Multi-scale Local approach to cleavage fracture and its applications

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Abstract. Key micro-mechanisms of cleavage fracture initiation, which control the statistical scale effect, the effect of temperature, strain rate, irradiation embrittlement, and which must be accounted in new improved versions of LA, are discussed. The effect of the crack nuclei generation rate with respect to plastic strain on the characteristics of fracture is analysed quantitatively. It is exhibited that along with radiation-induced hardening, increase in the crack nuclei generation rate in irradiated metal is an important micro-mechanism of radiation embrittlement of RPV steels.

Introduction

The Local approach to fracture (LA) develops significantly last three decades. This approach has enabled not only to clarify nature and micro-mechanism of metal fracture, but also to describe the effect of loading conditions on the fracture limit of structures, which cannot be easily realized with the conventional global approach [1-5]. However, recent results have demonstrated limitations of the conventional version of LA both in theoretical and applied sense. This is due, first of all, to unnecessarily oversimplified description of the fracture process in Beremin version of LA and it further modifications. Simultaneously with this version, the multi-scale approach to cleavage fracture was offered in [6-10]. Specific feature of this approach lies in possibility to describe regularities of a pre-cracked solid on macroscopic scale based on realistic physical models of the crack nuclei (CN) creation and instability in polycrystalline metals on micro-scale. This version of LA is more sophisticated for applications; however, it may be applied as theoretical basis for further development of the conventional LA.

Present report *is aimed at* presentation of key micro-mechanisms of cleavage fracture initiation in terms of multi-scale model. Physical interpretation of the main micro-mechanisms of radiation-induced embrittlement of RPV steels is given on this basis.

Generalized Criterion for Local and Global Fracture Initiation. The cleavage fracture is stochastic process, so, it is advisable to formulate the criterion of limit state of pre-cracked body in stochastic manner. In general case, the criterion for *global* fracture of pre-cracked body is the following:

$$P_{\Sigma} = 1 - \prod_{i=1}^{i=M} \left[1 - P_i\right], \tag{1}$$

where P_i is the probability of metal fracture in i^{th} elementary volume; M is the number of such volumes.

According to the "weakest link" concept:

$$P_i = 1 - [1 - P_0^i]^{\rho_i V_i}, \tag{2}$$

where P_0^i is the probability of instability of *one* crack nucleus within the i^{th} volume; ρ_i is the rate of CN generation with respect to plastic strain in the volume unit.

Eq. 2 is the criterion for initiation of *local* fracture because it describes the probability of fracture of i^{th} volume within the vicinity of a macroscopic crack tip. Weibull distribution is used for characterization of P_i, as a rule. Rigorously, it is correct in two cases: (*i*) independently on function P_0^i type at $\rho_i V_i >> 1$; (*ii*) if P_0^i is power function. At limit case $\rho_i V_i >> 1$ (great number of the CN):

, $$P_i \approx 1 - \exp\left[-\left(\frac{\sigma_f^i - \sigma_{th}}{\sigma_u^i} \right)^m \right] \quad (3)$$

where σ_f .is the local fracture stress; σ_{th} is the value of threshold stress; σ_u and m are the scale and shape parameters of a Weibull distribution:

$$\sigma_u^i = \left(1/C\rho_i V_i \right)^{1/m}, \quad (4)$$

where C is the constant.

In conventional versions of LA it is supposed that σ_u and m are the constants of material, which don't depend on temperature and loading rate, as well as on the value of local plastic strain [4]. Experimental evidence obtained recently indicates that such suppositions for wide temperature range are inappropriate [11,12]. For instance, neglect of the temperature dependence of σ_u gives rise to impossibility of correct prediction of a fracture toughness temperature dependence.

Peculiarity of LA version suggested in [6-10] is utilization of two-scale model of the CN formation and instability in polycrystalline metal. Besides, to predict probability P_i, the explicit expression (Eq. 2) was used. The essence of two-scale model is that the value of macroscopic stress of cleavage fracture initiation is determined from the conditions of formation and instability of the CN on -scopic scale with accounting for micro-stress fluctuations, stochastic distribution of grain sizes and orientations in polycrystalline metal. According to this model, the expression for probability of instability of the CN of a random size a and an orientation θ at the given macro-stress level σ_f is the following:

$$P_0(\sigma_f) = \frac{1}{2} \int_{\xi_C^{\min}}^{\xi_C^{\max}} g(\xi_C) \left[1 - \mathbf{erf}\left[\frac{\xi_C - \sigma_f}{\sqrt{2D_{\xi_{11}}}} \right] \right] \mathrm{d}\xi_C, \quad (5)$$

where $g(\xi_C)$ is the density distribution function for the critical micro-stress ξ_C.

$$g(\xi_C) = \frac{2}{k_a} \int_{\eta_{\min}}^{\eta_{\max}} g(\eta) \left[\int_{\theta_{\min}}^{\theta_{\max}} g(\theta) g(a) \frac{a^{3/2}}{\varphi(\theta,\eta)} \mathrm{d}\theta \right] \mathrm{d}\eta, \quad (6)$$

where

$$a = \frac{k_a}{\xi_c / \varphi(\theta,\eta) + \xi_{11}^{ef}}, \quad (7)$$

$g(\theta)$, $g(\eta)$ and $g(a)$ are the density distribution functions for orientation angles θ (an angle between the normal to a crack plane and the direction of tensile micro-stress ξ_{11}), respectively; $\eta = \xi_{22}/\xi_{11}$ is the parameter of a micro-stress state mode.

Eq. 7 is presented for 2D approximation. In 3D approximation the ratio ξ_{33}/ξ_{11} should be accounted as well. In general case, function $g(\eta)$ is determined from the condition that micro-stresses ξ_{11}, ξ_{22} and ξ_{33} are distributed by Gauss law with mean values $<\xi_{11}>=\sigma_1$, $<\xi_{22}>=\sigma_2$, $<\xi_{33}>=\sigma_3$ (where σ_1, σ_2 and σ_3 are the principal macro-stresses) and variances:

$$D_{\xi_{11}} = D_I\sigma_1^2 + D_{II}(\sigma_2^2+\sigma_3^2) + 2[\mu_I(\sigma_1\sigma_2+\sigma_1\sigma_3)+\mu_{II}\sigma_2\sigma_3] \quad (8a)$$

$$D_{\xi_{22}} = D_I\sigma_2^2 + D_{II}(\sigma_1^2+\sigma_3^2) + 2[\mu_I(\sigma_2\sigma_1+\sigma_2\sigma_3)+\mu_{II}\sigma_1\sigma_3] \quad (8b)$$

$$D_{\xi_{33}} = D_I\sigma_3^2 + D_{II}(\sigma_2^2+\sigma_1^2) + 2[\mu_I(\sigma_3\sigma_1+\sigma_3\sigma_2)+\mu_{II}\sigma_1\sigma_2] \quad (8c)$$

where $D_I = 1.7\cdot10^{-2}$, $D_{II} = \mu_{II} = 0.66\cdot10^{-2}$, $\mu_I = 0.72\cdot10^{-2}$ for polycrystalline iron and steels. In the first approximation, ξ_{ij} are uniformly distributed within the grain and change from one grain to another. In Eq. 7, k_a is the critical micro-stress intensity factor for CN instability in the lattice (for iron $k_a = 2.0\text{MPa m}^{0.5}$ is used). Expression for $\varphi(\theta,\eta)$ is the following:

$$\varphi(\theta,\eta) = 1\Big/\sqrt{\cos^2\theta + \eta\sin^2\theta} \quad (9)$$

The value of ξ_{11}^{ef} in Eq. 7 characterises the effect of dislocation *micro-stresses* on the CN instability. This micro-stresses are induced by dislocations within the regions where the CN form. The micro-stresses are distributed strongly non-uniformly within these regions, so, their effect on the CN instability is described by the *effective* tensile micro-stress ξ_{11}^{ef}. The value of ξ_{11}^{ef} depends non-monotonically on the plastic strain magnitude, reaching it maximum at critical value of equivalent strain e_c (for structural steels $e_c \approx 0{,}02$ [7]).

At $\bar{e} \le e_c$:

$$\xi_{11}^{ef} = k_{\xi_1}\cdot\sqrt{\frac{\bar{e}}{d}}, \quad (10)$$

where k_{ξ_1} is the coefficient, d is the average grain size.

At $\bar{e} \ge e_c$:

$$\xi_{11}^{ef} = k_{\xi 1}\cdot\sqrt{\frac{\bar{e}}{d}} - k_{\xi_2}\left(\frac{\bar{e}}{e_c}-1\right), \quad (11)$$

where $k_{\xi 2}$ is the coefficient (for iron and structural steels $k_{\xi 1} \approx 16{,}8\ \text{MPa}\cdot\text{m}^{0.5}$, $k_{\xi 2} \approx 40{,}0\ \text{MPa}\cdot\text{m}^{0.5}$).

Respectively, the expression for the value of a micro-stress ξ_C of one crack nucleus instability is:

$$\xi_C = \left[\frac{k_a}{\sqrt{a}} - \xi_{11}^{ef} \right] \cdot \varphi(\theta, \eta) , \qquad (12)$$

According to Eq. 12, the effect of dislocation-induced micro-stresses is the reason for the effect of plastic strains on the value of the critical micro-stress of ξ_C. It gives rise to non-monotonic dependence of the value of critical macro- stress, σ_f, on the value of local plastic strain $\bar{e}$ ahead of a notch or a crack tip. Unfortunately, in the most versions of LA this effect is not accounted. In the model [5] change in σ_f at strains greater than critical one e_c is considered.

In the considered two-scale model, the micro-stresses are separated on two components: (*i*) micro-stresses ξ_{ij} induced by elastic interaction of grains (grain-to-grain misfit); (*ii*) micro-stresses ξ_{11}^{ef} created by dislocations within the region where the CN form. It should be emphasized the essential difference between micro-and macro-stress state. For instance, at uniaxial *macro-stress* state ($\sigma_1 \neq 0$, $\sigma_2 = \sigma_3 = 0$) *micro-stress* state is triaxial ($\xi_{11} \neq 0, \xi_{22} \neq 0, \xi_{33} \neq 0$). Here, in the case of uniaxial tension of iron ($\sigma_1 \geq 0$, $\sigma_2 = \sigma_3 = 0$) maximum principal micro-stresses ξ_{11} change within the range [$0.60\sigma_1 \ldots 1.40\sigma_1$], and the values ξ_{22} and ξ_{33} change within the limits [$-1.24\sigma_1 \ldots +1.24\sigma_1$]. Neglect of this effect in existing models gives rise to the error in estimation both of the scatter limits of macroscopic fracture stress, the value of scale effect, as well as of calibration constants. Particularly, it is one of the reasons for overestimation of specific energy of fracture of metal in the microcrack tip.

Suggested dependences for probability of initiation of local fracture are much more sophisticated than usually employed Weibull distribution. However, the proposed approach enables us to simulation the effect of so important factors of metal microstructure, as grain size or carbide particle distribution and crystallographic texture on the probability of fracture. Distribution of sizes of grains or carbides specify unambiguously $g(a)$. Crystallographic texture governs distribution of orientations of the CN opening planes, $g(\theta)$. Without accounting for both crystallographic texture and the effect of dislocation micro-stresses (Eq. 10), it is impossible to simulate correctly the effect of warm plastic pre-strain on K_{jC}.

Cracks Nucleation. Prediction of a number of micro-cracks nucleating during plastic deformation is one of the most difficult and less investigated problems of LA. In classic models, estimation of the micro-crack length and, respectively, of the value of critical stress of it instability, is usually accentuated. However, as it follows from Eq. 2 or Eq. 3, the number of CN influences significantly the value of fracture probability and, respectively, critical local stress σ_f. Besides, in the most models the main peculiarity of the CN behaviour in metals is not accounted. It is the fact that only freshly nucleated micro-cracks may result in global fracture of the metal. If at the moment of micro-crack nucleation, Griffith' condition for it is not hold, then this micro-crack blunts and now can't "compete" with "fresh" sharp crack nuclei, which permanently generate during the plastic deformation. This specific feature CN behaviour in a metal was taking into account in the improved statistical model of fracture proposed in [3,5,6]. Therefore, the composition ρV in Eq. 2 and Eq. 3 is not the totality of CN accumulated in metal during its loading before certain value of plastic strain $\bar{e}$ is reached, but it is the number of CN, which *arises* at that strain value. It means that ρ is the rate of CN generation with respect to strain.

Inhomogeneity of micro-plastic deformation, which gives rise to plastic deformation incompatibility on grain and/or interphase boundaries, is a general reason for the CN formation in polycrystalline solids. Dislocation pile-up is the most suitable mean to describe such incompatibility, and it is widely used in conventional models of the CN formation. Generalized criterion for the CN formation on the grain boundary or interphase boundary is offered in [13]. Distinctive feature of this model is that stochastic nature of formation of *micro-plastic* incompatibilities in polycrystalline aggregate is accounted. Here, the following factors are considered: (*i*) fluctuation of shear micro-stresses ξ_{ns} acting in the slip plane where dislocation pile-up forms; (*ii*) possible relaxation of stresses in the vicinity of the head of a blocked pile-up; (*iii*) the level of *macroscopic* strain of polycrystalline aggregate. In such approach, formation of the pile-up may be described as follows:

$$CL\left[\overline{\sigma}\left(k_\sigma t - M\right) + \xi_\tau\right]^2 \geq \tau_c \quad , \tag{13}$$

where C is the constant depending on elastic constants of the lattice (for iron C =0.0336 *N/m*); d is average grain size; L is the pile-up length; $\overline{\sigma}$ and $\bar{e}$ are equivalent macroscopic stress and macroplastic strain, respectively; k_σ is coefficient ($k_\sigma = \sqrt{D_{\xi_{ns}}} / \overline{\sigma}$, where $D_{\xi_{ns}}$ is the variance of shear micro-scopic stresses ξ_{ns} in the slip systems, (for slip systems {110}<111> in iron k_σ =0.225); t is the dimensionless value of shear stresses ξ_{ns} "applied" to pile-up ($t = \xi_{ns} / \sqrt{D_{\xi_{ns}}}$); M is the orientation factor averaged over the all these slip systems orientation (for b.c.c. crystals M = 0.36); τ_c is the critical shear stress for a crack nucleus formation.

In Eq. 13, ξ_τ specifies the value of shear micro-stresses caused by interaction of a grain of averaged orientation M with plastically deformed to strain $\bar{e}$ surrounding matrix. At $\bar{e} \leq e_c$:

$$\xi_\tau = k_{\tau 1} \cdot \sqrt{\frac{\bar{e}}{d}} \,, \tag{14}$$

If $\bar{e} > e_c$:

$$\xi_\tau = k_{\tau 1} \cdot \sqrt{\frac{\bar{e}}{d}} - k_{\tau 2}\left(\frac{\bar{e}}{e_c} - 1\right), \tag{15}$$

where d is the average grain size; $k_{\tau 1}$ and $k_{\xi 2}$ are the coefficients (for iron and structural steels $k_{\xi 1} \approx 16{,}8\ \mathrm{MPa} \cdot \mathrm{m}^{0.5}$, $k_{\xi 2} \approx 40{,}0\ \mathrm{MPa} \cdot \mathrm{m}^{0.5}$). .

The value of fluctuation of stresses in a slip system where a pile-up has formed, ξ_{ns}, is specified by the expression $\overline{\sigma} \cdot (k_\sigma t - M)$.

A condition for the pile-up blocking is formulated as follows:

$$\sqrt{\frac{L}{r}}\left[\overline{\sigma} \cdot \left(k_\sigma t - M\right) + \xi_\tau\right] \leq m\tau_Y \,, \tag{16}$$

where r is the distance from the grain boundary to a dislocation source in the neighbouring grain ($r << L$); τ_Y is the critical shear stresses of such source activation. The parameter m characterizes influence of a slip system orientation of the dislocation source on the value of shear stress acting in this system. If variations of values τ_Y and r are neglected, then the expression for probability of CN formation is the following:

$$P_{nucl} = 2\int_{t_c}^{t_{max}} g(t)\left[\int_{m}^{m_{max}} g(m)\,\mathrm{d}m\right]\mathrm{d}t\,, \tag{17}$$

where $$m = [\bar{\sigma}\cdot(k_\sigma t - M) + \xi_\tau]\big/[\tau_Y \cdot \sqrt{r/L}]. \tag{18}$$

The distribution density function $g(m)$ is determined based on distribution of a scalar angle of misorientation of grain boundaries [14][1]. According to Eq. 13, an expression for critical value t_c is described by the dependence:

$$t_c = \frac{1}{k_\sigma}\left[M + \frac{1}{\bar{\sigma}}\cdot\left(\sqrt{\frac{\tau_c}{CL}} - \xi_\tau\right)\right]. \tag{19}$$

Density distribution function $g(t)$ is determined as:

$$g(t) = \frac{1}{\sqrt{2\pi}}\exp\left(\frac{t^2}{2}\right). \tag{20}$$

In some cases, at calculations it is reasonably to use an approximate expression for P_{nucl}. It may be obtained if the value of m variation is neglected. In this case:

$$P_{nucl} \approx P(t_c < t < t_r) = 2\int_{t_c}^{t_r} g(t)dt\,. \tag{21}$$

Accounting for Eq. 21, P_{nucl} is the following:

$$P_{nucl} \approx 2[\Phi(t_r) - \Phi(t_c)]\quad, \tag{22}$$

where $\Phi(t_r)$ and $\Phi(t_c)$ are the values of Laplace function at the corresponding magnitude of parameter t.

The expression for parameter t_r that characterises relaxation conditions is the following:

$$t_r = \frac{1}{k_\sigma}\left[M + \frac{1}{\bar{\sigma}}\cdot\left(\tau_Y m\cdot\sqrt{\frac{r}{L}} - \xi_\tau\right)\right]. \tag{23}$$

In general case, the rate of CN generation in the metal volume unit may be specified as:

$$\rho = k_\rho \cdot P_{nucl}\quad, \tag{24}$$

where k_ρ is the coefficient depending on the density of carbide particles or grain facets. The value of this coefficient can be estimated using experimental evidence by a calibration procedure.

The approach proposed enables to simulation the effect of many factors on the rate of CN

[1] $g(m)$ is somewhat cumbersome expression, so it is not presented in explicit form.

generation, such as *metallurgical* factors (average grain size d and maximum grain size $L \approx (0.5 \div 1.0) d_{\max}$), *loading* condition (temperature and loading rate (parameters τ_Y and $\overline{\sigma}$), *crystallographic texture* (function $g(m)$), the value of *plastic strain* $\bar{e}$.

Fig. 1 presents dependence of ρ on the value of plastic strain at different test temperatures for reactor pressure vessel steel 2*Cr-Ni-Mo-V*. Specific feature of these dependences is non-monotonic change of ρ with $\bar{e}$ growth. This agrees well with data of work [15]. This data analysis shows that rate of carbide cracking grows up to a certain level, after which it decreases monotonically.

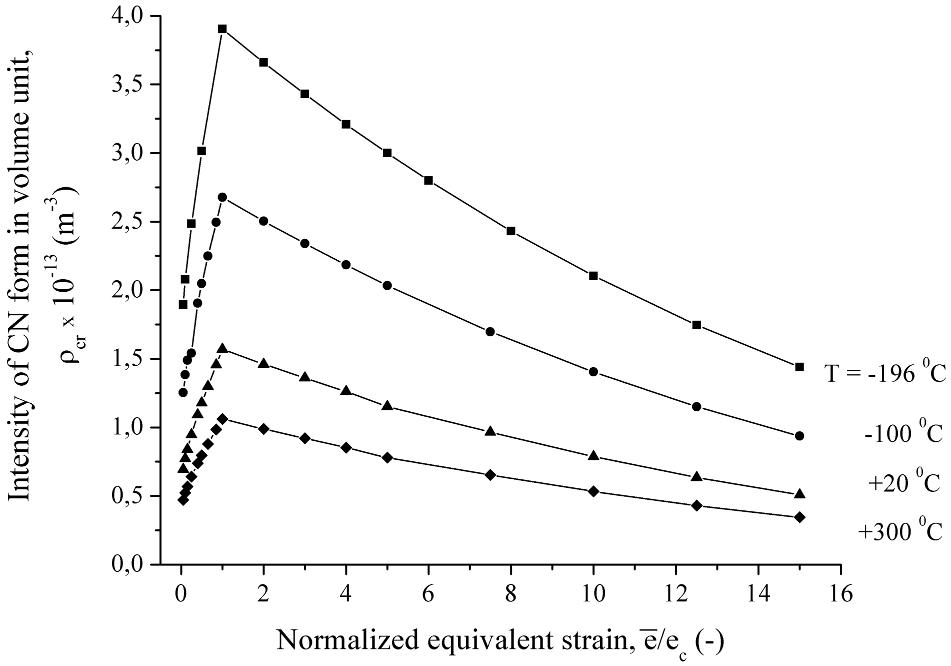


Fig.1 Dependence of the CN density in RPV steel, ρ, on the value of plastic strain and temperature: $\bar{e}$ and e_c are equivalent plastic strain and its critical value, respectively.

Effects Related to the Influence of the CN Generation Rate on the Brittleness of Metal.
Dependence of σ_f *on temperature*. Consideration of the temperature T influence on the value of cleavage stress is usually reduced to the analysis of possible temperature effect on the value of critical stress of instability *one* CN. It is supposed that specific energy of lattice fracture in the micro-crack tip must increase with temperature growth. At the same time, estimations show that for $\alpha - Fe$ these effects may be observed only at temperatures higher than $\approx 300^0 C$.

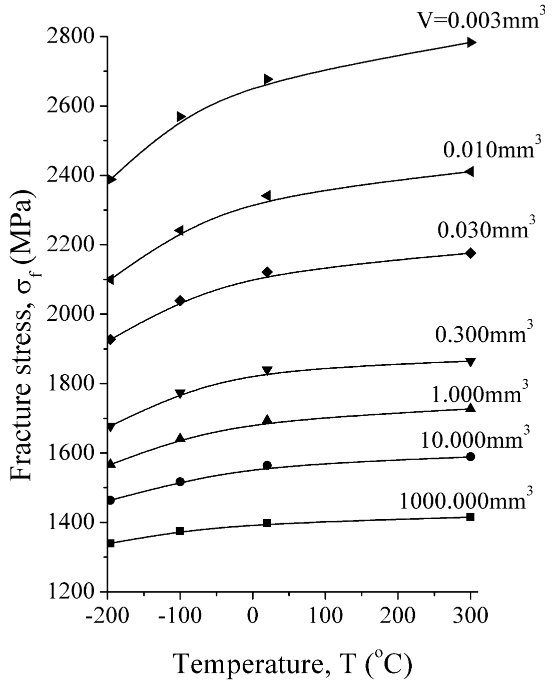
From the statistical fracture criterion (Eq. 2) it follows that decrease in the rate of CN generation within the volume unit ρ gives rise to increase in the average value of fracture stress of a specimen. It can occurs as a result of diminution in critical stress τ_Y, at which plastic relaxation of micro-stresses in the tip of pile-up is possible. According to Eq. 21, Eq. 23 and Eq. 24, decrease in τ_Y gives rise to reduction in the CN generation rate. From the physical point of view, it means that growth of temperature facilitates relaxation processes, which diminish incompatibilities of micro-plastic strain on the grain boundaries. According to Eq. 4 the dependence ρ on temperature, strain rate and a notch radius must lead to influence these factors on the value of a shape parameter σ_u. It is a good agreement with the experimental data [11, 12].

Fig. 2 presents temperature dependences of σ_f calculated for RPV steel 2Cr-Ni-Mo-V. In this calculations influence of the temperature on τ_Y was described as follows:

$$\tau_Y = 0{,}5 C_1 \exp[-(C_2 + C_3 \ln \dot{e})T] + \tau_a \tag{25}$$

Were τ_a is the athermal component of τ_Y, $\dot{e}$ is the strain rate, C_1, C_2, C_3 are the constants.

Fracture at the value of residual strain $\bar{e} = 2\%$[2] was modeled, where $C_1 = 1033\text{MPa}$, $C_2 = 0.00698$, $C_3 = 0.000415$, $\dot{e} = 1\cdot 10^{-4} c^{-1}$, $\tau_a = 230\text{MPa}$. According to these data, cleavage fracture at uniform uniaxial tension increases with growth of the test temperature from $-196^0 C$ to $+300^0 C$. Here, the rate of increase in σ_f grows with decrease in the volume of loading metal. This due to non-linear dependence of fracture probability P on the CN number in Eq. 2.



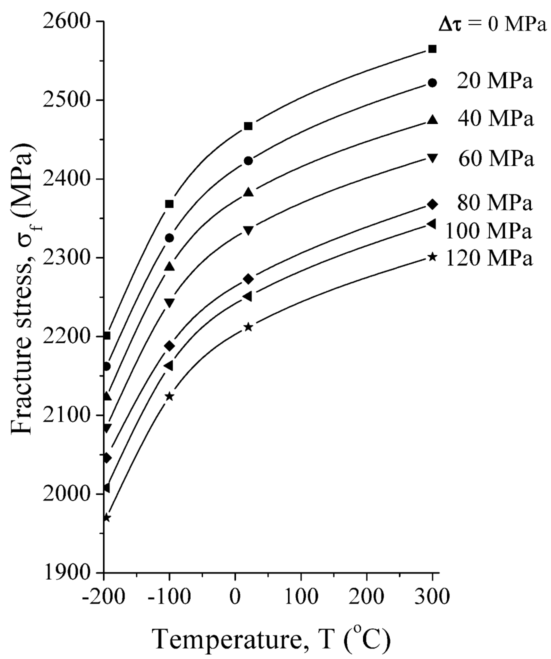


Fig. 2 Dependences of fracture stress σ_f on the temperature T at different values of the volume V of loading metal.

Fig. 3 The effect of temperature T on the level of fracture stress σ_f depending on the value of radiation-induced increment of critical relaxation stress τ_Y for the volume of "process zone" V=0.003 mm^3 ahead of macro-crack tip.

As a result, at great ρV magnitudes change in ρ doesn't virtually effect the value of probability P. In contrast, at decrease in the value of ρV, sensitivity to ρ change increases. Consequently, for standard tensile specimen ($V \approx 1000\,\text{mm}^3$) change in σ_f doesn't exceed 4% over the whole temperature range. It enables to substantiate conventional thesis that the value of brittle strength $R_{MC}(S_F)$, determining as minimum fracture stress within the ductile-to-brittle transition temperature range, is really independent on temperature. According to experimental evidence, the value of local stress ahead of notch or macro-crack, σ_f remains practically unchanged at low temperatures, or slightly decreases. This doesn't contradict obtained results because the volume of "process zone" ahead of notch/crack doesn't remain constant but increases with temperature growth. As a result, increase in V compensate decreasing in the CN generation rate ρ [13].

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[2] At simulation, there was assumed that critical stress of the CN instability, ξ_c, (12) remains constant till $T = 300^0 C$.

Simulation results presented on Fig. 2 also enable to explain the nature of local scale effect, which appears as increase in critical cleavage stress σ_f with decrease in the notch radius, and it 1.5-2.0 times excess over the level of brittle strength R_{MC} at uniaxial tension [10]. According to the model proposed, this is due to decrease in the "process zone" volume, which value is determined by the notch radius.

The effect of neutron irradiation on the value of cleavage stress σ_f. Recently, at analysis of radiation embrittlement, the focus is on reduction of the value of fracture toughness K_{jc} or on shift of the reference temperature T_0 as a result of radiation-induced hardening of metal. In terms of the micro-model of fracture, the effect of embrittlement by radiation-induced hardening is explained by decrease in stability of the CN due to growth of the level of tensile stresses, at which they form. On macroscopic scale it appears as lateral shift of the temperature dependence of fracture toughness to the region of higher temperatures. However, in many cases, decrease in fracture toughness of irradiated metal is related to decrease in the value of cleavage stress σ_f. According to Eq. 21 and Eq. 23, increase in the value of critical relaxation stress, τ_Y ,of irradiated metal must give rise to ρ growth, and, respectively, to reduction of σ_f. Precipitates or segregation of phosphorus and other impurities near grain boundaries are the reason for τ_Y growth after neutron irradiation[3]. On macroscopic scale this micro-mechanism must appear as change in shape of the temperature dependence of fracture toughness, namely, in it slope decrease. Fig. 3 presents the dependence of σ_f on temperature for certain values of radiation-induced athermal component of τ_Y. According to these data, the degree of change in σ_f after irradiation is determined by competition of two processes: (*i*) increase in σ_f due to decrease in thermal component of τ_Y at higher temperature T_0^{ir}; (*ii*) decrease in σ_f because of increase in τ_Y due to precipitates and segregations on grain boundaries.

Conclusions:

1. Analysis of fracture on micro- and macroscopic scales enabled to ascertain main peculiarities of fracture initiation mechanism, which must be accounted directly or indirectly in advanced version of Local Approach. They are:

- The effect of local plastic strain on the value of local fracture stress σ_f.
- Density of the crack nuclei forming in metal is not constant. The rate of CN generation in the unit volume of metal depends on temperature, plastic strain rate and other factors that influence the value of critical stress of relaxation of plastic incompatibilities on grain boundaries and/or inter-phase boundaries.

2. The number of forming CN within the "process zone" is a key factor that control the value of statistical scale effect. It is also the reason for temperature and strain dependence of the critical local stress and scale parameter σ_u in Weibull distribution. It governs the shape of the temperature dependence of fracture toughness as well.

3. Increase in density of the CN due to precipitates and impurities on grain boundaries is one of the micro-mechanisms of radiation-induced embrittlement of RPV steels and its welds. In contrast to radiation-induced hardening, at this micro-mechanism, decrease in fracture toughness K_{Jc} is related not to the lateral shift of K_{Jc} temperature dependence, but to change in the shape of the .fracture toughness curve (decrease in the slope of K_{Jc} temperature dependence).

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[3] Here, the case of the relatively low concentration of impurities is considered, when transcrystalline fracture has not yet changed by intercrystalline one.

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