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TRANSFORMATION OF STRUCTURE DURING FRICTION STIR WELDING

Summary. In comparison with low carbon steels, there is increased interest in the use of aluminium-based alloys as materials for the manufacture of welded structures rolling stock of railway transport. During friction stir welding aluminium-based alloy, against the background of the analysis structural transformations, issues of development hardening processes are considered. Under conditions of existence, a temperature gradient at zone of weld formation, shown degree approximation alloy to the conditions of superplastic flow and influence from presence particles of the second phase on grain size of matrix is estimated. Evaluation of the separate influence grain size of matrix and state of solid solution at total hardness of the weld showed dependence of their contributions on temperature of hot plastic deformation. As the temperature of plastic

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deformation of alloy at area of the weld increases, contribution to the total hardness from grain size increase and on state of the solid solution decreases.
Keywords: chemical compound, grain size, hardness, recrystallisation, temperature

1. INTRODUCTION

The high activity of aluminium-based alloys to oxygen atoms significantly limits their use for the manufacture of structures using the electric arc welding technology. In the last few decades, for the manufacture of welded structures from aluminium alloys in aviation [1], automobile [2] mechanical engineering and railway transport [3], friction stir welding (FSW) [4-7] has become widespread. A distinctive feature of this technology is the absence of change at state aggregation of the metallic material. The technology of FSW is based on the achievement of the area of the connect edges high degree mixing of the metal material from the simultaneous action of hot plastic deformation and rate of diffusion [8-10]. Indeed, if degree at mixing of the metal and density of the weld is determined by the plastic properties, the characteristics of strength and crack resistance would depend on the development of structural transformation after finish active phase connected of edges. For most aluminium alloys with a multiphase structure, the volume fraction, dispersion and morphology of the phase components influence the development of recrystallisation processes and the final grain size of the matrix. On the other hand, most chemical compounds of aluminium alloys are partially interacting with matrix [11-13], which involves their dissolution with increasing heating temperature of the alloy. Consequently, with increasing heating temperature and duration work tool, the greatest possible degree plasticisation of the alloy for the implementation of FSW technology [14-16], dissolution of the chemical compound should increase mobility boundaries of the matrix grains with large angles of disorientation and increase size of grain [12, 14]. The above provisions are confirmed by the high sensitivity of stir friction welding to the temperature and speed of hot plastic deformation [18]. Moreover, small deviations from their optimal ratio significantly complicate the process of forming a quality welded joint [9, 14, 19].

2. STATE OF PROBLEM

The special shape of the working tool ensures transition of the metal along the entire thickness of the connecting edges in such a plastic state, which should be sufficient for its quality mixing during weld formation. With increasing temperature, decrease resistance to the propagation of plastic deformation is not always accompanied by an increase at plasticity of the metal material. This position is due to the implementation of a certain ratio between the development of the hardening and softening processes during hot deformation. To achieve high ductility, in addition to proposals to maintain a uniform distribution of deformation [18, 19], it would, analysing the mechanism of structural changes, achieve and maintain stable conditions for the propagation of super plastic deformation at alloy. Further, in addition to increasing the plastic properties by tens to hundreds times [19], there is a decrease at deformation stress to abnormally low levels. In general, the conditions for achieving a superplastic state correspond to the relationship:

$$m = \frac{\sigma - \sigma_1}{2\sigma}, \tag{1}$$

where σ - deforming stress, m - a coefficient sensitivity to rate of deformation, σ_1 - characteristic, which by absolute value and physical meaning corresponds to stress of non-inverse motion of dislocations [20]. On the other hand, the value σ has a certain dependence on the degree of plastic deformation (ε):

$$\sigma = \sigma_1 + K \cdot \varepsilon^n, \tag{2}$$

where: K - coefficient of proportionality, n - the exponent (coefficient of strain hardening [21]). After substitution (2) in (1), by the ratio:

$$\sigma = \frac{K \cdot \varepsilon^n}{2m} \tag{3}$$

it is possible to determine the contribution of strain hardening on deviation from the achievement conditions of the superplastic state in metallic material. Thus, when approaching the superplastic flow, the structural transformations during hot deformation should compensate for the negative contribution of deformation hardening. As you approach the specified state of the metal material σ will be decrease, the corresponding decrease will be for σ_1 and n. Thus, when $\sigma_1 \to 0$ and $n \cong 0$, K will be equal to the operating stress. By the relation (3) maximum value of m cannot exceed 0.5. In fact, given the high sensitivity to small deviations from the optimal ratio parameters of hot deformation (temperature, strain rate, grain size, invariance, etc.), this characteristic will be at range $0<m<0,5$. In addition to the temperature and deformation conditions of the superplastic flow, the issues of achieving optimal structural state metallic material and maintaining it constant during deformation period are of great importance. At process of hot plastic deformation, effective stress depends on the grain size (d) according to the ratio:

$$\sigma = Ad^a, \tag{4}$$

where: a is an exponent (for most metallic materials, it is approximately equal to 1 [22]), A is constant [23]. By (4) the decrease in σ will be approach to superplastic state will be provided by a decrease d. Then, it is quite reasonable to obtain a stable structure during hot deformation, with the small grain of polyhedral shape [18, 21], in sizes to several microns. The need to maintain a constant grain structure is due to a change in the nature of the deformation distribution from grain size during the superplastic flow. The smallest grains are predominantly deformed along near boundaries with large angles of disorientation, at low level σ. In large grains, the deformation is distributed throughout the volume, which requires a continuous increase applied stress [20]. Such different conditions can only be met by maintaining a certain ratio between the hardening and the softening processes during hot plastic deformation. Indeed, at temperatures below optimum value, when the plastic properties of the metal become insufficient to achieve desired level of mixing, traces of the working tool become coarser, with obvious signs of exfoliation dispersed fragments [15, 17]. Under conditions when the metal is overheated to temperatures higher than optimal value, as shown in [14, 16], the zone influence of thermal and region of the weld have strength properties of a much lower level compared to the initial state. Then, preservation of the partial influence from development of deformation hardening processes during the propagation of plastic flow can be a factor of increasing strength of metal in the weld to level by initial state [10, 14]. The question of mechanism propagation of plastic deformation under conditions superplastic state remains open. According to [9, 19], it is believed that over plastic flow should be accompanied by rotation of randomly oriented grains relative direction of the applied stress or alternation of slippage in volumes near the grain boundaries, with subsequent migration of boundaries.

3. MATERIAL AND RESEARCH METHODS

As a material for research, used alloy based on aluminium with a content of Mg 6.5%, Mn 0.85%, Fe 0.13%, a rest of Al. Plates 2.5 mm thick was butt welded by FSW technology, on a specially designed research stand, using a working tool of known size and shape of the shoulder and pin [8, 14, 19]. Investigations of the welding process were carried out at different ratios of rotational speed working tool (ν) and force of its pressing to the connecting edges (*P*). The speed of movement working tool along connecting edges was constant and was 40 mm/min, the change interval ν was 800-1600 min^{-1}, and *P* was 100 - 1420 N. The temperature distribution (*T*) by heating edges of the metal was measured using thermocouples such as chromel - alumel, with location at different distances from the connecting edges. The structure of the alloy was examined using a light microscope "Epiquant". Size of grain alloy was determined by the use of methods quantitative metallographic. Measurements of microhardness (H_μ) were performed on a device type PMT-3, under load on the indenter 0.05N. Measurement data H_μ was used as a characteristic of strength properties for micro volumes of the alloy.

4. RESULTS AND DISCUSSION

In general, the process of FSW can be divided into two successive stages. From the beginning, a pin and shoulder are immersed at alloy under load, without moving the tool along the edges. According to the scheme (Figure 1a), the action of pressing force for a certain duration leads to a corresponding increase at heating temperature of the edges on the opposite side from the surface shoulder of the instrument (Figure 1b). The first stage ends with heating the edges to a temperature that provides the required level of plastic properties for high-quality mixing of the metal. The second stage starts from the moment the tool is moved along the connecting edges. In proportion to the speed of the weld formation, the energy consumption for continuous heating of cold volumes of metal first with the shoulder, and then with the tool pin leads to the emergence gradient of temperature and decrease of average temperature on the thickness of the edges. This is confirmed by the decrease in temperature from extremes in Figure 1b. To analyse the combined effect of the main technological parameters (*P* and ν) on heating temperature metal of the connecting edges, dependences $P \sim f(T)$ were constructed using experimental data $P,T = f(\tau)$ (Figure 1a, b). According to the curves (Figure 1c), it is determined that regardless of the value ν, moment of violation directly proportional relation $P \sim f(T)$ corresponds to approximately the equal temperature of 90-95°C. Generally, it should be assumed that at this temperature, the process of softening begins to compensate for a significant part resistance of the edges to immersion of the tool (strengthening alloy from hot plastic deformation). Indeed, with the increase of hot deformation temperature, processes moving of dislocations on different crystallographic sliding systems are significantly accelerated, which promotes their annihilation and redistribution at alloy matrix [20]. As a result, dislocation subgrains of polygonal shape are formed, separated by subboundaries of different types, with different angles of disorientation [19, 22]. It is experimentally determined that the sharp decrease at deforming force (Figure 1b), especially at elevated ν, is actually due to the rapid heating of the thin layer of metal under by shoulder, with simultaneous occurrence of a high temperature gradient by thickness of connecting edges. To determine factors of influence that lead to a violation of proportional nature at curves $P \sim f\,(T)$ (Figure 1c), the possibility of recrystallisation at alloy was evaluated. For this purpose, the known relation $T_R \approx 0{,}4T_S$ was used, where T_R and T_S - is the temperature of development static recrystallisation and melting of metallic material, respectively. After substitution T_S investigated alloy, the calculated value of T_R was 100°C. Thus, under certain assumptions, the moment of deviation from proportional of curves $P \sim f\,(T)$ equal 90-95°C should correspond to the conditions at the beginning of the development of the recrystallisation processes. In comparison with single-phase alloys, the presence of structure particles at the second phase can significantly change the influence of hot plastic deformation on the development processes of primary and secondary recrystallisation. On the other hand, the effectiveness of this effect on the grain size of the matrix can be limited stability of existing particles second phase at hot deformation temperatures (0.75-0.85 at FSW [8-10]. Given the presence of alloying chemical elements at alloy, the option of forming only binary chemical compounds was considered to simplify evaluation. According to the state diagrams of *Al-Mg* [11], *Al-Fe* [12] and *Al-Mn* [13] possibility of dissolving particles of the second phase at FSW temperatures was analysed. Thus, by the *Al-Mg* diagram (Figure 2a), a chemical compound Al_2Mg (β-phase) must be interacting with the matrix. For a *Mg* concentration of 6.5%, the volume fraction of β-phase, considering the solution of 2.95% of *Mg* atoms in α-solid solution, should be approximately 9-9.2%. In the process of isothermal exposure, at temperatures of approximately 375-380°C, one should expect beginning solution Al_2Mg particles (Figure 2a). However, the small degree of overheating from the phase equilibrium temperature indicates the need for a long enough exposure to complete dissolution. Analysis of the state diagrams of *Al-Mn* [10] and *Al-Fe* [12] (Figure 2b) showed the presence volume fraction of particles of Al_6Mn - 3.7%, with a solution above 500°C [13] and Al_3Fe - up to 0.6% the stably existing up to melting points of the alloy [12] (Figure 2b). At a speed of movement working tool along connecting edges of 40 mm/min, overcoming distance of 12 mm (diameter of a shoulder) occurs in 17 s. Thus, the total duration alloy will be under influence at weld temperatures should be divided into effects action from the shoulder and the rod. At the beginning, heating from a shouldered with a high temperature gradient, only one surface each of the edges with a width of 6 mm for 8-9 s to a temperature of about 440°C. The temperature opposite surface of the edges is lower by about 25%. Action of the pin will be lead to additional heating (especially root of the seam) and rapid mixing alloy on the thickness connecting edges to equalise temperature to an average value of 380-375°C at a distance 3-4 mm from the pin.

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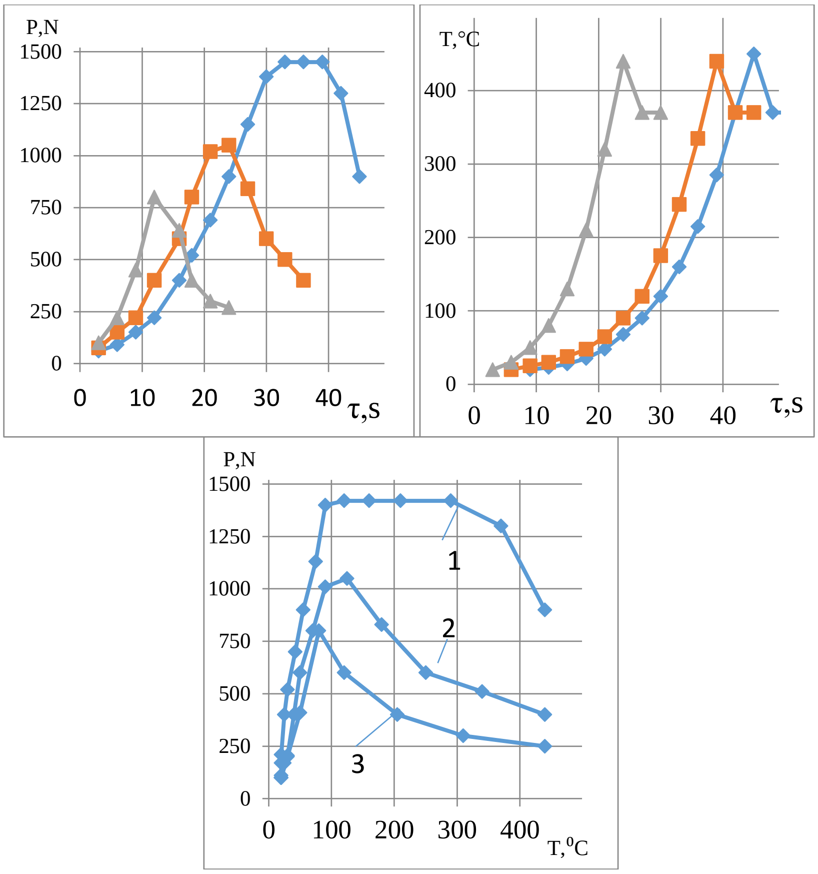
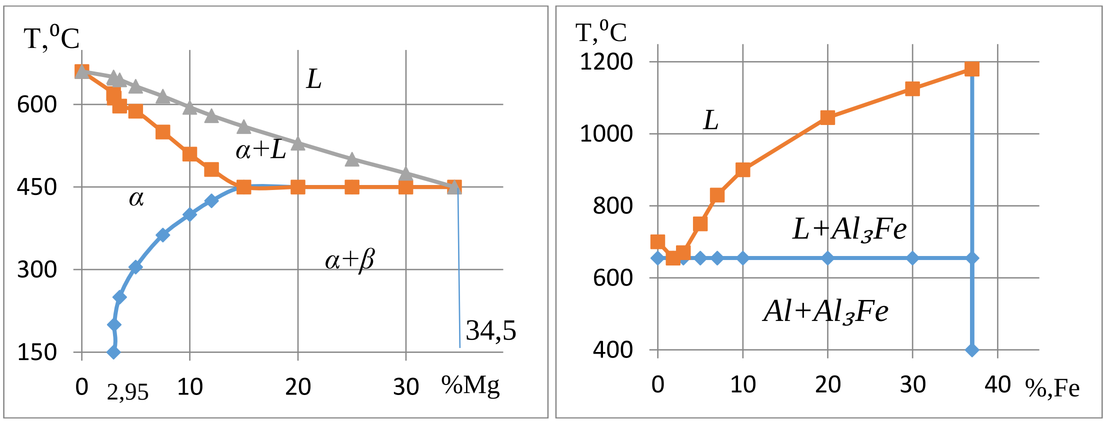


a b c

Fig. 1. Influence ν (♦ -800, ■ -1240, ▲ -1600 min $^{-1}$) on the change at resistance edges to immersion of the tool (a), the heating temperature edges on the opposite side of the shoulder (b) depending duration of the tool action and the ratio $P \sim f\,(T)$ for different ν (1-800, 2-1240, 3-1600 min $^{-1}$) (c)



a b

Fig. 2. Parts of state diagrams binary alloys *Al-Mg* - (a) and *Al-Fe* - (b)

Then for the same 8-9 s, there is a further decrease at the temperature gradient along the cross-section of the seam due to the distribution of metal from the shoulder. It follows from the above analysis that during the period of tool operation time, there is not enough heat for a complete dissolution of the particles of the chemical compounds. The presence of the second phase at structure should be considered as a factor that will help maintain a constant grain size of the alloy matrix during FSW. Stability of the structure during the development of secondary recrystallisation is determined by the conditions of compensation driving force of grain boundary movement by effect of inhibition on their motion by particles of the second phase. By these conditions, grain size (*d*) of alloy matrix on particle diameter second phase (*D*) and volume fraction (*f*) will be determined by the dependence [24]:

$$d = A\frac{D}{f}, \tag{5}$$

where *A* is a coefficient.

For alloys, when $f > 8\text{-}10\%$ $A = 2/3$ [24]. Using relation (5), it is possible to determine the factors influencing *d* under the conditions of FSW development. The stability of the structure as the alloy approaches the superplastic state will be determined by the coalescence of the particles of the second phase, their volume fraction and the possible solution in the matrix. In its original state, the grain size of alloy matrix was equal to 15-17 μm. Areas with different degrees mixing of metal edges were selected for the analysis of structural transformations of FSW. Figure 3 shows the structures of alloy volumes in which temperature, microhardness and grain size were measured. When forming a weld, the occurrence of a significant difference in temperature and degree of plastic deformation near the tool rod, lead to the formation of very complex gradient structures. In this case, there is a significant change not only at average grain size, but also its shape. When approaching surface near tool rod, the temperature 360°C and the degree of plastic deformation is sufficient for formation in alloy grains of relatively small size, with a shape close to a polygon (Figure 3d). On the other hand, a slight increase in distance leads to the expected decrease in temperature, for example, to 345°C (Figure 3c), and accordingly, degree of hot plastic deformation. A characteristic feature of this area is the increased structural heterogeneity of the formed grains. One of the reasons for the existence of such a structure to be considered incomplete is the completion of the recrystallisation processes [20]. For volumes of alloy with lower temperature (Figure 3b), a structure with an even larger interval of grain size change is observed. The location studied micro-volume relative to the curved portion of weld (Figure 3a) indicates a more significant influence on structure transformations from tool pin to shoulder. For average values of *d* (10-16 μm) and temperature range 330-360°C at $f \approx 13\%$, the conditional diameter particles of the second phase was estimated by relation (5). The calculated *D* values were in the range of 2-3 μm, which coincides with known data [20, 22]. Given the possible solution in particles of chemical compounds at studied temperature range (Figure 2) in FSW, one should expect an increase at concentration of alloying chemical elements in α - solid solution after cooling alloy. Then, after formation weld and its cooling, in the layers of alloy with a high degree mixing should change the effect of hardening from the state of solid solution. To confirm this is used the Hall-Petch type ratio [25], with H_μ as a characteristic of strength:

$$H\mu = H\mu_i + k_y d^{-0,5}, \tag{6}$$

where $H_{\mu i}$ is the expected microhardness of alloy matrix at absence grain boundaries with large disorientation angles [25], k_y is the angular coefficient. Given the lack of a clear boundary in the structure between the three zones (Figure 2b), it was decided to analyse only for two areas with guaranteed temperatures of 330 and 360°C. The result of applying H_μ against the corresponding values of d is shown in Figure 4. Execution of this ratio indicates that level H_μ is determined by the grain size of alloy matrix and the state of the solid solution, with virtually no contribution from substructure elements, etc. The effect on level H_μ from the particles of the second phase is almost absent, due to their location along the grain boundaries with large angles of disorientation. The fact is that when the particles of the second phase have a predominantly location at middle of grains, conditions of relation (6) will be violated due to transition of the role of the main structural element from d to distance between the particles [20, 21, 24]. According to estimates H_μ and k_y, it is determined that when temperature changes from 330 to 360°C, contribution to the total level of microhardness from solid solution (H_μ) decreases from 400 to 110 MPa, and resistance of deformation propagation from boundaries of grain (k_y) will increase more than twice (from 2,5 to 6 N mm$^{-1,5}$). Thus, with increasing temperature at FSW effect of hardening from the state of solid solution decreases, while the grain boundaries increases. Comparative analysis H_μ for the equal d but at different temperatures at FSW shows a violation of the known ratio, when under normal conditions of deformation decrease temperature increases the hardening effect, and reduces the degree of softening alloy. Thus, for d = 8 µm, at 330°C, H_μ is less than 700 MPa, and for 360°C on 15% higher ($H_\mu^{330} \prec H_\mu^{360}$). For a big d, different relationship is obtained. For example, at d = 28 µm, H_μ = 520 MPa for a temperature of 330°C, and for 360°C, H_μ = 480 MPa ($H_\mu^{330} \succ H_\mu^{360}$). Given that with increasing d, the parameters of strain hardening increase [21], to the known requirements for achieving over superplastic flow, in addition to maintaining a stable structure with over fine grain, it should further minimise influence from the state of solid solution.

5. CONCLUSIONS

1. At temperatures above phase equilibrium, the partial solution of particles chemical compounds is a source of supersaturation solid solution by atoms alloying elements under FSW.
2. At the formation of weld, the presence in structure particles of the second phase is a factor in inhibiting movement grain boundaries at the period of structure formation.
3. Increasing temperature during FSW, accompany by decrease contribution from state of solid solution and increased role of grain boundaries to the overall hardness level of the weld.

Gratitude

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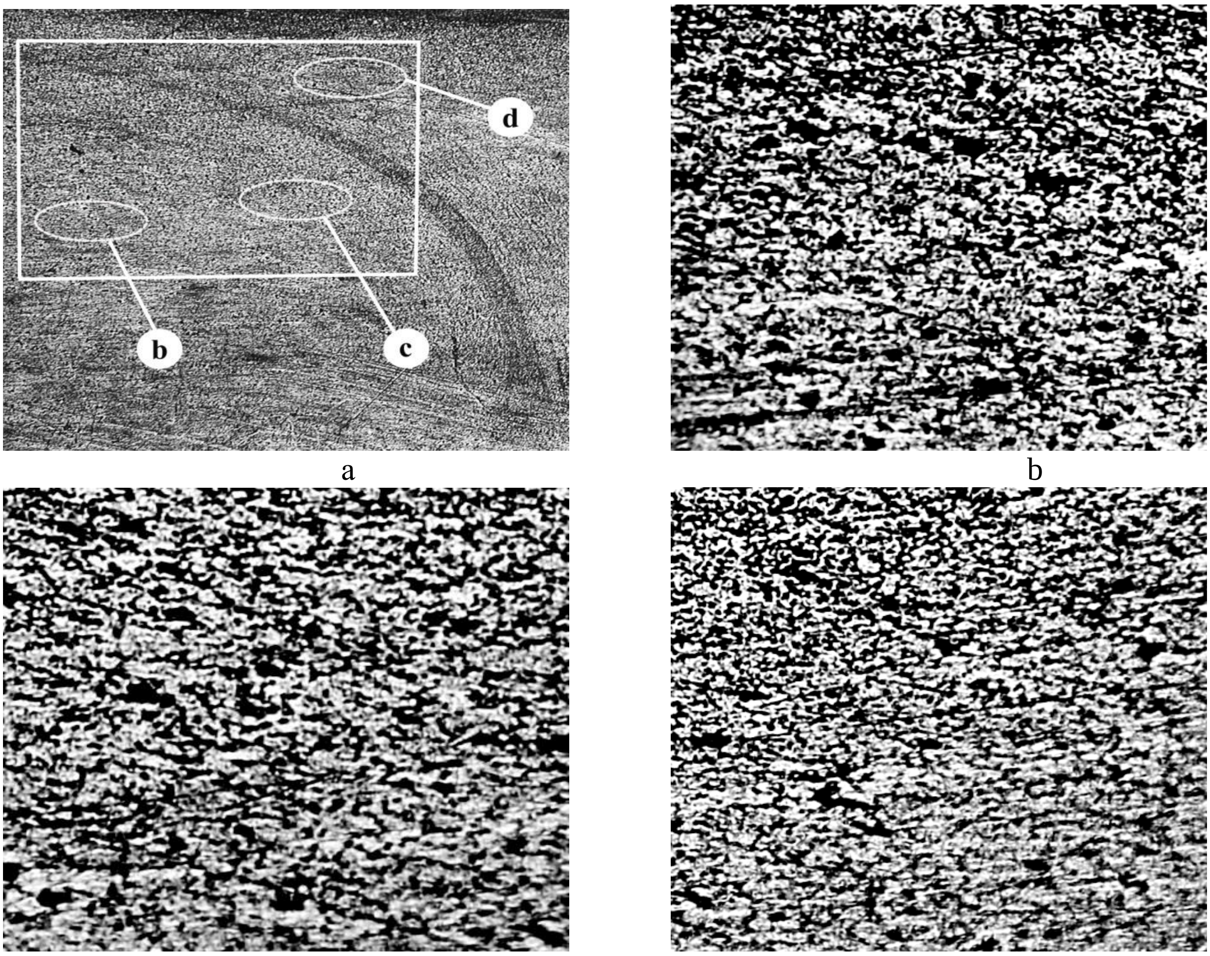

Fig. 3. General view of the area intensive mixing after FSW (ν = 1250 min $^{-1}$) (a) and enlarged image of the selected fragment depending on heating temperature: (b) -330, (c) -345 and (d) -360°C. Magnification: a - 100; b-d - 350

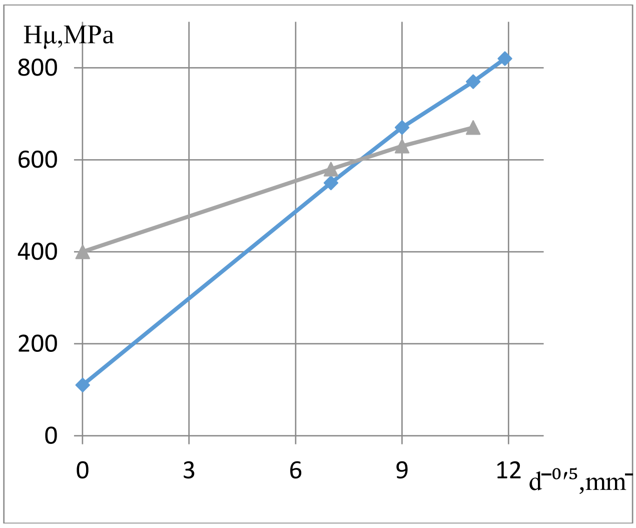

Fig. 4. Influence of the size of grain α - solid solution on microhardness, depending on the temperature of the microvolumes alloy at FSW (▲ - 330, ♦ -360°C)

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