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THE MODEL OF STRUCTURE REFINEMENT IN METALS AT LARGE DEFORMATIONS AND FACTORS EFFECTING GRAIN SIZES

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Abstract. The increase in the curvature of sample's bending and torsion at large cold deformation leads to the decrease in the sizes of the formed fragments and grains in pure metals, and to the increase of angular misorientations of their boundaries. Bending results in the change of the specific area of the deformation center surface ΔA_s, i.e. the parameter responsible for an average curvature of a sample and crystal lattice of grains. Extremely fine fragments are formed when the above parameter increases during deformation to the value equal to the maximum tensor density of dislocations β. This is provided by the increased value of the initial specific surface of the deformation center A_i/V – the ratio between area and volume. The peculiar features of deformation center and structure refinement in quasi-monotonic and non-monotonic processes are considered. It is shown that structure refinement is dependent not only on strain but also on the field of angular velocities of fragments' rotation in the deformation center.

1. INTRODUCTION

Processing ultra-fine grain (UFG) structure (grain size being less than 1μm) is the basis of one of the approaches to produce metals and alloys with high structural, functional and technological, e.g. superplastic, properties. In bulk materials, UFG structure can be processed using special techniques, such as equal-channel angular (ECA) pressing, torsion under pressure and a number of other ones that have received the generalized name – methods of severe plastic deformation (SPD) [1].

Their peculiar feature is the accumulation of large deformations in materials. For processing nanocrystalline structure, deformation is performed at the temperature below that of recrystallization without intermediate annealing operations between passes.

As known, the formation of micro-regions – fragments that are generally divided by low-angle and/or high-angle boundaries [2] – inevitably occurs in metals at large deformations. In some cases, the angular misorientations of fragment boundaries increase slightly with strain, remaining within the values corresponding to sub-grains, while in other cases the misorientations increase up to the values typical for grains. If the character of flow is close to monotonic, i.e. it is quasi-monotonic, in the deformation method used, a sub-grain structure is mainly formed, while a granular structure is generated at non-monotonic flow [3]. In the mentioned methods

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of SPD, the deformation character is non-monotonic; the direction of material's deformation is changed so that, as a result, its shape, dimensions, and area of outer surface remain close to the starting ones.

For the same materials, the sizes of minimum refined fragments (sub-grains and /or grains) depend on deformation technique, as well as on the deformation mode and regimes (strain, strain rate, temperature, hydrostatic pressure) and dimensions of samples. The lower is homologous deformation temperature and the higher are the strain and hydrostatic pressure, and the smaller are dimensions of samples, the finer sizes are acquired by the minimum refined fragments. The effect of strain rate on refinement is ambiguous. An increase in strain rate makes for the intensive division of initial grains into fragments because of activation of twinning and slipping of dislocations along crystallographic systems that are usually unfavorable at low strain rates [4]. At the same time, high strain rates are capable to warm-up the zones of deformation localization, leading to boundaries migration and growth of fragment sizes.

In this study, we show that the sizes of refined fragments in pure metals depend on such geometric parameters of the deformation center as the ratio between the starting surface area and its volume and on the variation of this value with deformation (note that all other factors remain the same here). As these parameters increase, an average size of the refined fragments decreases approaching the minimum value.

It should be pointed out that the sizes of refined fragments in metallic materials generally depend on their chemical composition and stacking fault energy (SFE). But, according to [5], surface impurities in pure metals during deformation do not affect the structure evolution and sizes of refined grains deep within samples at torsion under pressure. As for the effect of the second factor, by comparing [1,6] the sizes of grains in metals with different SFE, refined using various SPD methods (e.g. Al, Fe(α), Ni, Cu) we would like to note the following.

Though a considerable grain refinement is achieved in the above metals after ECA pressing, the attempts to produce a uniform nanocrystalline structure were not successful. After torsion under pressure of the same metals, there occurs more profound refinement of grains up to sizes corresponding to nano-meter range. These peculiar features cannot be attributed to the difference in SFE, cannot those be ascribed to the influence of the accumulated strain values.

The strain is brought to satiation in both methods of SPD, i.e. to such values, at which further deformation increment does not lead to noticeable refinement of grains.

It would be shown below that the depth of grain refinement depends not only on the above geometric parameters closely associated with the used deformation mode, but also on the rotation tensor, that together with deformation tensor, determines the distortion of a crystalline material in the deformation center.

2. FRAGMENTATION AS RESPONSE OF A CRYSTAL LATTICE TO SAMPLE' BENDING

Geometry of bending and mass transfer

A sample (billet) in some way or other undergoes bending in many deformation methods. At ECA pressing without backpressure, the deformation center in the zone of channels intersection has usually a fan shape, and, in this case, sample's bending is clearly pronounced. In this zone, it is seen from the distortion of a grid of orthogonal hairlines scribed in a sample's longitudinal plane that the distances between transverse hairlines decrease when bending radius is small, and increase when this radius is large. With increasing the bending radius, longitudinal hairlines acquire the shape of circular arches. Not counting typical distortions caused by friction forces, such change in grid's shape during ECA pressing corresponds to the picture of its change during pure bending.

In pure form, bending can be presented as an operation of twisting a flat sample (of t thickness) around a rigid cylinder with radius R through bend angle ϕ (Fig.1).

According to Polyani-Taylor principle, the grains in a polycrystal change shape and sizes during deformation in the same manner as a sample, and, hence, should acquire the curvature that more or less corresponds to sample's bending. It should be noted that the change in longitudinal sizes of low-angle fragments (sub-grains) generated during deformation is not in good agreement with this principle [7]. At the same time, the change in longitudinal sizes of banded structures, possessing low-angle fragments, formed at straining $\varepsilon > 0.2$-0.3 [8], is usually in conformity with this principle.

At relatively small bends of a sample, its associated curvature in crystal lattice of grains is produced by individual dislocations distributed in grains. When the bending is large, the crystal lattice of grains is fragmented into micro-regions – polygons [9]. In

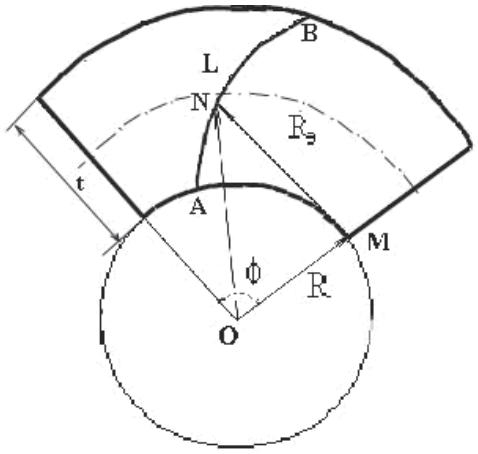

Fig. 1. The scheme of bending of a plate twisted at cylinder with radius R, ϕ – bending angle, R_e – the local radius of AB elolvent in the point N.

this case, the bending angle is mainly compensated by the sum of angular misorientations of boundaries that divide them.

At bending, changes occur in length values, surface areas, and volumes* of extended and compressed portions of a sample. These values decrease from the side of a small radius, while from the side of a large radius they increase. According to [10], transverse stresses push out from sample's volume the dislocations of one sign and make the dislocations of another sign to move to a neutral line (plane). The direction of dislocation movement corresponds to the family of evolvents, in particular, to evolvent AB shown in Fig. 1.

The effect of bending curvature on the sizes and misorientations of fragments

Let us assume that the crystal lattice of a grain is divided into N identical fragments with an average diameter d_f during bending in the direction of the family of evolvents.

Since N >>1, the number of segments – boundaries of fragments N_b crossing an evolvent with length L can be taken equal to:

$$N_b \approx \frac{L}{d_f}. \tag{1}$$

The length of an evolvent will be:

$$L = \frac{R\phi^2}{2}. \tag{2}$$

The local radius of evolvent curvature is:

$$R_e = R\phi. \tag{3}$$

Since the circumference R is an evolute (the curve of curvature centers' location) of an evolvent, then the triangle OMN is rectangular, hence:

$$R + t = \sqrt{R^2 + R_e^2}. \tag{4}$$

where t is the current cross size (thickness) of a sample.

Solving Eq.(3) with regard to ϕ and taking into account Eq. (4), one will have:

$$\phi = \left(\frac{2t}{R} + \frac{t^2}{R^2}\right)^{\frac{1}{2}}. \tag{5}$$

An average angle of boundary misorientation between fragments that are transverse to evolvents will be determined as:

$$\theta_a = \frac{\phi}{N_b} = \frac{d_f\phi}{L}. \tag{6}$$

With respect to Eqs. (2) and (5) and accounting for the fact that R<<t for ECA pressing, one will have:

$$\theta_a \approx 2d_f k_e. \tag{7}$$

where k_e is an average curvature of an evolvent.

In a similar way, the average misorientation angles of fragment boundaries that are transverse to the longitudinal bending lines of a sample (circular arches) by which the mass transfer also occurs, will be determined as:

$$\theta_a^{'} \approx d_f k_b. \tag{8}$$

where k_b is the curvature of longitudinal bending line.

It follows from Eqs. (7) and (8) that the larger sample's bending, and, consequently, its curvature, the smaller sizes and larger angular misorientations are acquired by fragments.

Maximum values of local curvature of evolvents and circular arches are typical for a zone with a minimum radius of sample's curvature. The formation of the finest fragments with larger misorientations can be expected in this zone, while the fragments formed in the zone with larger bending radius will

* On the whole, for a sample the volume $V \approx$ const.

have larger sizes and smaller angular misorientations.

It should be noted that the resulting curvature of sample shape can be close to zero if the direction and bending angle have been changed during deformation. But, the value of the accumulated deformation in a sample increases in this case; adequately increases the curvature of crystal lattice.

Thus, when bending is large, the change in the grain shape and size results in the formation of micro-regions – fragments, that, as known, are observed during electron microscopy and X-ray investigations in a form of coherent scattering areas. Low-angle fragment boundaries are arranged transversely to curvature lines and compensate the bending angle in the aggregate. The larger is the sample's bending and, consequently, its curvature, the smaller are sizes and larger angular misorientations are acquired by fragments.

3. THE EFFECT OF SURFACE ON FRAGMENTATION AND FRAGMENTATION ON DEFORMATION

The change in the surface area of deformation center*

When a sample undergoes bending, a mass transfer occurs between the areas of tension and compression, and, consequently, the areas of their outer surface change. The absolute change in the outer surface area is caused by dislocations, which penetrate into to the surface and go away from it:

$$|\Delta A| = NbI_a, \tag{9}$$

where N, b and l_a are, respectively, number, Burgers vector and average length of dislocations.

By dividing both parts of Eq. (9) by V – the deformed volume, one will receive:

$$\Delta A_s = b\rho, \tag{10}$$

where ρ is the scalar density of dislocations**, and ΔA_s – the change in the surface specific area.

According to [11], an average lattice curvature at bending can be expressed through the density of dislocations $k = b\rho$. Therefore:

$$k_a = \Delta A_s. \tag{11}$$

It is known that $b\rho \geq \beta$, where β is the tensor density of dislocations. By definition, $div\beta=0$ [10]. Hence, the tensor density of dislocations in the deformed volume can be changed only by making dislocations to go out and to enter through its surface.

At ECA pressing, sample's bending is constrained and localized in a narrow (with regard to the whole sample) volume – deformation center, the surface of which incorporates outer (external) and inner portions. Some factors are known that facilitate the generation of dislocation sources at the outer surface of deformation center [11,12], especially at the direct action of pressure exerted by tooling. The inner surface of deformation center, i.e. the border that separates the regions, where deformation is significant from the region where it is insignificant or is totally absent, is also an active source or discharge of dislocations [13,14]. That is why, there are the outer and inner portions of deformation center surface that exert influence on b; so, Eqs. (9-11) can be considered with respect to the whole surface of the deformation center.

Extremely fine sizes of fragments

From Eqs. (7) and (11) one will have:

$$d_f = \frac{\theta_a}{2\Delta A_s}. \tag{12}$$

For low-angle boundaries of fragments $\theta_a = b/h$, where h is the distance between dislocations in a wall. The h value is limited by the fusion of dislocation cores: $h \geq (2\text{-}3)r_c$. In turn, $r_c = (2\text{-}3)b$, where r_c is the radius of dislocations' core. Hence, $h \approx k_c \cdot b$, where coefficient $k_c \approx 4\text{-}9$. Thus, an average fragment diameter (ascribed to Burgers vector) can be determined as:

$$\frac{d_f}{b} \approx \frac{K_A}{\Delta A_s}, \tag{13}$$

where coefficient $K_A = 1/(2k_c b)$ characterizes an average curvature of dislocation boundary.

Minimum fragment sizes should, at least, be several times larger than b. As for many metals $b \sim (300\text{-}400)\cdot 10\ m^{-12}$, then $K_A \approx (1.4\text{-}4.1)\ 10^8\ m^{-1}$. According to various estimations, the maximum lattice density of dislocations is $\rho_{max} \sim (10^{15} - 10^{16})\ m^{-2}$,

*The changes in the areas of surface irregularities of all orders, from micro- to nano- values, are not taken into account because in many respects they are determined by the quality of tooling surface that can be assumed as constant.

** The initial density of dislocations is not taken into account, since in a non-deformed metal this value is low.

because $|\beta| \leq \rho\, b$; then $|\beta|_{max} \leq 4 \cdot 10^6$ m^{-1}. Hence, $d_{f\,min}/b \approx (0.4\text{-}1)\ 10^2$ or $d_{f\,min} \approx (40\text{-}100)b$.

The obtained assessment is in agreement with known experimental data on the sizes of low-angle fragments – sub-grains and nano-grains produced in metal powders at grinding in ball mills [8].

ΔA_s parameter as the characteristic of material curvature

The change in the specific area of deformation center surface ΔA_s is a macroscopic parameter; at the same time it characterizes the micro- and meso-scale level, i.e. the tensor density of dislocations. Usually, the volume of averaging $(\Delta x)^3$ in tensor density of dislocations corresponds to the interval $\rho^{1/2} << \Delta x < d_g$ [11], where d_g is grain size. The lower limit characterizes an average distance between dislocations. The upper limit is restricted by grain size, because dislocations do not cross a grain boundary forming a system of loops or grids in them. But, with the onset of fragmentation, starting with strains ~ 0.2-0.3, and further with its development, the initial grain boundaries and grains become practically indistinguishable in the formed fragmented structure. Instead of them, banded structures of various types are formed in the appropriate regions; their length oftenly exceeds the sizes of initial grains.

Towards the onset of fragmentation, the scalar density of dislocations practically reaches its maximum, and, further, does not increase with increasing deformation [2]. At large deformations, the ΔA_s parameter characterizes not only the proper tensor density of dislocations, but also its conventional value that might have been reached in the absence of the re-arrangement of dislocations into boundaries. But, since such re-arrangement does occur, then ΔA_s characterizes not the curvature of crystal lattice caused by the distribution of individual dislocations in it, but the curvature produced by their complexes – dislocation walls.

In accordance with Polyani-Taylor principle, the ΔA_s parameter should be approximately equal to sample's curvature. If a sample is bent around a cylinder with a small radius, then the ΔA_s value could exceed $|\beta_{max}|$, but such bend would not lead to the growth of dislocation density, and, consequently, to further division of fragments into smaller portions. With $d_f = d_{f\,min}$ = const the relationship (13) can be fulfilled only due to the proportional growth of ΔA_s and misorientation angle θ_a. In turn, it means the increase in the density of stationary and mobile grain boundary dislocations resulting in grain boundary sliding. In other words, when lattice dislocations move from one fragment to another within the boundaries dividing them, the dislocations of orientation mismatch are formed; this statement support the results reported in [2]. Sessile components of these dislocations, with Burgers vector normal to boundary plane, result in the increase of angular misorientations of boundaries, while sliding components, with Burgers vector coplanar to a boundary, lead to the relative rotational displacements of fragments.

Peculiar features of shear during fragmentation

If the crystal lattice of grains is bent continuously, i.e. without formation of fragments, then a conflict occurs. The shear of crystal lattice γ_g with size d_g is equal to $\gamma_g = \rho b\, d_g = \beta d_g$. Alternatively, the shear of grain's crystal lattice can be expressed through its curvature k_l equal to the tensor density of dislocations:

$$\gamma_g = k_l d_g. \tag{14}$$

On the other hand, the curvature depends on the bending radius $k_l = 1/R_a$. In particular, an average radius of sample's bending at ECA pressing is equal to $R_a \sim (d_s/2)\tan(\phi/2)$, where d_s is the transverse size of a sample, while f - the angle of channels' intersection. Expressing the shear of grain's crystal lattice through the curvature of a sample during ECA pressing, one will obtain:

$$\gamma_g = 2\left(\frac{d_g}{d_s}\right)\cot\left(\frac{\phi}{2}\right). \tag{15}$$

This expression satisfies the Polyani-Taylor principle only if $d_g = d_s$ (one grain in the deformation center). In this case it degenerates into a known formula for determining the shear intensity at ECA pressing:

$$\gamma_g = 2\cot\left(\frac{\phi}{2}\right). \tag{16}$$

Since polycrystalline metals are usually deformed, the grain sizes here are significantly smaller than the sizes of deformation centers; so, the shear in grains determined by Eq. (15) will be considerably less than the macroscopic shear, which is not in conformity with the mentioned principle. The conflict is resolved not only by the fact that a crystal lattice of grains is fragmented into micro-regions at bending, but also by that the boundaries separating them become the zones of deformation localization. Acquiring extremely fine sizes and shape close to a rounded one, the fragments are not further divided into portions. The activity of dislocation slip in them

decreases, formation of dense pile-ups of lattice dislocation and their subsequent transformation into boundaries are not observed. That is why, according to Eq. (15), the shear in crystal lattices of fragments or grains becomes small. It is the inter-crystalline shear that becomes the main mechanism of deformation, this being provided by the increase in the area of boundary surfaces and their misorientations. In this case, the Polyani-Taylor principle is fulfilled in a peculiar manner, in the same way as at superplastic deformation (SPD') via cooperative grain boundary sliding (CGBS) in planes close to the direction of operation of maximum shearing stresses [15].

Thus, plastic deformation is inevitably accompanied by the change in the surface area of the deformation center. The absolute change of this area is the sum of the absolute values of increasing and decreasing surface portions as a result of discharge and generation of dislocations in it. The value of the ΔA_s parameter corresponds to sample's curvature and tensor density of dislocations in the deformation center, and, consequently, to the curvature of crystal lattice. At large deformations, the ΔA_s parameter characterizes the curvature of crystal lattice within the deformation center caused by angular misorientations of fragments' boundaries. Minimum sizes of fragments with the average values being estimated as (40-100)*b* are formed in metals if, due to deformation, the ΔA_s parameter in deformation center attains the value equal to maximum tensor density. As the ΔA_s parameter approaches the value of maximum tensor density, the deformation mechanism changes. The shear in metals with strongly fragmented structure, as well as at SPD', occurs rather along the boundaries of crystallites than in crystallites.

4. DEFORMATION CENTER AND GRAIN REFINEMENT IN MONOTONIC AND NON-MONOTONIC PROCESSES

Let us use the above established approximate equality of the ΔA_s parameter to sample's curvature and crystal lattice for approximate evaluation of average sizes of fragments and grains produced using various deformation modes. For this purpose it is necessary to establish: (i) the value of strain at which fragmentation is completed; (ii) the dependence between variation of the ΔA_s parameter and strain; (iii) average sizes of fragments and grains depending on ΔA_s.

Monotonic and quasi-monotonic deformations

The first task is convenient to be considered using the above deformation modes as examples. Unlike non-monotonic processes, the structure evolution in such deformations is not veiled by variations introduced by the change in the straining direction. Rolling and drawing processes, including multi-pass ones, are examples of quasi-monotonic* deformations [3,16]. The peculiar features of deformation center in these methods are that it is developed in the direction of material's extension and its appropriate size exceeds the transverse size of a billet. In tool' groove, a billet undergoes equal bending in opposite directions and acquires a relatively small curvature, the layers close to billet's axis moving in a more uniform and rapid manner than outer layers.

These peculiar features of the deformation center along with non-uniform distribution of values and strain rates along billet's cross section lead to the fact, that in many metals, in particular, with cubic lattice and not low SFE, starting with $\varepsilon \sim 0.3$, the formed fragments are combined into layers – banded structures of various modifications, differing in transverse sizes. The largest are the sizes of deformation bands filled with enlarged fragments (in the form of cells or blocks), while the smallest are the sizes of micro-bands. The formation of such banded structures in pure nickel after rolling is reported in [17], see in Figs. 2a and 2b. Micro-bands are generated in the zones of intense deformation localization, while low-angle fragments in them have minimum sizes. With increasing the number of passes and accumulation of deformation, micro-bands fill out the whole material volume. As a result, the uniform fragments divided mainly by low-angle boundaries are formed. The fraction of high-angle boundaries in such a structure is relatively small, and, such structure is essentially sub-granular.

In the recent work [8], we have derived the equation determining the change in the volume fraction V_f of the formed fragments with extremely small sizes in the deformation center with volume V. This equation has the form

*In the strict sense, multi-pass processes are non-monotonic because of non-holonomic relation between stress and deformation. But the deformation character is close to a monotonic process at rolling performed without widening and drawing with small strains ~ (0.2-0.3) at passes.

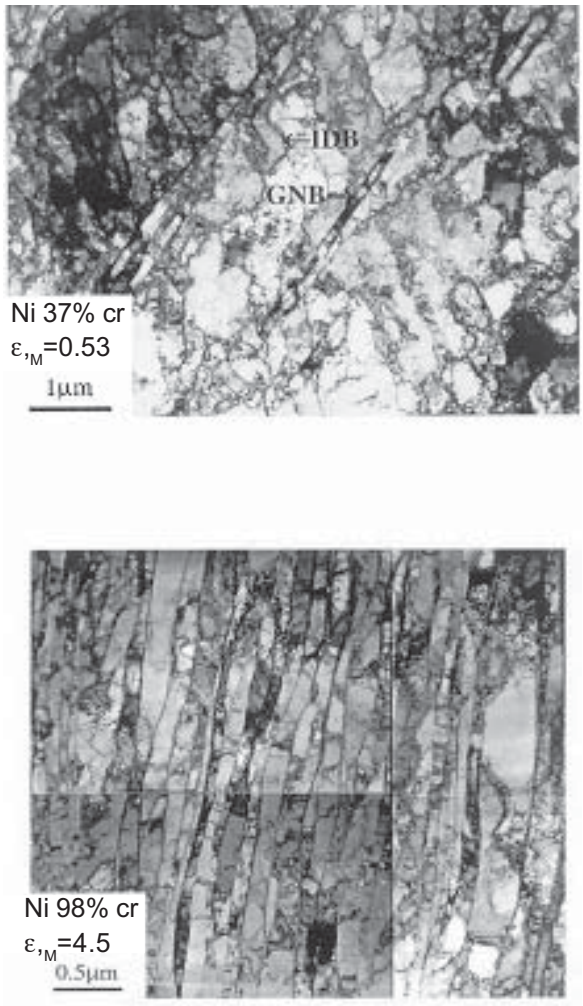

Fig. 2. Banded structures in the high purity nickel (99.99%) after cold rolling with strain 0.53 (left) and 4.5 (right), GNB – geometrically needed boundaries (boundaries between bands), IDB – random boundaries (boundaries of fragments within deformation bands) [8].

$$\frac{V_f}{V} = \left[1 - \exp\left(-\frac{3\varepsilon}{2}\right)\right]. \qquad (17)$$

The curve plotted from this equation is in good agreement with the experimental kinetic dependence reported in [18], which shows that the fragmentation in iron is practically completed after subjecting a wire to drawing with true strains ~ 3-4. As a result, fragments with size 0.1-0.3 mm are formed. The sizes of fragments located in micro-bands (see Fig. 2b) are also close to the mentioned values**.** In principle, the acquired estimate of strain value sufficient for the completion of fragmentation can be generalized for the case of producing sub-granular – granular structure. Such structure was produced in many materials, in particular, in Al and its alloys [19]. Mainly granular structure was formed when, during ECA pressing up to 4 passes ($\Sigma\varepsilon \approx 4$), the orientation of samples, made from these materials, was changed between passes; in doing so, a non-monotonic process was realized. On the contrary, mainly low-angle fragments – sub-granular structure were formed after pressing samples without changing orientation, i.e. after less non-monotonic deformation.

The second task will be also considered using monotonic deformation; uniform extension of a cylindrical billet can be considered as an example of this process*. In this case, the problem solution is facilitated because the area of the deformation center coincides with the area of sample's outer surface. The change in the surface area depending on true strain e for the indicated deformation is easily derived from the expression reported by the present authors in [8]:

$$\Delta A^e = (A_i)\left[\frac{me^{\frac{\varepsilon}{2}} + e^{-\varepsilon}}{m+1} - 1\right], \qquad (18)$$

where e – base of natural logarithm, ε – true strain, m – coefficient equal to L/R, L and R – length and radius of a starting cylindrical sample, respectively, A_i – initial area of the deformation center of a cylindrical sample, equal to $2\pi R (L+R)$.

Dividing both parts of Eq. (18) by V – the deformed volume with account that $A_i/V = (2/R)[(1+m)/m]$, and taking into account that at large deformations m>>1, one obtains:

$$\Delta A_s^e \approx \left(\frac{2}{R}\right)\left[e^{\frac{\varepsilon}{2}} - 1\right]. \qquad (19)$$

At drawing and rolling, large deformations are usually accumulated in thin, including flat, samples. That is why R can be set equal to half-thickness of a flat sample – $h/2$**.** Hence, integrating the expression $\Delta A_s^e \approx (4/h)\,[e^{\varepsilon/2} - 1]$ within the range from 0 to e and accounting for m>>1, one obtains:

*Though this ideal process differs from rolling and drawing by lesser local curvature of a billet in the deformation center, nevertheless, this process, as well as industrial processes, leads to uniform extension of a billet.

$$\Delta A_s = \int \Delta A_s^e d\varepsilon \approx \left(\frac{4}{h}\right)\left(2e^{\frac{\varepsilon}{2}} - \varepsilon - 2\right). \quad (20)$$

The third task. Eq. (20) was determined for the special case – uniform extension of a cylindrical sample (monotonic deformation). Nevertheless, it is also true for the case of uniaxial non-monotonic deformation. Thus, two-step uniform deformation of a cylindrical sample, that includes extension followed by compression of a sample for the same strain value, will lead to a zero resultant change of its dimensions and outer surface area. Meanwhile, the accumulated deformation and the absolute value of change in the area of outer surface will be equal to the doubled amount of the corresponding values at any of the considered stages. Moreover, Eq. (20) can be used for the general case of plastic deformation. Indeed, at any plastic deformation there occurs the change in the area of deformation center. Since divβ=0, see above considerations, no dislocation transformations, including duplication, reaction, and annihilation of dislocations, within the volume of deformation center, will change the β_{ij} value (dislocations come into as loops, for which Σb=0, and for any reaction of dislocations Σb=const) [11]. That is why there exists one-to-one correspondence between strain and change in the area of deformation center surface. To a first approximation, it has the form of the obtained relationship (20). Using this relationship and the earlier derived dependence (12), one can determine a tentative average size of fragments and/or grains for any methods of large deformations, accomplished at temperatures below recrystallization temperature.

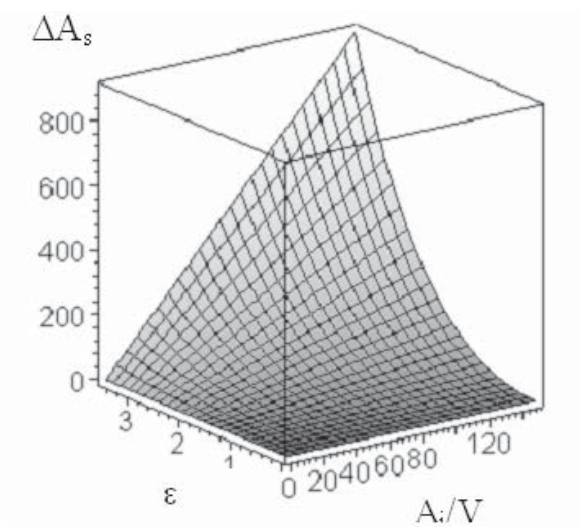

Fig. 3. The dependence of changing of the specific surface of deformation center – (parameter) ΔA_s (in mm^{-1}) with the initial specific surface A_i/V (in mm^{-1}) and true strain ε.

The dependence (20) in Fig. 3 shows that ΔA_s value depends to a larger extent on A_i/V, rather than on ε. Thus, at ε=3 (according to previous estimations and experimental data, such deformation practically results in completion of fragmentation) for samples with the starting $h/2$ = (1–10) mm, ΔA_s parameter will be ≈ (10-1) mm^{-1}. Then, one will have $d_g \approx$ (1-10) μm. For samples with $h/2$ = (0.1-0.05) mm the ΔA_s parameter ≈ (100-1000) mm^{-1}, hence $d_g \approx$ (0.1-0.01) μm. These estimations, at least, are in qualitative agreement with the experimental data. Thus, after cold multi-pass rolling of metallic foils, their thickness decreases from ~ 0.1 mm to ~ 0.01 mm and even less [20]. In the order of magnitude, the corresponding values of A_i/V will be ~ $1/h \approx 10$ mm^{-1}, and ~ 100 mm^{-1}, while $d_g \geq 0.1$ μm.

Non-monotonic deformations – ECA pressing

At ECA pressing, the geometry of a deformation center might have the shape of a fan or a equiconvex lens [21]. For the case of a lens-like center, its size along the lines of sample's bending vary from the values close to zero (at the vertex of the angle of channels' intersection) to ~$0.1d_s$ along the neutral line, while for the fan-like center – from 0 to $d_s\phi$. For the lens-like center $A_i/V \approx 2\pi R^2/(\pi R^2 \times 0.1 d_s) = 20/d_s$, while for the fan-like center $A_i/V \approx 2/d_s$. The diameter of typical rods subjected to ECA pressing is d_s ~ 20 mm. The corresponding values of A_i/V, for the cases of fan-like and lens-like shapes of deformation centers, are ~1 mm^{-1} and 0.1 mm^{-1}. Though these values of A_i/V differ by an order of magnitude, it follows from Fig. 3 that this does not lead to a significant increase in the ΔA_s value. That is why, pressing of rods with backpressure (lens-like center) and without it (fan-like center) does not result in a considerable (differing by an order of magnitude) difference in the sizes of the produced grains, even if very large deformations are used. This conclusion is confirmed experimentally. Thus, after subjecting pure Cu ($\Sigma\varepsilon \approx 2$) to pressing for two passes with backpressure and without it, the values of d_g were 0.4 μm and 0.6 μm, respectively; these grain sizes decreased towards 0.19 μm and 0.25 μm after 16 passes ($\Sigma\varepsilon \approx 16$) [22].

The maximum value of curvature is achieved in sample's surface layers adjacent to the vertex of the angle of channels' intersection. Due to techno-

*Formally, for an evolvent point that belongs also to a circumference, the curvature is equal to infinity.

logical reasons, this angle cannot be made sharp. Instead of an angle, a blending with radius ~ 0.1 mm is formed. Hence, the local curvature of a sample in this zone will be k ~ 10 mm^{-1}. Maximum curvature of an evolvent* will be determined as $k_e = 1/R\phi$, where ϕ, in terms of the order of magnitude, will be assumed to be equal to the angle of misorientation of low angle boundary, i.e. ϕ ~ 0.1 rad. As the result, one will have k ~ 100 mm^{-1}. Accordingly, the transverse sizes of fragments in the examined sample's layers will be ~ 0.1 μm. At farther and farther distance from the specified surface, k and k_e values decrease in proportion to sample's thickness, and the sizes of the refined fragments accordingly increase. For this reason, the efficiency of grain refinement in sample's layers far from the vertex of the angle of intersection will be much lower. As the finest fragments are formed in a strongly curved zone of the deformation center, in order to achieve more uniform refinement of grains during ECA pressing, it is vital not only to change the route of pressing through sample's reorientation, but also to use design and technological factors providing the enhancement and equalization of sample's curvature in the deformation center, by, for example, narrowing the channel from the side of external angle. To increase the fraction of high-angle boundaries, it is important to change the planes of sample's bending, i.e. to have it strained according to a spatial trajectory with larger curvature [3].

Non-monotonic deformations – torsion under pressure

Torsion under pressure is characterized by two stages. The first stage is torsion accompanied by upsetting. As shown in [3], even at turning of a thin (0.1 mm) disc for an quarter of turn, the deformation on the diameter of 5 mm in this case reaches the value of ~ 7, the average value for a sample being ε~ 3-4. So, it is of little wonder that grain refinement was observed at this stage at relatively small turning angle [1]. The second stage is restricted torsion that occurs without change of sizes and resulting (visible) variation in the surface area of deformation center, that being evidently due to the equality of dislocations coming into the surface and leaving it. Correlating the swirl angle Θ of disc's upper base relative to its lower base with the turning angle of the generator g at the lateral surface, one will obtain $\gamma = \mathrm{Arctg}(\Theta R/h)$. It follows from this equation that the angular deformation γ, close to the limit $\pi/2$ for discs with a large relation R/h is achieved even after turning for a rather small angle. For example, for a disc with R = 10 mm and thickness h = 0.2 mm at Q ~ 1.57 rads (quarter of turn), the value $\gamma \approx 1.44$ rads. Accordingly, $\Sigma\varepsilon$ also increases with radius. At the same time, the curvature $k_c = R/[R^2 + (h/2\pi n)^2]$ and torsion of disc's screw line $k_t = h/2\pi n / [R^2 + (h/2\pi n)^2]$ (where n is the number of rotations) are likely to increase at $R \to 0$. Since the distortion of crystal lattice is determined by the sum of deformation and rotation tensors, this allows clarification of the experimentally observed fact [23] of producing more or less uniform UFG structure in nickel along the disc radius*. A significant factor, that promotes the formation of dispersed fragments during torsion of thin discs, is the high value of the parameter $A_i/V = 1/h$ ~ 10-100 mm^{-1}. In combination with high degrees of accumulated strain and better, as compared with ECA pressing, conditions for abstracting heat to the tooling for the case of torsion under pressure, this leads to the formation of fine grains of about 0.1 μm in diameter and less. Let us note that the involute of a screw line is the screw line that contributes to the growth of the portion of the produced granular fraction.

The presented interpretation of the influence of curvature on the UFG structure formation during deformation is in conformity with the comparative data on the degree of grain refinement achieved under various modes of severe plastic deformation, reported by the present authors in [10]. These data show that the larger value of the A_i/V parameter of a sample, the finer grain size is produced in it at cold deformation. Thus, the smallest powders (lamellas) have $A_i/V \sim \beta_{max}$, hence grains, d_g ~ 0.01 μm, are produced in them.

5. THE EFFECT OF DEFORMED STATE ON STRUCTURE REFINEMENT

It is known that the deformed state (DS) of a sample depends on the geometry of the deformation center and affects the structural state (SS) acquired by a metal. As stated above, the deformation of about 3-4 is sufficient to refine structure to minimum sizes of low-angle boundaries or grains, peculiar to a method applied (rolling, drawing, ECA pressing) and a given material. But usually, for producing uniform UFG structure, a sample is subjected to large deformations. Thus, at ECA pressing, the number of passes and accumulated deformation are to be brought up to the values of 8-12 and more. A discrepancy between the calculated and experimental values is due to the non-uniform distribution of de-

* Uniformity is also provided by the changed position of sample's axis of symmetry that occurs during torsion.

formation in a sample, and to the influence of rotational deformation mode on structure formation. Hence, it is necessary to note that the notions of realization of simple and/or pure shear at ECA pressing leading to the formation of uniform structural state in metals, reported in [21,24] do not agree with a number of necessary conditions.

The uniformity of DS and, as a consequence, of SS follows from the Polyani-Taylor principle. As mentioned above, in accordance with this principle, the grains in a polycrystal are deformed (change their shape) in the same matter as a sample. To provide the uniformity of DS and SS, the grains are necessarily to change not only their shape (become elongated, compressed, and bent), but also are to be turned (rotated) in the same manner. In other words, to produce uniform UFG structure, it is essential that the equality of distortions is fulfilled at macro- and micro- scale levels, both in deformation and rotation tensors, and also their velocities. As known, the first condition requires the presence of 5 active dislocations slip systems in grains. Usually, not more than 2-3 slip systems are activated. The second condition, as reported in [2], is associated with the realization of reactive turns of grains opposite to Taylor rotation. This is hindered by the tightness of grains – presence of neighbors. As a result, there appear discrepancies that are accumulated in deformations of grains during deformation; in turn, they lead to fragmentation.

Usually, DS is estimated only by the first tensor of distortion, i.e. deformation tensor, because its components are responsible for changes in a material shape. The value of accumulated (true) strain is, finally, determined according to Odqvist, as:

$$\varepsilon = \int_0^t \xi_e dt = \int_0^t \sqrt{2\xi_{ij}\xi_{ij} / 3}\ dt, \qquad (21),$$

where ξ_e and ξ_{ij} are, respectively, the intensity and components of strain rate deviator, t – time of deformation.

It is generally supposed that if deformation, accumulated in different material points, is the same, then, its DS is also uniform. This is more or less valid for monotonic, or close to monotonic, deformations. But, at large non-monotonic deformations, structure formation in metals depends also on the second tensor (on rotation velocity, in particular) that usually is not taken into account.

Let us use the processes of rolling (quasi-monotonic deformation) and ECA pressing (non-monotonic deformation) as examples to consider to what extent the latter condition is realized in various methods, and how it affects the structure refinement.

Rolling

At rolling of a flat sample without widening, the change in its shape can be considered as the result of summation (superposition) of the two practically equal and oppositely directed bends, occurring under the simultaneous action of the upper and lower rollers. Therefore, the resultant bend of a sample in the deformation center is small. Consequently, the cumulative moment, produced by external stresses, striving to initiate the reciprocal rotation of the fragments formed in grains is small. Rotation of fragments occurs under the action of relatively small internal (moment) stresses [2], therefore, they are small in value. It has been pointed out above (see text and Fig. 2) that the fragments are being joined into a family of bands, oriented in one direction. The generation of bands is due to the compatibility of fragments' deformation, located in one band, and the difference in the flow rates in the direction of drawing between fragments, located in different bands. Sliding along the boundaries, that separate the bands, compensates the difference in the flow rates of bands that is responsible for their generation. But, at rolling of thin samples, this sliding is small because of a relatively minor difference in the rates of bands' movement. Therefore, the angular misorientations of fragments' boundaries increase weakly. The deformation accumulated in bands, determined according to Eq. (21), is approximately the same, and this results in the formation, on the whole, of the uniform structure consisting predominantly of low-angle fragments. Only some boundaries of such fragments, namely, boundaries between micro-bands, have larger angular misorientations as compared with their transverse boundaries [17].

ECA pressing

At ECA pressing, a material acquires different deformed and structural states. The results of computer computations of DS in sample's material after one pass of ECA pressing show that the values of the accumulated deformation in the material points, migrating in the deformation center along the arches of different radii, are practically the same*. This is due to the fact that integration of Eq. (21) is fulfilled

* The exception is a relatively narrow layer of material, migrating from the side of external angle, in which the accumulated deformation is approximately half less than in the other portion of a sample, that is caused by a strong retardation of material flow by channel walls.

by the paths – arches with different length values. But structure investigations of the deformed samples have revealed considerable differences: at small radii of rotation, the refined fragments are significantly larger in number and their sizes are finer that at large radii.

It is not difficult to show that this has been due to the non-uniform field of angular velocities in the deformation center. Actually, in the cylindrical coordinates of R and ϕ with the center in the vertex of an angle of channels' intersection, the velocity of material points in the center may be expressed in the form of two components: radial υ_r and peripheral (tangential) υ_p. At inner boundaries of the deformation center $\upsilon_r = 0$, while υ_p being equal to the punch speed. Due to the symmetry of deformation center (friction is not taken into account here), maximum value of υ_r corresponds to the angle $\phi_{max}/2$. Such character of υ_r changing results in relative displacements between the formed fragments and in the formation of deformation boundaries (oriented by the radius) that divide the fragments into bands, similar to those that have been considered above for the case of rolling.

Regarding the other component of velocity υ_p, it should be pointed out that, when punch velocity is uniform, there are no reasons to surmise that its value in the deformation center (but not the direction) changes depending on coordinates. The condition of entirety conservation requires that the material points should move along each arch with the same velocity; that is why, it can be assumed that υ_p = const. Hence, the change of the angular velocity grad$\Omega \approx d\omega/dR$:

$$\frac{d\omega}{dR} = -\frac{\upsilon_p}{R^2}. \qquad (22)$$

It follows from this equation that for two neighboring material points with their vicinities, located at a small distance from the center of coordinates (rotation), at, for example, R_1~0.1 mm and R_2=0.2 mm, the difference between ω_1 and ω_2 will be large. While for the two points located at the same distance from each other, but far from this center, in, for example, area R~10 mm, the difference will be small. If the material points with their vicinities are associated with the refined fragments, integrated into banded structures, then it becomes obvious that the transverse sizes of bands and the sizes of fragments, confined in these bands, should differ significantly according to radius. The difference occurs because each band combines fragments rotating in the deformation center as a unit, i.e. with the same angular velocity. The closer the formed fragments are to the rotation center, the larger is the gradient of angular velocities between the neighboring (by radius) fragments, the lesser transverse fragments are acquired by bands and fragments confined in bands. Due to the significantly differing angular velocities of rotation, the angular misorientations of boundaries q should be changed in a different manner, dividing fragments in the neighboring bands. As, according to [2,8] $\theta \propto \Delta\omega$, and since $\Delta\omega \propto 1/R^2$, then it is evident, that the fragments formed in the zone with a minimum radius will acquire large angular misorientations.

Thus, at ECA pressing, the fragments form two families of intersecting banded structures: radially and tangentially directed, in conformity with the components of field velocities and rotation of metal in different areas of the deformation center. The bands, located at a small distance from the rotation center, will acquire the minimum transverse sizes. In the same zone the fragments acquire lesser sizes and large θ.

6. MAIN STAGES OF STRUCTURE REFINEMENT AT LARGE MONOTONIC AND NON-MONOTONIC DEFORMATIONS

The banded structures can be considered as deformation channels emerging at a meso-scale level (within one or several initial grains) that provide externally induced directional displacement of crystal lattice defects in a stress field. At large quasi-monotonic deformations such channels retain more or less stable positioning that corresponds to the direction of material drawing under the action of external forces. At the same time, there also occurs the intersection of individual bands connected with local changes in internal stresses due to material strengthening. But in this case, the number of intersections of banded structures is considerably smaller than at non-monotonic straining, in which the intersection results from the changed direction of external forces' operation. Since longitudinal boundaries of fragments that coincide with the bands' boundaries have larger angular misorientations than their transverse boundaries, one can assume that the resultant formation of fragments with the increased values of misorientation angles, i.e. grains at a meso-scale at non-monotonic deformation, can be attributable to the increased quantity of intersections of micro-bands. But such a presentation

of the process of refined grains formation requires a more detailed consideration of structure evolution at a micro-scale level: the interaction of crystal lattice defects, in particular, dislocations and fragment boundaries.

In the well-known model presentation of structural changes, occurring in metals at micro-scale level during severe deformation [1], three specific stages are distinguished. (1) First, a cellular structure is formed, in which cellular walls present the intricately interlaced pileups of dislocations. (2) Then, the thickness of cellular walls starts to decrease. In such a case some dislocations annihilate, but, on the whole, the dislocation density increases. In cells' walls there remain excessive dislocations of two types: with Burgers vector perpendicular and lying (sliding) in the plane of boundaries. (3) With further straining, the increased density of dislocations with Burgers vector, perpendicular to a boundary, leads to their increased misorientations and transformation of cells into grains. At the same time, the increased number of sliding dislocations leads to the increase of the long-range fields of elastic stresses.

Thus formed grain boundaries are strongly non-equilibrium. There exist areas of considerable distortion of crystal lattice about these grain boundaries caused by the fields of elastic long-range stresses, introduced grain boundary dislocations, some of which (sliding grain boundary dislocations), when moving, result in grain boundary sliding and relative displacement of grains.

Judging from the results of numerous studies, including the data reported by the authors of the mentioned model [1], the presented structure evolution should be supplemented by the stage of formation of sub-grains. The cellular boundaries, isolated by the intricately interlaced pileups of dislocations, are not surfaces. The decrease in their wall thickness due to the introduction of new lattice dislocations, the annihilation of dislocations of opposite signs and rearrangement of single-valued and oppositely directed dislocations, might result in the degeneration of walls into a surface – low angle boundary, and, consequently, cells into a sub-grain. The probability of such structure evolution for materials possessing higher stacking fault energy is especially high. That is why, generally, the sequence of stages of structure evolution at fragmentation consists of the following stages: dislocation pileups (DP) $\rightarrow$ cells (C) $\rightarrow$ sub-grains (SG) $\rightarrow$ grains (G).

The above-stated model is in conformity with the experiments. The grains with greatly differing sizes are formed during deformation of such metals as Al, Cu, and Ni [1,25,26]. The presence of lattice dislocations is not observed in grains with minimum sizes, while separate chaotically located dislocations or fragmented cellular or sub-grain structure are observed in grains of larger sizes. These facts indicate not only the presence of the stage of sub-grain formation, but show also that the formation of fine grains and sub-grains is a multi-cycle process. With increasing deformation, the evolution of defects according to the scheme DP $\rightarrow$ C $\rightarrow$ SG $\rightarrow$ G is repeated resulting in the division of coarser grains into finer fragments. But this continues up to some size that is typical for a given material and conditions of its processing, on the attainment of which, the intersection of dislocations becomes unlikely. Further, the evolution develops on the way of increasing misorientations θ between low angle fragments.

The condition necessary for the increase of θ can be expressed as:

$$N_{in} b_{in} >> N_{out} b_{out}, \tag{23}$$

where N_{in} and b_{in} are the number and Burgers vector of lattice dislocations (LD) getting into a boundary, respectively, and N_{out} and b_{out} are the same variables for the dislocations leaving a boundary. A power W necessary for LD toenter a boundary should be

$$W = (f\,\nu)\cos\alpha \geq \Sigma W_t + \Delta\gamma_b, \tag{24}$$

where α – the angle between force vectors f acting on the dislocation and its velocity v; ΣW_t – the total energy due to the forces impeding LD's getting into a boundary; $\Delta\gamma_b$ – increments of boundary energy.

The surface area of a material A_{outer} increases at large monotonic and quasi-monotonic deformations. This becomes possible due to a favorable texture that provides coming out of a large flow of dislocations to the surface. Therefore $\cos\alpha \approx 1$ for W sufficient for dislocations to overcome the fragment boundaries. Only a portion of mobile dislocations is spent on the θ increment between fragments. As a result, a low-angle fragmented (sub-grain) structure is predominantly formed.

There is no noticeable change of billet's surface area In the considered modes of non-monotonic deformation. Larger portion of dislocations is spent on the formation of grain boundaries, and not on the increment of outer surface.

This is provided by the change in the direction of the deforming force operation at non-monotonic deformation, due to which the value of Schmid factor becomes low, and so the kinetic energy becomes insufficient for overcoming the fragment boundaries.

In such a situation the lattice dislocations trapped by fragment boundaries will lead to the increase of their angular misorientations, i.e. grains will be formed.

7. THE EFFECT OF DEFORMATION WARMING UP

It is well known that the larger portion of deformation energy transforms into heat. At severe deformation, in the zone of its localization this might result in a considerable short-term heating of a metal. In [18,19], it has been shown by high rate infrared microscopy that the temperature in thin slip bands increases by hundreds degrees, up to the melting temperature, at angular deformation $\gamma \geq 1$. According to the estimation reported in [20], the accelerated diffusion migration of atoms might come to 10 nm for the time of cooling of such warmed-up band. Since at ECA pressing the deformation center is relatively narrow and mostly limited by the inner surface, the cooling rate will depend on thermal conductivity of a metal. If the deformation localizes along the fragment boundaries, then, according to the estimations reported in [21], the deformation warming-up, even in Cu, and, especially in Ti, is capable to produce a noticeable migration of boundaries and growth of grains. Obviously, it is one of the reasons preventing the formation of nanocrystalline structure in samples during ECA pressing.

As indicated earlier, at torsion under pressure of thin discs, the conditions for heat diffusion are more favorable due to the developed surface of contact between material and tooling; the high pressure evidently provides forcing out of adsorbing gases from the contact zone.

8. DISCUSSION

Practically, any mode of plastic deformation, to some extent or other, leads to bending and/or torsion of a metal sample and, consequently, its crystal lattice. At a meso-scale level, bending and torsion are characterized by a curvature determined by the tensor density of dislocations. The curvature of crystal lattice increases proportionally to the increasing scalar density of dislocations and Burgers vector. At small deformations, the curvature localizes within more or less uniformly distributed dislocation cores, while at large deformations – within the boundaries of the formed fragments, and grows in proportion to their quantity and misorientations.

At plastic deformation, a directed streamflow of defects occurs in the stress field leading to the migration of atoms; this being inevitably accompanied by the change of the surface area of the deformation center due to the discharge and generation of dislocations in it. As shown in this paper, the absolute change in the deformation center area normalized to the corresponding volume, i.e. ΔA_s parameter, corresponds to sample's average curvature and tensor density of dislocations, and, therefore, to the crystal lattice curvature. At large deformations, the ΔA_s parameter characterizes the curvature of the turn of fragmented lattice produced by angular misorientations of fragment boundaries.

Minimum sizes of fragments, estimated in this study by the size of about (40-100)*b*, are formed in metals, if ΔA_s parameter resulting from the deformation will reach the value equal to the maximum tensor density of dislocations.

The refinement of fragments and their transformation into fine grains depends not only on the value of accumulated strain, but also on relative turns of fragments. The kinetic equation of fragmentation shows that, in principle, the accumulated strain of ~ 3-4 is sufficient for fragmentation completion, if the whole material volume acquires more or less uniform distortion, i.e. the equality of the deformation tensor and, also, of the rotation tensor. The uniform granular structure is formed at large values of accumulated strains because of the non-uniformity of distortion in metals under such methods of severe plastic deformation as ECA pressing and torsion under pressure. With increasing the number of pressing passes and turns at torsion, the accumulation of the rotational component of distortion occurs, that is important for forming a uniform refined granular structure.

Besides, the change in the direction of the main vectors of deformation and rotations in space, i.e. non-monotony of deformation, is important to form a uniform refined structure of granular type, that provides the uniform satiation of the boundaries of the formed fragments by stationary grain boundary dislocations. The role of the deformation center surface in refinement seems to be rather important too, since it acts as a dislocation source, capable of their efficient generation at fine values of fragments and grains. This is in conformity with the observed correlation between the sizes of refined fragments and grains, and the ratio of the area of deformation center surface to its volume.

The presented relationships, connecting the parameters of fragments, curvature and strain, do not limit the possibility of increasing the density of crystal structure (curvature) to a critical value, at which

the crystalline state of metals become unstable. In such case, the failure of metal will occur or it will become amorphous. The first is retarded by a high hydrostatic pressure, while the second – by the sliding along the boundaries of grains and fragments that increases at structure refinement.

At the same time, the amorphous state in metals is formed by its rolling to produce a very thin cross-section. One can assume that at some thickness values of samples, grain boundary sliding and rotations become non-efficient mechanisms for relaxation of inner stresses that increase with straining. The similar situation occurs during deformation of thin samples under superplasticity conditions: the formation of bands of cooperative grain boundary sliding in a fine grain material becomes difficult if the number of grains in a sample's cross section approaches to 10 in order of magnitude. In this case, the totality of grains, capable to form such bands, sharply decreases, and, therefore, such samples do not manifest high plastic properties [15]. It is likely that the mentioned scale factor is also of importance at large cold deformations of thin samples, the surfaces of the deformation center of which are rather developed and capable of being the active sources of defects resulting in the amorphous state of a metal.

9. CONCLUSION

1. At large plastic deformations, the sizes of refined fragments and samples depend on the distortion acquired by a metal in the deformation center that is determined generally by the summation of the deformation and rotation tensors. Total (accumulated) strains equal to, at least, 3-4 are necessary to refine a metal structure. If the gradients of angular velocities of rotation of the formed fragments in the deformation center are small, the structure of sub-granular type is formed, while at large gradients of angular velocities, induced by outer forces the structure is granular type.
2. The curvature acquired due to bending and torsion is the complex parameter characterizing the deformed (distortion) and structural states of a metal. At the micro- and meso- scale levels, the curvature-torsion are determined by the tensor density of dislocations, and its average value corresponds to the macro-parameter ΔA_s, i.e. the absolute change in the area of the deformation center normalized to its volume. The misorientation angles between the formed fragments are in direct proportion, while their sizes are in inverse proportion to the curvature. The accumulated curvature-torsion should reach the values corresponding to the maximum tensor density of dislocations to form the extremely refined fragments and grains, estimated in this paper by the value (40-100)*b*. The dependence between the macro-parameter of curvature ΔA_s and the value of accumulated strain is established.
3. The change in the surface area of the deformation center inevitably occurs at plastic deformation. Under conditions of cold deformation, this surface plays an important role in structural changes of a metal, being an efficient source and discharge of crystal lattice defects. Large values of the specific area of the deformation center surface contribute to the formation of finer fragments and grains.
4. At quasi-monotonic deformations, such as, for example, drawing or rolling, the bends of samples in the deformation center are inverse in direction and equal in value; a sample and grains acquire a minor resulting curvature in passes. Correspondingly, minor turns are acquired by crystal lattices of fragments. Significant increase in curvature and corresponding decrease in fragment sizes at such deformations are achieved due to large values of the initial specific surface of the deformation center, i.e. at reductions of samples with small transverse sizes.
5. At ECA pressing, the specific surface and curvature depend on the angle of channels' intersection and, significantly, on sample's transverse sizes. That is why, the formed fragments have rather different sizes ($\geq 0.1 – 10\ \mu m$) and angular misorientations at one-pass pressing of relatively small samples (10-20 mm in diameter). It is important to use the constructively technological factors contributing to the increase of curvature and its uniform accumulation in the cross-section of a sample to produce uniform ultrafine grain structure.
6. At torsion under pressure of thin (~ 0.1 mm) metal discs, the production of more or less UFG structure with grain sizes $\leq 0.1\mu m$ is provided by a large area of specific surface, curvature-torsion of a sample in the central zone of the deformation center, and also spatial change in the position of curvature centers during straining.

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