2. PLASTIC DEFORMATION

During the deformation of a solid the measurable macroscopic magnitudes are: the force $F$ applied to the sample and linear dimensions of the sample. The combination of these magnitudes allows us to calculate the average stress $\tau$ and average deformation $\varepsilon$:

$$\tau = \frac{F}{A} \quad (A \text{ is the sectional area that changes during the deformation process and to which the force } F \text{ is applied}), \quad \varepsilon = \frac{\Delta l}{l} \quad (\text{where } \Delta l, l \text{ are length variation and initial length of the test sample accordingly}).$$

If stretching (pressing) of the crystal occurs at constant deformation speed ($\varepsilon = \text{const}$), then we speak about dynamic regime of deformation (and this very regime is studied further in the paper). It is supposed that in dynamic regime of tests the material preserves its elastic properties in the area of plastic deformation and it is believed, in this case, that the deformation consists of two parts [Rab79]:

$$\varepsilon = \varepsilon_{\text{plast}} + \varepsilon_{\text{elast}}. \quad (2.1)$$

The purpose of the most works dedicated to plastic deformation is to determine interactions between macroscopic magnitudes that characterize the properties of test material during deformation and microscopic magnitudes that characterize the movement of dislocations. In connection with this in the first part of the present chapter the deformation curve is analyzed qualitatively, i.e. macroscopic parameters of the deformation are examined. Then our attention switches to qualitative and quantitative models of the microprocesses that flow in different deformation stages. Finally models of kinetics of the origination of point defects are scrutinized.

2.1 Deformation curve during stretching

Deformation curve during stretching in dynamic regime conditions can be divided into several areas [Die56]. Figure 2.1 represents the qualitative dependence of stress $\tau$ from deformation $\varepsilon$. It should be noted that in real experiments these or that areas do not necessarily become apparent (in the figure 2.1 the most general situation is depicted).

Area of the first curve rise corresponds to the area of elastic deformation or the Hooke area. It is characterized by the linear dependence of stress from deformation and reversibility of the deformation process.

The area of elastic deformation borders with the area where the decrease in stress at the increase in stretching is observed. There the plastic deformation begins and the area is called liquidity area, $\tau^*_h$ and $\tau_h$ are accordingly called the upper and lower yield stress. It should be mentioned that the nature of dependence of $\tau$ from $\varepsilon$ in that area to a considerable degree is determined by the initial state of the sample. For example it may be determined by the concentration of dislocations before the deformation. Moreover the dependence may take different values, e.g. constant stress can be traced during the deformation.

The area of easy glide denominated as stage $I$ adjoins the liquidity area. In this phase dislocations are free to move almost in any direction in single slip system increasing the deformation without significant rise in stress [Lak80]. This phase is sometimes called the phase of insignificant hardening. It is also possible that the stress in that area remains the same. In the time of strong deformation the phase of multiple slip is seen meaning that dislocations move in two or more systems [Lak80] (phase $II$). In this phase the dislocation structure becomes very complicated and density of dislocations increases in comparison with its initial state.
2. Plastic deformation

Because of the dislocations interaction the resistance to their movement grows sharply and so the stress should also rise to ensure their motion. Phase \( \text{II} \) is called the hardening phase.

Phase \( \text{III} \) is called the phase of dynamic recovery. Such term was chosen to describe it because in this phase dislocations disappear and then appear again.

At high temperature deformation additional stages of hardening (\( \text{IV} \)) and recovery (\( \text{V} \)) may appear.

### 2.2 Movement of dislocations

**Geometry.** The plastic deformation is occurs due to dislocation motion. The simplest types of dislocations are edge and screw dislocations. The edge dislocation model is shown on figure 2.2. If we imagine extra crystalline half plane (on the picture it coincides with the upper half plane \( y, z \); the \( z \) axis is directed vertically with respect to the picture plane) that is inserted into the lattice (its section is shown on the picture) then the edge line of that half plane (it coincides with the \( z \) axis on the picture) represents the edge dislocation. After the entire bypass about the dislocation line the lattice points displacement vector gains the increment \( \vec{b} \). This vector is called the Burgers vector and it is equal to one of the lattice vectors. In case of the edge dislocation the Burgers vector is perpendicular to the dislocation line, in case of the screw dislocation the vector is parallel to the dislocation line.

In [Rab79] it is shown that if there are no defects in the crystal (vacancies and interstitials) then dislocations are able to move only in directions where the following is true:

\[
(d\vec{u} \times d\vec{\ell}) \vec{b} = 0. \tag{2.2}
\]

Here \( d\vec{\ell} \) is the dislocation line element, \( d\vec{u} \) – vector of dislocation shift. It results from the (2.2) correlation that the edge dislocation is able to move only in the plane in which the dislocation is found and where the Burgers vector is present. On the contrary, according to the (2.2) equality the screw dislocation is able to move in any plane that contains the dislocation line because in this case the Burgers vector is parallel to the dislocation line. This means that in the process of its movement the dislocation line is able to move from one plane to another.

Theoretically possible directions of movement of dislocations were discussed above. However, the slip (displacement of separate crystal parts with regard to each other) in the crystal lattice is mainly realized along planes and directions with higher density of atoms where the magnitude of shear resistance is minimal.
2.2 Movement of dislocations

This is because the distance between such neighboring atomic planes reaches its maximum and thus the bonds between them are extremely weak. The slip plane and the slip direction that lie in these planes constitute the slip system. It is possible that more than one slip system act in crystals simultaneously. For instance in crystals with face-centered cubic lattice the slip is effected in planes (111) [Lak80]. The more slip planes and directions exist in a crystal, the more it is exposed to plastic deformation. It must be noted, however, that the slip process should not be considered as a process of simultaneous displacement of one crystal part with relatively to the other. Such simultaneous shear would require far greater stress, probably hundred times as great as that at which the process of plastic deformation runs in reality. The real shear mechanism is quite different. The movement of dislocations at distances that are significantly greater than interatomic distances leads to the following. Two crystal parts that are divided by a plane in which dislocation moves come closer at the distance that equals to interatomic distance; and visible shear is detected after the repeated movement of dislocations on the slip plane.

Tangential stress $\sigma$ that influences the dislocation in a certain slip system can be calculated by the formula $\sigma = m_s \tau$, and the shear $\gamma$ of a crystal in the slip plane for the case of little deformations ($\varepsilon << 1$) by $\gamma = \varepsilon / m_s$. Here $m_s$ is the Schmidt factor and $m_s = \cos \Phi \cos \Psi$ ($\Phi$ is the angle between the deformation axis and the unit vector that is perpendicular to the slip plane; $\Psi$ is the angle between the deformation axis and the Burgers vector). The Schmidt factor takes into account, in particular, direction of the influence of the external load on the crystal (using for this purpose the deformation axis) and the orientation of the slip plane (by means of orientation of the unit vector that is perpendicular to the slip plane).

Peierls potential. For the movement of the dislocation line along the lattice vector it is necessary to successively (as the dislocation moves) break the bond between the adjacent atoms on all dislocation length (see figure 2.2). To effect such movement the dislocation has to absorb large quantity of energy and that casts doubt on the mere possibility of the movement of dislocations itself. But in fact the dislocation can move consuming little energy. Peierls [Pei40] suggested the model that describes the movement of the dislocation according to which it moves in the periodic potential by the law:

$$W = W_0 \sin^2 \left( \frac{\gamma y}{a_p} \right)$$

(2.3)

At that the magnitude $a_p$ is equal to the Burgers vector. The scheme on picture 2.3 shows the movement of the dislocation line in the potential (2.3). Kinks appear on the dislocation that lies in the slip plane. These kinks are able to appear as a result of thermogeneration.
2.2 Movement of dislocations

The movement is effected by the successive overcoming of barriers by segments (elements of the dislocation line that include two kinks). This process of segments flow over the barrier (like a wave) is much more beneficial from the energy point of view than the process of barrier overcoming by the whole dislocation at once. Such dislocations motion regime is generally called the fracture movement regime.

The studied qualitative model of the movement of dislocations – when dislocations cover large distances without any difficulties – can be used to explain phase I of crystal deformation (see figure 2.1).

**Obstacles in the slip process.** As a rule, crystals contain defects of different types that impede the movement of dislocations. These are impurity atoms, point defects and their clusters (existing both in the initial crystal and appearing during the deformation process), dislocations themselves.

The figure 2.4 shows the model that illustrates the stretching of the dislocation line that is fixed on obstacles (for example, on impurity atoms). In the first position the dislocation line is fixed.

If the external stress $\sigma$ effecting the dislocation at the location of obstacles is strong enough then the dislocation stretches and tears off from the obstacles (position 2). In the time of further stretching the dislocation tears off from the layer of impurity atoms and hits...
2. Plastic deformation

upon the next layer of impurity atoms that are located more to the right along the \( y \) axis. Such dislocation motion regime is called the stretching regime. Other dislocations can also be obstacles for the movement of dislocations due to their interaction. Interaction force between the two parallel screw dislocations related to the dislocation length unit near the isotropic medium is [Lan87]:

\[
F_s = \mu \frac{b_1 b_2}{2\pi d_1}. \tag{2.4}
\]

The similar magnitude for the two parallel edge dislocations that lie in the single slip plane is [Rab79]:

\[
F_E = \mu \frac{b_1 b_2}{2\pi(1-\nu)d_i}. \tag{2.5}
\]

At this \( \mu \) is the shear modulus, \( \nu \) is the Poisson coefficient, \( d_1 \) – distance between the dislocations, \( b_1, b_2 \) – absolute value of the Burgers vector of the first and the second dislocation accordingly. From (2.4) and (2.5) results that dislocations draw if Burgers vectors \( (b_1 \times b_2 < 0) \) are pointed into opposite directions but if vectors \( (b_1 \times b_2 > 0) \) are pointed into one direction then dislocations push off.

Mutual influence of dislocations on their movement can be more complicated. A crystal may have dislocations located in different slip planes and not in the single plane. Thus dislocations located in one slip plane are influenced by other dislocations located in other slip planes. Such influence can be illustrated in the following manner. Let a certain slip plane \( S_1 \) (figure 2.5) contain infinite number of identical parallel rectilinear edge dislocations removed on equal distance \( h \) one from another (dislocation lines are directed along the \( z \) axis).

![Fig. 2.5: Dislocation wall scheme. \( S_1 \) is the plane of arrangement of dislocations (dislocations are directed along the \( z \) axis and are shown by straight lines). \( S_2 \) is the plane of observation. \( x_0, h \) are the distances between the planes and dislocations accordingly. \( \sigma \) is the tangential stress.](image-url)
2.3 Multiplication of dislocations

Such dislocation system is called the \textit{dislocation wall}. In the plane \( S_2 \) (the plane \( S_2 \) is parallel to the plane \( S_1 \) and is removed from it on the distance \( x_0 \)) this dislocation wall creates tangential stress \( \sigma \) that works parallel the \( y \) axis. If there are dislocations in the plane \( S_2 \) then this stress effects their movement. In approximation the isotropic medium and there where the condition \( x_0 \gg h \) is realized the expression for \( \sigma \) is \[ \text{Lan87} \):

\[
\sigma(y) = 4\pi^2 B \frac{b x_0}{h^2} \exp\left(-2\pi x_0/h\right) \cos(2\pi y/h).
\]

(2.6)

At this \( B = \mu / 2\pi(1-\nu) \), \( b \) is the Burgers vector length for the edge dislocation. Dislocation walls are comprised, as a rule, from dislocations of little mobility and represent obstacles for mobile dislocations. Dislocations of little mobility (as well as dislocation walls) located in different slip planes constitute the system of almost immobile \textit{tree-like dislocations} that influence a lot the movement of mobile dislocations. Tree-like dislocations can appear both in the process of crystal growth and during crystal deformation. Increase in the concentration of the tree-like dislocations influences crystal strengthening that is noted in stage \( II \) of the deformation curve during stretching (see fig. 2.1).

\section*{2.3 Multiplication of dislocations}

The concentration of dislocations is determined as the ratio of all dislocations length to the volume of the sample. This means that change in concentration of dislocations is not always connected with the change in number of separate dislocations because the change in total length of dislocations also means change in their concentration. In this aspect increase both in number of dislocations and their total length means multiplication. Let us focus on several mechanisms of multiplication of dislocations.

The dislocation line cannot simply end in a crystal: it rather appears on the surface of the crystal or forms a closed loop. For such dislocations the Burgers vector has constant value and direction throughout the entire dislocation line but the angle between the Burgers vector and the dislocation line element changes along the dislocation line. Proof of this statement can be found, for example, in \[ \text{Lan87} \]. Mechanisms of multiplication of closed dislocations can be explain with help of the Frank-Read mechanisms.

\textbf{Frank – Read Source.} The figure 2.6 illustrates the multiplication of closed dislocation loops (closed dislocation cycles). The segment A of the dislocation line

\begin{figure}[h]
\centering
\includegraphics[width=0.5\textwidth]{fig2_6.png}
\caption{The Frank-Reed deformation source. Segment A of the dislocation stretches between the phases B and C. A new dislocation appears on the phase D.}
\end{figure}
2. Plastic deformation

(the initial dislocation itself is already a closed loop) that is fixed on two obstacles stretches under the influence of the external stress applied to it. As the stress increases the segment A progresses through several phases. The figure 2.6 represents two such phases B and C. The process enters into the phase D after the specific stress value has been achieved and there both dislocation line bends merge forming a single closed dislocation. The remaining part of the segment A of the initial dislocation line can again stretch and form a new dislocation.

The multiplication of dislocations and their movement in the Peierls potential serves to explain the nature of dependence of stress from crystal deformation in the liquidity area (see fig. 2.1). The stress in the elasticity area is high enough to allow dislocations to overcome obstacles there where the elasticity area borders the plasticity area [Ale68a].

Cross Slip. As has been described above (see chapter 2.2), the Burgers vector and direction of the dislocation line determine only one slip plane of the edge dislocation. During its movement the edge dislocation stays within the borders of the slip plane. Theoretically, only planes can be a slip plane for screw dislocations but in practice only planes with the densest concentration of dislocations participate in the slip process. It’s known that, theoretically, a screw dislocation can move freely in any direction so there are no restrictions for it to change slip planes. Such movement from one slip plane to another occurs under the influence of neighbouring dislocations and other obstacles on the viewed dislocation. In this case we observe what is called double cross slip of dislocations. The figure 2.7 illustrates this double cross slip.

Fig. 2.7: Double cross slip of dislocations. MN- initial dislocation line. LE, ND-dislocation line segments. The segment C can act as at the Frank-Read source.

The initial dislocation moves in the slip plane (1). Here MN is the dislocation line. At a certain moment a kink can appear in the screw dislocation. Kink represents a dislocation segments (LE, ND are dislocation segments) in the form of a step. The segment slip plane (2 is the segment slip plane) is perpendicular to the slip plane of the initial dislocation. The slip process in the plane (2) causes the segment to pass from this plane to the plane (3) which is parallel to the slip plane of the initial dislocation (1). The segment C can act here as the Frank-Read source. After the segment has moved to the plane (3) it can again return to the initial slip plane (1) and so the initial dislocation will continue its movement in the plane (1). In this very manner segments and dislocations slip in lateral planes (planes 2 and 1 respectively).
Destrengthening of the stress curve during stretching seen in the stage V (see fig. 2.1) associates with more rapid cross slip because, on the one hand, dislocations density increases and, on the other hand, the velocity of dislocations also increases as the cross slip helps to avoid obstacles during the slip process.

One must remember, however, that speaking about the multiplication of dislocations one should consider the possible reduction in density of dislocations due to annihilation. If the total Burgers vector of some of the interacting dislocations takes zero value then these dislocations will disappear.

### 2.4 Kinetics of the movement of dislocations

Qualitative analysis of the causes that influence the behaviour of the dependence of stress from crystal deformation has been already considered above (see chapters 2.2 and 2.3). To sum up we can say that the behaviour of dislocations, i.e. their movement and multiplication, determines significantly the behaviour of the deformation curve. This statement concerns both the dislocations that were present in the crystal long before the deformation and those that were emerged during the deformation process. To be able to analyze the behaviour of dislocations in a deformed crystal one needs certain equations that connect experimentally observed magnitudes, as well as those parameters that characterize the conditions at which the deformation process runs like deformation values, deformation velocity, and crystal temperature, with the magnitudes that characterize kinetics of the movement of dislocations, i.e. velocity and density. This chapter deals with these equations.

#### Effective Stress Model

It is very important to know what stress affects the dislocation line. Apart from the external stress on dislocation line is affected, stress fields of crystal dislocations. In [Ale68a] got expressions for stresses caused by dislocations with the density $N_d$:

\[
\sigma_i^E = \left[\mu b / (2\pi(1-\nu))\right]N_d^{1/2}.
\]

(2.7)

\[
\sigma_i^S = \left[\mu b / 2\pi\right]N_d^{1/2}.
\]

(2.8)

The first equation serves to measure the stress in edge dislocations, the second is used in case of screw dislocations. To determine the stress applied to the dislocation line one should substitute the effective stress $\sigma_{\text{eff}}$ for the external tangential stress $\sigma$ that influences dislocations in a certain slip system [Haa62]:

\[
\sigma_{\text{eff}} = \sigma - \sigma_i = \sigma - AN_d^{1/2}.
\]

(2.9)

Here the constant $A$ includes different types of dislocations. The minus sign appears in the formula because dislocations strive to reduce overall stress in a crystal. Using (2.9) the only thing we can do is measure the maximum possible density of dislocations. Assuming $\sigma_{\text{eff}} = 0$ ($\sigma_{\text{eff}} \geq 0$) the maximum value of dislocations density will be $(\sigma / A)^2$. As a rule, it’s better to use (2.10) to determine the density of dislocations at a given external stress applied to the sample [Ger86]:

\[
\tau = \alpha \mu b N_d^{1/2}.
\]

(2.10)
At this $\alpha$ – is a constant determined by the geometry of the crystal structure and takes its values between 0.5 and 1 [Mec81].

**Orowan Equation.** Orowan [Oro40] was the first to view plastic deformation as a dynamic process. He supposed that plastic deformation is similar to other transport processes that run in a solid, for example, the process of movement of charge carriers. The difference is that in this case dislocations act as charge carriers. The following correlation was obtained to define the velocity of crystal shear along the dislocations slip plane:

$$\dot{\gamma} = bN_{dm} \nu .$$  \hspace{1cm} (2.11)

Here $\nu$ is mobile dislocations velocity, $N_{dm}$ is mobile dislocations density.

**Dislocations Velocity.** There are two models that relate between dislocation velocity to parameters that characterize experimental conditions: stress and temperature of the sample [Sch65], [Ale68a].

Schoeck [Sch65] followed thermodynamic understanding of the phenomenon and studied the changing of the thermodynamic Gibbs potential $G$ (the Gibbs enthalpy) of a crystal during its deformation\(^1\), i.e. studied the enthalpy difference $\Delta G$ before and after crystal deformation. In [Zon94] the equation for $\Delta G$ was given in its commonest form:

$$\Delta G = U - V_A \sigma_{\text{eff}} .$$ \hspace{1cm} (2.12)

Here $U$ is the energy barrier that dislocations have to pass during gliding. $V_A$ is the *activation volume* of the movement of dislocations [Wee83a]:

$$V_A = bdl .$$ \hspace{1cm} (2.13)

Here $l$ is the length of a dislocation element fixed on the two nearest obstacles (see fig. 2.4); $d$ is the stretching amplitude of that dislocation element in lateral direction with respect to the element line at the moment it tears off from obstacles under the influence of the external stress.

In [Wee83a] relying on the Schoeck model [Sch65] the following was obtained to define dislocation glide velocity:

$$\nu = lf \exp(-\Delta G / k_B T).$$ \hspace{1cm} (2.14)

At this $f$ – is the oscillation frequency of the dislocation line, $k_B$ is the Boltzmann constant. Later, assuming that the activation volume does not depend on the external stress

\(^1\) The enthalpy $G$ for deformed crystal is $G=E-TS-\sigma_{\text{ik}u_{\text{ik}}}$ [Lan87]. Here $E$, $T$, $S$ are internal energy, temperature and enthalpy of a deformed crystal accordingly; $\sigma_{\text{ik}}$ is the external stress tensor, it functions in the $x_k$ coordinate direction in the plane perpendicular to the $x_i$ coordinate; $u_{\text{ik}}$ is the deformation tensor which is $u_{\text{ik}} = \frac{1}{2} \left( \frac{\partial u_i}{\partial x_k} + \frac{\partial u_k}{\partial x_i} \right)$; $u_j$ shows the crystal points displacement along the $x_j$ axis.
and taking into consideration 2.9, 2.12 and 2.14 the following expression is derived to connect dislocation velocity $\nu$ with stress $\sigma$: $\nu \sim \exp(\sigma)$. However, it was experimentally found out that the functional dependence between $\nu$ and $\sigma$ is expressed like $\nu \sim \sigma^m$, where rate of stress indicator $m$ takes values from 1 to 2 (for semiconductor materials) [Sch67].

In [Ale68a] Alexander and Haasen suggested a model for movement of dislocations velocity assuming that dislocations move in the fracture regime (see fig. 2.3). Taking into consideration this supposition and with regard to the assumption that fractures appear due to thermal activation (by means of passing some of oscillating atoms’ energy to dislocations) dislocations velocity is given by:

$$\nu = B\sigma^m \exp(-U/k_B T). \quad (2.15)$$

Here $U$ is the activation energy of movement of dislocation which is the sum of formation energy and double fractures migration energy. $B$ is the empiric constant. Rate of stress indicator takes values from 1 to 1.5.

If we use the Schoeck model to describe movement of dislocations velocity, then $V_A$ and $U$ will be the parameters to characterize the movement of dislocations. But if, on the contrary, we take the Alexander and Haasen model [Ale68a] for this purpose, then $m$ and $U$ will be the parameters to characterize the movement of dislocations.

Let’s determine the activation volume $V_A$ of the movement of dislocations. Using the Orowan equation (2.11) for expressing $\nu$ we'll have: $\nu = \gamma / (bN_{d_a})$. Given that $\gamma = \varepsilon / m_s$ we come to $\dot{\gamma} = \dot{\varepsilon} / m_s$. Now for movement of dislocations velocity we arrive at $\nu = \varepsilon / (m_s bN_{d_a})$. Substituting this expression for velocity into (2.14) and finding the logarithm of the expression we get:

$$\ln \varepsilon = \ln(m_s bN_{d_a} blf) - \Delta G / k_B T.$$

Using (2.12) for $\Delta G$ and (2.9) for $\sigma_{eff}$ we see that:

$$\ln \varepsilon = \frac{\sigma V_A}{k_B T} \left( U + AV_A N_{d_a}^{1/2} \right) + \ln(m_s bN_{d_a} blf).$$

Assuming that the dislocations density $N_{d_a}$ does not depend on $\sigma$ the following is obtained:

$$\frac{V_A}{k_B T} = \left[ \frac{\partial \ln \varepsilon}{\partial \sigma} \right]_{T=\text{const}}.$$

Taking into account that $\sigma = \tau m_s$ ($\tau$ is the external stress applied in a certain direction) we arrive at:

$$V_A = \frac{k_B T}{T}.$$

(2.16)
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Here

\[ I = m \left( \frac{\partial \tau}{\partial \ln \varepsilon} \right)_{T=\text{const}}. \]  

(2.17)

Let’s express rate of stress indicator \( m \). Substituting the above found expression for velocity \( \nu = \varepsilon(l(m, bN_d)) \) into (2.15) and taking the logarithm we’ll finally have:

\[ \ln \dot{\varepsilon} = m \ln \sigma + \ln(m, bN_d, B) - \frac{U}{k_B T}. \]

Assuming that the density of dislocations is a constant value and taking into account that \( \sigma = \pi m_j \), we arrive at the expression for \( m \):

\[ m = \left( \frac{\partial \ln \dot{\varepsilon}}{\partial \ln \tau} \right)_{T=\text{const}}. \]

(2.18)

It is possible to determine \( V_A \) and \( m \) but first we need to determine experimentally the dependence of \( \tau \) from \( \varepsilon \). But to do this it’s necessary for the value of dislocations density to be a constant. Changing the stretching velocity the density of dislocations can also change. This means that at the chosen experimental conditions for determining the values \( V_A \) and \( m \) stretching velocity should be changed abruptly. Then the abrupt changing of \( \varepsilon \) prevents the changing of dislocations density because the necessary processes have very little time to run smoothly till the end.

Unfortunately, it’s impossible to determine the value of \( U \) in this case. For example, if the above described method to determine \( U \) is used then, assuming that the Alexander and Haasen model [Ale68a] is right, we have:

\[ \left[ \frac{\partial \ln \dot{\varepsilon}}{\partial \left( \frac{1}{k_B T} \right)} \right]_{N_d=\text{const}} = m \left[ \frac{\partial \ln \tau}{\partial \left( \frac{1}{k_B T} \right)} \right]_{N_d=\text{const}} - U. \]

As the strain rate \( \dot{\varepsilon} \) does not depend on parameter \( T \) under the experimental conditions, we arrive at:

\[ U = m \left[ \frac{\partial \ln \tau}{\partial \left( \frac{1}{k_B T} \right)} \right]_{N_d=\text{const}}. \]
The assumption that the Schoeck model [Sch65] is right gives the following expression for $U$:

$$U = V_A m \left[ \tau + \frac{1}{k_B T} \frac{\partial \tau}{\partial \left( \frac{1}{k_B T} \right)} \right]_{\varepsilon = \text{const}} .$$

It results from these formulae, deduced for activation energy $U$ of the movement of dislocations, that to define the value of $U$ one must change abruptly the crystal temperature. Such change, however, should be made so that the density of dislocations retains its constant value, but it’s technically impossible as the speed of crystal warming has its limit. So a different method is used to define the activation energy $U$ of the movement of dislocations. In [Ale68b] the relation between the liquidity lower limit $\tau_{ly}$ (see fig. 2.1) and $U$ was derived as:

$$\tau_{ly} = C \varepsilon^{1/(2+m)} \exp \left\{ \frac{U}{(2+m)k_B T} \right\} .$$

Here $C$ is an empiric constant. Using (2.19) we come to:

$$\ln \tau_{ly} = \frac{1}{(2+m)} \ln \varepsilon + \frac{U}{(2+m)k_B T} + \ln C .$$

Hence, for $U$ and $m$ we have:

$$U = (2+m) \left[ \frac{\partial \ln \tau_{ly}}{\partial \left( \frac{1}{k_B T} \right)} \right]_{\varepsilon = \text{const}} .$$

Thus, from experiments for dependence of the lower yield stress $\tau_{ly}$ on the strain rate $\varepsilon$ at a maintained temperature we can define using (2.21) the rate of stress indicator $m$ value. Later, having defined the experimental dependence of $\tau_{ly}$ from the crystal temperature $T$ at a constant strain rate $\varepsilon$, it is possible to find $U/(2+m)$ using (2.20). Finally, from these experiments we are able to find $U$ and $m$ but not the value of the activation volume $V_A$ of movement of dislocations. To determine $V_A$ one must first experimentally determine the dependence of $\tau$ from $\varepsilon$ changing abruptly the value of $\varepsilon$ during one experiment. At the same time the temperature of a sample must be kept constant during the experiment. Then the value of $V_A$ is derived using (2.16) and (2.17).
Moreover, these very experiments help to define the value of \( m \) using (2.18); in this case it's unnecessary to conduct experiments with abrupt changing of \( \varepsilon \) for determination of \( m \) because as has been stated above the value of \( m \) may be obtained from other experiments. However, the comparison of the values of \( m \) obtained by two above described methods is of great interest.

### 2.5 Formation of point defects

#### 2.5.1 Models

**Dislocations Climb.** When (2.2) is true a dislocation moves conservatively in a slip plane. But when the (2.2) correlation doesn't apply dislocations may come out of the slip plane by means of climb and such movement of dislocations is considered to be nonconservative. Such kind of movement can be realized as long as a crystal lattice contains vacancies (defects) and interstitials capable of moving freely due to disproportional distribution of the thermal oscillation energy between atoms. We can imagine a defect located within a dislocation neighbourhood to move, such movement having diffusion inherent characteristics. The dislocation starts moving and follows the defect leaving its own slip plane. However, in most cases such movement of dislocations can be seen only at extremely high temperatures.

As a matter of principle, dislocations climb is a mechanism that may be used for generation of point defects because such movement of dislocations may be effected by means of emission or absorption of point defects. The emission of point defects, in its turn, may be explained the following way. A crystal lattice vacancy adjoining the dislocation can be rapidly transported along the dislocation line with help of diffusion and drifting movement in the mechanical stress field created by the dislocation [Cui96]. The dislocation line is a ‘narrow gully’ for vacancies [Mec80]. This can be explained by creation of alternate mechanical stress in anisotropic environment by the dislocation. The stress causes the dislocation to move to those areas containing the dislocation line where the crystal is compressed. As a result, crystal stress there where the dislocation rests now relaxes a little. The emission of crystal lattice vacancies can be affected only when reduction in the energy of the system (comprising vacancies with the dislocation) is higher than the energy required to form crystal lattice vacancies. The reduction is possible due to partial relaxation of stress in the locality of vacancies.

In [Wee83b] the following expression was obtained for the concentration of a point defect near dislocations that move according to the climb mechanism:

\[
C - C_0 = C_0 \exp(-E / k_B T) \exp(\pm \sigma b^3 L / dk_B T).
\]

Here \( C \) is the concentration of vacancies near the dislocation; \( C_0 \) is the concentration of vacancies at the thermodynamic equilibrium of the crystal away from dislocations (there where there’s no influence of dislocations on the concentration of vacancies); \( E \) is the vacancy formation energy; \( \sigma \) is the mechanical stress caused by the dislocation; \( L, d \) denote the average way of the dislocation line before annihilation with another dislocation during dislocation glide and climb accordingly. The plus (minus) sign in the second exponential member corresponds to the emission (absorption) of crystal lattice vacancies by the dislocation.

Phase III on the deformation curve during stretching (see fig.2.1) is explained by dislocations climb [Haa89].

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**Fracture Resistance.** In case a screw dislocation is studied, there is one more reason (in addition to that considered above) that causes emission and absorption of defects. During its movement a screw dislocation experiences a kink in certain cases. Such kinks on the dislocation line can be of different nature. Some appear due to double cross slip of dislocations (see fig. 2.7). The influence of mechanical stress fields of other dislocations on the given dislocation can, in particular, be the reason why a screw dislocation travels to a neighbouring slip plane, such movement being caused by double cross slip.

But there is another reason for appearance of kinks. Kinks may appear by virtue of crossing of two screw dislocations. Figure 2.8 (the scheme was taken from [Hub98]) illustrates how kinks appear.

![Fig. 2.8: Crossing of two screw dislocations (t<sub>1</sub> is the time moment before crossing, t<sub>2</sub> is the time moment after crossing).](image)

As after the passage of a dislocation one-crystal part (that one above the dislocation slip plane) displaces with regarding to the other part of the crystal (that one under the slip plane) on the Burgers vector, the dislocation line itself also moves, i.e. a kink appears.

Mott [Mot60] suggested a model explaining how a screw dislocation moves having gained impulse of a kink during emission or absorption of point defects. A kink in a dislocation line is a graded dislocation segment. The kink slip plane is perpendicular to the screw dislocation slip plane. The screw dislocation kink affects nonconservative climb motion. Such nonconservative motion can be realized by emission or absorbing point defects and such dislocation movement mechanism is called fracture resistance. Series of defects may be formed by means of fracture resistance. But such defects configuration (series) is sometimes unstable. As a result, two-dimensional or three-dimensional defects clusters may originate. In particular, three-dimensional vacancies clusters are constituted.

### 2.5.2 Kinetics of formation of point defects

Both processes that lead to defects emergence and processes that are responsible for their annihilation should be taken into consideration when analyzing the kinetics of formation of point defects. There are different ways of annihilation of vacancies and interstitials (interstice atoms):

1. Recombination of crystal lattice vacancies with interstitials.
2. Interaction of vacancies and interstitials with dislocations and atoms of different impurity (in this case other point defects emerge).
To make the annihilation of vacancies and interstitials possible it is required that they be mobile at crystal deformation temperature. Further on, for formation of stable defects – complex of a vacancy (interstitial atom) and an impurity atom – it’s necessary that the annealing temperature of such complex be higher than the deformation temperature, or else these complex will not be formed. The difficulty to conduct analysis of the annihilation process in this case lies in the impossibility to use experimental data obtained for an undeformed crystal. The point is that dislocation mechanical stress may significantly influence the defects formation processes in a deformed crystal (there is no such influence in an undeformed crystal).

**Formation of point defects velocity**

Two completely different approaches are used to determine the velocity of formation of vacancies and interstitial atoms during the movement of dislocations. The Mott [Mot60] and Popov [Pop90] models study macroscopic processes, while the Estrin-Mecking model [Mec80] uses energetic considerations rather than macroscopic ones.

A kink generates point defects during screw dislocation glide and it is believed to be the initial moment of the formation of point defects velocity model. Mott was the first who paid attention to screw dislocations glide that contain kinks [Mot60]. Taking into consideration processes of kinks creation resulting from crossing of gliding screw dislocations with noncoplanar slip system dislocations (tree-like dislocations), and their annihilation or merger, Mott proposed the following:

\[
\frac{\partial C_V}{\partial \gamma} = \frac{\xi^{1/2}}{2\sqrt{2}} bN_d^{1/2}.
\] (2.23)

Here \( C_V \) is the relative concentration of vacancies (ratio of the true concentration of vacancies to the concentration of lattice atoms), \( \xi \) – is the ratio of tree-like dislocations density to the general density of dislocations. Popov proceeds from the same considerations as Mott but examines the conservative movement of kinks along the screw dislocation in more detail [Pop90]. In that model it’s assumed that to form point defects it is necessary for the following to be true:

\[
\nu > v_j b/\lambda_j.
\] (2.24)

If (2.24) is not right, defects cannot be formed. Here \( v_j \) is the kinks glide velocity, \( \nu \) – is the movement of dislocations velocity, \( \lambda_j \) is the distance between kinks on the dislocation line. The (2.24) condition was obtained proceeding from the following suppositions. To form defect it is necessary that a dislocation has covered the distance \( \Delta l \) before kinks annihilation. This distance is larger than the Burgers vector length. The time required for two neighbouring kinks to annihilate can be measured up by \( t = \lambda_j / v_j \). During this time the dislocation has to cover \( \Delta l = v\lambda_j / v_j > b \). Hence we have (2.24). When (2.24) is true we get [Pop90]:

\[
\frac{\partial C_{Vj}}{\partial \gamma} = \frac{1}{2} p_j^{1/2} \xi^{1/2} bN_d^{1/2} (\nu / v_j)^{1/2}.
\] (2.25)
2. Plastic deformation

Here $C_{1,V}$ is the relative concentration of interstitial atoms ($I$) and vacancies ($V$), $P_t$ is density of tree dislocations that create kinks on a screw dislocation to the general density of dislocations. Unfortunately, the dependence of $\frac{\partial C_{1,V}}{\partial \gamma}$ from parameters that characterize experimental conditions (from strain rate $\varepsilon$ and temperature of the sample) cannot be valued by (2.23) and (2.25). It’s quite difficult to use these formulae to make practical evaluations. To determine the formation of point defects velocity the authors of [Mec80] and [Mil93, Mil94] proceeded from energetic considerations. It is assumed that the formation energy $E_k$ of the type $k$ point defect (a crystal lattice vacancy or an interstitial) is proportional to $\mu b^3$. The starting point in the Estrin-Mecking model is the fact that a certain amount of work done during the deformation of a crystal is used to form point defects. The equation for the formation of point defects velocity $P$ is noted as:

$$P = \frac{\chi \tau \varepsilon}{\alpha_1 \mu b^3}.$$  \hspace{1cm} (2.26)

Here $\alpha_1$ is a constant, $\chi$ is the portion of the overall work done that is spent on formation of point defects. This formula is preferential to (2.23) and (2.25) when making practical evaluations because here we have the apparent dependence of $P$ from $\tau$ and $\varepsilon$.

**Point defects annihilation velocity**

The already formed crystal lattice vacancies and interstice atoms can react with impurity atoms during the diffusion process thus constituting new point defects. These processes together with vacancies and interstitials generation processes determine the value of concentration of vacancies and interstitials. Generation velocities have been treated before in the present paper. Here let’s consider interaction velocities of vacancies and interstitials with impurity atoms.

A number of scientific works were dedicated to studies of velocities of diffusively controlled reactions in a solid state. These are reactions velocities of which are limited by the diffusion process of the reaction components. Notable are [Vin72a, Vin72b, Vin75], [Koz81] works. The [Koz81] findings are given below because this paper takes account of the other mentioned works’ findings. Velocities of diffusively controlled reactions in semiconductor materials between two components (at least one of them must be mobile) were examined in [Koz81]. The coulomb interaction between the components having charge condition spectrum in the semiconductor band – gap was taken consideration of in the work. The following was obtained for velocity of reactions:

$$\frac{1}{t} = \frac{1}{4\pi} \int_0^{r_i} \frac{\exp \left[ \frac{U(r)}{k_B T} \right]}{D(r) r^2} dr.$$ \hspace{1cm} (2.27)

$$U(r) = \sum_{i,j} U_{ij}(r) f_{ij}(r).$$ \hspace{1cm} (2.28)
2.5 Formation of point defects

\[ D(r) = \sum_{i,j} (D_i + D_j) f_{ij}(r). \]  
\[ f_{ij}(r) = F_i f_{i,j-1}(r) + \Phi_j f_{i-1,j}. \]  
\[ \sum f_{ij}(r) = 1. \]

Here \( t \) is the reaction velocity; \( r \) is the distance between the reacting defects; \( r_0 \) is the reaction radius (the distance at which the components must come closer so that the reaction will run and a new defect will form); \( D_i \) is the diffusion coefficient of the reaction first component in the charge condition \( i \); \( D_j \) is the diffusion coefficient of the reaction second component in the charge condition \( j \); \( U_{ij} \) is the coulomb interaction energy between the reacting components having \( i e \) and \( j e \) charges (\( e \) is the electron charge); \( f_{ij}(r) \) – is the probability that at the distance \( r \) between the interacting components one of them is found in the charge condition \( i \) and the other one is found in the charge condition \( j \); \( U(r) \) – is the average energy of electrostatic interaction between the reaction components removed on the distance \( r \); \( D(r) \) – is the average total diffusion coefficient of the two components removed on the distance \( r \); \( F_i \) and \( \Phi_j \) are the functions of the distance \( r \) between the reaction components, electrons and holes concentration, crystal temperature, energy levels in the semiconductor band – gap (of the charge condition spectrum) for each reaction component; all charge conditions of the reaction components are summed.

Qualitatively the reaction process can be explained the following way. The reaction components are found initially at the distance \( r \gg r_0 \). This distance is taken equal to the infinity, so in the (2.27) integral the integration upper limit is infinity. The components have \( i e \) and \( j e \) charges. For instance, if the first component is in the single negative charge condition and the second one is in the single positive charge condition, then \( i = 1 \) and \( j = -1 \). As defects come closer, it may turn out that at a certain distance \( r \) between the components the condition having different \( i \) and \( j \) values will be more beneficial from the energetic point of view. If the components continue their approaching motion the values of \( i \) and \( j \) can again change. So \( f_{ij} \) depends on the distance \( r \) and the components’ mutual diffusion coefficient and the interaction energy between them is characterized by average values (see (2.28) and (2.29) equations). In case the reaction components are neutral (when \( U_{ij} = 0 \) and, hence, \( U = 0 \) and \( D = \text{const} \) ) we get the common formula for the velocity of diffusionally controlled (diffusionally limited) reaction:

\[ t = 4\pi(D_1 + D_2)r_0. \]

Here \( D_1 \), \( D_2 \) are the diffusion coefficients of the first and the second components accordingly. From (2.27) results that if one component attracts the other (\( U_{ij} < 0 \) and hence \( U < 0 \) ) the reaction velocity value is higher then the similar magnitude in case the components are neutral. And vice versa: if one component repulsion the other (\( U_{ij} > 0 \) and hence \( U > 0 \) ) the reaction velocity value is less then the similar magnitude in case the components are neutral. In other words, attraction between the components speeds the reaction up, while repulsion slows it.
2. Plastic deformation

The discussed reaction velocity model can be applied to the case of formation of defects located in the field of dislocation mechanical stress. First, however, the influence of dislocation stress on energy levels values and on activation energies of the reaction components migration in various charge conditions must be considered in the \((2.27 \pm 2.30)\) equations. The model as it is cannot be applied to practical estimation of reactions velocity values that run in the field of dislocation mechanical stress. In [Mec80] a model of annihilation velocity of vacancies that are situated in the dislocation neighbourhood was proposed:

\[ A = \beta (C - C_0). \] (2.33)

Here \( A \) is the annihilation velocity, \( C - \) is the relative concentration of vacancies, \( C_0 \) is the relative concentration of vacancies (away from dislocations) in the crystal thermodynamic equilibrium. The constant \( \beta \) is calculated by the following expression [Mec80]:

\[ \beta = \frac{D_v}{C_0 b^\gamma \lambda^\gamma}. \] (2.34)

Here \( D_v \) is the vacancy diffusion coefficient, \( \lambda \) – is the average distance between annihilation centers.

**Stationary concentration**

The stationary concentration of vacancies is reached when vacancies emission velocity is equal to annihilation velocity. In the framework of the Estrin-Mecking model [Mec80], the stationary concentration can be determined from \( P = A \) assuming that only dislocations constitute annihilation centers (see \((2.26)\) and \((2.34)\) equations). Then:

\[ \Delta C = C_0 \frac{\chi^2 \tau \varepsilon}{\alpha_1 D_v \mu} \] (2.35)

Here \( \Delta C = C - C_0 \) is the surplus relative concentration of vacancies.

Let’s multiply the numerator and the denominator in the right part of the equation by the density of dislocations \( N_d \). After that, let’s express the density of dislocations in the denominator with help of \((2.10)\): \( N_d = (\tau / \alpha mb)^2 \). Then the following expression is obtained for \( \Delta C \):

\[ \Delta C = C_0 \frac{\chi^2 \tau \varepsilon}{\alpha_1 D_v} \left( \frac{\mu}{\tau} \right)^\gamma \sim \lambda^2 N_d \frac{\varepsilon}{\tau}. \] (2.36)

From \((2.15)\) results that \( \tau \sim \nu^{1/m} \) (\( m \) is the rate of stress indicator) and from \((2.11)\) and \( \gamma = \varepsilon / m \) results that \( \nu = \varepsilon / (m b N_{d_a}) \). Thus \( \tau \sim (\varepsilon / N_{d_a})^{1/m} \). Finally we come to:
2.6 Dislocations in the diamond structure

\[ \Delta C \sim \lambda^2 N_d N_{\text{d},m} \varepsilon^{\text{\text{\(\frac{1}{m}\)}}}. \]  

(2.37)

Here \( N_d \) – is the density of mobile dislocations. From (2.35) results that at a constant deformation velocity

\[ \Delta C \sim \tau \lambda^2. \]  

(2.38)

From (2.37) results that, given that the density of dislocations has a constant value, the dependence of \( \Delta C \) from \( \dot{\varepsilon} \) is:

\[ \Delta C \sim \dot{\varepsilon}^{\text{\text{\(\frac{1}{m}\)}}}. \]  

(2.39)

2.6 Dislocations in the diamond structure

The diamond lattice can be viewed as superposition of two face-centered cubic lattices displaced with respect to each other in the direction of the volume diagonal for \( \frac{1}{4} \) of its length. If, in this case, the diamond lattice contains atoms of two different types, then we deal with zinc blende structure. The GaAs lattice is the blende type lattice.

Plastic deformation occurs as a result of shear in a certain crystallographic plane in a certain direction in that plane. The combination of the slip plane and the slip direction in that plane is called the slip system. Slip plane is a plane with the densest packing of atoms (because distances between such planes are maximal) and slip direction is a direction for which distances between adjacent atoms centers in a slip plane are minimal. The diamond lattice does not possess all the properties of cube symmetry. However, by its macroscopic parameters a diamond crystal has cube symmetry [Ans78]. That’s why here let’s consider slip systems and as an example take face-centered lattice. Figure 2.9 represents the plane (111) and all the possible slip directions in it. Slip planes are planes NAC, MDB, EDB. So we have four systems of physically equivalent planes: \{111\} slip systems each having three slip directions. Thus, in the general case a face-centered lattice can have up to twelve slip systems.

Fig. 2.9: Slip systems and slip directions scheme. The ALC plane is the (111) slip plane. QK, KG, GQ directions are the slip directions in the ALC plane (atoms in the cube vertices and in the middle of the AEND, DNMC, ELMN facets are not shown on the scheme).
The real quantity of slip systems under specific experimental conditions may be less than their maximum theoretically possible number. This can be explained in the following way. It has been discussed above that the relation between the external stress (\( \tau \)) and the tangential stress (\( \sigma \)) affecting a dislocation (and causing its movement) in the slip plane is expressed like: \( \sigma = m_\tau \). The Schmidt factor (\( m_\tau \)) value is determined by direction of the unit vector which is perpendicular to the slip plane, the Burgers vector plane and by stretching direction. If we have such stretching direction that for any slip systems \( m_\tau = 0 \) then in this slip system \( \sigma = 0 \) and there’s no dislocation glide in it. Hence, the quantity of real slip systems will be less than their theoretically possible quantity. For example, in a diamond lattice there are 8 slip systems for the [100] stretching direction (four slip plane systems and two slip directions in slip planes), four slip systems for the [110] direction. Since the GaAs lattice is nothing more nor less than superposition of two face-centered lattices with different atoms (Ga and As), two types of dislocations exist there: 60°-degree \( \alpha \)-dislocations (in Ga atoms lattice) and 60° -degree \( \beta \)-dislocations (in As atoms lattice) [Mat74]. There are two possible positions of slip planes in the diamond structure [Ale68a]: slip planes can be situated between plane couples, or, in the other case, between planes within one couple. That makes the diamond structure so peculiar. In the slip regime complete dislocations can dissociate for two partial dislocations. As a result, the atomic structure deformation reduces. The greater part of dislocations in the diamond lattice decomposes (dissociates) for partial dislocations. During dissociated screw dislocation cross slip dislocation line segments that stay in the cross slip plane constitute 60° – degree dislocations. If two screw dislocations cross we have the similar situation: 60° – degree jumps appear on both screw dislocations after their crossing.

Conclusions

Proceeding from the above-explained material, we may conclude that two principal factors – changing in the density of dislocations (both mobile and immobile) and their velocity – determine the nature of experimentally observed deformation curves during stretching of a sample. In all probability we can state that nowadays there are clear and understandable notion concerning reasons that lead to changes in the values of the discussed dislocation characteristics during plastic deformation of crystals. To be more precise, impurity, point defects and their clusters, and other dislocations influence the movement of dislocations velocity. The fact that defects impeding the movement of dislocations appear during the proper movement of dislocations (in the process of kinks climb in screw dislocations) complicates studies of the matter. The concentration of dislocations changes as a result of movement of dislocations as well (double cross slip of dislocations). It’s comprehensible that to be able to quantitatively interpret experimental data and to make prognosis concerning the behaviour of deformation curves under these or that experimental conditions (like sample temperature and stretching velocity) we need equations that connect together concentration of defects, macroscopic parameters (\( \tau, \varepsilon, \dot{\varepsilon}, T \)), and parameters that characterize the movement of dislocations (\( \nu, N_d, N_{da} \)). Such equations are those discussed earlier in this work: (2.10), (2.11), (2.15 ÷ 2.17), (2.19 ÷ 2.21), (2.38), (2.39). Unfortunately, these do not form a closed system. For example, there’s no evident connection between concentration of defects and movement of dislocations velocity. Later, if in the (2.15) equation for movement of dislocations velocity we substitute \( \tau \) for \( \sigma \) using \( \sigma = m_\tau \) and then make use of (2.10), we’ll arrive at:
\[ \nu \sim N_d^{m/2} \exp(-U/k_B T). \]

This means that \( \nu \) grows as \( N_d \) (\( m > 0 \)) grows. In reality, however, \( \nu \) should decrease with the increases of \( N_d \) because dislocations are obstacles for movement of mobile dislocations. Thus, the movement of dislocations activation energy \( U \) should rise as density of dislocation \( N_d \) increase. The activation energy should rise if concentration of defects increases, too, because they are obstacles for the movement of dislocations. In this situation the use of \( V_A, m, U \) parameters for description of the movement of dislocations is justified only if these parameters depend very little on those that characterize experimental conditions \( (\epsilon, T) \).

[Hub98] provides us with findings of systematic experimental research of GaAs deformation curves at the crystal temperature range of \( = T(400 \div 800)^0C \) and strain rate of \( \dot{\epsilon} \approx (10^{-5} \div 10^{-3}) s^{-1} \). In particular, for an undoped GaAs during stretching the following results were obtained (the crystal was stretched in the [110] direction) [Hub98]: in the range of \( \dot{\epsilon} \approx (10^{-5} \div 10^{-4}) s^{-1} \) at \( T = 400^0C, 500^0C,600^0C,800^0C \) the research gave \( 2+m = 4.5 \pm 0.6, 3.0 \pm 0.4, 3.7 \pm 0.3, 3.9 \pm 0.3 \) values accordingly; \( U/(2+m) \approx 0.44 eV \) \( (\epsilon = 8 \cdot 10^{-5} s^{-1}) \). Thus, very weak relation between rate of stress indicator \( (m) \) value and the value of the activation energy \( (U) \) of the movement of dislocations was determined. These experimental data prove that the use of \( m \) and \( U \) parameters and, perhaps, \( V_A \) is justified. It’s sad, however, that in [Hub98] and in other works there are no data for the values of \( V_A \) in a vast range of \( \epsilon \) and \( T \) parameters.

[Hub98] also proved that there is influence from part of doped impurity on the value of \( U \), i.e. defects influence significantly the processes of movement of dislocations. It was stated above, however, that the system of expressions describing processes of movement of dislocations doesn’t directly connect the values of \( U \) and concentration of defects. This fact does not allow us to determine the value of concentration of defects (leaving alone their type) directly from the experimental data obtained for crystal deformation. That’s why to determine the concentration and type of defects in deformed crystal scientists use additional experimental methods. The PAS method (that is further treated in detail) was used in the present paper. Using experimental data obtained by the PAS method it will be possible to make a conclusion concerning the applicability of (2.38) and (2.39) when determining functional relation between concentration of defects and \( \tau \) with \( \epsilon \).