The influence of texture on the mechanical properties of a superplastic magnesium alloy

O. A. KAIBYSHEV, I. V. KAZACHKOV, N. G. ZARIPOV
Institute of Metal Superplasticity Problems, USSR Academy of Sciences, Ufa 450001, USSR

Superplastic (SP) characteristics of a MA15 magnesium alloy subjected to tensile deformation in both textureless and textured states with identical fine grain structures have been studied. The existence of crystallographic texture is found to be the cause of the anisotropy of flow stress and plasticity. The shift of SP optimum strain rate towards higher values due to the strengthening of texture is observed. A special experiment has indicated that the difference of the alloy behaviour in textureless and textured states results from the operation of grain boundary sliding stimulated by the intragrain dislocation slip.

1. Introduction

The influence of crystallographic texture on the superplastic deformation (SPD) of metals and alloys is of great interest. One of the main effects which attracts the attention of the investigators is mechanical anisotropy. Some of the authors who have studied the plastic flow anisotropy and the superplastic deformation conditions on Zn–Al system alloys [1], bronze [2, 3], Sn–Pb eutectic [4], and Ti–6Al–4V alloy [5, 6] are of the opinion that crystallographic texture does not influence the SP effect, mechanical anisotropy being conditioned by the orientational heterogeneity of the microstructure (the grain elongation, directed arrangement of phases and microparticles along grain boundaries).

Under ordinary deformation conditions the effect of anisotropy is associated as a rule with the influence of the texture-conditioned orientational factors on the intragranular dislocation sliding (IDS). Under SP flow conditions, when grain boundary sliding (GBS) is the main deformation mechanism, such an explanation seems to be neither exact nor complete, especially in the framework of the traditional conceptions of GBS. That is why even in those works where the SPD mechanical anisotropy is unambiguously connected with the influence of the crystallographic texture, the nature of this effect is not revealed.

Special experiments, investigating the influence of texture on the SPD have been conducted in order to study the textured and textureless alloys Zn 22% Al and BT6 [12–14]. In the course of the experiments new facts have been found testifying to the substantial influence of the preferred orientation on the SPD. When the microstructures were identical the existence of the crystallographic texture made the flow tension decrease, plasticity increase, and the optimum rate interval shift to higher strain rates than those of the textureless state.

Most of the studies concerned with the investigation of the crystallographic texture influence on SPD were conducted on two-phase alloys, which certainly complicated the interpretation of the results. That is why to reveal the effect of the preferred grain orientation on the mechanical properties in more detail it would be useful to investigate materials which are closer in their microstructure to one-phase alloys.

A suitable material for the solution of such a problem are alloys with a hexagonal lattice, as it is comparatively easy to obtain various texture states in them, and the interpretation of the results is not so complicated due to the limited character of the operative slip systems.

2. Experimental procedure

The magnesium alloy MA15 (wt %: 3.1 Zn, 0.65 Zr, 1.6 Cd, 0.95 La) was used. Specimens were produced from 110 mm diameter hot-compacted rods by upsetting at 450° C and the strain rate $\dot{\varepsilon} = 3.3 \times 10^{-3}\,\mathrm{sec}^{-1}$ (state I), and by annealing at 450° C for 90 min (state II).

In order to obtain comparable results, materials that underwent different initial processing were brought to the same microstructural states by the choice of the proper temperature and time of additional annealing. The metallographic analysis of the microstructure was carried out on the optical structure analyser "Epiquant"; finer investigations were conducted on the electron microscope "Tesla BC 540". The crystallographic texture was studied by means of plotting complete pole figures using X-ray structural diffractometer DRON-2.0. The porosity of the specimens was determined by means of hydrostatic weighing [15].

The mechanical properties of the specimens produced from upset (state I) and annealed (state II) rods were tested on the Instron testing machine. The gauge length of the specimens was 25 mm with 5 mm diameter.

3. Results

The MA15 alloy microstructure and pole figures (0001)

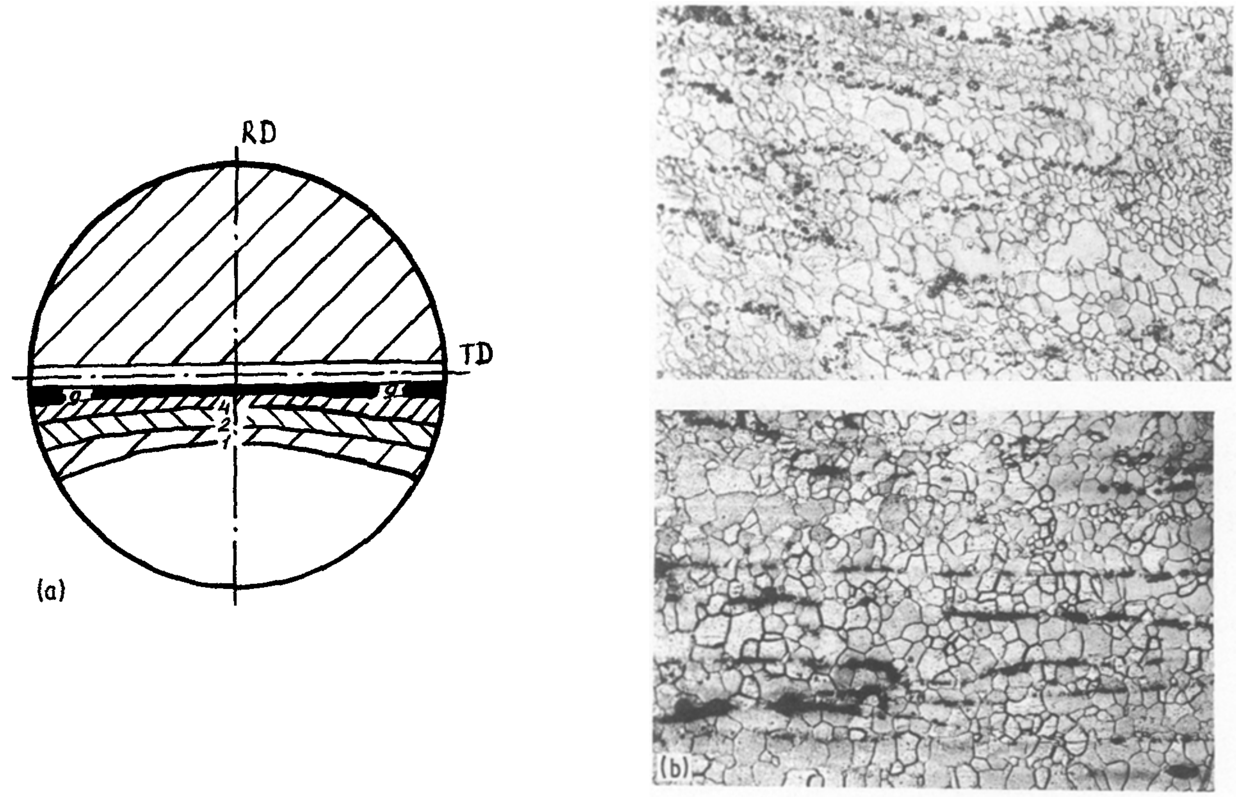

Figure 1 Microstructures and (0001) pole figures of the MA15 alloy: for (a) state I and (b) state II, ($\times$ 300).

in states I and II are shown in Fig. 1. The microstructures of both the states are obviously identical: the grains are equiaxed with the average size of 9 μm, there are lines of intermetals in the direction of pressing and the size of intermetals is commensurable with the size of the grains. Under certain temperature conditions such an alloy displays all the features of the superplastic flow [16].

The state I alloy is devoid of the crystallographic texture, while the state II alloy is characterized by the sharp axial texture with the maximum in the transversal direction.

Let us consider the changes of the MA15 alloy mechanical properties in different initial states under the hot strain.

The results of the MA15 alloy mechanical tests in states I and II are shown in Fig. 2. One can see that in state I the alloy is devoid of both the flow stress anisotropy and the relative elongation one. The isotropy of properties practically does not alter with the change of the strain temperature.

The textured alloy flow stress in the longitudinal direction is higher than that in the transversal one over the whole testing temperature interval. The flow stress anisotropy decreases with the increase of the strain temperature, but the difference between the flow stresses of the longitudinal and transversal specimens is preserved up to 500° C.

The alloy plasticity is anisotropic too. The relative elongation of the longitudinal specimens with a

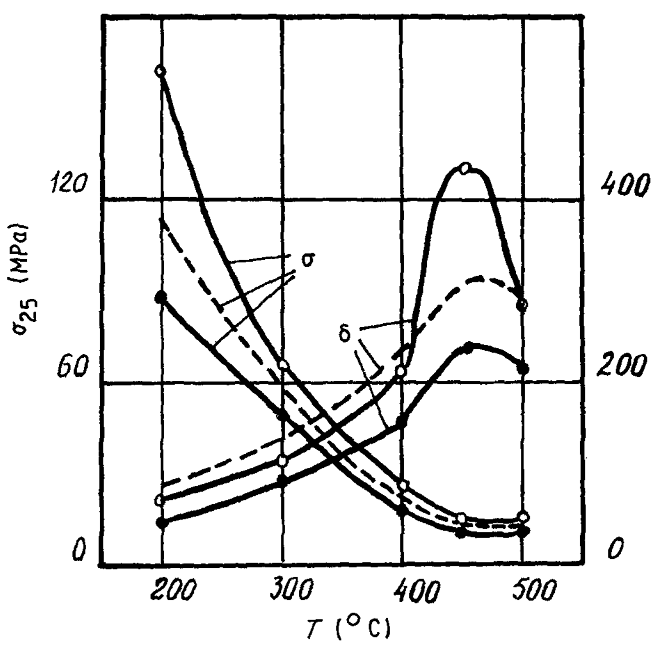

Figure 2 Temperature dependence of σ and δ measured at $\dot{\varepsilon} = 2.1 \times 10^{-3}\,\mathrm{sec}^{-1}$; --- state I, —— state II.

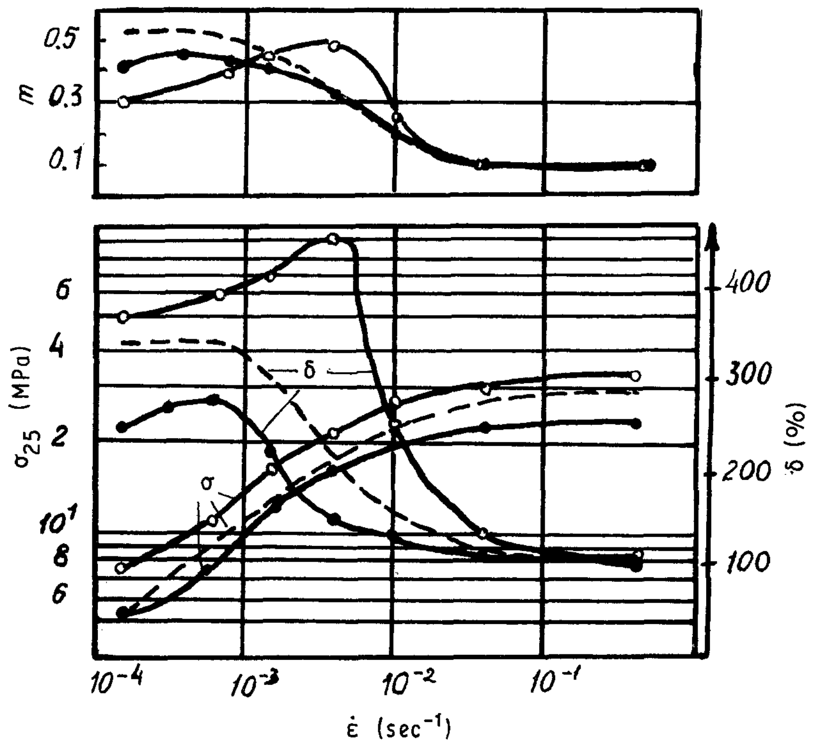

Figure 3 Strain-rate dependence of σ, δ, and m measured at 450° C; - - - state I, —— state II; ○ longitudinal specimens, ● transversal specimens.

greater flow stress in much higher than that of the transversal ones. The difference in δ is especially high at the temperature of the alloy's superplastic state. The comparative analysis of the strain-rate dependence of the flow stress, relative elongation, and factor m for MA15 alloy longitudinal and transversal specimens in state I and state II shows that in the textureless alloy mechanical isotropy is preserved in all testing directions (Fig. 3). At the specific strain-rate factor m and relative elongation δ have no clearly expressed maximum. The optimum SP strain rate for state I lies in a wide range of strain rates from $\dot{\varepsilon} = 10^{-4}\,sec^{-1}$ up to $\dot{\varepsilon} = 10^{-3}\,sec^{-1}$.

The investigation of the strain-rate dependence of the SP parameters during the tensile tests of the states II longitudinal and transversal specimens has shown a considerable difference in the alloy's behaviour as compared with the textureless state. The flow stress of the longitudinal specimens proved to be higher than that of the transversal ones throughout the strain-rate testing range. A strong dependence of the flow stress on the strain rate is observed for the longitudinal specimens in the strain rates range $\dot{\varepsilon} = 2 \times 10^{-3} + 5 \times 10^{-3}\,sec^{-1}$. The change of the m value correlates well with the value of the relative elongation.

Maximum m and δ for the transversal specimens are observed under lower strain rates ($\dot{\varepsilon} = 5 \times 10^{-4}\,sec^{-1}$), the rate range of the SP flow for the transversal specimens being higher than that for the longitudinal ones.

The σ–ε ratios of the alloy in different initial states show that the flow stress of the textureless state under superplastic conditions is similar for both directions. The flow stress isotropy is preserved with the increase in the rate and degree of deformation (Fig. 4a). Contrary to the results obtained for Zn 22% Al [12, 17] the existence of the initial texture leads to the flow stress anisotropy on the initial stage of the SPD (curves I, Fig. 4b). The strain grows with the increase of the strain rate.

The flow stress anisotropy decreases with an increase of the strain degree. Above a strain of about $\varepsilon = 70\%$ the state II longitudinal and transversal specimens have the same σ values which are practically the same as the σ value of the textureless state.

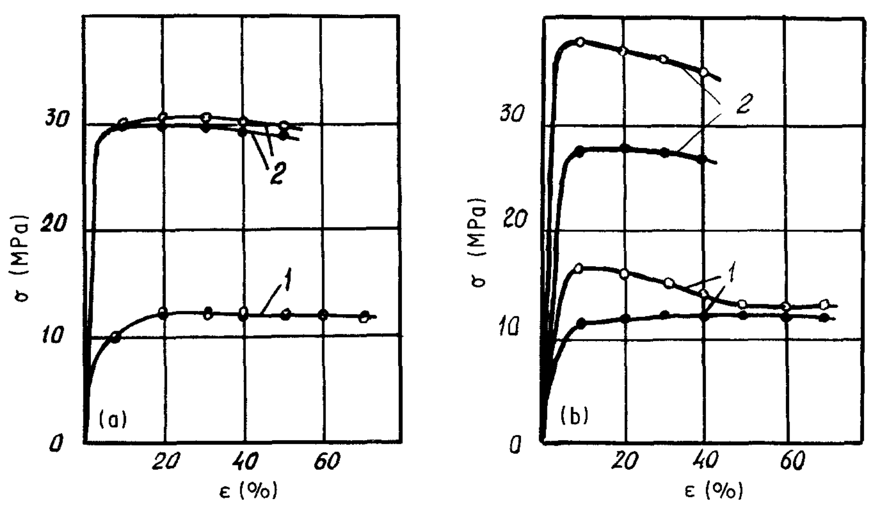

Figure 4 Stress–strain curves of the MA15 alloy at 450° C; 1, $\dot{\varepsilon}_{opt}$, 2, $\dot{\varepsilon} > \dot{\varepsilon}_{opt}$; (a) state I, (b) state II; ○ longitudinal specimens, ● transversal specimens.

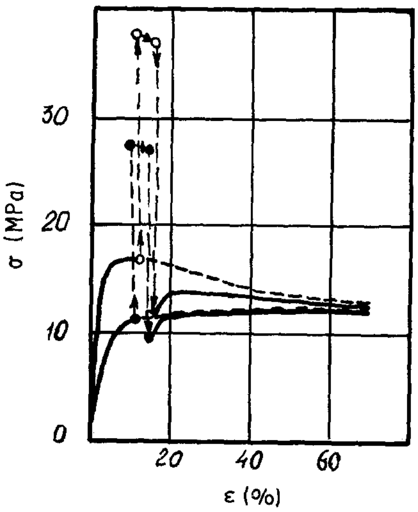

Figure 5 Stress–strain curves for the state II MA15 alloy with strain rate switched by steps at 450° C; ○ longitudinal specimens, ● transversal specimens; --- the switching of the strain rate.

The analysis of the specimens texture after ε = 70–100% strain has indicated that while straining a new axial texture is formed in the textureless state under any strain rate. In the area of a higher strain-rate sensitivity of the flow stress the texture is considerably weaker ($I_{max}/I_1 = 3$). The rise in the strain rate of up to $\dot{\varepsilon} = 5.1 \times 10^{-2}\,sec^{-1}$ causes the strain to increase ($I_{max}/I_1 = 6$). In the alloy with a sharp texture (state II) the SPD leads to a considerable weakening of the initial axial texture ($I_{max}/I_1 = 3$). With the decrease or increase of the strain rate (areas I and III of the superplastic curve) the texture weakens to a lesser extent.

The process of the void formation in states I and II specimens of the MA15 alloy after the tensile test was found to be essentially different for the longitudinal and transversal specimens both with and without texture. Under the optimum SP flow conditions the specific volume fraction of voids in the state II longitudinal specimens is higher than that in the transversal ones and amounts to $\gamma = 0.5\%$ at $\varepsilon = 50\%$, and $\gamma = 1.2$ and 1.8% at ε = 100 and 150% respectively. For the transversal specimens γ is 0.11%, 0.24%, and 0.27%, the degree of the strain remaining the same.

The specific volume fraction of the voids for the textureless specimens has intermediate values.

The strain-rate dependence of the voids formation correlates accurately enough with the changes of δ and m.

The influence of the prestraining on the textured alloy mechanical properties under the SPD was also studied.

The specimens were strained in both directions at the rate of $\dot{\varepsilon} = 8.3 \times 10^{-4}\,sec^{-1}$, but in the initial stage the strain rate was switched by steps two orders higher than the given rate, with the strain reaching ε = 2–5% (Fig. 5). This results in a more intensive generation of the lattice dislocations both in the longitudinal and transversal specimens. After the straining the specimens were quenched in water, with no significant microstructural changes being found after the straining. The reverse switching of the strain rates to the initial level results in various flow stress changes, depending on the type of specimens.

The flow stress in the longitudinal specimens decreased to the level of the flow stress in the transversal ones, then it started to increase, but up to $\varepsilon = 80\%$ it failed to reach the flow stress value of the longitudinal specimens strained without the rate switching.

The transversal specimens flow stress also decreased after the rate switching, but after $\varepsilon = 10\%$ it increased up to the flow stress level of the same specimens strained without the rate switching. From there on the flow stresses of the transversal specimens strained either with or without switching were identical.

The electron–microscopic investigation of the longitudinal and transversal specimens after the straining at the optimum rate of the SPD has indicated that in grain boundaries there are linear defects, whose contrast indicates their dislocational nature (Figs 6 and 7). In the literature there are different views on the origin of these defects. However in recent years, with the foils being strained directly in the column of an electron microscope, it has been definitely proved that they are nothing other than trapped lattice dislocations (TLDs). Their density after $\varepsilon = 10\%$ is practically the same for both longitudinal and transversal specimens and amounts to $\varrho = 5.0 \times 10^4 \pm 1.2 \times 10^4\,cm^{-1}$ and $\varrho = 5.7 \times 10^4 \pm 10^4 \pm 1.4 \times 10^4\,cm^{-1}$ respectively. The peculiarity of both the states is that after the SPD TLDs have a diffuse image. This must be connected with their relaxation at the SPD temperature [18, 20].

The switching of the strain rate to a higher $\dot{\varepsilon}$ ($\dot{\varepsilon} = 2.1 \times 10^{-1}\,sec^{-1}$), characteristic of the ordinary hot straining, considerably changes the dislocational structure. In the transversal specimens the density of TLDs rises insignificantly ($\varrho = 5.5 \times 10^4 \pm 1.4 \times 10^4\,cm^{-1}$). Sliding is observed only in one slip system. Under these conditions TLDs have a distinct contrast. The flow stress in the longitudinal specimens with a high strain rate is much higher, with the multiple sliding starting and TLDs of several slip systems being observed in grain boundaries. Their density is much higher than that of the transversal specimens and amounts to $\varrho = 1.4 \times 10^5 \pm 1.4 \times 10^4\,cm^{-1}$.

The reverse switching of the strain rate to the initial level results in a sharp fall in the flow stress, and only then with the use of surplus TLDs the flow stress reaches its initial level. The TLD density fixed for the steady-state stage of the flow stress is close to the value of the TLD density before switching and amounts to $\varrho = 4.7 \times 10^4 \pm 1.1 \times 10^4\,cm^{-1}$. After the reverse switching of the strain rate in the transversal specimens the flow stress decreases to a lesser extent and after the 3–5% straining it already reaches the initial stress level. From there on the flow stress grows monotonously. The TLD density in this case is a bit higher and amounts to $\varrho = 6.0 \times 10^4 \pm 1.5 \times 10^4\,cm^{-1}$.

The grain boundaries crystallographic parameters were defined on the given structures typical for each case (Figs 6 and 7). In all cases they had random orientations without any special boundaries.

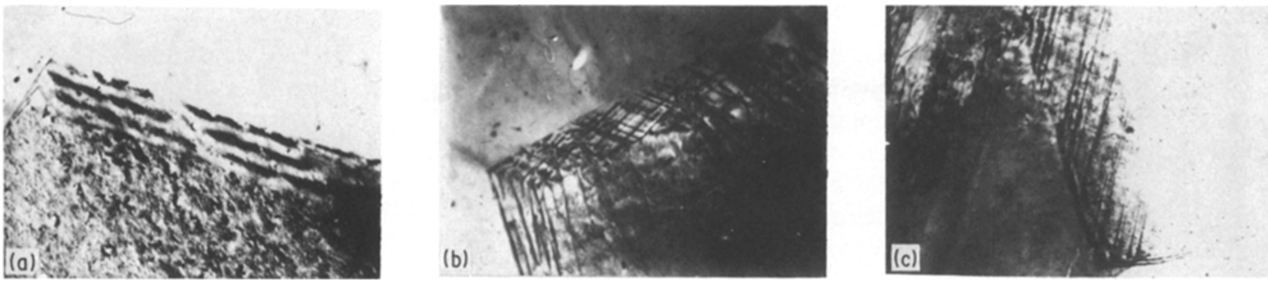

Figure 6 Microstructures of the MA15 alloy (longitudinal specimens); (a) $\dot{\varepsilon}_{opt}$, (b) $\dot{\varepsilon}_{opt} \rightarrow \dot{\varepsilon} > \dot{\varepsilon}_{opt}$, (c) $\dot{\varepsilon}_{opt} \rightarrow \dot{\varepsilon} > \dot{\varepsilon}_{opt} \rightarrow \dot{\varepsilon}_{opt}$ ($\times$ 22 000).

4. Discussion

The flow stress anisotropy under ordinary straining is accounted for easily taking into consideration the preferred orientation of the basic planes with reference to the effort applied.

It is more difficult to explain why the flow stress anisotropy is preserved under the SPD conditions, for in this case the main deformation mechanism is GBS [17, 21], and it seems that its effect must not depend on the grain orientation with respect to the applied effort. Besides the high m value indicates that GBS controls the flow stress of the alloy under the SPD. The attempt to connect mechanical anisotropy with the lines of intermetals in the initial state contradicts the experimental results. In spite of the lines of intermetals the anisotropic behaviour of the textureless alloy appears rather slight (Figs 2 and 4a).

It is more probable that the flow stress anisotropy under the SPD is conditioned by the development of the GBS of the "stimulated" IDS [22, 23]. According to this conception, along with the effect of "pure" GBS conditioned by the generation and motion of the grain boundary dislocations (GBD), the development of IDS is needed in order to provide for the SPD. The results of the strain-rate switching experiment have indicated that the influence of IDS on the GBD development under the SP flow is doubtless. The introduction of dislocations in grain boundaries due to IDS reduces the flow stress, indispensable for the GBS development and lowers the SPD flow stress. The IDS development in its turn depends on the grain orientation with reference to the actual stresses. This accounts for the preservation of the flow stress anisotropy under the SPD.

Thus the difference in the longitudinal and transversal specimens flow stresses is conditioned by the difference in the ease of the IDS development which, due to the peculiarities of its interaction with grain boundaries, influences GBS.

The decrease of the flow stress anisotropy with the increase of the strain degree is connected with the texture changes under the SPD. As has been already noted above, in the course of the SPD in the textured state a considerable weakening of the original texture occurs. This results in the reduction of the grains with preferred orientation in a certain direction, and decreases σ and m anisotropy. One should note that the SPD is observed in the said alloy when the grains of comparatively large size are available, which is not the case with other metals and alloys. This results in the reduction of the total length of the grain boundaries and the lack of the clearly expressed maximum m in the specimens without crystallographic texture (Fig. 3).

The difference of the optimum strain rates under SP conditions of both longitudinal and transversal textured specimens gives the possibility to show the influence of the texture on the position of the strain-rate range which corresponds to a maximum m. These specimens are characterized by one and the same grain boundary structure and the decisive role here is played by the orientation of the crystals with respect to the effort applied. This is the reason why the stress level of the longitudinal specimens, needed to maintain IDS, is lower than that of the transversal ones and is reached at a lower strain rate. The maintenance of IDS in the transversal specimens needs the increase of flow, which is attained with the increase of the strain rate. As the kinetic correspondence between IDS and GBS for the specimens of different types is achieved at different strain rates, the strain-rate value corresponding to maximum m also changes.

The anisotropy of the textured specimens is essential. It may be the result of the difference in the accommodation procedures and voids formation [17, 18, 21, 24]. In some cases porosity facilitates GBS [24], so that large values of the relative elongation can be obtained. This seems to be the reason of the diversity of δ in the longitudinal and transversal specimens.

Figure 7 Microstructures of the MA15 alloy (transversal specimens); (a) $\dot{\varepsilon}_{opt}$, (b) $\dot{\varepsilon}_{opt} \rightarrow \dot{\varepsilon} > \dot{\varepsilon}_{opt}$, (c) $\dot{\varepsilon}_{opt} \rightarrow \dot{\varepsilon} > \dot{\varepsilon}_{opt} \rightarrow \dot{\varepsilon}_{opt}$ ($\times$ 22 000).

5. Conclusions

The crystallographic texture exerts considerable influence on the mechanical properties in the course of the SPD of magnesium alloys. Mechanical anisotropy is observed in the textured state and is absent in the textureless one. The existence of the texture results in the shift of the optimum strain-rate range of the SPD depending on the testing direction, the SP flow parameters being different in different directions.

The influence of the texture on the anisotropy of properties under the SPD is conditioned by its influence on the development of the IDS which in its turn influences the development of GBD – the main type of deformation conditioning the SP flow.

The influence of the texture on the development of the void formation under the SPD and, accordingly, on the alloy's plasticity in different directions has been established.

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