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Crystons: basic ideas and applications

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A brief review of the cryston concept and its applications is given. By definition, crystons are shear carriers arising in the contact interaction of dislocations with intersecting slip planes. The notion of standard orientations of habits (SOH) is introduced for shear bands. On the example of fcc crystals with SOH of the {hhℓ}-type, the conditions for generation and propagation of crystons, as well as for a discrete change of habit orientations, are considered. The reasons for deviations from SOH are indicated. With regard to the formation of martensite crystals, crystons can act as dislocation nucleation centers, carriers of threshold deformation and relaxation shear. The areas of promising investigations are mentioned.

Keywords: crystons, shear bands, standard orientations of habits, martensitic transformations

1. Introduction

At the present time, the need to consider plastic deformation as a hierarchical, multi-level process, where the meso-level plays a significant role (see, for example, [1—4]), is clearly realized. The universal character of the phenomenon of deformation localization does not cause any doubt either. In our opinion, the issue concerning the spectrum of deformation carriers at the meso-level has not been worked through with sufficient completeness. For instance, one has started to identify a rather general notion of superdislocation, applicable for describing a shear along arbitrary crystallographic planes, with dislocation pile-ups on typical slip planes.

It seems natural to introduce carriers of displacement vectors **S** during shear deformation, larger (an arbitrary number of times) in value than the lattice parameter, and having in the general case an arbitrary, but crystallographic, orientation. For the sake of brevity, such carriers have been termed [5] «crystons», in order to emphasize their closest connection with the crystalline medium.

In order to avoid ambiguity, we should note that talking about a cryston, we mean a shear carrier of the superdislocation type in which the displacement field is distributed in the carrier volume, and only in the limiting (degenerate) case, when the defects glide on one and the same plane, it comes down to a dislocation pile-up. It should also be stressed that we are referring to quite specific shear carriers that appear during interaction of dislocation groups belonging to slip systems with intersecting planes.

The aim of this paper is to present a brief review of the results achieved in the framework of the concept of crystons, focusing the attention on the ideology and illustration of the interpretations of complex processes, which will allow researchers (especially beginners) to perform quickly a preliminary analysis on the basis of the morphological attributes observed during formation of shear bands and deformation-induced martensite.

2. The basic postulates of the cryston model of formation of shear bands with boundaries of the {hhℓ} type (using single crystals with an *fcc* lattice as an example)

The experiments on deformation of single crystals [6—12] demonstrate that formation of shear bands with a boundary orientation different from easy slip planes, takes place in the regions where at least two slip systems with intersecting planes are active. Therefore it is natural, when describing a shear carrier, to determine the effective magnitude of its Burgers vector **b** as a characteristic obtained by a certain superposition of the Burgers vectors $\mathbf{b}_1$ and $\mathbf{b}_2$ of interacting dislocations. Then the vivid fact of the discreteness of the observed orientation spectrum of the planar boundaries of shear bands during the developing plastic deformation is naturally accounted for by a change in the fractional contributions of interacting dislocations. By analogy with the terminology accepted for description of martensite morphology, shear band boundaries are hereinafter referred to as habits. In addition, by definition, we shall introduce standard orientations of habits (SOH) connected with the interaction of pairs of easy slip systems with intersecting planes.

Let us consider the SOH (hhℓ) in crystals with an fcc lattice. This is the simplest case, as in fcc crystals there is only one type of easy slip planes {111}. Let us select, to be specific, the system with the slip plane (111) as the basic one, and the system with the slip plane $(11\bar{1})$ as the conjugated one. Since the planes intersect along the direction $\mathbf{\Lambda} \| [1\bar{1}0]$, the contact interaction of dislocations forms, in the vicinity of the segments $[1\bar{1}0]$, barriers characterized by the vector **b** which is composed by the Burgers vectors $\mathbf{b}_1$ and $\mathbf{b}_2$ of the dislocations belonging to the basic and conjugate slip systems. The cylindrical region of localization of the pile-up dislocation lines is characterized by a diameter of about 0.1 μm and a generating line length prescribed by the length of the barrier segment $[1\bar{1}0]$. Such a barrier, containing n vectors $\mathbf{b}_1$ and m vectors $\mathbf{b}_2$, is a meso-concentrator of stresses and is characterized by the sum Burgers vector **b**:

$$\mathbf{b} \| \, n\,\mathbf{b}_1 + m\,\mathbf{b}_2. \qquad (1)$$

Natural appears the scheme of cryston generation by a generalized Frank-Read source (GFRS), in which the function of a pinned segment of an individual dislocation is now performed by a dislocation bundle (see Fig.1), and the result of the generation is a cryston (superdislocation) loop which can be considered as a set of closed loops localized in the region bounded by a surface that is topologically similar to a torus with a cross-sectional diameter ≤0.1 μm.

Let us make a non-essential simplification assuming that the vectors $\mathbf{b}_1$ and $\mathbf{b}_2$ are equal in value and have a purely edge orientation with respect to the operating segment of the GFRS, *i.e.*

$$\mathbf{b}_1 \| [11\bar{2}]. \qquad \mathbf{b}_2 \| [112] \qquad (2)$$

Let us demand that the superpositional Burgers vector (1) should lie in the plane (hhℓ), which is equivalent to the orthogonality **b** to the direction of the normal $\mathbf{N} \| [hh\ell]$, from Eq.(1),(3) follows the relationship

$$h/\ell=(n-m)/(n+m), \qquad (4)$$

connecting the relation of the indexes h/ℓ with the numbers n and m, prescribing the contributions of the interacting dislocation systems. In particular, at m=0, we obtain h/ℓ=1, *i.e.* the slip is effected on the octahedral plane (111). At n=m, we obtain the orientation (001) corresponding to a cubic slip.

It is evident that an addition to the vector b of an arbitrary-value vector $\mathbf{b}_{||}$ collinear to $[1\bar{1}0]$, does not affect the fulfillment of the condition (3), since the vector $\mathbf{b}_{||}$ is orthogonal to any

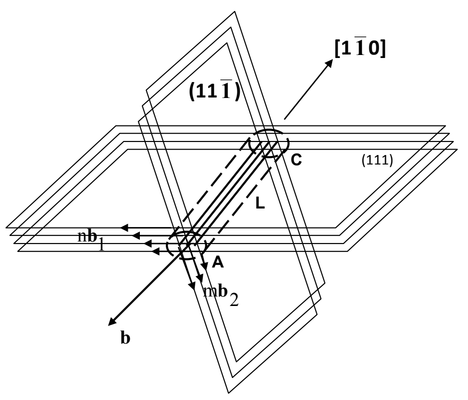

Fig. 1. Formation of an operating segment of the generalized Frank-Read source [13-15]

direction of [hhℓ]. This means that in defining the vector **b** lying in the plane (hhℓ), there is an additive ambiguity:

$$\mathbf{b} \rightarrow \mathbf{b}' = \mathbf{b} + \mathbf{b}_{||} \qquad (5)$$

Let us mention briefly the consequences that derive from the proposed cryston model.

1. Let us remind that in the case of a shear on the plane with the normal **N** in the direction **b**, a material turn around the axis $\boldsymbol{l}$ is effected:

$$\boldsymbol{l} \,\|\, [\mathbf{b},\mathbf{N}], \qquad (6)$$

where the symbol [,] denotes the operation of vector multiplication. It is clear that in the case of a purely edge orientation of the vector b in respect to the line $[1\bar{1}0]$, a shear on the plane (hhℓ) will be accompanied by a material turn around the axis $\boldsymbol{l} \| [1\bar{1}0]$, *i.e.* around the dislocation line. A transition to the vectors **b**' (5) having a screw component implies a change in the turn axis orientation. The larger is the fraction of the screw component $\mathbf{b}_{||}$ in the superpositional vector b', the more the axis will deviate from the direction $[1\bar{1}0]$, when staying in the plane (hhℓ).

2. If the angle between the vectors $\mathbf{b}_1$ and $\mathbf{b}_2$ is obtuse, then

$$b^2<(n\,b_1)^2+(m\,b_2)^2 \qquad (7)$$

and, in accordance with the Frank criterion (see, *e.g.*, [16]), a cryston is stable. It is easy to make sure that the inequality h<ℓ meets this case. Of course, there is no strict prohibition of realization of the variant h>ℓ (this issue is discussed in more detail in [15]).

3. Already in the framework of the kinematic analysis focusing attention on the «structure» of crystons, it is possible to give a simple interpretation of the jump change of orientations, in particular, for the chain of habits observed in [8]:

(232325) → (111113) → (557) → (112).

Table 1 lists the orientations (hhℓ) and the values n/m derived from the relationship (4) and uniquely corresponding to the former.

Table 1. Orientations of (hhℓ) and the corresponding structure of crystons (n/m)

(hhℓ)	(23 23 25)	(11 11 13)	(5 5 7)	(1 1 2)
n/m	24/1	12/1	6/1	3/1

The correlation of the values n/m with each other (the lower line of the table) demonstrates that a change of the orientations can be interpreted as a result of successive modification of the shear carrier structure, with an addition of an additional dislocation from a conjugate slip system at each stage. Indeed:

$$\frac{12}{1} \equiv \frac{24}{2};\ldots\frac{6}{1} \equiv \frac{12}{2};\ldots\frac{3}{1} \equiv \frac{6}{2}.$$

It is evident that the largest number of changes in the discrete orientations of habits (hhℓ) in the process of plastic deformation could be achieved in the case of $n = 2^k$, where k is an integer, since the doubling of m allows after k steps to reach the relation n/m=1. In the discussed case, $n = 24 = 2^3 \cdot 3$, therefore after three stages of doubling of m, the relation 3/1 is realized.

3. The critical stress of cryston generation

The formation of the operating segment of the GFRS is connected with a strong contact interaction of dislocations. Therefore, as the maximum transverse size of the segment's section by a plane orthogonal to **Λ** (in the direction lying in the plane of the conjugate system), in the general case the quantity k·d can be accepted, where d is the width of the stacking fault, m≥1. It is clear that at k=1, the width of the dislocation bundle is immediately prescribed by the quantity d. At large values of d, the probability of capture of dislocations gliding on the package of planes of the basic system is larger that at small values of d. This means that in materials having a lower stacking fault energy (SFE), for shear bands formed by crystons, a larger number n of dislocations of the basic system are included in a cryston, as compared with materials having a larger SFE. This conclusion is in agreement with the data on the habit orientations which are the first to appear at the stage of developed plastic deformation. For instance, in the Al3%Cu alloy with an fcc lattice, which refers to materials with a high SFE value, the first to be observed were the boundary orientations close to (111113). At the same time, for the Ni_3Fe alloy characterized by an average value of SFE, the first to be observed were the boundary orientations close to (232325). In accordance with Eq. (4), to the orientations (111113) corresponds a cryston with a structure of n=12, m=1, and to the (232325) corresponds a cryston with a structure of n=24, m=1.

At a value of k considerably exceeding 1, quite probable are the processes of division of the GFRS operating segment into an aggregate of segments characterized by the Burgers vectors with smaller contributions of dislocations of the conjugate system. The bowing of a system of such segments will be accompanied by the formation of a package of shear bands. As is known, the existence of shear band packages is rather typical for the stage of developed plastic deformation. It is clear, however, that the formation of such packages is not necessarily connected with the multiplication of sources owing to a double cross slip.

The probability of cryston generation grows in the process of plastic deformation. It develops against the background of the continuing formation of dislocation loops in the planes of the basic slip system (for instance, (111)) and, especially, the conjugate system (for instance, $(11\bar{1})$) that entered the process of plastic deformation later. This leads to an increase in the density of barriers and, correspondingly, stress meso-concentrators. An increase in the number of dislocation barriers in the basic and conjugate planes of a crystal fragment means, first, growth of the number of potential sources of shear carriers on planes different from closely-packed ones, and second, enriches the spectrum diversity of these sources, understood as the possibility to select different variants of the planes (hhℓ) and the Burgers vectors (1).

It is natural to believe that while the cryston density is not very high, an efficient stress relaxation can take place due to the propagation of a large number of crystons with the minimum value of the Burgers vector edge component in relation to the direction $[1\bar{1}0]$ of the plane (hhℓ). An additional consideration in favor of this can be the well-known evaluation of the critical value of stress τ_c for the action of the Frank-Read source (see, for example, [16,17]):

$$\tau_c \sim \frac{Gb}{L} \qquad (8)$$

where b is the Burgers vector's modulus, G is the shear modulus, and L is the length of the dislocation segment pinned at the ends, whose periodic bowing is accompanied by the formation of dislocation loops. In the case of cryston generation, in criterion (8) b corresponds to the length of the superpositional Burgers vector (1) of a cryston capable of slip on the planes (hhℓ), and L corresponds to the length of the Lomer-Cottrell barrier, in the vicinity of which forms a dislocation bundle consisting of dislocations of interacting slip systems. Note should be made that if we consider a fragment of a single crystal with two interacting slip systems (and, hence, with one barrier variant, to be specific, with the line $[1\bar{1}0]$), as the L value, a value can be selected that satisfies the inequality $L \le L^{\perp}$ where, $L^{\perp}$ is the distance between the barriers:

$$L^{\perp} \sim \frac{1}{\sqrt{\rho^{\perp}}} \qquad (9)$$

where $\rho^{\perp}$ is the barrier density in the direction perpendicular to $[1\bar{1}0]$. The latter remark takes into account that an unimpeded bowing of the superdislocation segment to a critical radius ($\rho \ge (L/2)$) presupposes the presence of a free volume with a size of about $L^{\perp}$.

4. On the evaluations of the Peierls stress for crystons, using as an example the shear of (hhℓ) $[\ell\ell 2\bar{h}]$

Since the movement of a cryston is realized in the periodic potential of the lattice, it is important to know the stresses required to overcome the lattice energy barrier without thermal activation. Such a type of stresses is an analogue of the Peierls stress τ_p for an individual dislocation (see, *e.g.*, [16]), therefore it is natural to use for it the notation τ_{pc}.

Currently, there are efficient experimental methods to determine τ_p for dislocations (see, *e.g.*, [18]). For instance, in the case of octahedral slip in metals with an *fcc* lattice (such as Cu, Al, Au, Ag, Pb), the reduced value of $\tau_p/G<10^{-5}$ (it should be noted that in bcc metals the value of τ_p/G is higher by two orders of magnitude).

Analytical estimates of τ_p were also made in [15,19] for a model of a cryston core displayed in Fig.2, whose propagation is accompanied by deformation of simple uniform shear.

If shear deformation on the plane (hhℓ) is prescribed by the value of tgψ, such a shear can be technically viewed as a result of a coordinated movement of partial dislocations on each of the (hhℓ) planes spaced by a distance:

$$d_{hhl} = \frac{a}{\sqrt{2h^2 + \ell^2}} \tag{10}$$

in Fig.2 the value of $\mathrm{tg}\psi = \sqrt{2}/4$.

The sum of the Burgers vectors of partial dislocations should be equal to the magnitude of the superpositional Burgers vector:

$$\mathbf{b} = \sum_{j=1}^{N} \mathbf{b}_{j\,[\ell\ell 2\bar{h}]} \; . \tag{11}$$

The number N of the summands in Eq.(11) is given by the relation

$$\mathrm{N} = \frac{\mathrm{d}}{\mathrm{d}_{\mathrm{hhl}}} = \frac{\mathbf{b}}{\mathrm{tg}\psi \mathrm{d}_{\mathrm{hhl}}} , \tag{12}$$

where d is the size of a shear band in the direction [hhℓ].

It is essential that in [15,19] a more consistent, than in the original Peierls-Nabarro variant [20], derivation of the τ_{pc} formula is discussed, which allows to consider also lattices different from a simple cubic one, on the basis of the conclusions made in the work [21], containing an obvious dependence of τ_p on the interplanar distance *a′* on the shear plane in the direction perpendicular to the dislocation line.

As a result, it is shown that the orientation dependence of τ_{pc} containing an increasing (quadratically with respect to the indices h and ℓ) pre-exponential factor, is effectively cut by a decreasing exponential (with an exponent that is also quadratic in respect to the indices h and ℓ) factor. Therefore the account of τ_{pc} does not impose insurmountable limitations on cryston generation. For instance, the processing of data for Ni_3Fe provides an estimate of τ_{pc}~3.9×10^{-3}G for slip on the planes (23 23 25).

It is also useful to mention independent physical considerations. First of all, let us note that the slip of quasi-planar crystons on the planes (hhℓ) with large (and almost coinciding) values of h and ℓ should be physically indistinguishable from slip on the octahedral plane (111). This means that $(\tau_{pc})_{(hh\ell)}$ at the limiting transition $h\rightarrow\ell\rightarrow\infty$, $d\rightarrow d_{hh\ell}\rightarrow 0$ should be small (on the order of τ_p for octahedral slip of dislocations). Of course, there is no need to realize the limiting transition $d_{hh\ell}\rightarrow 0$, since there is a natural lower bound on the value of $d_{hh\ell}$ that allows to establish how true is the concept of the distinguishability of the nearest crystallographic planes (hhℓ). For that purpose, let us use the uncertainty relations for the position Δx_i and the momentum Δp_i:

$$\Delta x_i \cdot \Delta p_i \geq \frac{\hbar}{2} .$$

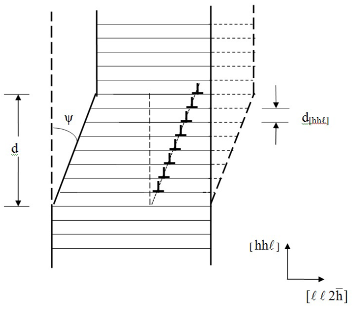

Fig. 2. Formal dislocation scheme of a cryston - carrier of simple shear $[\ell\ell 2\bar{h}]$ (hhℓ).

Let us accept that Δx_i~$d_{hh\ell}=\sqrt{2}\, d_{\ell\ell 2\bar{h}}$ and take into account that the minimum kinetic energy $E_k=((\Delta p)^2)/(2m)$. Then, for example, at an atom mass of m~10^{-25} kg, for $\Delta x_i\approx d_{\ell\ell 2\bar{h}}$ the energy E_k corresponds to the level of the absolute temperature 210 K at the indices h=95 and ℓ=97. It is appropriate to stress that the carried out estimation refers, strictly speaking, to the case of absolute zero temperatures, *i.e.* the value of E_k corresponds to the zero-point oscillation energy. At T>0, as was demonstrated in [22], the uncertainty relation is modified by multiplying the right-hand side by $\coth(\hbar\omega_0)/(2k_Б T)$, where ω_0 is the typical oscillator frequency. Therefore, in reality separate planes (95 95 97) are physically indistinguishable already at room temperature. Evidently, this conclusion is the more so correct for planes with large indices.

Taking into account that the plane (95 95 97) forms an angle $\varphi\approx 0{,}565^0$ with the closely-packed plane (111), it can be expected that slip on such planes (and on planes even closer to closely-packed ones) is associated with stresses slightly exceeding $(\tau_p)_{111}$. Let us note additionally that at a large (but fixed) magnitude of the Burgers vector **b** of a cryston, a limiting deformation of simple shear, still having a physical meaning, is described by the quantity $\mathrm{tg}\psi=b/d_{hh\ell}$ and is localized in a layer with a thickness $d_{hh\ell}$. Conversely, an increase in the transverse (in respect to the shear direction) size of a cryston core, at a fixed Burgers vector of a cryston **b**, should be accompanied by an increase in τ_{pc}.

5. Factors controlling the selection of GFRS, and the mechanism of transition from standard to arbitrary habit orientations

It is clear from the preceding analysis that at the first stage of deformation, most probable is the formation of habits with standard orientations. For the common case of uniaxial loading, an important role is played by the Schmid factor M, proportional to the product of the cosines of the angles between the loading axis and the shear directions (the Burgers vector b) and the normal to the shear plane. In essence, the factor M prescribes the angular dependence of the level of the external shearing stress τ for each GFRS. To determine the critical value for cryston generation τ_c, which we shall consider, by definition, exceeding τ_{pc}, according to Eq.(8), a consistent account of the orientation dependence is required, including the knowledge of not only the magnitudes of the Burgers vectors, but also the anisotropy of the shear modulus G, as well as the values of $L\leq L^{\perp}$. In the case of uniaxial compression, also possible is the mechanism of dislocation loop generation, connected with the bending instability of a dislocation bundle [15].

As was shown in [23], using such information, one can evaluate the deformation levels corresponding to a change in the habits listed in Table 1. Indeed, since the numbers m and n in (1) determine the concentrations of dislocations of the primary and conjugate systems (n/(n+m) and m/(n+m),

Table 2. The values of $(L^{\perp})_i$, calculated using formula (14) for Ni_3Fe single crystals

i	$(n/m)_i$	$(h/\ell)_i$	$(L^{\perp})_i$, μ	ε_i, %
1	24/1	23/25	1.155	9
2	12/1	11/13	0.833	13.5
3	6/1	5/7	0.611	19
4	3/1	1/2	0.462	25.9
5	2/1	1/3	0.4	47

respectively), the barrier density $\rho^{\perp}$ in the averaged lattice of meso-concentrators should be proportional:

$$\rho^{\perp}\sim m/(n+m)=1/(1+(n/m)) \tag{13}$$

Let the experimental values $L^{\perp}_{ej}$ and $L^{\perp}_{ek}$ correspond to the values of plastic deformation ε_j and ε_k Then, on the basis of Eq.(9) and (13), one can expect a quantitative correspondence for correlations of the following type:

$$\frac{L^{\perp}_{ek}}{L^{\perp}_{ej}}=\sqrt{\frac{\rho^{\perp}_j}{\rho^{\perp}_k}}=\sqrt{\frac{1+(n/m)_k}{1+(n/m)_j}} \tag{14}$$

where $(n/m)_k$ satisfies the relation of dislocation fractions of the basic and conjugate systems under deformation ε_k, and $(n/m)_j$ - under deformation ε_j.

Let us take into account that under large deformations ($\varepsilon \gtrsim 50\%$), the saturation effect is observed, manifesting itself in the existence of the minimum non-altering (with increasing deformation) interband distance, equal to 0.4 μm for Ni_3Fe single crystals.

Then, following [24], let us accept the minimum value of $L^{\perp}=L^{\perp}_5 = 0.4$ μm, and assuming that to the latter corresponds $(n/m)_5=2$, using Eq.(14), let us restore the values of $(L^{\perp})_i$ (i<5) satisfying $(n/m)_i$, which are listed in Table 2 for all the observed boundary orientations of shear bands. The values of ε_i in Table 2 have been found using the experimental dependence $L^{\perp}(\varepsilon)$ described in [24].

The found values of ε_i play the role of some reference points on the σ−ε curve.

Certainly, the same factors are essential also during realization of shear bands with habits of the general type. It is clear that during propagation, a cryston loop segment may be fixed (for instance, owing to the interaction with particles of the previously precipitated phase or the interaction with cryston loops with different slip planes) in a position for which the Schmid factor increases. Then there will start the generation of crystons forming bands with habits containing a new GFRS segment. As demonstrated by analysis [14], this scenario that is realized during the formation of bands with the habits {123}. Thus, it is evident that, in principle, habits of an arbitrary crystallographic orientation are possible.

A similar analysis can be performed for crystals with *bcc* and hcp lattices containing more representative sets of SOH [15, 25, 26].

An attractive feature of the cryston model is the possibility to predict the expected habit orientations of shear bands at the subsequent stage of deformation, in case the stress meso-concentrators formed at the preceding stages are known.

6. Main areas of the cryston concept application to the formation of martensite crystals

At the present time, cooling-induced martensite (CIM), stress-induced martensite (SIM) and deformation-induced marteniste (DIM) are distinguished. CIM appears spontaneously during cooling of the initial phase (austenite) at a certain temperature M_s which is below the temperature T_0 of phase equilibrium. SIM is realized under the action of the external elastic stress field σ, in a temperature range of $M_s<T<M_{s\sigma}$, where $M_{s\sigma}$ corresponds to the temperature of the MT start during cooling in case σ is equal to the elastic limit. SIM does not differ in principle from CIM, since the wave mechanism for controlling the growth of separate martensite crystals [27—30] is similar. Therefore, the influence of external stress decreasing the symmetry of the system (crystal plus the external field) comes down to a reduction (in comparison with CIM) of the number of the observed martensite reaction variants. The only exception is the axis orientation of the external stress along the directions <111>, not disturbing the equality of 24 possible variants. To be specific, information about the γ−α MT in iron-based alloys is used, and the crystallographic designations, as well as those above, refer to the fcc lattice basis. Let us remind that the most important specific feature of the heterogeneous nucleation mechanism for CIM and SIM is the emergence of the initial excited state in the elastic fields of dislocation nucleation centers (DNC).

By DIM (in a broad sense of the word) is understood martensite forming in a temperature range of $M_{s\sigma}<T<M_d\leq T_0$, in the conditions in which plastic deformation of a material takes place. In this paper, we use the term «DIM» is a narrow sense. In particular, it is believed that during the formation of a DIM crystal, the controlling process is the process of cryston propagation, which carries shear deformation and stress fields exceeding, in the vicinity of the cryston core, the level of macroscopic elastic limit. In relation to crystons as carriers of threshold deformation, it is appropriate to distinguish between two variants. 1. The fulfillment of the threshold condition is achieved due to the influence of the meso-elastic field σ^{mel}, propagating with the cryston outside the cryston core. Let us designate the crystals of such DIM as DIM1. 2. DIM (let us designate it DIM2) is formed in the region where the main role is played by intra-core deformation components of a cryston. To reveal the peculiarities of DIM2, it is necessary to analyze the elastic field created by a cryston, which in the first approximation can be regarded as a superposition of the elastic fields created by the distribution of prismatic dislocation loops, modeling the cryston core (see, for example, Fig.3).

Calculation of the elastic field of such a distribution represents a separate task, since the nearest vicinity of the cryston core with a high level of σ^{mel} is sensitive to the configuration of the core's deformation field. At distances noticeably exceeding the size of the cryston core, while evaluating σ^{mel} one may confine oneself to the extrapolation of data for the cryston's elastic fields in an approximation characterizing the cryston as a dislocation (dislocation loop) with a prescribed Burgers vector (equal to the cryston's Burgers vector).

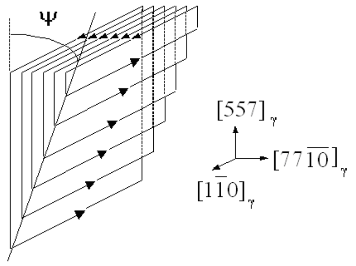

Fig. 3. Example of the dislocation model of a cryston capable of performing the functions of DNC for CIM crystals and a carrier of simple shear[77$\overline{10}$](557)$_\gamma$.

Thus, crystons can be viewed as DNC for CIM and SIM crystals. As compared with separate loops, crystons have a modified angular distribution of the elastic field (primarily, due to a change in the Burgers vectors), as well as a cumulative effect ensuring strengthening of the field. This variant of the influence of crystons on the formation of CIM illustrates in pretty much detail the results described in [31].

Since σ^{mel} accompanies a shear, when analyzing the orientations of habit planes (HP) of DIM crystals one does not need to distinguish between the cases of DIM1 and DIM2, assuming that the shear is localized in a band whose planar boundaries determine the HP orientation of DIM crystals. It is evident that the trivial habit planes of DIM crystals in *fcc* crystals can be the planes {111}. It is also clear that the simplest non-trivial HP orientations coincide with {hhℓ}, although, in principle, HP orientations can be of the general type {hkℓ}. In this connection, it is reasonable to remind that the start of DIM crystals formation in materials with an *fcc* lattice is associated with an intersection of shear bands on the closely-packed planes of the γ-phase, either in the regions of intersection of ε-martensite crystals (with an *hcp* lattice) having, as known, the habits {111}, or at the intersection of the γ-phase twins, also appearing in the case of a shear on the planes {111} in the directions <11$\bar{2}$>. Consequently, the above discussed conditions are fulfilled for contact interaction of dislocations, leading to standard habit orientations {hhℓ}. One of the very impressive examples of this sort is the data [32] on the collision of two austenite twins accompanied by the started growth of a deformation-induced α'-martensite nanocrystal. Reduction of the data from the measurements [32] of habits from a twin basis to an initial austenite basis shows that the α'-crystal habit coincides with {441}. The possibility to extract additional information to confirm the cryston model is associated with the specifics of the interacting objects. Indeed, the identity of the shear value $\sqrt{2}/4$ in the colliding twins allows us to assert that the widths of twins in the region of the growth start of DIM crystals with the habits {441}, according to Eq.(4), should have the relation n/m=5/3, since the magnitude of the sum Burgers vector, correlated to a twinning shear, is directly proportional to the width of the twining plates. In Fig.4, a figure fragment from [32] is displayed, where we have marked the widths d_1 and d_2 of the intersecting twin plates, the interaction of which leads to the formation of one of the α'-crystals.

It can be seen that the relationship of the widths $d_2/d_1 \approx 5/3$ corresponds to the one expected in the framework of the cryston model.

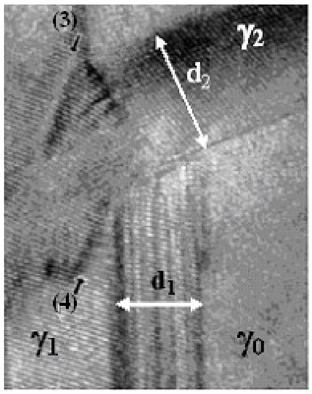

Fig. 4. Fragment of a figure from [32], clarifying the correlation of the fractions of interacting dislocations in the framework of the cryston model during formation of α'-crystals with the habits {441}$_\gamma$.

At the same time, during a detailed elaboration of the morphological description, certain differences can be expected. Namely, in the case of DIM1 crystals, the boundaries will most likely have a crystallographic faceting not so distinct as in the case of DIM2 crystals. It is clear a priori that formation of DIM crystals is possible, in which the central part represents a DIM2 crystal, while its peripheral regions should be referred to DIM1. Not excluded is the case when an insufficient value of threshold deformation inside a cryston leads to a morpho-type where a shear band borders with a DIM1 crystal. In the morphological description, such a variant can be perceived as the initiation of deformation-induced martensite by a shear band. In essence, all the above mentioned DIM variants are observed and were analyzed, with a different degree of completeness, in other papers by the authors (see, for example, [33—36]). Thus, the second important area in the application of the cryston approach in describing MT is related to the leading role of crystons in the formation of DIM crystals as shear carriers ensuring a loss of austenite stability in the propagation region.

Since the formation of all types of martensite crystals is accompanied by the appearance of internal stresses, quite expected is the participation of the cryston mechanisms of shear in the processes of stress fields relaxation at the stage of accommodation of the contacting phases. This is the third area of the application of the cryston approach in describing MT, which has confirmed its constructivity in explaining the interior dislocation structure of the laths [37] of the lath CIM.

7. Promising areas for development and application of the cryston concept

A successful interpretation of the results of the experiments conduced on single crystals creates a reliable basis for interpretation of the shear band formation in polycrystalline samples or, more generally, in a crystalline medium with a dislocation structure conditioned by the past history of the emergency of the sample under deformation. Note should be made that grain boundaries, like boundaries between fragments, are natural regions for localization of cryston sources. Besides, since boundaries are sinks for defects, such sources possess a rich potential for a change of their structure. It means that there is a potential possibility to create sources meeting the most optimum conditions for cryston generation (minimization of the value of τ_c and maximization of the values of M). As a result, the typical propagation mechanism of shear micro-, meso- and macro-bands should be the relay variant of cryston generation. Here, the selection between the intergranular and intragranular shear during the band propagation is governed by the existing distribution of sources and the local threshold conditions of cryston generation.

The concept of crystons looks promising for describing the textures of crystals under deformation. Here, one may expect also a description of the sets of characteristic spectra of misorientation angles of regions divided by high-angle boundaries, including, in particular, special boundaries.

The considered quasi-classical scheme of generation describes in a satisfactory manner the observed experimental features of shear band formation at relatively small strain rates. For these conditions, a slow (viscous, with an intensive energy dissipation) character of the movement of separate crystons should be typical, in which an account of their kinetic energy does not play a significant role. However, it is clear that if a system of meso-concentrators (not necessarily ordered in the form of a superlattice) is present, possible is a collective effect of generation of a cryston avalanche due to a synchronous, mutually induced functioning of GFRS. Such an effect, surely critical to the rate of external deformation, should be accompanied by a change in the character of cryston movement in the avalanche (from viscous to quasi-diabatic), and morphologically it should be reflected in the formation of a system of deformation meso-bands, whose synchronized (self-catalytic) occurrence will be accompanied by a macroscopic effect. It is obvious that a transition to a rapid (quasi-diabatic) character of cryston movement makes an account of its kinetic energy significant.

In the case of a maximally rapid deformation under shock-wave loading, it is nonlinear wave dynamics that should be adequate to the physical picture from the very beginning. Notably, if a loaded sample in its past history experienced plastic deformation, the shock action may lead to the formation of a system of shear bands (with boundaries not necessarily coinciding with slip planes), initiating at least a one-time cryston generation by the earlier created GFRS. And if a shock-loaded perfect single crystal (with a low dislocation density), alongside with shear bands corresponding to the slip systems typical for this crystalline lattice, contains shear bands with boundaries of different orientations, a successive interpretation of the formation of such bands is possible only within a nonlinear-wave description.

Finally, various combined scenarios may also be realized, with participation of the wave and cryston deformation carries. For instance, the formation of the basic component of a bainitic ferrite macro-plate is capable of stimulating the formation of twins of initial austenite in the wave mode, while in the emerged austenite, sub-laths of the additional component of bainitic ferrite start to grow [38].

It is understood that in all the cases, band formation is connected with the process of the propagation of deformation localized in the region of the growing band front, representing, in the mathematical aspect, a switching wave. Therefore, making a final summary, one can state that an interpretation of the dynamic structure of a displacement (distortion) field in the region of the front (as well as in the case of martensite crystals formation) is essential for understanding the details of the shear band formation mechanism.

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