Nuclear Materials and Energy 000 (2016) 1–6

Contents lists available at ScienceDirect

Nuclear Materials and Energy

journal homepage: www.elsevier.com/locate/nme

Low-temperature embrittlement and fracture of metals with different crystal lattices – Dislocation mechanisms

V.M. Chernov[a,b,*], B.K. Kardashev[c], K.A. Moroz[a]

[a] A.A. Bochvar High-technology Research Institute of Inorganic Materials (JSC "VNIINM"), 123098 Moscow, Russia
[b] National Research Nuclear University "MEPhI", Moscow, Russia
[c] Ioffe Institute, 194021 Saint-Petersburg, Russia

ARTICLE INFO

Article history:
Received 19 October 2015
Revised 22 December 2015
Accepted 2 February 2016
Available online xxx

Keywords:
Metals
Low-temperature embrittlement
Brittle fracture
Damaging irradiation
Dislocation models

ABSTRACT

The state of a low-temperature embrittlement (cold brittleness) and dislocation mechanisms for formation of the temperature of a ductile-brittle transition and brittle fracture of metals (mono- and polycrystals) with various crystal lattices (BCC, FCC, HCP) are considered. The conditions for their formation connected with a stress-deformed state and strength (low temperature yield strength) as well as the fracture breaking stress and mobility of dislocations in the top of a crack of the fractured metal are determined. These conditions can be met for BCC and some HCP metals in the initial state (without irradiation) and after a low-temperature damaging (neutron) irradiation. These conditions are not met for FCC and many HCP metals. In the process of the damaging (neutron) irradiation such conditions are not met also and the state of low-temperature embrittlement of metals is absent (suppressed) due to arising various radiation dynamic processes, which increase the mobility of dislocations and worsen the strength characteristics.

1. Introduction

At rather low temperatures (around and below the room temperature) in metals (mono- and polycrystals) with different crystal lattices (with body-centered-cubic lattice – BCC, some with hexagonal-close-packed lattice – HCP) the state of low-temperature embrittlement (LTE) or cold brittleness arises, characterized by the temperature (a narrow interval of temperatures) of the ductile-brittle transition $\mathbf{T_{dbtt}}$. Below this temperature a brittle fracture of a metal (product) is possible [1–8]. In metals with face-centered-cubic lattice (FCC) and in many metals with HCP crystal lattice the LTE state does not arise. The physical-mechanical mechanisms for formation of the LTE state and brittle fracture of metals in different states are still insufficiently defined. The researches in this field are conducted on the basis of the methods of the fracture mechanics, theories of dislocations, internal stresses and specific features of the interatomic interactions in metals of different crystallographic classes [1–18].

The work is devoted to discussion of the dislocation mechanisms for formation of the $\mathbf{T_{dbtt}}$ and brittle fracture of metals in the LTE state. It determines the conditions of their occurrence in metals (mono- and polycrystals) in different states (initial, neutron irradiated and during neutron irradiation) with different crystal lattices (BCC, FCC, HCP). On the basis of these conditions and by the nondestructive ultrasonic internal friction technique, the temperature conditions were determined for the LTE state ($\mathbf{T < T_{dbtt}}$) in BCC metals (reduced activation ferritic-martensitic steel Rusfer-EK-181: Fe–12Cr–2W–V–Ta–B, low activation alloy V–4Ti–4Cr) and the reasons for its absence in FCC metals (austenitic steel EK-164: Fe–16Cr–19Ni–2Mo–Nb–Ti–B).

2. Nucleation and growth of cracks

In metals, under loading by an uniaxial tensile stress $\boldsymbol{\sigma}$, a stress-deformed state appears characterized in the crack plane by a normal $\boldsymbol{\sigma_{22}}$ and shear $\boldsymbol{\sigma_{12}}$ stress components. Such a state determines the stiffness of the metal loading $\boldsymbol{\lambda=\sigma_{12}/\sigma_{22}}$. The stress components in the crack plane are determined in its top (concentrator of stress) by expression [6,16] $\boldsymbol{\sigma_{ij}(r)=\sigma_{ij}(1+(L/2r)^{1/2})}$, where $\mathbf{2L}$ is the length of a crack and $\mathbf{r}$ is the radius from the crack edge. In the top of the crack the components $\boldsymbol{\sigma_{22}}$ (mode I) and $\boldsymbol{\sigma_{12}}$ (mode II) can locally reach the theoretical values of the metal fracture stress on rupture $\boldsymbol{\sigma_{th}}$ (normal stress) and on shift $\boldsymbol{\sigma_{sh}}$ (shear stress). For opening of a crack with formation of a free surface in a metal, the normal critical stress in the top of the crack (mode I) is necessary

* Corresoponding author at: A.A. Bochvar High-technology Research Institute of Inorganic Materials (JSC "VNIINM"), 123098 Moscow, Russia. Tel.: +7 499 1903605.
E-mail address: vmchernov@bochvar.ru (V.M. Chernov).

http://dx.doi.org/10.1016/j.nme.2016.02.002
2352-1791/

Please cite this article as: V.M. Chernov et al., Low-temperature embrittlement and fracture of metals with different crystal lattices – Dislocation mechanisms, Nuclear Materials and Energy (2016), http://dx.doi.org/10.1016/j.nme.2016.02.002

equal to (or exceeding) $\boldsymbol{\sigma_{th} = (\gamma E/a)^{1/2}}$ [19]. Here **E** is the modulus of elasticity, $\boldsymbol{\gamma \approx 0.1Ea}$ [14] is the specific surface energy, **a** is the size of the crystal lattice parameter in the fracture direction. Close to unity coefficients, depending on the directions in the crystal lattice (for all crystal directions the lattice parameter is determined by size **a)** and type of the stress-deformed state of metal are omitted. The practically important plane-deformed state (high stiffness) of the ruptured metal determined by its volume (thickness) will be considered.

From the condition $\boldsymbol{\sigma_{22}(a) = \sigma_{th} \approx 0.32E}$, the critical value of a normal stress for crack opening $\boldsymbol{\sigma_{cr} = [2E\gamma/L]^{1/2} \approx 0.45E(a/L)^{1/2}}$ follows. It coincides by the value with Griffith's critical stress $\boldsymbol{\sigma_{Gr} = [2E\gamma/L]^{1/2}}$ for a crack opening (power criterion of Griffith [20]), obtained within the framework of the linear mechanics of fracture and connecting the density of the elastic energy of the loaded metal with its specific surface energy. Under action of the shear stress component in the top of the crack (mode II), a local zone of plasticity appears (of nucleation and movement of the edge dislocations) with $\mathbf{r_c}$ radius, in which the shear stress component exceeds the yield strength $\boldsymbol{\sigma_{ys}}$. The value of $\boldsymbol{r_c = (L/2)(\sigma_{12}/\sigma_{ys})^2}$ is determined from condition $\boldsymbol{\sigma_{12}(r) \geq \sigma(r_c) = \sigma_{ys}}$. The value of the theoretical shear strength $\boldsymbol{\sigma_{sh} = G/2\pi}$ [21] is determined in the top of the crack by stress $\boldsymbol{\sigma_{sh}(a) = 0.23G(a/L)^{1/2}}$ (**G** is the shear modulus). In the plasticity zone ($\mathbf{r < r_c}$), the deformation begins ensured by the nucleation (Frank-Read sources) and mobility of dislocations. The intragranular crack opening (rupture of interatomic bonds) under action of $\boldsymbol{\sigma_{cr}}$ with formation of a free surface should be ensured by a plastic (shear) deformation in the zone of plasticity of the crack. If the work for the plastic deformation (dislocation movement) exceeds the work for formation of a free surface, the crack becomes "viscous" (does not spread). If the work for the plastic deformation is less than the work for formation of a free surface, there is an opening and increase of the crack length. Spreading of the front of the crack cannot take the lead over the dislocation, moving in its plasticity zone. The speed of the leading dislocation defines the speed of the crack front. The plastic deformation (movement of dislocations) always precedes the growth of a crack and brittle fracture [22].

The basic models of cracks, their nucleation and spreading in metals (products) are based on dislocation conceptions [1,3,6,9–15]. The relief of a field of the internal stresses in the dislocation slipping plane (relief resistance to its movement) defines the mobility (slipping) of a dislocation (metal strength). The relief of the stresses is formed by a crystal lattice (slipping system, Peierls barrier) and its defects as heterogeneities and sources of the internal stresses (solid solutions, clusters, phases, dislocations, etc.).

The nucleation and prior-to-macroscopic crack growth (the first stage of fracture) is characterized by the nucleation and thermoactivated mobility (slipping) of the dislocations. The dislocation models of the process include a typical dimensional parameter, determining the distances between the structure elements in polycrystals (grain boundaries, substructure, phase particles, twins). This parameter determines the size of the plane dislocation accumulation and the number of dislocations in such an accumulation, pressed by a shear stress to the stopping structure barrier (hindered shift). The strengthened stress in the head part of a dislocation aggregate is a source of nucleation and prior-to-macroscopic crack growth. In the monocrystals, the characteristic mechanisms of nucleation and prior-to-macroscopic crack growth are a substructure and dislocation reactions in the crossed slipping planes. At the stage of intragranular spreading of the nucleated macrocrack (the second stage of fracture), the mechanism, which ensures its opening, is the nucleation and mobility of dislocations in the plasticity zone of the crack.

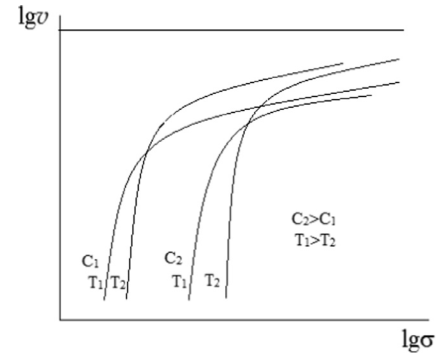

Fig. 1. Influence of the temperature ($T_1 > T_2$) and solid solution concentration of defects ($\mathbf{c_2 > c_1}$) on the stress dependence of the dislocation velocity v($\boldsymbol{\sigma}$) in thermoactivated (low stress) and dynamic (high stress) areas of movement (scheme). Horizontal line is the sound velocity (asymptotic level for the dislocation velocity).

3. Mobility of dislocations

Depending on the applied stress (shear component in the slip plane) and temperature, the mobility of dislocation is characterized by a thermoactivated (small stresses) and dynamic (big stresses) areas (Fig. 1) [12,13,15,23–27].

In the thermoactivated area, one can observe dependences of the dislocation speed on the stress (power law with exponent significantly greater than unity) and on the crystal structure (slip system, crystal relief, concentration of defects). The higher is the temperature, the higher is the dislocation speed, and the power exponent increases with an increase of the concentration of the defects. The structural changes of the conditions of dislocation movement, increasing the strength (low-temperature yield strength) of a metal, shift the thermoactivated area of mobility to higher stresses with an increase of the stress in the beginning of the dynamic area. In the dynamic area (the beginning is evaluated by the stress of $\boldsymbol{\sigma_{dyn} \approx 10\sigma_{ys}}$ [23–25]), one can observe a linear dependence of the speed on the stress, a relatively weak dependence on the state of a metal and an inverse dependence of the speed of dislocation on the temperature (the higher is the temperature, the lower is the speed). For FCC metals the thermoactivated and dynamic areas of the dislocation speed are significantly shifted towards the stress reduction in comparison with BCC metals. The speed of dislocations **v** in the dynamic area is determined by expression $\boldsymbol{v = (b \cdot \sigma)/B}$, where **b** is the size of Burgers vector of dislocation, $\boldsymbol{\sigma}$ is the shear stress, **B** is the coefficient of viscous drag of dislocations. The size **B** depends on the temperature (goes down with lowering of the temperature) and determined by the interaction of a dislocation with the phonons, electrons and magnons.

4. Conditions for a low-temperature embrittlement and brittle fracture of metals

The following conditions determining the formation of the LTE state with a possibility of fragile fracture of a metal should be met for opening and spreading of a crack at a stretching stress $\boldsymbol{\sigma}$:

(1) The stress-deformed state, forming the values of normal $\boldsymbol{\sigma_{22}}$ and shear $\boldsymbol{\sigma_{12}}$ components of the applied stress in the crack plane, should be stiffness enough (plain-deformed state) at large values of $\boldsymbol{\sigma_{12}}$ and $\boldsymbol{\sigma_{22}}$ components. The stage of nucleation and prior-to-microscopic crack growth (the first stage of fracture) is defined by the shear stress component, and its duration depends on the speed of nucleation and thermoactivated mobility of dislocations and strength (low-temperature yield strength). The second stage of fracture is

Please cite this article as: V.M. Chernov et al., Low-temperature embrittlement and fracture of metals with different crystal lattices – Dislocation mechanisms, Nuclear Materials and Energy (2016), http://dx.doi.org/10.1016/j.nme.2016.02.002

defined by the joint action of the normal (crack opening – rupture of the interatomic bonds) and shear (crack lengthening) components of the stress, strengthened in the crack top (stress concentrator).

(2) Normal stress at the front of the macroscopic crack (mode **I**) should exceed the critical metal fracture breaking stress $\boldsymbol{\sigma_{cr}} = [2E\gamma/L]^{1/2} \approx 0.45E(a/L)^{1/2}$, ensuring opening of the crack with formation of a free surface. This condition coincides with the Griffith's criterion for crack opening.

(3) The shear stress in the plasticity zone of a crack (mode **II**) $\boldsymbol{\sigma_{dyn}}$ should ensure nucleation and mobility of a dislocation in the dynamic area ($\sigma_{dyn} \approx 10\sigma_{ys}$). The relief of resistance to the dislocation movement and the temperature should ensure high starting stress for the nucleation of a dislocation and achievement of the dynamic area of its mobility. The speed of spreading (lengthening) of a crack is defined by the speed of a dislocation in the dynamic area in the plasticity zone of the crack. The crack front cannot leave behind the leading dislocation, moving in front of it in the plasticity zone. The distance between the moving crack top and the leading dislocation practically remains constant, but their speeds can increase with an increase of the crack length.

(4) Coefficient of viscous drag of dislocation **B,** which determines the speed and work during the movement of the dislocation in the dynamic area in the plasticity zone of the crack, should be less than critical value $\mathbf{B_{cr}}$ (mode of "dry" friction), in order to ensure a low level of the dynamic drag of the dislocation with a small work for its travel ($\mathbf{B_{cr}v} < 2\gamma$) in front of the opening crack. Condition $\mathbf{B} < \mathbf{B_{cr}}$ is met at rather low temperatures and it defines the possible temperature area of the LTE state of metals ($\mathbf{T} < \mathbf{T_{dbtt}}$).

The first two conditions can be fulfilled in all metals irrespective to their crystallographic class. To meet the third condition it is necessary to ensure the achievement of the dynamic area of the dislocation mobility. a rather high low-temperature yield stress defined by a high resistance to the dislocation movement. This requirement is executed by a high Peierls barrier and high energy of elastic interaction of the dislocation with defects. This third condition is met by the directions of dislocation slipping (Burgers vectors) along the crystal axes of symmetry of an odd order ($\langle 111\rangle$ for BCC and $< 11\bar{2}3 >$ for HCP crystals). Practically all BCC metals and alloys belong to the class with the odd axes of symmetry for the slipping direction $<111>$. Only with certain HCP metals (for example, beryllium, zirconium, zinc) the odd axes of symmetry for slipping directions can be realized ($< 11\bar{2}3 >$). All FCC and many HCP metals have slipping directions with even axes of symmetry ($\langle 110\rangle$ and $< 11\bar{2}0 >$), characterized by a small Peierls barrier and weak elastic interaction of the dislocations with point defects (rather low-temperature yield strength).

The fourth condition determines the work for the viscous dislocation movement in the dynamic area in the plasticity zone of a crack defined by the coefficient of viscous drag **B** and speed of the dislocation. In case of a bigger viscosity of movement (bigger work for dislocation movement, $\mathbf{B_{cr}v} > 2\gamma$), the crack does not grow.

In nanocrystalline metals (the grain size is less than 100 nm), the crack development is weakened (crack growth resistance raises) because of the reduction of the sizes of the flat aggregations of dislocations and the reduction of the quantities of dislocations in them, and stress relaxation in the top of the growing (prior-to-macroscopic) crack during its interaction with the grain border. In the highly texturized metals (a small axis of grain less than 100 nm) the fracture strength will also raise, as well as its anisotropy, depending on the mutual orientation of the crack plane and texture.

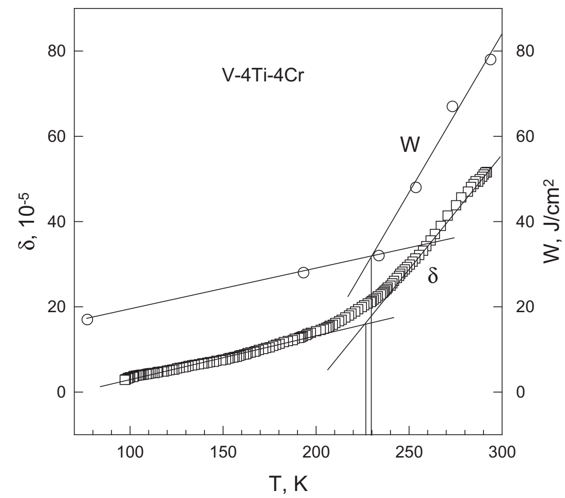

Fig. 2. Temperature dependences of the impact toughness W and the logarithmic decrement δ_i for the V-4Ti-4Cr alloy after plastic bending.

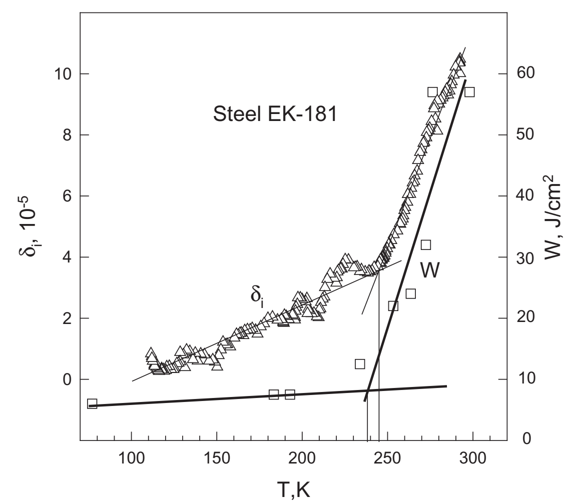

Fig. 3. Temperature dependences of the impact toughness W and the logarithmic decrement δ_i for the RAFMS Rusfer- EK-181 of high yield point $\boldsymbol{\sigma}_b$ after plastic bending.

5. Cold brittleness of a metal and internal friction technique

Amplitude-independent internal friction or logarithmic decrement δ_i, measured at vibration frequency of about 100 kHz (ultrasonic) is an effective technique for the investigation of the mobility of dislocations and estimation of the coefficient of their viscous drag in metals [28,29]. At 100–300 K, the BCC (V–4Ti–4Cr alloy and RAFMS RUSFER-EK-181 steel of various technologies with different yield strength σ_b) and FCC (EK-164 austenitic steel) metals developed as structural materials for the nuclear and thermonuclear power reactors [7,30,31] were investigated (Figs. 2–5). The results of internal friction experiments and their detailed analysis are published in [29]. We present here only the most important facts.

Fresh dislocations were introduced [29] in samples by a preliminary small plastic bending before the beginning of measurements. It is well known that the bending introduces into the sample an excess of edge dislocations of one mechanical sign. Only for bent samples, one can notice a good correlation (Figs. 2 and 3) of the temperature dependencies of the impact toughness W and the internal friction (logarithmic decrement δ_i). It means that edge dislocations mobility (viscous dragging of edge dislocations) is responsible for the low temperature embrittlement.

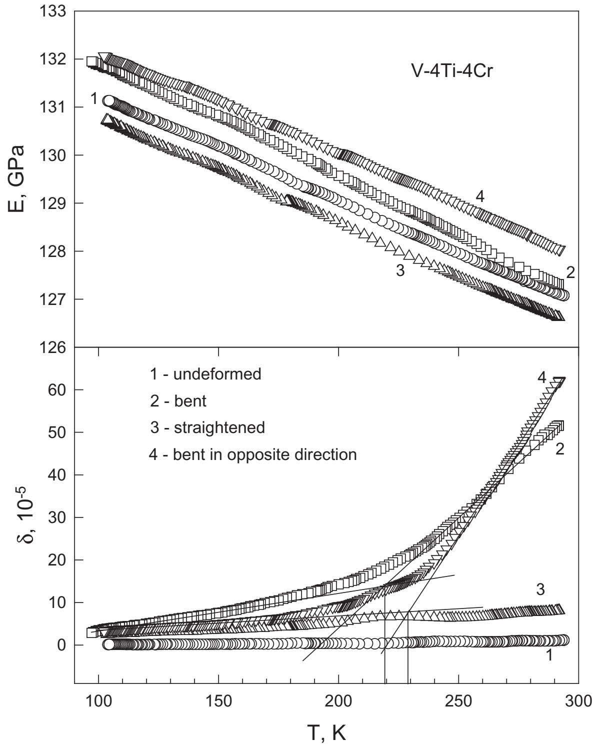

Fig. 4. Temperature dependences of the Young modulus E_i and the logarithmic decrement δ_i for the alloy V-4Ti-4Cr.

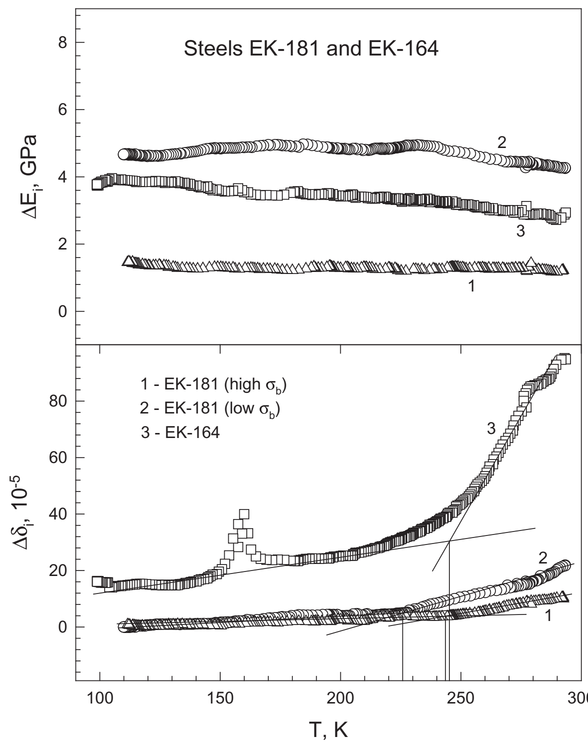

Fig. 5. Temperature dependences of the increments of the Young modulus $\mathbf{\Delta E}_\iota$ and the logarithmic decrement $\mathbf{\Delta\delta}_\iota$ for the RAFMS Rusfer-EK-181 of high (1) and low (2) yield points σ_b and for the austenitic EK-164 steel (3) due to plastic bending.

After a straightening of the specimen, the decrement decreases considerably and the curvature of the $\delta_i(T)$ dependence near $\mathbf{T_{dbtt}}$ disappears almost completely (Fig. 4). Note that the straightening introduces into the sample almost equal quantity of dislocations of opposite mechanical sign. It leads to annihilation of edge dislocations and the level of internal friction decreases. If the sample is bent in opposite direction there are in the sample an excess of edge dislocations of another mechanical sign.

The temperature of the sharp increase of the amplitude-independent internal friction in BCC metals is close to $\mathbf{T_{dbtt}}$ determined from the impact experiments (Figs. 2 and 3). This temperature depends (Fig. 5) on the structural-phase state of a metal (low temperature yield stress σ_b). In BCC metals at temperatures below $\mathbf{T_{dbtt}}$, there is a lower, compared to FCC metals, level of the internal friction ("dry" friction). It is clearly seen in Fig. 5. The internal friction in FCC metal (steel EK-164) is considerably higher (an order of magnitude) than in BCC steel EK-181. High viscosity of the dislocation movement in FCC metal demands high energy consumptions. It ensures a relaxation of the elastic stresses at the front of the crack. So, the crack opening stops and brittle fracture does not occur.

6. Dislocation mobility and cold brittleness of metals under damaging (neutron) irradiation

Low-temperature (up to ~700 K) damaging (neutron) irradiation of the metals showing cold brittleness without irradiation (BCC, etc.) results in an increase of the initial (before irradiation) values of temperature $\mathbf{T_{dbtt}}$ in the irradiated metals. It is the phenomenon of the low-temperature radiation embrittlement – LTRE (depending on the irradiation temperature and dose) [6,7,32–37]. Such an increase of $\mathbf{T_{dbtt}}$ in the irradiated metals (the strongest at small doses) is defined by an additional radiation hardening after an irradiation. At high irradiation temperatures (over ~700 K), an increase of $\mathbf{T_{dbtt}}$ in the irradiated metals is practically not observed. This effect makes the basis for the methods of annealing (over ~800 K) of the products for nuclear technologies irradiated at low temperatures to recover the initial (before irradiation) functional properties [32,36,37].

In the works studying the LTRE state and its consequences (rise in $\mathbf{T_{dbtt}}$), it is assumed that the LTRE state defined during the tests of the irradiated samples (without a neutron irradiation) remains in a metal during the irradiation also. However, there are no grounds for such a conclusion. The properties of metals during and after irradiation differ considerably. Researches (theoretical, experimental, modeling) of the mechanical properties and dislocation mobility show essential qualitative and quantitative distinctions before, in-the-process and after irradiation of metals [23–29,32–52]. In the irradiated (additionally strengthened) metals, it is the same mechanisms of mobility of dislocations, nucleation and growth of cracks that are realized but in the irradiation-changed (depends on the irradiation temperature and dose) structure of the samples (concentration of defects and their state, radiation-induced segregations, etc.). Research of the mechanical properties and dislocation mobility in the irradiated samples demonstrates considerable changes only in the thermoactivated areas of the dislocation mobility. There an increase of the speed level shift in the direction of a higher stress with an increase of the dose of the low-temperature irradiation is observed. In the dynamic areas of dislocation mobility, the character of dislocation movement in the irradiated metals changes insignificantly. With an increase of the irradiation dose, the level of the stress raises for beginning of the dynamic area in the irradiated samples. This corresponds to an increase of the temperature $\mathbf{T_{dbtt}}$.

The damaging (neutron) irradiation in metals causes dynamic processes determined by the radiation defects (generation and loss, radiation "jolting", long-range action) and accompanied by emergence of the dynamic phenomena (elastic waves of shifts and stresses). They can influence considerably the mobility of the dislocations (increase) and mechanical properties of metals (weakening) under irradiation, strengthening with lowering of the temperature of irradiation. The dynamic waves of displacement (stress) abate with the distance **r** more slowly ($\sim$**1/r**), than the static ones ($\sim \mathbf{1/r^2}$), which defines the long-range influence of the dynamic displacements on the evolution (mobility) of the dislocations and on the functional properties of metals under irradiation. As a whole, it was pointed out that under the damaging (neutron) radiation influences: (1) in the metals, irradiated at low temperatures, the dislocation mobility decreased (the yield strength raised), (2) during irradiation, the dislocation mobility raised (the yield strength went down), (3) elastic modules decreased under irradiation, (4) during irradiation the creep rate increased, and the lower was irradiation temperature, the higher was the influence of irradiation, (5) the higher was the flux of a neutron irradiation, the stronger were the changes in the deformation rates (creep), (6) the radiation creep rate did not depend on the irradiation dose.

Such changes of the characteristics of metals and dislocation mobility under the low-temperature damaging (neutron) irradiations allow one to draw a conclusion, that the specified conditions of the LTE state (the $\mathbf{T_{dbtt}}$ formation) and brittle fracture of metals (Section 4) in the process of an irradiation will not be met (essentially weakened). The phenomenon of LTRE in metals in the process of a low-temperature damaging (neutron) irradiation is absent (suppressed) because of the radiation dynamic processes. After a low-temperature damaging irradiation (including pulse neutron irradiations) of metals and alloys, which demonstrated a cold brittleness state before an irradiation (BCC, etc.), the danger of their brittle fracture (LTRE) increases because of the residual radiation hardening (depending on the temperature and the irradiation dose), raising the initial (before the irradiation) temperature $\mathbf{T_{dbtt}}$.

7. Conclusions

Concerning the mechanisms of the LTE state (cold brittleness) and brittle fracture of metals (products) with different crystal lattices (BCC, FCC, HCP) under action of the external stretching stress and damaging irradiation, it is possible to draw the following conclusions:

1. The LTE state (formation of temperature $\mathbf{T_{dbtt}}$) with a possibility of brittle fracture is a characteristic property of the metals (mono- and polycrystals) with high values of Peierls barrier and high energy of the elastic interaction of dislocations with point defects (high low-temperature yield strength). The conditions are met for the directions of dislocation slipping (directions of Burgers vectors) along the odd axes of symmetry of the crystal lattices (type ⟨111⟩ in BCC and $<11\bar{2}3>$ in HCP metals). Practically all BCC metals (steels and alloys) belong to the class with the odd axes of symmetry ⟨111⟩. It is only in some HCP metals (for example, beryllium, zirconium, zinc) and alloys on their basis for which the odd axes of symmetry $<11\bar{2}3>$ can be realized and the effect of the cold brittleness be revealed. All FCC and many HCP metals have slipping directions with the axes of an even order (⟨110⟩ and $<11\bar{2}0>$), distinguished by a small Peierls barrier and weak elastic interaction of the dislocations with the point defects (relatively small low-temperature yield strength).
2. The critical conditions appear, the simultaneous fulfillment of which determines the LTE state and brittle fracture of a metal (product):

2.1. Realization of the plain-deformed state of the loaded metal characterized by a high stiffness at rather big values of the normal and shear components of the external stress in the crack plane. The first stage of fracture (nucleation and prior-to-macroscopic crack growth) is defined by the shear component of stress and the thermoactivated mobility of the dislocations. The second stage of fracture (an avalanche spreading of a crack) is determined by the normal and shear components of the stress in the crack plane, providing the conditions necessary for the crack opening and spreading.

2.2. The normal critical stress in the crack top (mode I), defining the condition necessary for its opening (rupture of the interatomic bonds), should reach the values of the metal fracture breaking stress $\boldsymbol{\sigma_{cr}} = \mathbf{[E\gamma/L]^{1/2} \approx 0.45E[a/L]^{1/2}}$ (equal by value to Griffith's stress $\boldsymbol{\sigma_{Gr}}$).

2.3. The shear stress in the crack top $\boldsymbol{\sigma_{dyn}}$ (mode II) should exceed essentially the yield strength $\boldsymbol{\sigma_{ys}}$ ($\boldsymbol{\sigma_{dyn} \approx 10\sigma_{ys}}$) for creation of a plasticity zone and achievement in it of a dynamic area for the dislocation mobility. The front of the spreading crack cannot overtake the dislocation moving in the plasticity zone of the crack and defining the speed of its spreading. The distance between the moving top of a crack and a leading dislocation will be practically constant. The factor of viscous drag of the dislocations **B** should reach (less) the value of $\mathbf{B_{cr}}$ defining the dislocation mobility ("dry" friction mode) in the dynamic area with small work for its movement in the plasticity zone ($\mathbf{B_{cr}v < 2\gamma}$), ensuring a crack opening with formation of a free surface (state of cold brittleness and brittle fracture). The condition $\mathbf{B < B_{cr}}$ determines the temperature area of the LTE state ($\mathbf{T < T_{dbtt}}$).

3. Non-destructive amplitude-independent ultrasonic internal friction allows one to estimate the temperature of $\mathbf{T_{dbtt}}$ defined from the impact toughness experiments. At temperatures above $\mathbf{T_{dbtt}}$, there is a sharp growth of the internal friction. This is due to increasing viscous drag of the edge dislocations.
4. At a low-temperature damage (neutron) irradiation (up to $\sim$700 K) the temperature $\mathbf{T_{dbtt}}$ in the irradiated BCC metals and alloys increases because of their radiation hardening, which remains after the irradiation (the LTRE state is formed). Additional post-radiation annealing of irradiated metals at temperatures over $\sim$800 K lowers their $\mathbf{T_{dbtt}}$ because of reduction of the concentrations and states of the radiation defects (reduction of the low-temperature yield strength).
5. In the process of a low-temperature damaging (neutron) irradiation of metals (steels and alloys), which demonstrated the LTE state before the irradiation (BCC, etc.), temperature $\mathbf{T_{dbtt}}$ does not raise and the fracture strength (plasticity) increases because of many arising radiation dynamic processes ("jolting", reduction of the elastic modules, long-range action, etc.), which reduce the critical shear stress for beginning of the movement of dislocations (reducing of the low temperature yield stress), increasing the dislocation mobility and relaxation of stresses in the crack's plasticity zone. In the structural metals and alloys (BCC, etc.) of the operating nuclear and thermonuclear reactors (in the process of a neutron irradiation) the state of LTRE with the fragile fracture practically will not be realized (essentially suppressed).

Acknowledgment

The work was supported by the State Corporation "Rosatom" (the State Contract no. M.4x.44.90.13.1082).

References

[1] A.V. Stepanov, Osnovy Prakticheskoy Prochnosti, Nauka, Moskva, 1974, p. 132.

[2] Ya.B. Fridman, Mekhanicheskie Svoistva Metallov. Chast' 2. Mekhanicheskie Ispytaniya. Konstruktsionnaya Prochnost', Mashinostroenie, Moskva, 1974, p. 368.
[3] A.Ya. Krasovskiy, Khrupkost' Metallov Pri Nizkikh Tempetaturakh, Naukova Dumka, Kiev, 1980, p. 340.
[4] G.N. Yermolaev, Vopr. At. Nauki I Tekhniki. Ser. "Mater. Novye Mater." 2 (63) (2004) 267–272.
[5] A.B. Kut', Phys. Metals Metall. 106 (2) (2008) 195–201.
[6] M.A. Stremel', Razrushenie, in 2 books. Book 1 – Razrushenie Materiala, MISiS, Moskva, 2014, p. 670. Book 2 – Razrushenie struktur, MISiS, Moskva, 2015, -976 p.
[7] V.M. Chernov, M.V. Leonteva-Smirnova, M.M. Potapenko, N.I. Budylkin, Yu.N. Devyatko, A.G. Ioltoukhovskiy, E.G. Mironova, A.K. Shikov, A.B. Sivak, G.N. Yermolaev, A.N. Kalashnikov, B.V. Kuteev, A.I. Blokhin, N.I. Loginov, V.A. Romanov, V.A. Belyakov, I.R. Kirillov, T.M. Bulanova, V.N. Golovanov, V.K. Shamardin, Yu.S. Strebkov, A.N. Tyumentsev, B.K. Kardashev, O.V. Mishin, B.A. Vasiliev, Nucl. Fusion 47 (2007) 839–848.
[8] V.M. Chernov, G.N. Yermolaev, M.V. Leontieva-Smirnova, Z. Tekhnicheskoy Fiz. 80 (7) (2010) 72–77.
[9] V.M. Finkel', Fizika Razrusheniya, Metallurgiya, Moskva, 1970, p. 376.
[10] Atomistics of Fracture. Ed. R.M. Latanision, J.R. Pickens. New York, Plenum Press, 1983, 1074 p.
[11] V.V. Rybin, Bol'shie Plasticheskie Deformatsii I Razrushenie Metallov, Metallurgiya, Moskva, 1986, p. 224.
[12] J. Friedel, Dislocations, Pergamon Press, Oxford, 1964, p. 587.
[13] J. Hirth, J. Lothe, Theory of Dislocations, second ed., Elsevier, Great Britain, 2011, p. 600.
[14] A. Kelley, Strong Solids, Clarendon Press, Oxford, 1973, p. 261.
[15] T. Sudzuki, Kh. Yesinaga, S. Takeuchi, Dinamika Dislokatsii i Plastichnost', Mir, Moskva, 1989, p. 296.
[16] V.Z. Parton, E.M. Morozov, Mekhanika Uprugo-Plasticheskogo Razrusheniya: Osnovy Mekhaniki Razrusheniya, LKI, Moskva, 2008, p. 352.
[17] V.A. Likhachev, V.G. Malinin, Strukturno-analiticheskaya Teoriya Plastichnosti, Nauka, S.-Perersburg, 1993, p. 471.
[18] V.E. Panin, A.V. Panin, T.F. Elsukova, Yu.F. Popkovaл, Fiz. Mezomekh. 17 (6) (2014) 7–18.
[19] E. Orowan, Rep. Prog. Phys. 12 (1949) 185.
[20] A.A. Griffith, Philos. Trans. R. Soc. 221 (1920) 163 Ser. A.
[21] J. Frenkel, Z. Phys. 37 (1926) 572.
[22] A.W. Stepanov, Z. Phys. 92 (1934) 42.
[23] S.V. Lubenets, Fizicheskie Protsessy Plasticheskoy Deformatsii Pri Nizkikh Temperaturakh, Naukova Dumka, Kiev, 1974, pp. 220–252.
[24] V.I. Al'shitz, V.L. Indenbom, Sov. Phys. 18 (1) (1975) 1–20.
[25] E.M. Nadgorny, Dislocation Dynamics and Mechanical Properties of Crystals, Oxford Pergamon Press, 1988, p. 530.
[26] V.M. Chernov, Radiatsionnoe Materialovedenie, 2, Kharkov, 1990, pp. 17–25.
[27] V.L. Indenbom, V.M. Chernov, V.L. Indenbom, J. Lothe, Elastic Strain Fields and Dislocation Mobility, Elsevier, Amsterdam, 1992, pp. 517–570.
[28] B.K. Kardashev, Kristallogr. Rep. 54 (6) (2009) 1021–1032.
[29] B.K. Kardashev, V.M. Chernov, Mater. Sci. Eng. A 521-522 (2009) 329–334.
[30] A.V. Vatulin, A.V. Tselischev, Metall. Term. Obrab. Metall. (11) (2004) 13–19.
[31] J.M. Chen, V.M. Chernov, R.J. Kurtz, T. Muroga, J. Nucl. Mater. 417 (2011) 289–294.
[32] B.A. Gurovich, E.A. Kuleshova, Yu.A. Nikolaev, Ya.I. Strombakh, J. Nucl. Mater. 246 (1997) 91–120.
[33] V.V. Rybin, I.P. Kursevich, A.N. Lapin, J. Nucl. Mater. 258-263 (1998) 1324–1328.
[34] R.L. Klueh, D.R. Harris, High-Chromium Ferritic and Martensitic Steels for Nuclear Applications, ASTM stock Number: MONO3, 2001.
[35] G.S. Was, Fundamentals of Radiation Materials Science, Springer, Berlin, 2007, p. 827.
[36] Radiatsionnoe Materialovedenie I Konstruktsionnaya Prochnost' Reaktornykh Materialov, ed. I.V. Gorynin. – S.-Petersburg: CNII KM "Prometey", 2002, 270 p.
[37] E. Gaganidze, J. Aktaa. Fusion Eng. Des. 88 (2013) 118–128.
[38] A.I. Bondarenko, V.M. Chernov, Radiatsionnaya Polzuchest' I Skol'zhenie Dislokatsii FEI-1475, FEI, Obninsk, 1983, p. 22.
[39] A.S. Kruglov, V.N. Bykov, M.G. Bul'kanov, V.D. Dmitriev, Yu.V. Pevcikh, V.A. Fazkulin, O.G. Yurin, Radiatsionnye Defekty V Metallakh, Nauka KazSSR, Alma-Ata, 1981, pp. 136–145.
[40] V.M. Troyanov, M.G. Bulkanov, A.S. Kruglov, E.A. Kryuchkov, M.P. Nikulin, J.M. Pevchikh, A.E. Rusanov, A.A. Smirnov, S.N. Votinov, J. Nucl. Mater. 233-237 (1996) 381–384.
[41] V.V. Andreev, E.S. Aitkhodzhin, V.V. Kirsanov, G.A. Naumenko, S.S. Ogorodnik, A.S. Oryschenko, V.D. Yaroshevich, Radiatsionnye Defekty v Metallakh, Nauka KazSSR, Alma-Ana, 1981, pp. 156–162.
[42] V.K. Gorbatov, O.A. Kozhevnikov, A.N. Lapin, N.P. Losev, A.M. Parshin, A.Ya. Rogozyanov, B.V. Samsonov, V.D. Yaroshevich, Radiatsionnye Defekty v Metallakh, Nauka KazSSR, Alma-Ana, 1981, pp. 174–178.
[43] V.A. Baranov, Akusticheskie Izmereniya v Yadernoi Tekhnike, Energoizdat, Moskva, 1990, p. 320.
[44] D. Tomson, V. Pare, U. Mezon, The Effects of Imperfections, vol. IIIA, Academic Press, N.Y. and London, 1966, pp. 347–427.
[45] V.L. Indenbom, Pis'ma v ZhTF 5 (9) (1979) 489–492.
[46] M.P. Zhetbaeva, V.L. Indenbom, V.V. Kirsanov, V.M. Chernov, Pisma v ZhTF 5 (19) (1979) 1157–1161.
[47] Radiatsionnaya akustika, ed. L.M. Lyamishev, 1987, Moskva, Nauka, 136 p.
[48] A.V. Korchuganov, K.P. Zolnikov, D.S. Kryzhevich, V.M. Chernov, S.G. Psakhie, Nucl. Instrum. Methods Phys. Res. B 352 (2015) 43–46.
[49] N.P. Oparina, M.I. Guseva, B.N. Kolbasov, S.N. Korshunov, A.N. Mansurova, Yu.V. Martynenko, I.V. Borovitskaya, L.I. Ivanov, Probl. At. Sci. Technol. Ser. Thermonucl. Fusion (3) (2007) 18–27.
[50] V.V. Ovchinnikov, Uspekhi Fiz. Nauk 178 (9) (2008) 991–1001.
[51] O.V. Kamaeva, V.M. Chernov, Vopr. At. Nauki Tekhniki, Ser. Mater. Novye Mater. 1 (62) (2004) 310–318.
[52] A.B. Tsepelev, Vopr. At. Nauki Tekhniki, Ser. Mater. Novye Mater. 2 (65) (2005) 62–73.

Please cite this article as: V.M. Chernov et al., Low-temperature embrittlement and fracture of metals with different crystal lattices – Dislocation mechanisms, Nuclear Materials and Energy (2016), http://dx.doi.org/10.1016/j.nme.2016.02.002