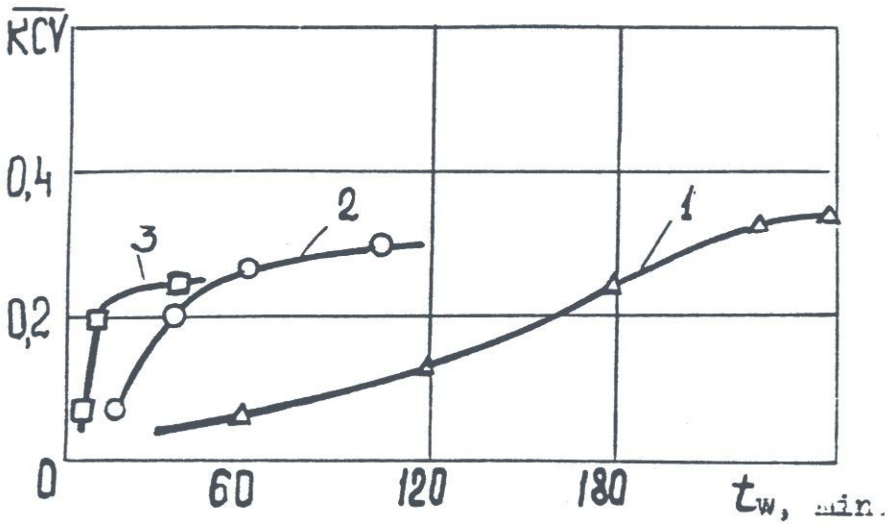


Fig. 25

Dependence KCV(t_W) in case of DC welding BT16 + OT4-1 (R_Z 0,05 + R_Z 10,0) at ε , C^{-1}: $1,22\cdot10^{-5}$ (1), $7,0\cdot10^{-5}$ (2), $27,0\cdot10^{-5}$ (3)

area is the most efficient.

In case of DC welding characterized by low ΔC_{β} but sufficient ΔC (OT4-1 + BT3-1, OT4 + BT3-1) the influence of heteroalloying may be different. With comparatively high R_{Z} (10,0 mcm) and low deformation rate the chemical heterogenuity slows down a physical contact formation causing Frenkel pores-formation.

In addition discontinuities remain in the contact area even after a full macrocontact formation. Furthermore an additional deformation promotes their removal and obtaining qualitative joints with high enough KCV_{DC} in case of a tough fracture (fig. 26a, 27a). On the other hand, unlike SC a deformation rate increase results in obtaining qualitative joints with less accumulated deformation (fig. 26a, 27a,b). In this case KCV_{DC} values increase may happen more intensive than of DC (fig. 26a). This tendency also evident for lower R_{Z} (fig. 26b). Consequently to obtain higher KCV_{DC} values with less accumulated deformation it is necessary to increase a class of purity of welded surfaces and to reduce time of thermodeformation influence. With substantial R_{Z} after macrocontact formation in $\alpha+\beta$ -area the annealing of welded joint in β-area would be desirable.

The above-mentioned recommendations for different groups of DC will permit to diminish the negative concequencies connected with chemical heterogenuity in the contact area. However, a better weldability effect will not be realized in this case.

Its realization is possible only in case of short contact interaction of surfaces with high reaction activity (fig. 1, table 2), moreover for combinations with maximum ΔC and ΔC_{β} this is the only possibility of obtaining qualitative joints with limited accumulated deformation.

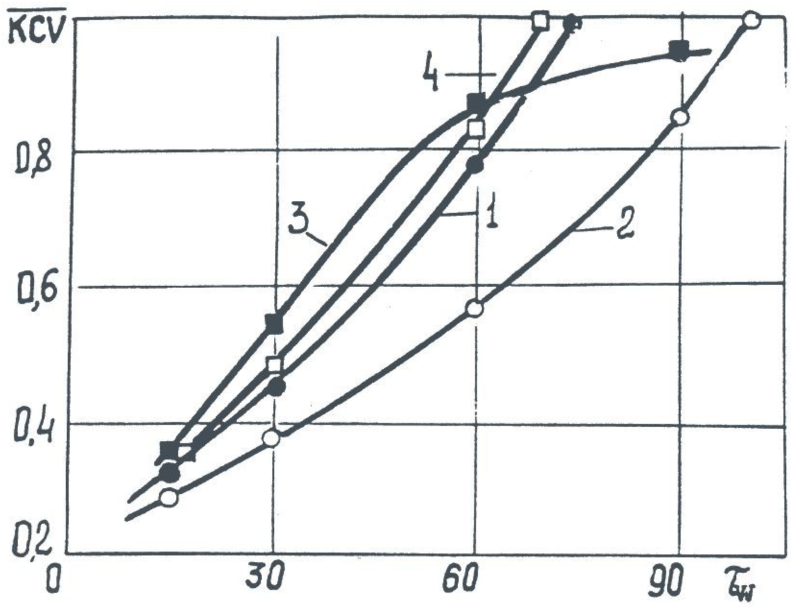


Fig. 26a

Dependence KCV(t_W) for DC OT4 + BT3-1 (2,4) and SC OT4 + OT4 (1,3) at T_W = 940 °C: a- R_z 10,0 + R_z 0,05 at ε, C^{-1}: $2\cdot10^{-5}$ (1,2) and $3,5\cdot10^{-5}$ (3,4); b- R_z 1,5 + R_z 0,05 at ε = $0,8\cdot10^{-5}$ s^{-1}

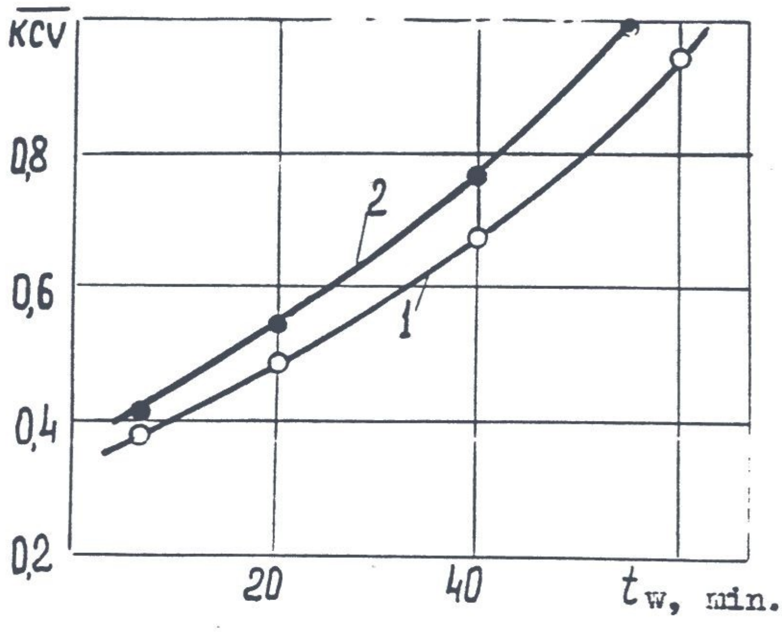


Fig. 26b

Dependence KCV(t_W) for DC OT4 + BT3-1 (2,4) and SC OT4 + OT4 (1,3) at T_W = 940 °C: a- R_z 10,0 + R_z 0,05 at ε, C^{-1}: $2\cdot10^{-5}$ (1,2) and $3,5\cdot10^{-5}$ (3,4); b- R_z 1,5 + R_z 0,05 at ε = $0,8\cdot10^{-5}$ s^{-1}

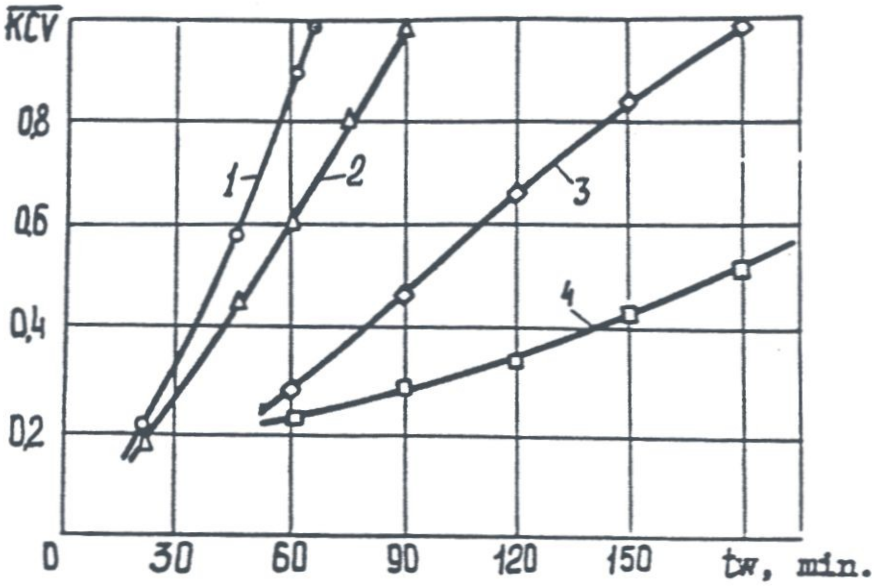
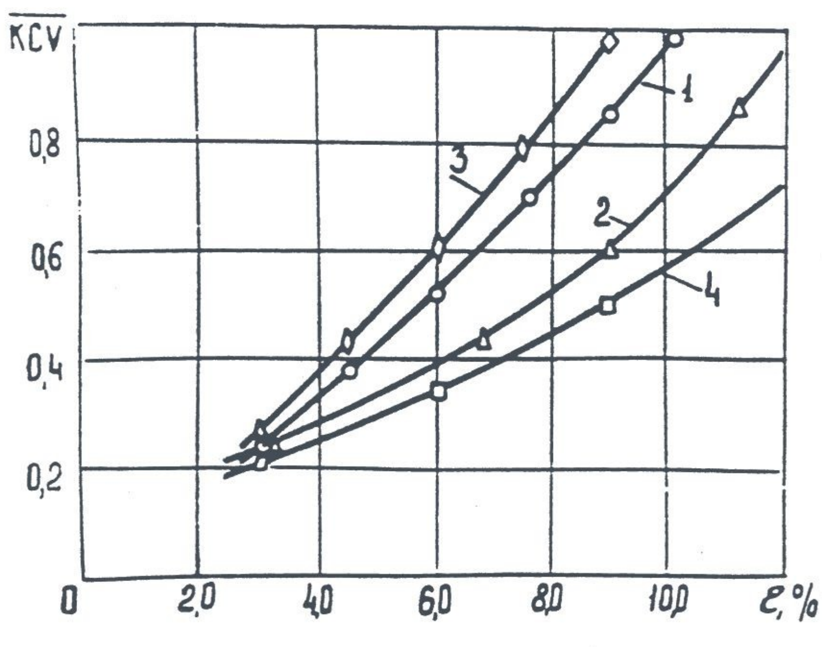


Fig. 27a,b

Dependence KCV(t_W) (a) and KCV(ε) (b) for combinations (R_Z 10,0 + R_Z 0,05) OT4-1 + OT4-1 (1,3) and BT3-1 + OT4-1 (2,4) at T_W = 900 °C and ε= 0,82·10^{-5} s^{-1} (3,4) and 2,5·10^{-5} s^{-1} (1,2)



VI. Conclusion

The analysis of special features of realization of volumetric interaction stage has determined that weldability of different titanium alloys has substantial differences in comparison with similar titanium alloys.

There have been analyzed the special features of a physical contact formation for various groups of different combinations of titanium alloys in comparison with similar combinations.

It has been shown that the process rate is lower for DC than for SC. This results in the growth of accumulated macrodeformation in the first case.

The determined tendency caused by realization of Frenkel effect and structure herdening in the contact area is intensified in accordance with the growth of chemical heterogenuity in the contact area and also depends on plasticity of welded materials at welding temperature.

There has been studied the influence of heteroalloying on the welded joints properties in case of different duration of thermal influence in the contact area. The obtained data were analyzed in accordance with the special features of a physical contact formation and the recommendations on optimization of thermodeformation cycle of welding for different groups of different titanium alloys were given.

Captions

Fig. 1 Dependence KCV(t_W) for combinations OT4-1 + OT4-1 (1) and OT4-1 + BT3-1 (2)

Fig. 2 Structure of welded joint area of SC (a) and DC (b) after welding at T_W = 900 °C, t_W = 75 min

Fig. 3 Welded specimens

Fig. 4 Dependence KCV (t_{an}) for combinations OT4+OT4 (1', 2', 3') and OT4+BT4 (1, 2, 3) at T_{an} °C: 3,3' – 850, 2,2' – 900, 1,1' – 940

Fig. 5 Dependence KCV – 1/T for DC (1) and SC (2)

Fig. 6 Dependence F(t_W) (a) and F(ε) (b) in case of welding of combinations (R_Z 10,0 + R_Z 0,05) OT4-1 + OT4-1 (1,3) and OT4-1 + BT3-1 (2,4) with different deformation rate ε', C^{-1}; $0,82\cdot10^{-5}$ (1,2)

Fig. 7 Welded joint area structure OT4-1 + OT4-1 (a,b) and OT4-1 + BT3-1 at t_W = 30 min. (a,b) and t_W = 150 min. (c,d), x1000

Fig. 8 Dependence F(t_W) in case of welding (T_W = 940°C, ε= $2,0\cdot10^{-5}\ c^{-1}$) of combinations (R_Z 10,0 + R_Z 0,05) OT4 + OT4 (1) and OT4 + BT3-1 (2)

Fig. 9 Dependence F(t_W) in case of welding (T_W = 940°C, ε= $1,0\cdot10^{-5}\ c^{-1}$) of combinations (R_Z 1,5 + R_Z 0,05) OT4 + OT4 (1) and OT4 + BT3-1 (2)

Fig. 10 Welded joint areas structure (T_W = 857°C, t_W = 30 min.) BT1-0 + OT4-1 (a), BT1-0 + BT3-1 (b) and BT1-0 + BT16 (c) x500

Fig. 11 Dependence F(R_Z) (where R_Z – middle height of microprojections of turned surface) in case of welding T_W = 940 °C, t_W = 40 min., P = 2,0 MPa) of OT4 alloy

with different combinations of contacting of small grain (S) and large grain (L) structures:
1- ST+SP; 2- LP+LT; 3- SP+LT; 4- ST+LP

Fig. 12 Dependence $F(t_W)$ in case of welding (T_W = 875°C, ε = $1,0\cdot10^{-5}$ s^{-1}) of combinations (R_z 10,0 + R_z 0,05) BT3-1 + BT3-1 (1) and BT3-1 + BT1-0 (2)

Fig. 13 Dependence $F(t_W)$ (a) and $F(\varepsilon)$ in case of welding (T_W = 900°C) of combinations (R_z 10,0 + R_z 0,05) OT4-1 + OT4-1 (1) and OT4-1 + BT16 (2,3,4) with different deformation rate ε, s^{-1}: $1,22\cdot10^{-5}$ (3), $27,0\cdot10^{-5}$ (4)

Fig. 14 Welded joint structure BT3-1 + OT4-1 at T_W = 940 °C, t_W = 30 min., x 1000

Fig. 15 Welded specimens: 1- specimens; 2- welded seams (welds); 3- titanium capsules

Fig. 16 Dependence $KCV(t_{an})$ for BT1-0 + BT5 combination after welding and annealing at T_W = T_{an} = 875 °C

Fig. 17 Dependence $KCV(t_{an})$ for combinations OT4-1 + BT3-1 (1-4) and OT4-1 + OT4-1 (5) after welding (T_W = 900 °C) and annealing at T_{an} °C: 1 – 850; 2,4,5 – 900; 3 – 940; BT3-1 -initial small grain structure; OT4-1 -small grain (1-3,5) and large grain (4) structure

Fig. 18 Dependence $KCV(t_{an})$ for combinations OT4 + BT14 (1) and OT4 + BT3-1 (2) after welding and annealing at T_W = T_{an} = 940 °C

Fig. 19 Welded joint structure of OT4-1 + BT14 combination after welding (T_W = 900 °C, x 500 – a) and annealing T_{an} = 900 °C, t_{an} = 64, x 1000 – b)

Fig. 20 Dependence $KCV(t_{an})$ for OT4 + OT4 (1) and BT14 + BT3-1 (2) after welding and annealing at 900 °C (1) and 940 °C (2)

Fig. 21 Dependence $KCV(t_{an})$ for BT1-0 + BT3-1 combination after welding (T_W = 875 °C) and annealing at T_{an} °C: 1- 850; 2- 875; 3- 900.

Fig. 22 Welded joint zone structure of BT1-0 + BT14 (a,b) and BT1-0 + OT4-1 (c,d) combinations after welding and annealing $T_W = T_{an}$ = 850 °C, t_{an} = 6h; a,c (x100) – etching: b, (x 1000), d (x 500) – electropolishing

Fig. 23 Dependence KCV(t_{an}) for combinations BT1-0 + BT16 (1), OT4-1 + BT16 (2,3), OT4 + BT16 (4), BT15 + BT16 (6). Welding and annealing at T °C: 850 (3), 875 (1), 900 (2,4,5,6).

Fig. 24 Welded joint zone structure OT4-1 + BT16 (R_Z 0,05 + R_Z 10,0) at T_W = 900 °C, t_W = 2h; ε = 10%; x1000.

Fig. 25 Dependence KCV(t_W) in case of DC welding BT16 + OT4-1 (R_Z 0,05 + R_Z 10,0) at ε, C^{-1}: $1,22 \cdot 10^{-5}$ (1), $7,0 \cdot 10^{-5}$ (2), $27,0 \cdot 10^{-5}$ (3)

Fig. 26 Dependence KCV(t_W) for DC OT4 + BT3-1 (2,4) and SC OT4 + OT4 (1,3) at T_W = 940 °C: a- R_Z 10,0 + R_Z 0,05 at ε, C^{-1}: $2 \cdot 10^{-5}$ (1,2) and $3,5 \cdot 10^{-5}$ (3,4); b- R_Z 1,5 + R_Z 0,05 at ε = $0,8 \cdot 10^{-5}$ s^{-1}

Fig. 27 Dependence KCV(t_W) (a) and KCV(ε) (b) for combinations (R_Z 10,0 + R_Z 0,05) OT4-1 + OT4-1 (1,3) and BT3-1 + OT4-1 (2,4) at T_W = 900 °C and ε= $0,82 \cdot 10^{-5}$ s^{-1} (3,4) and $2,5 \cdot 10^{-5}$ s^{-1} (1,2)

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