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SUPERPLASTIC BEHAVIOR OF 1933 ALUMINUM ALLOY WITH BIMODAL STRUCTURE AT ELEVATED TEMPERATURES

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The features of the structural state, phase composition, and deformation relief of the working part of the alloy 1933 specimens are studied. It is determined that the grain-boundary sliding intensively occurs both along the boundaries of ultra-fine grains and along the boundaries of large polygonized grains parallel to the strain direction of the specimen.

INTRODUCTION

It is known that superplasticity is a structurally sensitive property [1, 2]. Among the structural factors influencing on the manifestation of this effect the main are the size and shape of grains, the relative cavity volume, texture, the presence and the kind of intermetallic particles, the state of the grain boundaries. Changing of these characteristics during superplastic flow can significantly effect on the strain rate of specimens and on the value of their elongation to failure. It is generally accepted that there is no special deformation mechanism under the conditions of superplasticity, but the same mechanisms act as under normal hot deformation conditions. Superplasticity is characterized, above all, by a special combination of the contributions of various mechanisms to the overall deformation. The greatest contribution to the development of superplastic deformation is made by grain-boundary sliding. Usually this term denotes a deformation process leading to the displacement of one grain to another along the common surface of the grain boundary. Therefore, grain boundaries are considered as the main structural element in the effect of superplasticity. It is generally considered that the process of grain-boundary sliding develops most actively in microcrystalline alloys at high-angle grain boundaries oriented in the direction of the maximum tangential stresses under the conditions of superplasticity. The main distinguishing feature of structural superplasticity is that it manifests when the grains are stably uniform and ultra-fine during the deformation process [1, 2]. However, there are a number of publications on the manifestation of superplasticity by coarse-grained aluminum alloys [3, 4], by aluminum alloys with elongated grains [5, 6] as well as by alloys with a bimodal structure [6, 7]. It is therefore is interesting to study the processes of grain boundary sliding development in aluminum alloys with different types of grain structure.

At present, multicomponent high-strength aluminum Al-Zn-Mg-Cu-Zr alloys are widely used in various industries. This is due to their higher specific strength as compared to conventional high strength materials [8]. High-strength forging alloy 1933 of system Al-Zn-Mg-Cu-Zr is widely used in aircraft industry. In particular, it was used for the manufacture of large forgings of the aircraft AN-225 "Mriya" [9]. This alloy has a high fracture toughness, and is superior in this respect to alloys of similar purpose 7050, 7175, and 7040 [8, 9]. Aluminum-based alloys are the main structural material not only in aircraft and rocket construction, but also in nuclear power engineering. In particular, centrifuges for enrichment of ^{235}U are produced from aluminum alloy B96c. On the base of alloy B96c, an alloy 1933 was developed. It is the most promising of the alloys of the Al-Zn-Mg-Cu-Zr series in combination with its corrosive and mechanical properties. Meanwhile, low technological plasticity of high-strength aluminum alloys significantly limits their wide application in the production of thin-walled products. This disadvantage can be eliminated by using the effect of superplasticity in the operations of the materials forming. In this paper, we studied the features of the development of superplastic deformation and, in particular, grain-boundary sliding in the aluminum alloy 1933 of Al-Zn-Mg-Cu-Zr system with a coarse-grained bimodal structure.

1. MATERIAL AND EXPERIMENTAL

The chemical composition of the investigated alloy is Al-6.35…7.2 wt.%Zn-1.6…2.2 wt.%Mg-0.8…1.2 wt.%Cu-0.1…0.18 wt.%Zr-0.15 wt.%Fe-0.1 wt.Si-0.05 wt.%Cr [8, 9].

In this study, specimens with sizes of the gage portion of 4.5×2.5×10 mm were mechanically tested. The specimens were mechanically tested by tension in the creep regime in air under a constant flow stress according to the technique described in [10]. The time of specimen heating to the test temperature was not higher than 25 min. The test temperature was held with an accuracy of 2 K.

The microstructures of the undeformed specimens were examined by the electron back scattered diffraction (EBSD) technique. The grain structure and morphology of cavities in the specimen were investigated by optical microscopy and by scanning electron microscopy (JEOL JSM 840 microscope), as well as by conventional methods of quantitative metallography. Energy dispersive X-ray microanalysis was held using the JEOL JSM 840 scanning electron microscope equipped with technique for energy dispersive X-ray microanalysis. Fracture mechanisms of the specimens were determined according to the type of fracture views.

2. RESULTS AND DISCUSSION

2.1. TENSILE DEFORMATION

The tensile tests were performed under the creep conditions at a constant flow stress that enabled to maintain equal deformation conditions during the experiment. The tests were performed at a temperature ranging from 753 to 813 K at stresses from 3.0 to 7.0 MPa.

To build the temperature dependence of the plasticity of the Al-Zn-Mg-Cu-Z alloy specimens for each of the investigated temperatures, the tests were performed to determine the maximum elongation to failure δ_{max}. The elongation to failure δ as a function of the applied stress σ was studied for each temperature.

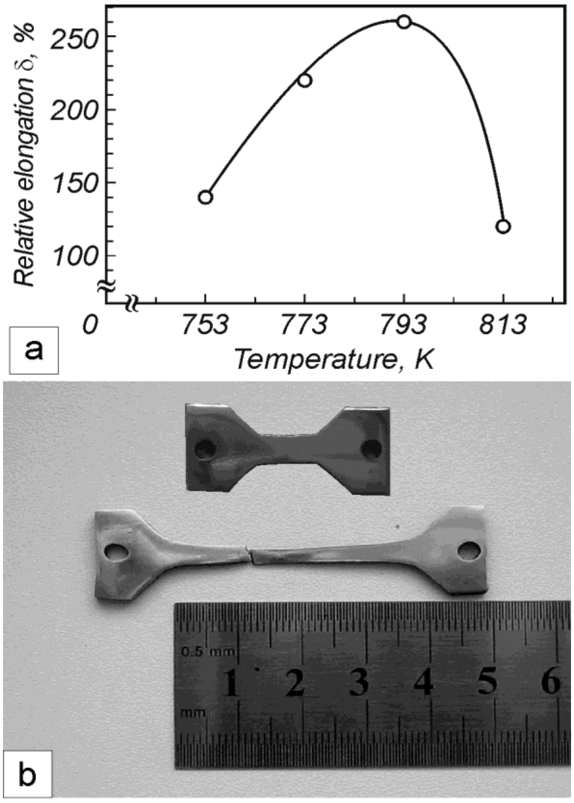

Fig. 1. Elongation to failure δ, as a function of the test temperature T for the Al-Zn-Mg-Cu-Zr alloy (a), the general view of the initial specimen and a view of the specimen deformed to failure under the optimal conditions (b)

The maximum values of the elongation to failure δ_{max}, corresponding to the optimum conditions of superplasticity, as a function of the test temperature T for the Al-Zn-Mg-Cu-Zr alloy are given in Fig. 1,a.

Elongation increased with an increase of the test temperature until it reached the maximum value at 793 K, and then decreased as the temperature increased. A maximum elongation of 260% was achieved at 793 K and true strain rate $1.2 \cdot 10^{-4}$ s^{-1}.

Fig. 1,b is a general view of the initial specimen of the Al-Zn-Mg-Cu-Zr alloy and a view of the specimen deformed to failure under the optimal conditions of superplasticity. It is seen that the specimen deformed without necking.

2.2. INITIAL MICROSTRUCTURE

Fig. 2 shows a micrograph of the typical view of the initial microstructure of Al-Zn-Mg-Cu-Zr alloy specimen obtained using light microscopy methods. It is found that the microstructure of the alloy is bimodal. It consists of sections containing a large number of recrystallized fine and ultrafine grains with an average grain size of 7 μm. It also contains a number of large elongated polygonized grains with an average grain size of 50 μm.

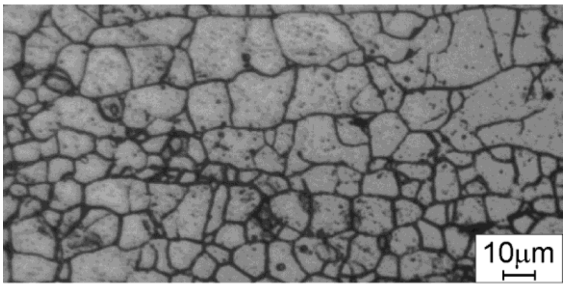

Fig. 2. Micrograph of the typical view of the Al-Zn-Mg-Cu-Zr alloy initial microstructure

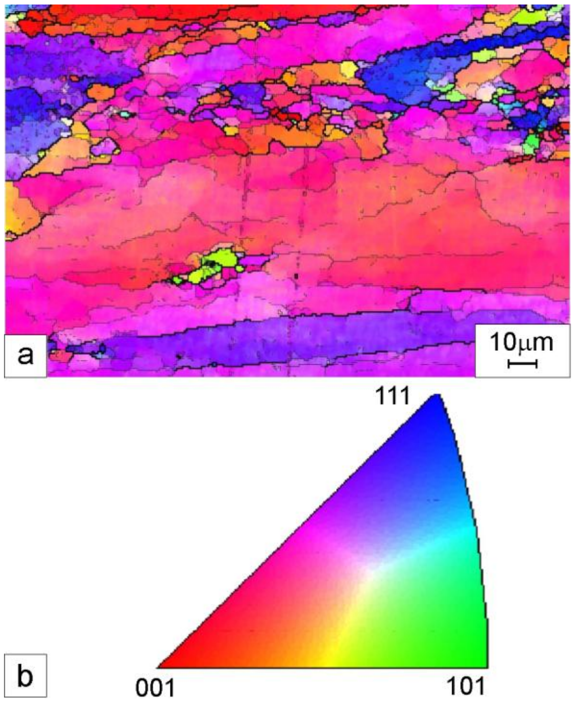

Fig. 3. The map of grain orientations in the investigated section of the Al-Zn-Mg-Cu-Zr alloy specimen (a), and an image of the legend to it, which were constructed in the space of inverse pole figures (b)

To determine the misorientation angles of the grain boundaries and to estimate their quantitative content in the bimodal structure of the Al-Zn-Mg-Cu-Zr alloy, a diffraction technique for backscattered electrons was used. Fig. 3,a shows a map of grain orientations in the investigated section of the Al-Zn-Mg-Cu-Zr alloy specimen, and Fig. 3,b shows an image of the legend to it, which were constructed in the space of inverse pole figures. It is seen that the grain orientation in this section is not uniformly distributed. For ultrafine grains orientations tending to (111) and (001) are prevailing, and for large polygonized grains orientations, tending to an orientation (001) are prevailing. There is also a certain number of grains, the orientation of which tends toward orientation (101) on the orientations map.

Fig. 4 shows the quantitative distribution of the grain boundaries on the misorientation angles. It was built as a result of taking into account of all the attested grain boundaries available in the investigated section of the Al-Zn-Mg-Cu-Zr alloy specimen.

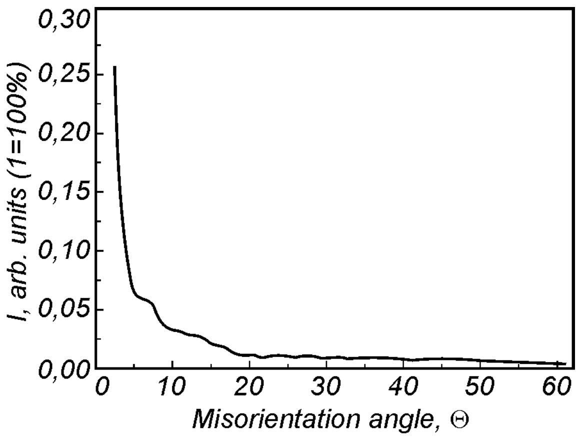

Fig. 4. The quantitative distribution of the grain boundaries on the misorientation angles

In building of this dependence, it was accepted to refer to the law-angle grain boundaries those grain boundaries that have a misorientation angle below 10°, and to the high-angle grain boundaries – those grain boundaries that are disoriented above 10°. It is found that the specific proportion of the low-angle grain boundaries for the tested specimens' area is 65.5% and the specific proportion of high-angle grain boundaries is equal to 35.5%. These data convincingly indicate that the initial structure of the Al-Zn-Mg-Cu-Zr alloy specimens is not fully recrystallized, and that a developed substructure is present in it.

From the data given on the chemical composition of the investigated alloy, it is seen that its major alloying components (zinc, copper and magnesium) are contained in an insignificant amount. Therefore, according to [11], they practically completely dissolve in an aluminum matrix. At the same time, chromium, iron and silicon are slightly soluble in aluminum and form different phases. It is known [11] that the Mg and Si atoms in an aluminum solid solution tend to form the Mg_2Si phase. Chromium causes some hardening, but its main role is to increase the corrosion resistance under stress.

X-ray investigations of the specimens prepared for the mechanical tests were performed for the diffraction angles 2θ lying in the range from 19 to 49°. Because of the small volume fraction of intermetallide phases in the alloy and their large dispersion, these studies did not allow to determine its phase composition fully. Reliably determined only that very intense diffraction peaks in the diffractograms correspond to aluminum based solid solution α_{Al} as well as the fact that the phase Mg_2Si is present in the alloy's specimens. In addition, low-intensity diffraction peaks corresponding to the η-phase ($MgZn_2$) and T-phase ($Mg_3Zn_3Al_2$) are detected. This indicates that the amount of η-phase and T-phase in the initial specimen of the alloy is insignificant.

2.3. MICROSTRUCTURE EVOLUTION

To understand the microstructure evolution during the superplastic deformation of the specimens of the Al-Zn-Mg-Cu-Zr alloy, studies of the features of the deformation relief formed on the surface of the working part of the specimens were carried out. These studies were performed using optical microscopy and scanning electron microscopy.

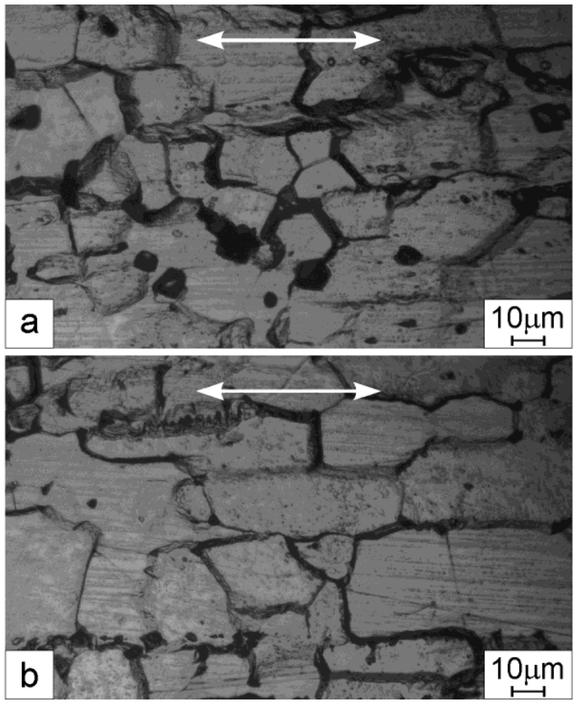

Fig. 5. The typical views of deformation relief formed on the surface of the working part of the Al-Zn-Mg-Cu-Zr alloy specimens (optical microscopy)

Fig. 5 (optical microscopy) and Fig. 6 (raster electron microscopy) show the typical views of deformation relief formed on the surface of the working part of specimens of the Al-Zn-Mg-Cu-Zr alloy deformed to failure under the optimal conditions of superplasticity.

It is seen that during the superplastic deformation grain boundary sliding developed in the specimens. Grain boundary sliding performed intensively along boundaries of both ultrafine and larger polygonized grains. This is evidenced by the formation of the developed deformation relief on the surface of the working part of the specimens, as well as the presence of characteristic displacements and ruptures of marker risks at the boundaries of the sliding grains (see Figs. 5 and 6).

Intensive rotation of the grains takes place during the grain-boundary sliding. This is confirmed by the fact that the marker risks applied to the pre-polished surface of the working part of the specimen before its deformation, after deformation, change their view.

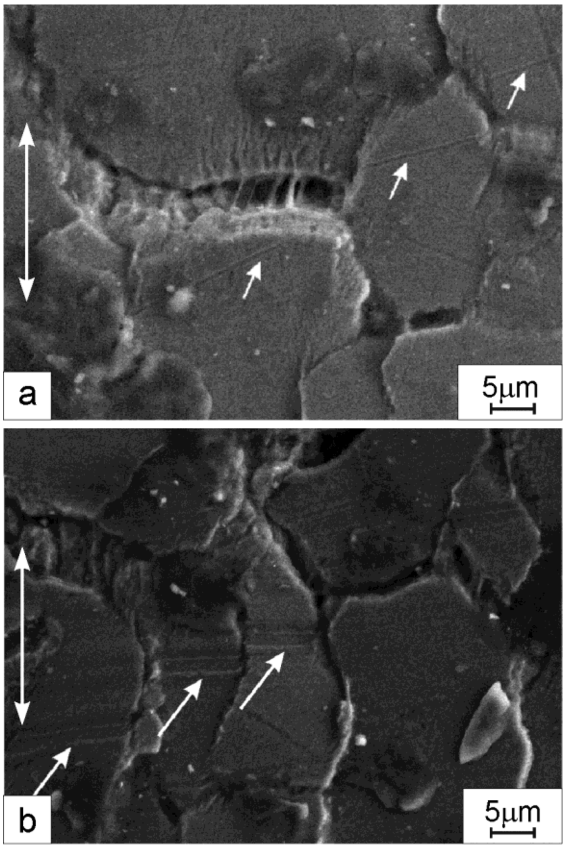

Fig. 6. The typical views of deformation relief formed on the surface of the working part of the Al-Zn-Mg-Cu-Zr alloy specimens (scanning electron microscopy)

In a deformed specimen, they are disoriented relative to each other and consist of individual segments which are displaced at a certain distance or have discontinuities while crossing the boundaries of neighboring grains. It should be noted that intense sliding of large grains is observed (see Fig. 5), which occurs along intergranular boundaries parallel to the strain direction of the specimen. Most of these boundaries, as shown by EBSD analysis, are low-angle. It is known that under tension the maximum tangential stresses act in planes located at an angle of 45° to the strain direction of the specimen [1, 2]. And, according to the existing classical ideas of superplasticity, grain-boundary sliding develops most intensively on the high-angle grain boundaries located at an angle of 45° to the strain direction of the specimen [1, 2]. Therefore, the observed development of grain-boundary sliding along the intercrystalline low-angle grain boundaries, parallel to the strain direction of the specimen, requires further clarification.

A study of the microstructure and topography of the working part surface of the specimens of the Al-Zn-Mg-Cu-Zr alloy superplastically deformed at $T = 793$ K allowed to find fibrous formations similar in shape to filamentary crystals in near-surface cavities and cracks (Fig. 7).

Whisker formation is already observed at engineering strain of specimen approximately 50%. The average fiber diameter is about 1 μm, and their length is determined by the longitudinal dimension of the discontinuity in which they are located, and can reach 50…100 μm. Fibers in the cavities are parallel to the direction of the tensile axis of the specimen and, basically, by their both ends are fixed on the internal surfaces of cavities and cracks perpendicular to the strain direction.

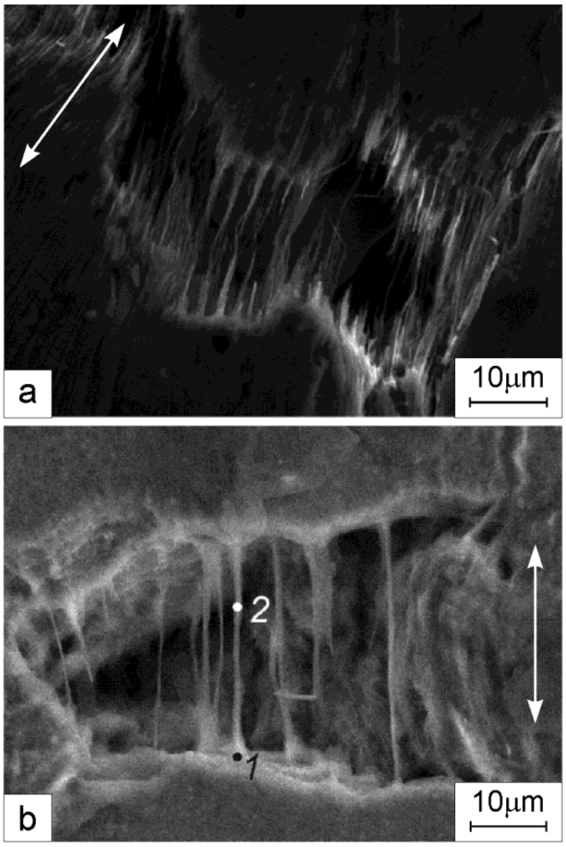

Fig. 7. Fibrous formations found in the working part of the specimen superplastically deformed under high homologous temperature

The energy-dispersive X-ray microanalysis of the chemical composition of the material from which fibers consist was performed. Fig. 8 shows the energy spectra of characteristic X-ray radiation at the point where the fiber is fixed to the cavity wall (near the fiber base) and in its middle part (see Fig. 7,b).

The spectra indicate the presence of aluminum, magnesium and oxygen in the fiber. It is determined that the concentration of Mg in these sections is slightly increased in comparison with its average concentration in the alloy and is equal, at the point indicated by figure 1 in Fig. 7,b to 3.77 wt.%. At the point indicated by figure 2 in Fig. 7,b – 7.04 wt.%.

On the surface of the fibers and grains with which they are connected, loose oxide films were found. This suggests that during the superplastic deformation of the specimens of the 1933 alloy at the test temperature $T = 793$ K, dynamic oxidation of inclusions of a viscous material intensively took place in its working part, which in a small amount was probably present at grain boundaries and on the boundary edges of sliding grains.

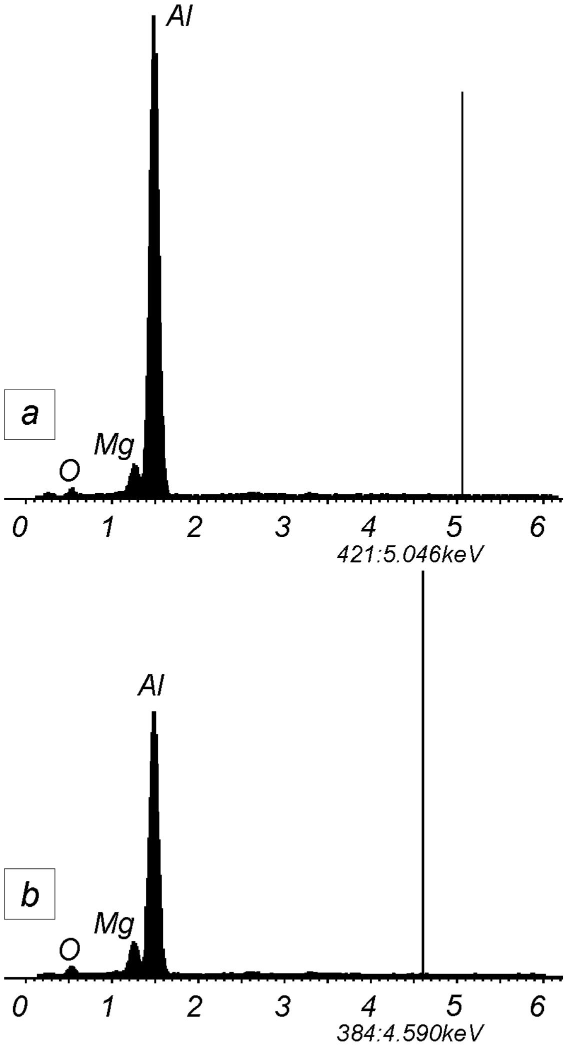

Fig. 8. The energy spectra of characteristic X-ray radiation at the point where the fiber is fixed to the cavity wall (near the fiber base) and in its middle part indicated by 1 and 2 in Fig. 7,b

2.4. FRACTURE OF DEFORMED SPECIMENS

Fig. 9 shows fractograms of fractures of specimens of alloy 1933 deformed to failure under the optimal conditions of superplastic deformation. Analysis of the types of fractograms gives grounds to assert that at the macroscopic level their failure has a quasi-brittle character, and at the microscopic level it has a mixed character.

Sections, characteristic for the fractures of the specimens, which were in a solid-liquid state during failure were observed on the fractograms [12].

On the fracture surfaces, facets of grains are observed, on which there are inclusions of the liquid phase crystallized during the cooling of the deformed specimen. On the fractograms of the fractures, grain boundary facets, representing the areas formed as a result of local intergranular fracture are also visible. This failure is probably caused by the high-temperature embrittlement caused by the partial melting of the grain boundaries and by the formation of thin extended films of the liquid phase on the grain surface. Loose oxide films are observed on the surface of some grains. On the surface of some grains (see Fig. 9,b) dendrites were revealed.

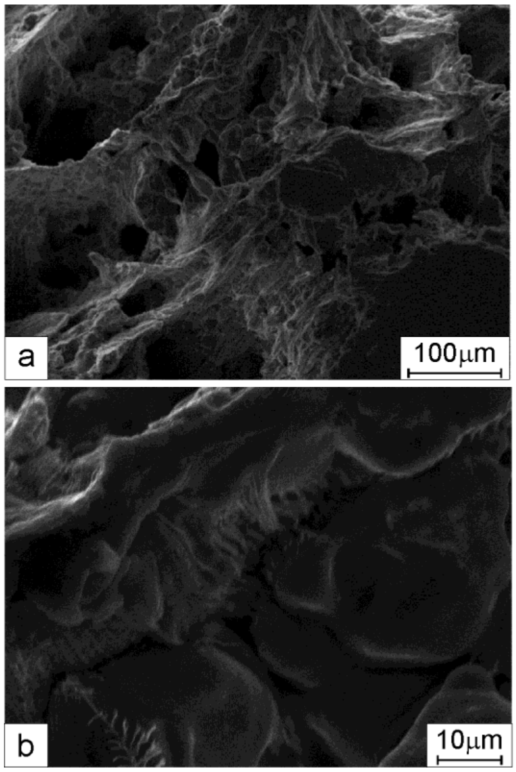

Fig. 9. Fractograms of fractures of specimen deformed to failure under the optimal conditions of superplastic deformation

It is determined that dendrites formed as a result of the crystallization of the liquid phase during cooling of the specimens. In some cases, the sharp edges of the facets of the separated grains contain fringe, which apparently is the crystallized melt streams of a solid solution based on aluminum formed during grain separation during the failure of the specimen.

2.5. DIFFERENTIAL THERMAL ANALYSIS

Thermal studies were performed to determine the temperature intervals in which partial melting of the alloy investigated in the work could take place.

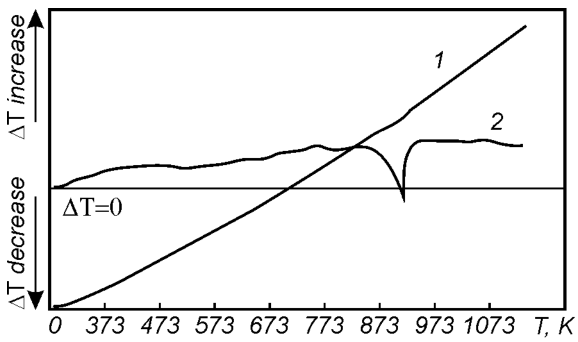

Fig. 10. The temperature curve (1) and the differential thermal analysis curve (2) of the Al-Zn-Mg-Cu-Zr alloy

Fig. 10 shows the temperature curve (1) and the differential thermal analysis curve (2), which were obtained as a result of differential thermal analysis of a specimen of alloy 1933 using “Derivatograph Q-1500”.

The alloy was heated from room temperature to a temperature of 973 K with a heating rate of 5 K/min.

It can be seen that in the temperature range that was used to study the superplastic properties of the 1933 alloy, phase transitions occur in the specimen, as a result of which heat is absorbed, that is, partial melting of the alloy takes place. According to approximate calculations, the volume of the liquid phase formed as a result of partial melting will be small and will be no more than 0.15% of the specimen volume.

3. DISCUSSION

Let us analyze the main causes of partial melting of specimens of alloy 1933 during superplastic deformation. Apparently, one of the most probable reasons for the formation of a liquid phase can be local melting of those grain regions and grain boundaries, which consist from aluminum based solid solution and contain an increased concentration of magnesium, zinc and copper. These elements reduce the melting temperature of the aluminum alloy. Therefore, partial melting of this alloy is most likely takes place because of heterogeneity in the distribution of magnesium, zinc and copper in those microvolumes at the boundaries or on the edges of grains, where their concentration due to the presence of segregations or because of intragrain liquation is increased in comparison with their average concentration in the alloy. The analysis of literature data on the phase composition of the alloys of the Al-Zn-Mg-Cu-Zr system [8, 9, 11] suggests that partial melting can also occur as a result of melting of such nonequilibrium eutectics: a double eutectic at 475 °C by the reaction: (L) $\Leftrightarrow$ (α_{Al} + $AlZn_2$), a double eutectic at 489 °C by the reaction: (L) $\Leftrightarrow$ (α_{Al} + $Mg_3Zn_3Al_2$); triple eutectic at 475 °C according to the reaction: (L) $\Leftrightarrow$ (α_{Al} + $AlZn_2$ + $Mg_3Zn_3Al_2$); quadruple eutectic at 475°C by the reaction: (L) $\Leftrightarrow$ (α_{Al} + $AlZn_2$ + + $Mg_3Zn_3Al_2$ + Al_2CuMg).

The fibrous formations found in the near-surface cavities and cracks (see Fig. 7) are also of interest. The authors of [13–23] observed the formation of fibers in cavities and cracks during the superplastic flow of specimens of aluminum alloys at high homological temperatures. They suggest that their formation is an evidence of a liquid phase presence along the grain boundaries in specimens of alloys deformed superplastically at high homological temperatures. This conclusion is justified by the fact that neither diffusion creep nor dislocation flow mechanisms can ensure the rate of deformation of fibers during superplastic flow [13]. Fibers in the cavities can not also be viscously deformed intermetallides, since, as studies of their microstructure showed, in all the observed cases [13–23] their body is a solid solution based on aluminum enriched with alloying elements and especially by magnesium.

Apparently, an important role in the formation and development of fibers is played by the dynamic oxidation of the inclusions of the liquid material that was present in the superplastically deformed alloy. Thus, in [19], the effect of atmospheric air on the process of fiber development during superplastic deformation was studied. The authors state that during the superplastic deformation of specimens of an aluminum alloy AA5083 in vacuum, no fibrous formations were observed at all, while in specimens deformed under the same conditions in air, a large number of fibrous formations were observed. In the chemical composition of the fibers oxygen is present. It should be noted that in another paper [20], during the deformation of the same aluminum alloy AA5083, the formation of fibers from surface oxide layers called oxide bonds was observed. The authors of [20] note that it was grain-boundary sliding that led to the formation of fibrous structures from oxides (bundles) at high temperatures.

Thus, it can be assumed that the formation and development of fibrous structures during the superplastic deformation of specimens of various aluminum alloys, including in the alloy 1933, was carried out by the same mechanism. In the working part of the specimen, as a result of intensive grain-boundary sliding, isolated grain-boundary cavities and magisterial cracks were formed. Their disclosure under the influence of tangential and normal stresses led to the manifestation of the so-called micro-superplasticity [21, 22]. The process of formation and growth of fibrous structures in superplastically deformed aluminum alloys was called as micro-superplasticity in [21]. The development of fibers is apparently takes place in accordance with the model described in [23]. According to this topological model, during the superplastic flow, the grain-boundary cavities begin to open at the boundaries perpendicular to the strain direction under the action of normal stresses. It is at these boundaries, if they contain cells of viscous liquid phase, that fibrous formations begin to develop. Fig. 11 is a scheme of the formation and development of fibers.

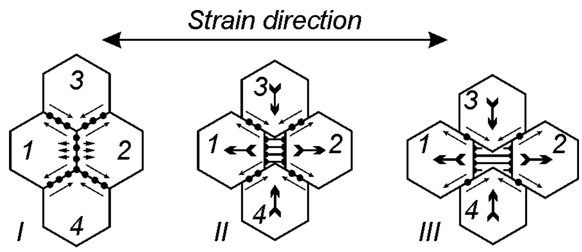

Fig. 11. Scheme of the formation and development of fibers during superplastic deformation in specimens containing a viscous liquid phase at intergranular boundaries. Arrows at the boundaries show the forces acting on the grains boundaries. Arrows in the centers of the grains (in positions II and III) show the direction of grain movement. The dots indicate the liquid phase at the grain boundaries

The edges of grains perpendicular to the strain axis, that is, the walls of the cavities and cracks, in this case play the role of captures to which the fibers formed as a result of viscous flow are fixed. Further development of the cavities mainly occurs due to grain boundary sliding of adjacent to cavity grains. The progressive movement of the grains at the boundaries of which a cavity was formed during their further removal from each other in a direction parallel to the strain axis leads simultaneously with the opening of the grain-boundary cavity to the growth of the fibers in a direction parallel to the strain axis. This development occurs by viscous flow of the liquid phase with rate equal to the rate of the grain boundary cavity opening.

Analyzing the appearance of the fibers, and taking into account the data on the chemical composition of the alloys, it can be seen that the pronounced "filamentary" type of fibers is characteristic for those aluminum alloys that contain Mg in their composition. It is known [24, 25] that if such elements as Cu, Si, Zn, Mn, Fe are present in a small amount in an aluminum alloy, then an oxide film formed on the surface of aluminum alloys with these elements when they are oxidized in air in the liquid state, is similar in structure to the film, which is formed on molten pure aluminum. At the same time, when the Mg content in the aluminum alloy exceeds 1.5 wt.%, as shown in [24, 25], the oxide film consists essentially of pure magnesium MgO. Thus, for an alloy 1933, which contains more than 1.5 wt.%Mg, the oxide film covering the surface of liquid-phase inclusions localized in the opening cavities will practically consist from pure magnesite, and therefore it will be loose [24, 26]. Since the resulting oxide film on the surface of the liquid-phase inclusion will be loose, it will not interfere with the shape of this inclusion in tension. It is one of the main conditions necessary for the formation and development of fibrous structures when the grain boundaries cavities open.

As stated above, intensive sliding of large polygonized grains is observed in the 1933 alloy (see Fig. 4,b), along intergranular boundaries parallel to the strain direction of the specimen. In addition, most of these boundaries are low-angle boundaries. All this requires further clarification.

As a rule, the structure of materials exhibiting superplasticity is preliminarily prepared according to certain technological cycles (see eg. [1, 2]) in order to obtain a ultrafine-grained equiaxed grain structure. It is also known that some industrial aluminum alloys with unprepared rather coarse-grained structure can exhibit structural superplasticity. However, the ultrafine-grained structure in them is formed at the initial stages of deformation as a result of dynamic recrystallization. That is, an ultrafine-grained grain structure with a predominance of high-angle grain boundaries is created, which allows the alloy to exhibit the effect of superplasticity. In this paper, the conditions and nature of the manifestation of superplasticity are of a different nature. Despite the relatively large grain size of the alloy in the study, deformation take place by the prevalence of grain boundary sliding. The process of dynamic recrystallization, which would lead to refinement of the grain structure in this alloy, doesn't occur during deformation. Apparently, heating of the specimen to the required temperature, namely, to the temperature at which partial melting of the alloy is possible, is the determining factor for the realization of superplastic behavior in the alloy. Apparently, activation of the grain boundaries to grain-boundary sliding occurs due to the presence of the grain boundary inclusions of the liquid phase in the material.

With regard to the model for the realization of high-temperature superplastic deformation in the presence of a liquid phase in the deformed object, a discussion is still going on as to how the distribution of the liquid phase along the grain boundaries should be: continuous or in the form of discrete inclusions. According to the model proposed in [27, 28], under optimal conditions of high-temperature superplasticity, the liquid phase at the grain boundaries should be concentrated in the form of a very thin continuous layer of liquid. In this case, the interlayer should be such that the atoms of the neighboring grains still experience the force of adhesion to each other. The increase in the thickness of the liquid phase with increasing temperature, according to [27, 28], will lead to liquid metal embrittlement. This point of view is supported in [29], where a view on the distribution of the grain-boundary liquid phase under conditions of superplasticity is also presented. According to [29], the liquid phase is either localized in triple junctions of grains, or evenly distributed along the grain boundaries, but the thickness of the liquid phase must be such that the atoms of the neighboring grains still experience the force of adhesion to each other. In [17] is reported that under the conditions of high-temperature superplasticity, when the processes of grain-boundary sliding develop intensively, the liquid phase can change its location by flowing under the action of compressive stresses from inclined and parallel to strain directions to boundaries perpendicular to the strain axis. If liquid has time to flow from the compressed portions of boundaries to the stretched portions, grain boundary cracks, opening at perpendicular to the strain direction grain boundaries will develop in the specimens. The inclined boundaries in this case, according to [17], will no longer contain inclusions of the liquid phase and, thus, the conditions for effective stress relaxation will be lost on them.

The presence of liquid at the boundaries apparently affects the character of dislocation sliding in grains, since under the conditions of superplasticity it is closely related to the interaction of lattice dislocations with boundaries whose state changes qualitatively in the presence of liquid. This is now not the boundary between two solid grains, but the boundary between solid grains and liquid. One can think that in such conditions, on the one hand, the number of sources of lattice dislocations at the boundaries will decrease, and on the other hand, the latter will become good drains for dislocations. The ratio of intragranular dislocation sliding in the total deformation will fall. In the presence of liquid at the boundaries, the role of intragranular dislocation sliding in the accommodation of the shape changing of the grains will also drop. It is the main accommodation process in the case of ordinary structural superplasticity. Grain-boundary sliding – the main deformation process of superplastic flow due to the presence of liquid at the grain boundaries becomes more intense, and its contribution to the overall deformation increases. Thus, the observed sliding of large grains (see

Fig. 5), which occurs along low-angle intergranular boundaries parallel to the strain direction of the specimen, is apparently possible due to the presence of local inclusions of the liquid phase on them.

The hypothesis about the positive effect of the liquid phase on the deformational and accommodative processes on the whole is confirmed by the experimental results. Thus, it is shown in [14] that the relatively coarse-grained IM/7475 alloy ($d \approx 20$ μm), in the presence of small inclusions of liquid phase along the grain boundaries, demonstrated an elongation to failure much greater than that for specimens with a smaller grains but without grain-boundary liquid phase. It was also found in [30] that in the case of superplastic deformation, when the partial melting is realized along the grain boundaries, more coarse-grained specimens of the Al-25%Si alloy reach higher elongation to failure, albeit at a lower deformation rate. These facts indicate the extremely important role of the liquid phase in the manifestation of superplastic properties by coarse-grained materials, including the alloy 1933.

CONCLUSIONS

1. The initial structural state of the industrial alloy 1933, which exhibited the effect of superplasticity, is investigated. It is shown that the initial structure of alloy 1933 is bimodal. It is found that the specific fraction of low-angle grain boundaries is 65.5%, and the specific fraction of high-angle grain boundaries is 35.5%.

2. It is found that when the specimens of alloy 1933 are heated to test temperatures at which it exhibits the effect of superplasticity, partial melting occurs in them. It leads to the formation of a small amount of the metastable liquid phase.

3. In the near-surface grain-boundary cavities and in cracks, which formed and developed on the surface of the working part of the alloy of 1933 specimens during the superplastic deformation, fibrous structures were found. The chemical composition of the material from which the fibers consist is investigated. It is suggested that the formation and development of fibrous structures during the superplastic flow of the alloy of 1933 specimens is associated with the intensive development of grain-boundary sliding, which occurs when there is a viscous liquid phase at the grain boundaries.

4. Data on the temperature conditions of the manifestation of superplasticity by the alloy 1933 specimens, the presence of whiskers in the structure of the alloy, and the presence in them of an increased concentration of magnesium, which reduces the melting point of the alloy, convincingly indicates that this alloy exhibits superplasticity in the solid-liquid state.

5. The effect of the initial structural state of alloy 1933 on the development of grain-boundary sliding is analyzed. It is found that grain-boundary sliding occurs intensively both along the high-angle boundaries of ultra-fine grains and along the low-angle boundaries of large polygonized grains parallel to the strain direction of the specimen. The mechanism of the development of grain-boundary sliding in the alloy 1933 with a bimodal structure is proposed.

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СВЕРХПЛАСТИЧНОЕ ПОВЕДЕНИЕ АЛЮМИНИЕВОГО СПЛАВА 1933 С БИМОДАЛЬНОЙ СТРУКТУРОЙ ПРИ ПОВЫШЕННЫХ ТЕМПЕРАТУРАХ

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Изучены особенности структурного состояния, фазового состава и деформационного рельефа рабочей части образцов сплава 1933. Установлено, что зернограничное проскальзывание интенсивно происходит как по границам ультратонких зерен, так и по границам крупных полигонизированных зерен, параллельных направлению растяжения образца.

НАДПЛАСТИЧНА ПОВЕДІНКА АЛЮМІНІЄВОГО СПЛАВУ 1933 З БІМОДАЛЬНОЮ СТРУКТУРОЮ ПРИ ПІДВИЩЕНИХ ТЕМПЕРАТУРАХ

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Вивчено особливості структурного стану, фазового складу і деформаційного рельєфу робочої частини зразків сплаву 1933. Встановлено, що зернограничне проковзування інтенсивно відбувається як по межах ультратонких зерен, так і по межах великих полігонізованих зерен, паралельних напрямку розтягування зразка.