The role of deformation twins in recrystallization texture formation of the FeGa alloy

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Abstract. It has been found that at cold axial deformation of the (Fe83.4Ga16.6)99.9(NbC)0.1 alloy by hydrostatic extrusion, along with the slip processes, deformation twins are formed. The orientations of these formations in the plane perpendicular to the deformation direction coincide with the crystallographic planes {100} or {110}. It is believed that in bcc alloys, the axial deformation texture <110> is stable and undergoes only a small scattering upon recrystallization. Using the EBSD technique in this work, we have shown that the desired <100> // DD grains can grow into a deformed <110> // DD matrix upon primary recrystallization. At the same time, recrystallization of as-deformed samples leads to almost complete destruction of the original texture. The <100> orientation probably develops in connection with the presence of nuclei in the form of the least deformed twins, which are not observed in ordinary iron alloys. This feature of Fe-Ga alloys gives principal possibility of creating necessary cube texture and achieve a high magnetostriction value.

1. Introduction
The need to create a cubic texture in bcc (body-centered cubic) iron-based alloys is dictated by their field of application. The most widely known representatives of this class of materials are FeSi alloys, which are widely used as materials for magnetic cores of transformers and for electric motors. To achieve the best magnetic properties, it is necessary that the magnetic flux flows in one of the directions of easy magnetization <100>. Fortunately, in the case of magnetic cores, this may not necessarily be the cube texture {100} <100>, but also the Goss texture {110} <100>, which also contains the directions <100> // rolling direction (RD). This texture component (<100> // RD) is also called η-fiber. FeGa alloys near the first magnetostriction peak (~20% at. Ga) with anomalously high magnetostriction [1] are another representative of the group of Fe-based materials in which the η-texture is desirable. This is due to the anisotropy of magnetostriction, which in the <100> single crystal can reach 400 ppm, and in the <111> single crystal, at equal conditions, its value approaches zero [2]. Technology of Goss texture creation during secondary recrystallization was successfully applied to FeGa by various authors [3,4]. However, large grains after secondary recrystallization with diameter up to centimeters led to significant deterioration of the alloy strength, which is an important property for many contact applications of FeGa. For this reason, it is necessary to find a technology of η-texture creation using primary recrystallization process. To develop the technology, it is necessary to detailed investigate texture formation processes in FeGa alloys. Last few years such studies are actively carried out [2,5]. In [6], we studied deformation behavior and texture in FeGa after cold axial deformation by hydrostatic extrusion and recrystallization annealing. Strong axial <110> texture that is characteristic for bcc alloys was observed after deformation. It is well known that such kind of deformation texture in bcc is stable, and undergoes only a small scattering upon recrystallization. It was shown for FeGa also [7]. The aim of this work is twofold: first, to study the texture of the FeGa
alloy after axial deformation in detail; second to study the recrystalization behavior of the $<110>$ axial texture of deformation and find the possibility of increasing of the $<100>$ // deformation direction (DD) orientations during primary recrystalization. It has both fundamental and practical importance because it can improve our knowledge about texture formations in bcc materials and it may help us to develop technology of desirable texture creation in FeGa.

2. Material and methods
A non-oriented polycrystalline ingot (Fe$_{83.4}$Ga$_{16.6}$)$_{99.9}$NbC$_{0.1}$, obtained by arc melting and conventional casting into a copper mold, was used as a starting material. A cylindrical ingot with a diameter of 12.2 mm was placed in a brass shell and subjected to hydrostatic extrusion at room temperature. This deformation was carried out in two passes: the first from 12.2 to ~ 10.5 (the exact diameter is unknown, since after the first pass the sample was not removed from the shell) and the second to a final diameter of 8.4 mm. The degree of deformation was 53%. Samples were cut from the deformed rod to study the structure and texture after deformation, as well as to carry out annealing in different modes. To remove residual stresses without structural transformations, we used annealing at 550°C 60 min. The purpose of this annealing was to improve quality of EBSD analysis of the texture of deformation. Recrystallization annealings of deformed samples were carried out at 800 or 950°C with a holding time of 40 minutes to get samples with different recrystallization stages. In addition, some of the deformed samples were annealed with a slow heating of 8°C/min from 400 to 1000°C without holding at the final temperature. To study the microstructure and texture of samples SEM FEI Quanta 200 and Tescan Vega 3 with EBSD attachments were used. Post processing of the data obtained, was realized using commercial EBSD software and MTEX toolbox [8]. Color-coding in the orientation maps always corresponds the directions along deformation direction (DD). In more detail this means that structure elements with $<100>$ // DD orientations are painted red, $<110>$ // DD are painted green and so on.

3. Results and discussion
The initial texture of the ingot is multicomponent without any prevailing orientations. The average grain size is about 600 µm. NbC particles are evenly distributed throughout the ingot body [6]. The structure after the deformation is shown in figure 1. Structure after 550°C 60 min annealing is also presented here (fig.1.b) In the course of such annealing, the processes of recovery and polygonization begin, and the quality of structure image, obtained by EBSD method, improves. It can be seen that the $<110>$ orientation prevails in the cross section of the deformed sample; axial deformation during hydrostatic extrusion, as for other bcc alloys, leads to the axial texture $<110>$. Besides the stable orientation $<110>$, there are scattered orientations near the direction $<113>$ (fig.1e and f). In bcc metals slip systems are $\{hkl\}<111>$. The most-density packed planes are the $\{110\}$, although the packing density is only slightly different from the planes $\{112\}$ and $\{123\}$ [9]. Plane $(110)$ has four equivalent directions $<111>$; two of them should be located symmetrically about the axis of deformation. This is possible for two crystallographic directions of the axis $<100>$ and $<110>$ [10]. Practically, $<110>$ direction is realized because angles between $<110>$ and $<111>$ are less than between $<100>$ and $<111>$. This is the case for all initial orientations except ones, which lie on the line of symmetry $<001>$ - $<111>$ of standard orientational triangle. Such orientations have a trend to align along $<113>$ [10]. In figures 1a and b we can see structural elements with parallel boundaries and with a sharp change in orientation relative to the matrix. Despite these elements differ from “canonical” twins – narrow straight areas, everything points to the fact that they are result of twining. Their irregular shape can be explained by the fact that if they were formed at the initial stage of deformation, then in the course of subsequent deformation, they undergo changes. Orientation of such elements depends on the surrounding matrix, one example presented in figure 2. For a more detailed study of the texture features, we carried out selective scanning with a higher magnification and a smaller scan step. One of these structure fragments is presented in figure 2.
Fig. 1. Orientation maps of the sample after 53% cold deformation by hydrostatic extrusion (a) and after 550°C 60 min annealing (b) with corresponding Inverse pole figures (IPF) (c,d) and distribution of the main components of the deformation texture by the angle of deviation from the ideal position (e,f).

Fig. 2. Orientation map (x500) of the sample after 53% deformation and 550°C 60 min annealing (a) with the corresponding image quality map (b).

Here we can see narrow, parallel-sided twins, between other structural elements of the deformed material. This areas have higher image quality compared with the surrounding matrix, it is may be result of lower level of internal stresses here. To realize partial recrystallization we carried out annealing at 800°C 40 min. The main axial texture components are shown in figure 3a. After such annealing, ratio of recrystallized and unrecrystalized (deformed) areas is about 50:50 (fig.3b). Unrecrystalized areas are presented by large grains with “cube of the edge” orientation {110}<100>. This orientation is known to be stable, hence it is still retained after annealing at 800°C. Not all green grains have the orientation {110}<100>, there are new recrystallized grains with different <110>//DD orientations also (fig.3a). It is seen, how recrystallized grains with different orientations grow into large <110> grains (black dotted circle in fig.3a). This feature of grain growth during recrystallization may be caused by presence of thin long twins inside grains after deformation. This twins can serve as recrystallization nuclei, a similar phenomenon was observer in FeSi [11]. The volume fraction of η-grains after such partial recrystallization is 14.4%.
Fig. 3. The main texture components <110>//DD and <100>//DD with grain boundaries after partial recrystallization at 800°C 40 min (a); volume fraction of recrystallized and deformed structure areas (b).

Fig. 4. The main texture components <110>//DD and <100>//DD after recrystallization at 950°C 40 min (a); volume fraction of recrystallized and deformed structure areas (b).

After high temperature annealing at 950°C 40 min, primary recrystallization almost completed (fig.4b); however, structure is still quite uneven-grained (fig.4a). Components of texture of deformation almost disappeared, whereas volume fraction of <100>//DD was increased to 18.6%. Both {100}<001> and {110}<001> orientations are present. Annealing at higher temperature led to increase in η-texture volume fraction. To study the effect of annealing conditions on the formation of the grains with the <100> // DD, we carried out additional annealing with gradual heating from 400 to
1000°C at 8°C/min. The starting point of 400°C instead of room temperature was chosen in order to reduce the annealing time, as we found that below 400°C structure transformation processes (recovery) have extremely low intensity.

![Orientation map of the sample after heating from 400 to 1000 at 8°C/min.](image)

**Fig. 5.** Orientation map of the sample after heating from 400 to 1000 at 8°C/min.

We can see that structure still contains large deformed grains with initial orientation. At the same time, η-grains are significantly bigger than after different annealing regimes, its volume fraction is 20%. We consider that annealing with slow heating can promote low stored energy nucleation process due to recovery prior recrystallization onset during gradual heating [12]. Accordingly well-known orientation dependency of stored energy in bcc metals [13]; η-orientations have low level of stored energy after deformation and, therefore, low ability to grow in frame of high stored nucleation conception. Obviously, cold axial deformation and slow heating of FeGa sample create conditions for the nucleation and growth of these orientations.

In summary, we can say that during deformation of the FeGa alloy by cold hydrostatic extrusion, not only slipping, but also twinning is observed. We consider that reason of twinning is low plasticity of this alloy and the limited ability to slip during cold axial deformation with high intensity. It is known that hot and warm drawing of FeGa led to axial texture <110> [7], which remains stable during recrystallization; only its slight scattering occurs. The similar phenomena observed also in other bcc materials, for example Fe-5%Ni [8]. Probably, this happens because as a result of slipping processes during such deformations there are no structural elements that are sufficiently misoriented with the matrix to be able to grow into surrounding matrix. The appearance of deformation twins upon cold deformation of an alloy with low plasticity makes it possible for them to be nuclei of recrystallization and leads to a sufficiently large volume of η-texture in the recrystallized material. This occurs even despite the fact that the twin boundaries have low-mobility.

**4. Conclusions**

Deformation by the hydrostatic extrusion of the alloy (Fe$_{83.4}$Ga$_{16.6}$)$_{99.9}$NbC$_{0.1}$ led to a pronounced axial texture <110>. In contrast to other cases of axial deformation of bcc iron-based alloys after hydrostatic extrusion of FeGa, deformation twins with different orientations including <100> // DD were observed in the texture of deformation. Recrystallization annealing of the deformed alloy led to significant volume of <100> // DD orientations in the texture. We consider that the reason is that twins with this orientations can play the role of recrystallization nuclei in the axial <110> deformation texture. At the same time, annealing with the slow heating enhances this tendency.
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