# Behavior of a Submicrocrystalline Aluminum Alloy 1570 under Conditions of Cyclic Loading

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**Abstract**—A study of the resistance to fatigue-crack growth in a submicrocrystalline alloy Al–6% Mg–0.3% Sc–0.4% Mn in combination with a precision analysis of the fracture surface of the samples has been performed. A comparison of crack resistance between coarse-grained and submicrocrystalline states of this alloy showed that only at the stage of near-threshold crack growth the velocity of fatigue-crack propagation in the submicrocrystalline state proves to be higher than that in the coarse-grained state. At the stage of linear crack growth, the fatigue-crack propagation becomes insensitive to the grain size. Upon transition to the stage of accelerated crack growth, the velocity of crack propagation in the submicrocrystalline alloy is retarded. A fractographic analysis of the fracture surface of the samples indicates that the retardation of the fatigue-crack growth in the submicrocrystalline alloy is connected with a gradual transition from the intercrystalline to the transcrystalline mechanism of fatigue fracture of the material.

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## INTRODUCTION

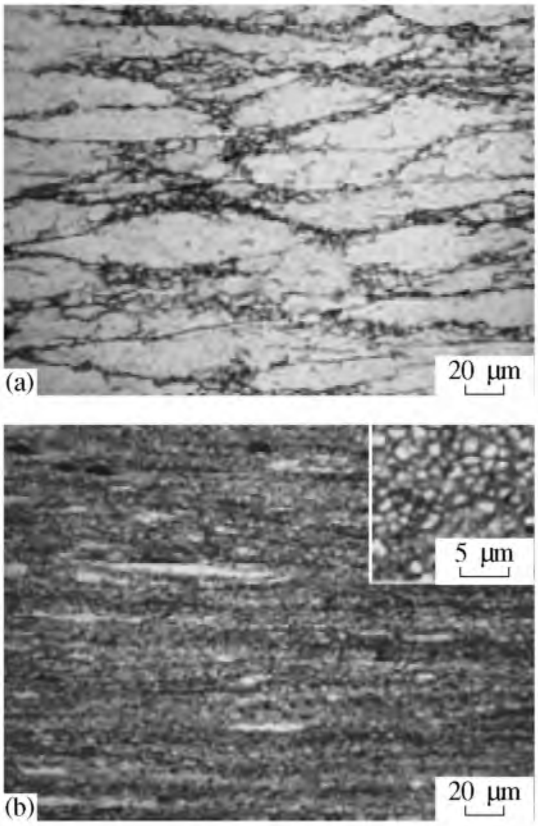
At present, numerous studies are performed to investigate the effect of severe plastic deformation (SPD) on the mechanical properties of materials. It is shown that the employment of SPD to form a submicrocrystalline (SMC) structure makes it possible to improve characteristics of static strength of metals and alloys [1–4] and to obtain high values of superplasticity at fairly high strain rates (~$10^{-2}$ $s^{-1}$) [5–7]. A large number of works (e.g., [8–17]) are also devoted to fatigue properties of materials subjected to SPD.

It is a common practice to subdivide the fatigue curves in materials into two stages, namely, the stage of fatigue-crack nucleation and the stage of its propagation. It is known that grain refinement by SPD methods prolongs the stage of crack nucleation, because of an increase in the strength of SPD-treated materials [18, 19]. Yet, the behavior of materials at the stage of crack propagation is the least understood property of SMC structures. It was shown in [8, 10, 11, 13, 14] that in the near-threshold range of crack growth the resistance to fatigue-crack growth of SMC polycrystals can be lower than that of coarse-grained materials, since the fatigue fracture in SMC structures occurs more easily by the intercrystalline mechanism and can be connected with a less wavy trajectory of crack propagation. At the same time, as was reported in [20], for some SMC aluminum and copper alloys the transcrystalline fatigue-crack propagation can prevail at both low and high values of the double amplitude of the stress-intensity factor. At present, unfortunately, the morphological features of fatigue fracture formed in a wide range of parameters of cyclic deformation and/or at different stages of fatigue-crack propagation have been investigated insufficiently. As a result, the precision mechanisms of fracture of SMC materials remain unclear.

The purpose of this work was to study the cyclic crack resistance of the Al–6% Mg–0.3% Sc–0.4% Mn alloy 1570 with an SMC structure formed in the process of SPD. It should be noted that this alloy became a promising structural material for civil aircraft construction, although one of its significant drawbacks is a relatively low crack resistance [21], which limits the wide application of this alloy. For this reason, the investigation of the SPD effect on the crack resistance of this material is of noticeable interest for practice. In this work, along with measurements of the velocity of fatigue-crack propagation in the SMC alloy 1570, a great attention was given to an analysis of morphological features of the fracture surface along the crack trajectory, which allow one to reveal the main mechanisms of its propagation in the SMC structure.

## EXPERIMENTAL

The aluminum alloy 1570 used in this work has the following chemical composition (wt %): Al–6% Mg–0.3% Sc–0.4% Mn–0.2% Si–0.1% Fe. The alloy was produced by the method of semicontinuos casting and




1 **Fig. 1.** Microstructure of the allloy 1570: (a) the as-delivered state and (b) the state after ECA pressing. Dark regions in Fig. 1b correspond to fine (~1 μm) grains, as is shown in the right upper corner at a larger magnification.

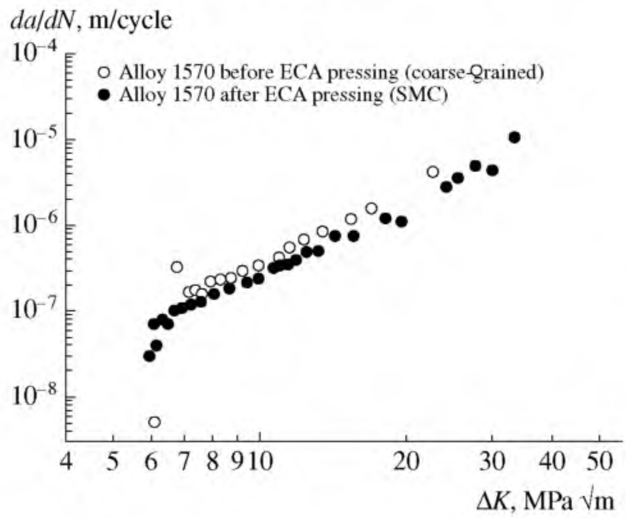



**Fig. 2.** Fatigue-fracture diagrams of the coarse-grained and SMC states of the aluminum alloy 1570.

was homogenized in air at 520°C for 48 h. Then, the alloy was subjected to pressing at a temperature of 390°C to a reduction of 50% and to annealing at a temperature of 400°C for 1 h. The equal-channel angular (ECA) pressing of the alloy 1570 plates with a dimension of 152 × 152 × 25 mm was performed at a temperature of 325°C to a true degree of deformation $\varepsilon \sim 9.2$ by route *Bcz* [22] in which upon transition from one ECA pass to another the plates were sequentially rotated through an angle +90° about the normal to the largest face of the workpiece. The tests for cyclic crack resistance were conducted on a Schenk hydropulse PSA hydraulic machine using compact samples 20 mm thick with an edge notch. The samples were loaded using a sinusoidal cycle with a frequency of 5 Hz and an asymmetry coefficient $R = 0.1$. The fatigue-crack profile was investigated in the plane perpendicular to the plane of the fatigue-crack propagation. The metallographic analysis was performed on a Nikon L-150 optical microscope. The fractographic investigations were carried out using a JEOL JXA-6400 scanning electron microscope.

## EXPERIMENTAL RESULTS AND DISCUSSION

### *Formation of an SMC Structure in the Process of ECA Pressing*

Typical microstructures obtained before and after ECA pressing are presented in Figs. 1a and 1b. In the as-delivered state (see Fig. 1a), the alloy had an inhomogeneous partly recrystallized structure consisted of large elongated grains whose sizes in the longitudinal and transverse directions were 170 and 70 μm, respectively, and regions of finer equiaxed grains ~4 μm in size with a volume fraction of ~0.35 that were located in the form of a "mantle" along initial grain boundaries. The ECA pressing ensured the formation of an SMC structure with a grain size of ~1 μm. The volume fraction of SMC grains was ~0.88 (see Fig. 1b).

### *Diagrams of Fatigue Fracture*

Figure 2 displays the dependence of the rate of the fatigue-crack growth $da/dN$ in the Al–6% Mg–0.3% Sc–0.4% Mn alloy with an SMC structure on the double amplitude of the stress-intensity factor $\Delta K$. For comparison, analogous data on the crack propagation in coarse-grained samples (see Fig. 1) are presented in this figure as well. It is seen in Fig. 2 that the dependence constructed for conventional polycrystalline material demonstrates three well-known stages of crack propagation: the stage of slow crack growth in a near-threshold region at $\Delta K \le 6$ MPa $\sqrt{\text{m}}$, the stage of linear growth (in the Paris regime, $da/dn \sim (\Delta K)^m$ [23]) at $\Delta K \sim$ 6.5–20 MPa $\sqrt{\text{m}}$, and the stage of unstable crack growth at high values of $\Delta K \ge 20$ MPa $\sqrt{\text{m}}$. Figure 2 likewise shows that the stage of linear fatigue-crack growth in the SMC state of the material expands toward large values of $\Delta K$, whereas the stage of accelerated growth in this state becomes less pronounced. The rate of crack growth in this state in the near-threshold range is higher than that in the normal polycrystalline state. At the stage of linear crack growth, however, the velocities $da/dN$ of crack propagation in both states reach much

2 the same valuesf. This indicates that, although in the near-threshold range the fatigue-crack behavior is determined by microstructure factors, the rate of fatigue-crack growth in this material becomes "structurally insensitive" at high values of $\Delta K$. It should be noted that such a behavior, when in the Paris regime the microstructure exerts a less influence on the fatigue-crack propagation, was reported previously in works [11, 12, 14, 20] devoted to the crack resistance of SMC materials. This effect was also demonstrated in conventional coarse/fine-grained materials [24–27]. Thus, such an effect of the microstructure on the crack resistance can be a common phenomenon in polycrystals having a wide spectrum of grain sizes. Unfortunately, as was noted above, only very limited information on the effect of SMC structure on the fatigue-crack development in the range of high values of $\Delta K$ is currently available in the literature.

It is seen from Fig. 2 that the resistance to crack growth in the SMC structure somewhat increases upon transition to the range of medium/high values of $\Delta K$ ($\Delta K \geq 20$ MPa $\sqrt{\text{m}}$). As a result, the final fracture of SMC samples occurs at $\Delta K \sim 33$ MPa $\sqrt{\text{m}}$, whereas a fracture of coarse-grained samples takes place already at $\Delta K \sim 27$ MPa $\sqrt{\text{m}}$. Such an ambiguous influence of $\Delta K$ on the crack propagation is most likely to be connected with a change in the mechanism of fatigue fracture in the SMC structure. In what follows, there will be given, according to the data presented in Fig. 2, a detailed analysis of features of the fatigue-crack propagation in the SMC samples of the alloy 1570 upon transition to the range of intermediate and high values of $\Delta K$.

*Fatigue-Crack Profile in the SMC Material*

Figure 3 displays typical pictures that show the fatigue-crack profile in the SMC material upon a transition from the stage of near-threshold crack growth to the stage of linear crack growth, i.e., at $\Delta K \sim 7$ MPa $\sqrt{\text{m}}$ (see Fig. 3a); at the mid-stage of linear crack growth, i.e., at $\Delta K \sim 9$ MPa $\sqrt{\text{m}}$ (see Fig. 3b); and at the end of this stage, i.e., at $\Delta K \sim 18$ MPa $\sqrt{\text{m}}$ (see Fig. 3c).

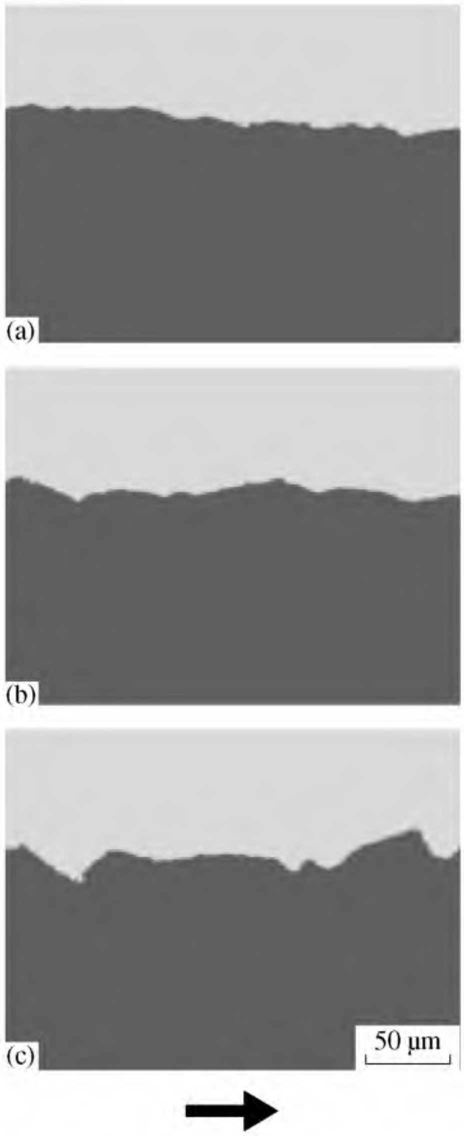



**Fig. 3.** Typical fatigue-crack profile in the SMC alloy 1570: (a) $\Delta K \sim 7$ MPa $\sqrt{\text{m}}$, (b) $\Delta K \sim 9$ MPa $\sqrt{\text{m}}$, and (c) $\Delta K \sim 18$ MPa $\sqrt{\text{m}}$. Arrow points to the direction of crack propagation.

It is seen in Fig. 3a that the crack propagation at the stage of near-threshold growth and upon transition to the stage of linear growth is characterized by a relatively weakly pronounced relief of the fracture surface. Yet, as the crack passes through the stage of linear growth, the crack trajectory becomes gradually more and more wavy and the crack acquires a zigzag shape (see Figs. 3b, 3c). As $\Delta K$ increases, the energy spent on the formation of the free fracture surface of the material enhances.

*Fractography*

Figure 4 shows typical micrographs of the fracture surface of the SMC samples of the alloy 1570 that correspond to the stages of near-threshold (see Fig. 4a), linear (see Figs. 4b–4d), and accelerated (see Figs. 4e, 4f) fatigue-crack growth. At small values of $\Delta K$ (~6.5 MPa $\sqrt{\text{m}}$), the morphology of fractures of the samples points to the intercrystalline character of fracture of the material; this follows from the faceted type of fracture with the size of facets close to the grain size (~1–2 μm) (see Figs. 4a, 1b).

At the stage of stable crack growth ($\Delta K \sim 7$–20 MPa $\sqrt{\text{m}}$), a gradual change in the mechanism of fracture was observed (see Figs. 4b–4d). On the fracture surface, there arise isolated regions that look as "conglomerates of facets" or "large facets" (see Fig. 4b) whose sizes increase gradually with crack growth (see Figs. 4c, 4d). This increase is accompanied by a pro-

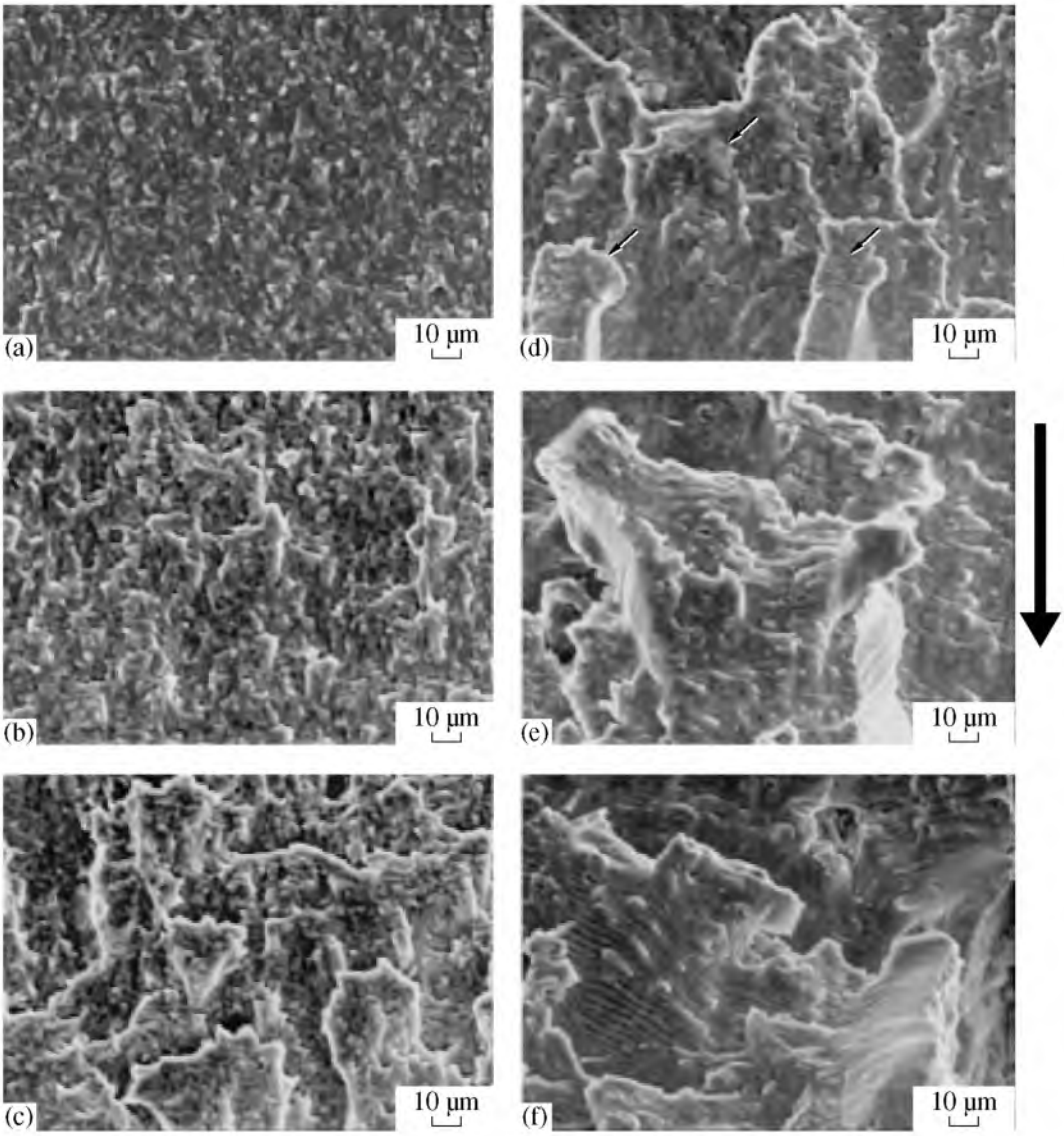
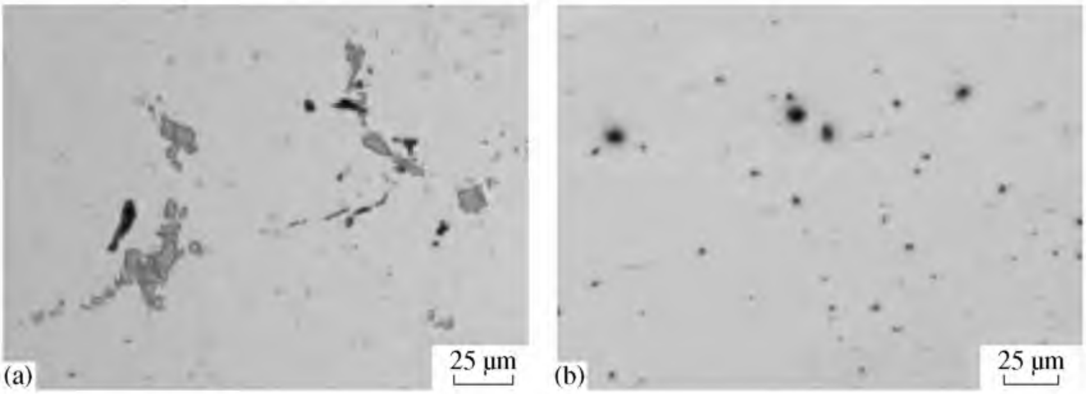



**Fig. 4.** Micrographs of the fracture surface of the SMC samples of the alloy 1570: (a) stage of near-threshold crack growth, $\Delta K \sim 6.5$ MPa $\sqrt{m}$; (b) stage of linear crack growth, $\Delta K \sim 9$ MPa $\sqrt{m}$; (c) stage of linear crack growth, $\Delta K \sim 12$ MPa $\sqrt{m}$; (d) stage of linear crack growth, $\Delta K \sim 18$ MPa $\sqrt{m}$; (e) transition to the stage of accelerated crack growth, $\Delta K \sim 22$ MPa $\sqrt{m}$; and (f) stage of accelerated crack growth, $\Delta K \sim 25$ MPa $\sqrt{m}$. The direction of crack propagation is indicated by an arrow in the right-hand part of this figure. Small arrows in Fig. 4e show weak grooves arising on the fracture surface at the stage of linear crack growth.

gressive coarsening of the relief of the fracture surface in the linear stage (see Fig. 3). The data shown in Fig. 4 make it clear that the fatigue crack gradually ceases "feeling" boundaries of SMC grains. This confirms a structurally insensitive character of its propagation with respect to the grain size at the stage of linear growth (see Fig. 2). In this SMC material such a structure insensitivity with enhancing $\Delta K$ can be caused by an increase in the size of a plastically deformed zone that is formed near the growing crack tip [18]. It may be assumed that at the initial stages of crack propagation, the size of the plastically deformed zone is comparable with the size of SMC grains. With an enhancement in the double amplitude of the stress-intensity factor, the zone size grows and, hence, the number of grains inside this zone increases; i.e., the crack can interact "simultaneously" with a group of grains. For these grains, there can arise "common" fracture planes whose size will be determined by the size of the plastically deformed zone [14].

Based on the data presented in Fig. 4, it can also be concluded that there takes place a gradual change in the character of the fatigue fracture of the SMC material from the intercrystalline ("faceted") mechanism operating in the range of near-threshold crack growth to the transcrystalline ("grooved") mechanism operating in the interval of high values of $\Delta K$. At the fracture surface, poorly resolvable fatigue grooves are frequently




1 **Fig. 5.** The polished surface of the allloy 1570 samples: (a) the as-delivered state and (b) the state after ECA pressing.

revealed, as, for example, is shown by arrows in Fig. 4d. This indicates that in the SMC alloy the crack begins growing by the classical mechanism of blunting of the crack tip [25]. As the crack propagates further, the fatigue grooves become more distinct and opening of secondary cracks occurs along them, whereas the regions of faceted and grooved fracture are scarcely revealed. As a result, the transcrystalline mechanism of fracture of the alloy with an SMC structure becomes dominant at large values of $\Delta K$ at the end of the linear stage and upon transition to the stage of accelerated crack growth (see Figs. 4e, 4f).

Thus, the results of these investigations show that the intercrystalline fracture of the SMC samples takes place at small values of the double amplitude of the stress-intensity factor (see Fig. 4a) and is accompanied by a "weak" waviness of the fatigue-crack trajectory (see Fig. 3a). This determines a smaller resistance to fatigue-crack growth, as compared to the coarse-grained state of the alloy (see Fig. 2). Analogous results were obtained earlier in [8, 10, 11, 13, 14]. At the same time, the retardation of the fatigue-crack growth in the SMC material at relatively high values of the double amplitude of the stress-intensity factor is connected with a gradual coarsening of the fracture surface of the samples caused by a transition from the intercrystalline to the transcrystalline mechanism of fatigue fracture.

### *Effect of Phase Composition*

One more important factor that affects the crack resistance can be the presence of second-phase particles in the material structure [28]. The results of metallographic analysis of the polished surface of the alloy 1570 samples before and after ECA pressing are presented in Fig. 5. It is seen that in the as-delivered state, i.e., after conventional thermomechanical treatment, the alloy contains relatively large primary particles (see Fig. 5a) that were identified in [29] as intermetallic compounds $Al_6$(Fe, Mn). These particles are nonuniformly distributed in the alloy structure, and their size ranges from 5 to 30 μm. In the process of tests for crack resistance, they can act as numerous stress concentrators that favor the rapid nucleation and development of secondary cracks in the Al matrix [28]. These latter can grow and emerge onto the surface of the main crack, thereby reducing the crack resistance of the material. After ECA pressing at high temperatures, the size of second-phase particles decreases considerably (see Fig. 5b). The dimensions of most of them decreases to the submicrocrystalline size (see also microstructures presented in [30]), and their distribution becomes more homogeneous; i.e., the particles experience fragmentation and/or are partly dissolved in the process of SPD [31–33]. As a result, their negative influence on the alloy behavior under conditions of cyclic loading decreases. In other words, the treatment of the alloy 1570 by ECA pressing can yield an additional enhancement in its crack resistance at the expense of fragmentation and/or partial dissolution of coarse primary phases during SPD.

## CONCLUSIONS

(1) It was established that at the initial stage of crack propagation (at small values of $\Delta K$) the rate of fatigue-crack growth proves to be higher in the alloy with SMC structure than in the initial coarse-grained alloy. This is caused by the dominant action of the intercrystalline (faceted) mechanism of fracture of the SMC material.

(2) At the stage of linear growth, the crack propagation in the samples with an SMC structure becomes insensitive with respect to the boundaries of SMC grains. The crack growth in this case is accompanied by a gradual change in the character of fracture from the intercrystalline (faceted) mechanism to the transcrystalline (groove) mechanism.

(3) The change in the character of fatigue fracture at the stage of linear growth is accompanied by a significant increase in the waviness of the fatigue-crack trajectory in the SMC material. This can be one of the main factors responsible for high crack resistance obtained in this work for the SMC material at large values of $\Delta K$.

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