# Stress-Controlled and Stress-and-Strain Controlled Criteria of Brittle Fracture in Local Approach

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**Abstract** Stress-controlled and stress-and-strain controlled criteria are compared in light of prediction of brittle fracture on a macro-scale and in light of adequate description of microstructural features of brittle fracture of irradiated RPV steels.

**Keywords** brittle fracture, local criterion, radiation defects, modelling, RPV steel

## 1. Introduction

At present, several brittle fracture models based on local criteria are developed. The local approach models differ from each other just in the formulation of local criteria. Basically, two types of local criteria may be defined: stress-controlled and stress-and-strain controlled criteria.

According to stress-controlled criterion proposed by Ludwik [1] and Yoffe [2] brittle fracture occurs when the maximum principal stress reaches some critical fracture stress $S_C$ independent of temperature. Later Davidenkov [3] and Knott [4] added necessary condition to initiate brittle fracture - a requirement of active plastic deformation. This stress-controlled criterion was initially used for low-strength steels. Application of this criterion for middle and high strength steels, in particular, for reactor pressure vessel (RPV) steels, showed that $S_C$ varies for specimens with various geometries. It means that stress-controlled criterion does not work, at least, in deterministic statement.

The Beremin model [5] based on stress-controlled criterion takes into account a stochastic nature of brittle fracture using the Weibull approach. This model allows one to describe the scale effect and provide the agreement of test results from notched and cracked specimens from RPV steel over narrow temperature range. However to predict the temperature dependence of fracture toughness with the Beremin model for RPV steels the introduction of additional a priori assumption is required. More difficulties arise when using stress-controlled criterion for irradiated RPV steels.

To overcome these difficulties, stress-and-strain controlled criterion was proposed near 20 years ago [6]. Application of this criterion allows one not only to provide the transferability of test results from specimens with various geometries, but also to predict known properties of brittle fracture.

The application of fracture modelling with a local cleavage fracture criterion (local approach) in principle allows the prediction of the fracture properties on a macro-scale without any additional assumptions. Local approach models may be enlarged to include in the consideration microstructural characteristics, in particular, irradiation-induced defects. Thus, fracture modelling with local fracture criterion unites various scales from irradiation-induced defects on a nano-scale to cleavage microcrack on micro-scale and to fracture on a meso- and macro-scales. This approach may provide reliable (both from mechanical and physical viewpoints) underpinning for engineering methods that are required for RPV integrity assessment. The predictable possibilities of fracture models appear to be substantially determined by local criterion used.

The present report compares stress-controlled and stress-and-strain controlled criteria in light of prediction of brittle fracture on a macro-scale and discusses which local criterion provides adequate description of microstructural features of brittle fracture of irradiated RPV steels.

## 2. Local brittle fracture criteria

Currently two local cleavage fracture criteria are mainly used in brittle fracture modelling. Traditional formulation is written as [4]

$$\sigma_{eq} \geq \sigma_Y \tag{1a}$$

$$\sigma_1 \geq S_C, \tag{1b}$$

where $\sigma_{eq}$ is the equivalent stress, $\sigma_Y$, the yield stress, $\sigma_1$, the maximum principal stress and $S_C$, the critical brittle fracture stress, which is independent of temperature, strain rate and stress triaxiality.

Another local criterion of cleavage fracture was formulated and verified in the papers [6-10]. This formulation is written in the form

$$\sigma_{nuc} \equiv \sigma_1 + m_{T\varepsilon} \cdot \sigma_{eff} \geq \sigma_d, \tag{2a}$$

$$\sigma_1 \geq S_C(æ), \tag{2b}$$

where the effective stress is $\sigma_{eff} = \sigma_{eq}-\sigma_Y$, $æ = \int d\varepsilon^p_{eq}$ is the accumulated plastic strain, $d\varepsilon^p_{eq}$ is the equivalent plastic strain increment, $\sigma_d$ is the critical stress for microcrack nucleation and $m_{T\varepsilon}$ is the concentration coefficient for the local stress near the microcrack-nucleating particles. This coefficient depends on temperature T and plastic strain and may be written as $m_{T\varepsilon}=m_T(T)\cdot m_\varepsilon(æ)$. From the physical viewpoint the parameter $\sigma_d$ is the strength of carbides or carbide-matrix interfaces or other particles on which cleavage microcracks are nucleated. The functions $S_C(æ)$, $m_T(T)$ and $m_\varepsilon(æ)$ are calculated as [6-10]

$$S_C(æ) = \left[C_1 + C_2 \exp(-A_d æ)\right]^{-1/2}, \tag{3}$$

$$m_\varepsilon(æ)= S_0/S_C(æ), \tag{4}$$

$$m_T(T) = m_0 \cdot \sigma_{Ys}(T), \tag{5}$$

where $C_1$, $C_2$, $A_d$ are material constants, $S_0 \equiv S_C(æ=0)$ is the stress of start for the nucleus microcrack, $m_0$ is a constant which may be experimentally determined and $\sigma_{Ys}$ is the temperature-dependent component of the yield stress.

From the physical viewpoint, Eqs. (1a) and (2a) are the conditions for cleavage microcracks nucleation, and Eqs. (1b) and (2b) are the conditions of their propagation. In criterion (1) the condition (1a) is the simplest requirement to reach a minimum plastic strain corresponding to yield stress that is usually equal to 0.2%. As distinct from condition (1a), cleavage microcrack nucleation according to condition (2a) depends on the maximum principal stress, plastic strain and temperature and is characterized by the critical stress $\sigma_d$. It is important that the plastic strain when microcrack is nucleated may exceed 0.2% and increases with the temperature growth [8, 10].

It may be noted that the connection of cleavage microcrack nucleation with plastic deformation seems to be quite clear from the physical point of view. Nevertheless, when formulating the local cleavage fracture criterion this connection was explicitly used only near twenty years ago [6, 11]. Now this consideration is widely used in other models, for example, [12]).

The important consequences follow from this difference between (1a) and (2a). In criterion (2) two critical parameters - $S_C$ and $\sigma_d$ may control cleavage fracture and this depends on material properties and loading conditions, mainly, on the ratio $S_C/\sigma_Y$, stress triaxiality and temperature. For example, the brittle fracture of smooth specimens is controlled by (2b) and, by contrast, the brittle fracture of notched or cracked specimens from RPV steels by (2a) [9, 10]. For smooth tensile specimens for that stress triaxiality is low, condition (2a) is satisfied earlier than condition (2b). Therefore condition (2b) controls the brittle fracture of smooth tensile specimens. When stress triaxiality is high that is typical for notched and cracked specimens, condition (2b) is satisfied for medium and high strength steels already at very small plastic strain when cleavage microcracks are still not nucleated (condition (2a) is not fulfilled). That's why brittle fracture occurs just after satisfaction of condition (2a). By other worlds, condition (2a) controls brittle fracture for this case.

According to criterion (1) the brittle fracture on a macro-scale is controlled practically by the only process – microcrack propagation, i.e. by condition (1b) as the microcrack nucleation condition (1a) is practically always satisfied earlier than condition (1b). In terms of mechanical parameters it means that criterion (1) is stress-controlled fracture criterion and criterion (2) is stress-and-strain controlled fracture criterion.

Prediction of brittle fracture on a macro-scale in a stochastic manner may be performed with the Beremin model [5] that uses criterion (1) and stochastic parameter $S_C$, and with the Prometey model [9, 13] that uses criterion (2) and one ($\sigma_d$) or two ($\sigma_d$ and $S_C$) stochastic parameters. Both models use the Weibull statistics for stochastic parameters and the weakest link model to predict the brittle fracture on a macro-scale.

## 3. Local criteria and microstructural features of brittle fracture of RPV steels

### 3.1 Microcrack nucleation sites for cracked specimens

The first interesting consequence from local brittle fracture criteria concerns localization of cleavage fracture initiation sites (i.e. sites where microcrack is nucleated and propagates) for cracked specimens. It follows from criterion (1) that cleavage fracture near the crack tip is always initiated where the peak stress is located as the probability of microcrack propagation is maximum at the peak stress, and the microcrack nucleation condition is satisfied over the whole plastic zone.

Brittle fracture of cracked specimens according to criterion (2) is mainly controlled by condition of cleavage microcrack nucleation (2a) as condition (2b) is satisfied practically over the whole plastic zone (excepting regions close to the crack tip and plastic zone boundary). As a result, both stress and plastic strain determine the cleavage fracture initiation sites.

Criterion (2) predicts that, firstly, site of the maximum probability of nucleation of propagating microcrack is located between the crack tip and the peak stress. This trend predicted with the model with one stochastic parameter $\sigma_d$ [9] is schematically shown in Figure 1. (The model with two stochastic parameters $\sigma_d$ and $S_C$ [13] predicts some strip for sites of the maximum probability of nucleation of propagating microcrack.) Criterion (2) shows also that this site moves to the peak stress as degree of embrittlement of a material increases. The latter follows from increasing the contribution of stress as compared with plastic strain in condition (2a) for the embrittled material. These results are confirmed to a larger degree by fractographic examination of fracture surfaces [12]. Criterion (2) and the developed models of the prestrain effect on cleavage microcrack nucleation [14] allows also an analysis of the WPS effect on localization of initiation sites for cracked specimens. WPS results in a shift of initiation sites from the crack tip as plastic prestrain decreases a number of possible initiators especially near the crack tip. This trend was observed in [15].

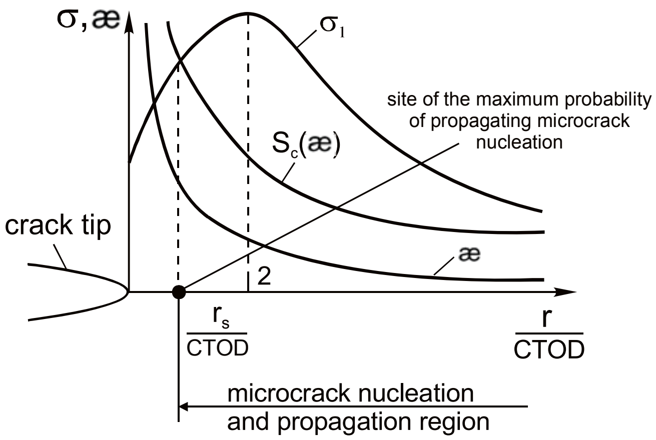


Figure 1. Microcrack nucleation sites near the crack tip according to criterion (2) for initial and not highly embrittled conditions of a material [10].

### 3.2. A link of local criterion with irradiation-induced defects

Local fracture criteria (1) and (2) contain internal parameters that, in principle, may be linked with the physical mechanisms of fracture and material microstructure. From viewpoint of fracture modeling for irradiated RPV materials, first of all, it is important to find how internal parameters in local criterion are linked with irradiation-induced defects.

For RPV steels three types of irradiation-induced damage are found: *matrix damage* caused by irradiation-induced lattice defects, such as clusters of point defects and dislocation loops, *precipitation* of various elements, namely, copper, nickel, manganese and other, and *segregation* of impurities, mainly phosphorus [16-18]. These radiation damages and mechanical properties of irradiated RPV materials are linked as follows. The matrix damage and precipitates result in an increase of $\sigma_Y$ as they affect the dislocation mobility. An increase of $\sigma_Y$ is caused by an increase of the athermal component $\sigma_{YG}$ of the yield stress. Segregation of impurities, as a rule, is not associated with changes in $\sigma_Y$ due to irradiation, at the same time these segregations may result in increase of the ductile-to-brittle transition temperature (DBTT), $T_{tr}$. [18]. Thus, segregation of impurities, in particular, phosphorus, results in so-called non-hardening mechanism of embrittlement.

When analysing the link of the above criteria with irradiation-induced defects it should be taken into account that the critical stress $S_C$ does not practically depend on neutron fluence (at least, for transcrystalline fracture) [10, 17, 18]. This follows from experimental and theoretical results. From a physical viewpoint, irradiation-induced lattice defects and precipitates result in not decreasing the critical stress $S_C$. It follows from the model [8] of propagation of Griffith's crack on cleavage plane through the microstress fields which are considered as barriers for microcracks. Then criterion (1) contains the only parameter $\sigma_Y$ that depends on fluence. Hence, criterion (1) describes radiation embrittlement as a result of the material hardening only and cannot describe non-hardening mechanisms of embrittlement, for example, caused by P segregation. Thus, criterion (1) describes radiation embrittlement through the mechanical factor only - increase of $\sigma_1$ due to increase of $\sigma_Y$.

Criterion (2) contains two parameters - $\sigma_d$ and $\sigma_Y$ that depend on fluence. It means that criterion (2) takes into account not only the material hardening but also a possible weakening of microcrack initiators that is described by decreasing $\sigma_d$. This process may be considered as the physical factor of embrittlement. It is clear that impurity segregation resulting in embrittlement without hardening may be explained with criterion (2). Indeed, as known the P segregation may occur on ferrite-carbide interfaces [16] and result in decreasing the interface strength (i.e. in decreasing $\sigma_d$). This means that the nucleation of cleavage microcracks becomes easier compared with unirradiated steel.

The mechanisms of the effect of radiation defects on cleavage microcrack nucleation were proposed in [13, 14]. It should be noted that for RPV steels in unirradiated condition the cleavage microcrack initiators are mainly globular carbides [19] and additional initiators do not arise under irradiation as practically all radiation defects are very small to nucleate a sharp cleavage microcrack. (For example, sizes of dislocation loops are near 5÷20 nm and the nucleus microcrack size estimated from the Griffith's condition is 100÷400 nm.) The proposed mechanisms explain how radiation defects affect cleavage microcrack nucleation on carbides. The proposed mechanisms allow the determination of the dependence of critical parameter $\sigma_d$ on neutron fluence [13, 14].

These mechanisms are divided into two groups. The first group includes the mechanisms [13, 14] connected with decreasing the strength of carbide-matrix interface. One mechanism of decreasing $\sigma_d$ is the impurity (P) *segregation* on ferrite-carbide interfaces caused by the P diffusion accelerated by irradiation. Another mechanism for decreasing $\sigma_d$ is *the arising of internal stresses* caused by irradiation-induced dislocation loops and precipitates on carbide-matrix interface. These internal stresses result in rupture of the interface at stress $\sigma_{nuc}$ being less than $\sigma_{nuc}$ for unirradiated material.

The second group includes the mechanism connected with easier formation of dislocation pile-ups near initial microcrack initiators (for example, globular carbides) due to increasing the concentration of radiation defects [13, 14]. This process may be described by increasing the probability of dislocation pile-up formation in material with high concentration of barriers for dislocations. The detailed consideration is given in [13, 14] where the probability of pile-up formation is described with the Weibull function and the Orovan stress used as a characteristic parameter for concentration of various barriers.

### 3.3. On the intergranular fracture mode

Brittle fracture of unirradiated RPV steels occurs, as a rule, by transcrystalline cleavage and microcleavage. Neutron irradiation and post-irradiation annealing may result in appearance of intercrystalline fracture mode. It is interesting that a fraction of intercrystalline fracture is not always correlated with mechanical properties. For irradiated steels the mechanical parameters (DBTT and $\sigma_Y$) speak about significant embrittlement, however a fraction of intercrystalline fracture is usually small (near 20% of fracture surface). After annealing, significant recovery of mechanical properties is observed, at the same time, a fraction of intercrystalline fracture may increase.

These findings may be explained with criterion (2) and the proposed mechanisms of the effect of radiation defects on the critical stress $\sigma_d$ [13, 14]. In deterministic statement, value of $\sigma_d$ is determined by minimum value of the critical stresses for nucleation of transcrystalline $\sigma_d^{tr}$ and intercrystalline $\sigma_d^{int}$ microcracks. The fracture mode may be trans- or intercrystalline that depends on which value is less if the critical stress $S_C$ is the same for both types of microcracks. In probabilistic statement, mixed fracture is expected if the difference of $\sigma_d^{tr}$ and $\sigma_d^{int}$ is not large.

The interpretation for variation of the parameters $\sigma_d^{tr}$ and $\sigma_d^{int}$ is schematically shown in Figure 2, for RPV steels in various conditions (unirradiated, irradiated and after post-irradiation annealing). Two steels are considered: 2.5Cr-Mo-V and 2Cr-Ni-Mo-V (steels for WWER-440 and WWER-1000 RPV respectively). For unirradiated steels (Fig. 2a) brittle fracture occurs mainly by the transcrystalline mechanism that means that $\sigma_d^{tr} < \sigma_d^{int}$ and $S_C^{tr} < S_C^{int}$, where $S_C^{tr}$ and $S_C^{int}$ are the critical stresses for propagation of transcrystalline and intercrystalline microcracks. The difference of $\sigma_d^{tr}$ and $\sigma_d^{int}$ is less for 2Cr-Ni-Mo-V steel for which some fraction of intercrystalline fracture is observed.

For irradiated steels (Fig. 2b) $\sigma_d^{tr}$ and $\sigma_d^{int}$ decrease. For carbides located on grain boundary, segregation of impurities and arising of internal stresses caused by dislocation loops occur more intensively. Intercrystalline phosphorus segregation caused by irradiation decreases also $S_C^{int}$ as sharp microcrack nucleation and propagation on phosphorus monolayer weakening atomic bonds, require less energy. Nevertheless, for irradiated 2.5Cr-Mo-V steel the conditions $\sigma_d^{tr} < \sigma_d^{int}$ and $S_C^{tr} \approx S_C^{int}$ are more possible, so that transcrystalline brittle fracture is more typical. For irradiated 2Cr-Ni-Mo-V steel the situation is possible when $\sigma_d^{tr} \approx \sigma_d^{int}$ (this steel is more sensitive to the P segregation) and $S_C^{tr} \approx S_C^{int}$ as so that mixed trans- and intercrystalline brittle fracture is observed.

After post-irradiation annealing at $T_{ann}$=475$^o$C (Fig. 2c) phosphorus segregations dissociate in a grain only, and do not dissociate on grain boundary [16, 17]. Radiation-induced dislocation loops and precipitates may be dissociated practically completely both in a grain and on grain boundary. Therefore the $\sigma_d^{tr}$ value increases up to the value for the unirradiated condition but the $\sigma_d^{int}$ and $S_C^{int}$ values remain less than for the unirradiated condition. As a result, for 2.5Cr-Mo-V steel a fraction of intercrystalline fracture may increase as compared with irradiated specimens as the values of $\sigma_d^{tr}$ and $\sigma_d^{int}$ become close, although this annealing results in full recovery of the mechanical properties ($T_{tr}$ and $\sigma_Y$). For 2Cr-Ni-Mo-V steel the situation is possible when intergranular carbides become "weakest link" and intercrystalline fracture predominates, and although $\sigma_Y$ may recover fully, $T_{tr}$ recovers not fully so that $(T_{tr})^{unirr} < (T_{tr})^{ann} < (T_{tr})^{irr}$. The reason is clear: for this case the brittle fracture resistance is controlled by $\sigma_d^{int}$ that does not recover fully.

Annealing at $T_{ann}$≈600$^o$C (Fig. 2d) results in full recovery of $\sigma_d^{int}$ as dissociation of grain boundary P segregations occurs at this temperature [16] and the situation become close to the unirradiated condition (Fig. 2a). Full recovery of the mechanical properties and the fracture modes is observed.

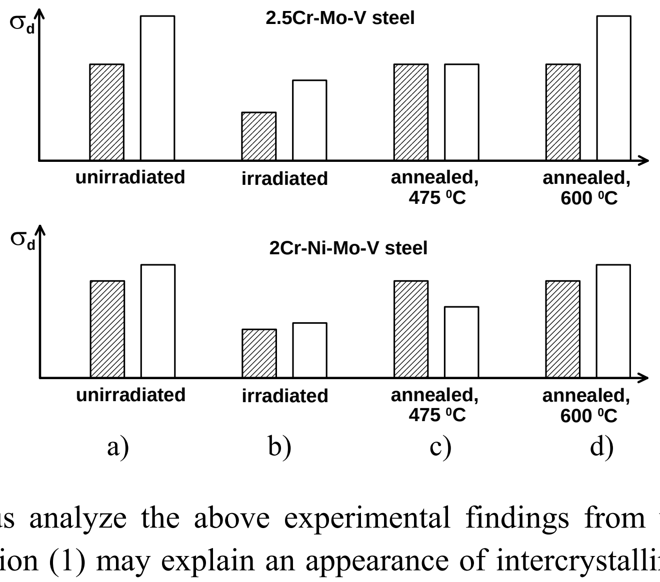


Figure 2. Variation of $\sigma_d^{tr}$ (shaded bar) and $\sigma_d^{int}$ (open bar) for RPV steels in various conditions: unirradiated (a), irradiated (b), after post-irradiation annealing (c) and (d) [14].

Let us analyze the above experimental findings from viewpoint of criterion (1). It is clear that criterion (1) may explain an appearance of intercrystalline fracture for irradiated steel but does not explain an increase of fraction of intercrystalline fracture after post-irradiation annealing. Indeed, brittle fracture according to criterion (1) is controlled by minimum value of the two values $S_C^{tr}$ and $S_C^{int}$. For unirradiated steels brittle fracture occurs mainly by the transcrystalline mechanism so that it follows from criterion (1) that $S_C^{tr} < S_C^{int}$ and $S_C^{tr}$ is the controlling parameter. Intercrystalline phosphorus segregation caused by neutron irradiation decreases the critical stress $S_C^{int}$. (The critical stress $S_C^{tr}$ does not decrease for irradiated RPV steels [18].) For irradiated RPV steels the relation $S_C^{tr} \approx S_C^{int}$ or $S_C^{tr} > S_C^{int}$ become possible and, hence, either both stresses $S_C^{tr}$ and $S_C^{int}$ or $S_C^{int}$ control brittle fracture according to criterion (1). As a result, intercrystalline fracture is predicted for irradiated steel. However post-irradiation annealing can not result in additional $S_C^{int}$ decrease (on the contrary, annealing may increase $S_C^{int}$) so that an increase of fraction of intercrystalline fracture after post-irradiation annealing can not be expected from viewpoint of criterion (1).

## 4. Comparison of local criteria in light of prediction of brittle fracture on a macro-scale

### 4.1 Prediction of $K_{JC}$(T) curve and its transformation for irradiated RPV materials

Application of the Beremin model [5] for prediction of $K_{JC}$(T) for various materials may be found in [20], application of the Prometey model for RPV steels in various conditions (initial, irradiated and highly embrittled) – in [9, 10, 13]. It was found that there are difficulties with the use of the Beremin model for medium and high strength steels, in particular, for RPV steel [6-10]. It was shown that the prediction of fracture toughness of irradiated RPV materials with this model is not correct. The reason consists in the fact that according to this model the dependence $K_{JC}$(T) is determined practically by the dependence $\sigma_Y$(T) as $S_C$ does not depend on temperature. The fact of the matter is that according to criterion (1) the rate of growth of $K_{JC}$ with temperature T is controlled by the parameter $\frac{1}{\sigma_Y} \cdot \frac{d\sigma_Y}{dT}$. For highly embrittled material the variation of $K_{JC}$ with temperature T occurs over the temperature range where this parameter $\to 0$. As a result, $K_{JC}$ does not practically depend on temperature that is in contradiction with test results.

Some attempts (for example, [21]) were undertaken to reform the Beremin model by introduction of the temperature dependence for the parameter $S_C$ (or the parameter $\sigma_u$ in the terms of the Beremin model). The parameter $S_C$ becomes not invariant relative to stress triaxiality and temperature. From physical viewpoint such "reformation" cannot be considered as reasonable. It has been found in [6-9] that difficulties with the use of the Beremin model for medium and high strength steels are connected with the use of microcrack nucleation condition in the form (1a).

When using the Prometey model there is no problem with prediction of $K_{JC}$(T) for embrittled RPV steels as the parameter $\sigma_{Ys}$(T) is used in criterion (2) as temperature dependent parameter (see Eq. (5)). As a result, over the temperature range where $\frac{1}{\sigma_Y} \cdot \frac{d\sigma_Y}{dT} \to 0$, the ratio $\frac{1}{\sigma_{Ys}} \cdot \frac{d\sigma_{Ys}}{dT} \neq 0$. That' why criterion (2) allows one to describe $K_{JC}$(T) adequately even for highly embrittled material.

Transformation of $K_{JC}$(T) curve for irradiated RPV steels as the test results [22, 23] show may be approximately described as a lateral shift to higher temperature range for small degree of embrittlement, and as a variation in the $K_{JC}$(T) curve shape for high degree of embrittlement [22]. Criterion (2) and the Prometey model provide a possibility to predict this transformation of the $K_{JC}$(T) curve as illustrated in Figure 3a where the decrease in $\sigma_d$ models increasing neutron fluence.

At the same time criterion (2) and the Prometey model may also predict a pure lateral shift of $K_{JC}(T)$ curve to higher temperature range [14]. It may be achieved by an increase of the parameter $m_0$ with increasing neutron fluence F, i.e. an increase of "driving force" $\sigma_{nuc} \equiv \sigma_1 + m_{T\varepsilon} \cdot \sigma_{eff}$ in condition (2a) (without the decrease in $\sigma_d$). For this case the calculated $K_{JC}(T)$ curves are shown in Figure 4b. It should be noted that the parameter $m_0$, being to some degree sensitive to material microstructure, may depend on the particularities of plastic deformation in steels with radiation defects. It is clear that as a common case, irradiation-induced dislocation loops and precipitates may affect the geometry of dislocation pile-ups arrested by carbides and, hence, the coefficient $m_0$. However, at present, the experimental data are too few to allow an analysis of this effect. Possible trends of this effect have been considered in [14]. In principle, the coefficient $m_0$ may increase as F increases due to decrease of the width and blunting of dislocation pile-up near microcrack initiator that is a result of an increase of the density of radiation defects.

Thus, criterion (2) provides a possibility to predict not only a change in $K_{JC}(T)$ curve shape but also pure lateral shift of $K_{JC}(T)$ curve to higher temperature range. It should be emphasize that any models based on the stress controlled criterion (1) predict a variation in the $K_{JC}(T)$ curve shape for any degree of embrittlement. This is because the $K_{JC}(T)$ dependence is determined according to criterion (1) by the $\sigma_Y(T)$ dependence. These models may predict lateral shift of $K_{JC}(T)$ only if the dependence of $S_C$ (or $\sigma_u$ in terms of the Beremin model) on temperature and neutron fluence is a priori introduced, for example, as it is made in [21]. Thus, it should be concluded that at present the Prometey model based on criterion (2) is the only model that allows the prediction of lateral shift of $K_{JC}(T)$ curves without additional assumptions.

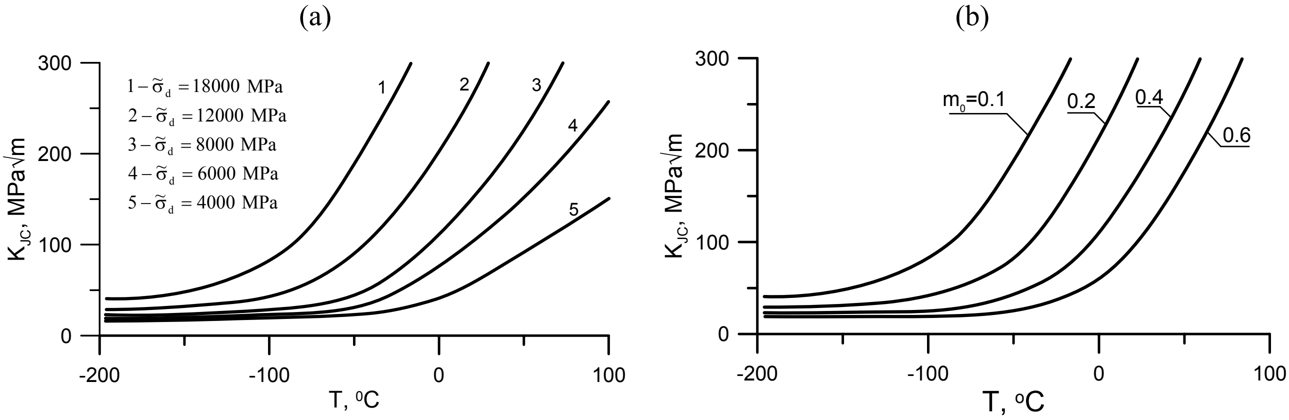


Figure 3. The $K_{JC}(T)$ curves calculated with the Prometey model for decreasing parameter $\sigma_d$ and $m_0$ = const (a) and for increasing function $m_0(F)$ and $\sigma_d$ =const (b). (Specimen thickness is 25 mm and $P_f = 0.5$.) [14]

## 4.2. On minimum value of fracture toughness for RPV materials

The Master Curve [24]] and Unified Curve [22] methods use the Weibull statistics to describe the scatter in $K_{JC}$ results and the effect of specimen thickness on $K_{JC}(T)$ curve. The fracture probability $P_f$ and $K_{JC}$ is connected by the Weibull distribution function [24]

$$P_f = 1 - \exp\left[-\left(\frac{K_{JC} - K_{min}}{K_0 - K_{min}}\right)^4\right], \tag{6}$$

where $K_0$ is a scale parameter depending on the test temperature and specimen thickness; deterministic parameter $K_{min}$ is the minimum value of fracture toughness.

As a common case, $K_{min}$ may depend on the test temperature. In [24] it was accepted that $K_{min}(T)$ = const and on the basis of large number of experimental data sets for RPV steels over low temperature range (on the lower shelf of $K_{JC}(T)$ curve) it was determined that $K_{min}$= 20 MPa√m.

In the present section the results are represented on calculation of $K_{min}$ as a function of temperature, $K_{min}(T)$, for RPV steels on the basis of local criteria (1) and (2). These results are calculated with the Beremin and Prometey models based on criteria (1) and (2) respectively.

To obtain $K_{min}(T)$ the following considerations are used. (1) When calculating with the Beremin model the value of $K_{min}$ is taken as stress intensity factor $K_I$ when conditions (1a) and (1b) are satisfied for the unit cell nearest to the crack tip.

(2) When calculating with the Prometey model the value of $K_{min}$ is taken as $K_I$ when condition (2a) is satisfied for the unit cell nearest to the crack tip. (For cracked specimens from RPV steels brittle fracture is controlled by microcrack nucleation condition (2a).)

(3) For both models the unit cell size $\rho_{uc}$ is taken to be equal to 50 μm. The dependence $\sigma_Y(T)$ was taken as obtained in [9] for RPV steel in initial conditions. For the Beremin model two values of the parameter $S_0$ are taken as more typical for RPV steels [9, 10]: $S_0$= 1300 and 1500 MPa. For the Prometey model three values of the controlling parameter $\sigma_{d0}$ are taken as $\sigma_{d0}$ = 1300, 1500 and 2000 MPa that corresponds to various estimations of $\sigma_{d0}$ in [9, 10]. It is important to note that when calculating $K_{min}(T)$ the critical parameters are taken as calibrating from specimens without cracks, i.e. from smooth tensile specimens.

(4) Stress-and-strain fields near the crack tip are calculated by approximated solution of elastic-plastic problem represented in [9].

The calculation results are shown in Figure 4 for cracked specimens with 50 mm in thickness from RPV steel in initial condition. (Here maximum temperature is restricted by T=100°C as for higher temperatures brittle-to-ductile transition occurs for this steel.) As seen from Figure 4a $K_{min}(T)$ decreases over low temperature range and becomes constant at T>0°C for curve 1 and T>-75°C for curve 2. This behavior is caused by different controlling conditions in (1) for these temperature ranges: Eq. (1a) gives decreasing part and Eq. (2) gives const part. As seen from Figure 4b according to the Prometey model $K_{min}(T)$ is practically constant or slightly increasing (from 18 to 25 MPa√m for curve 3).

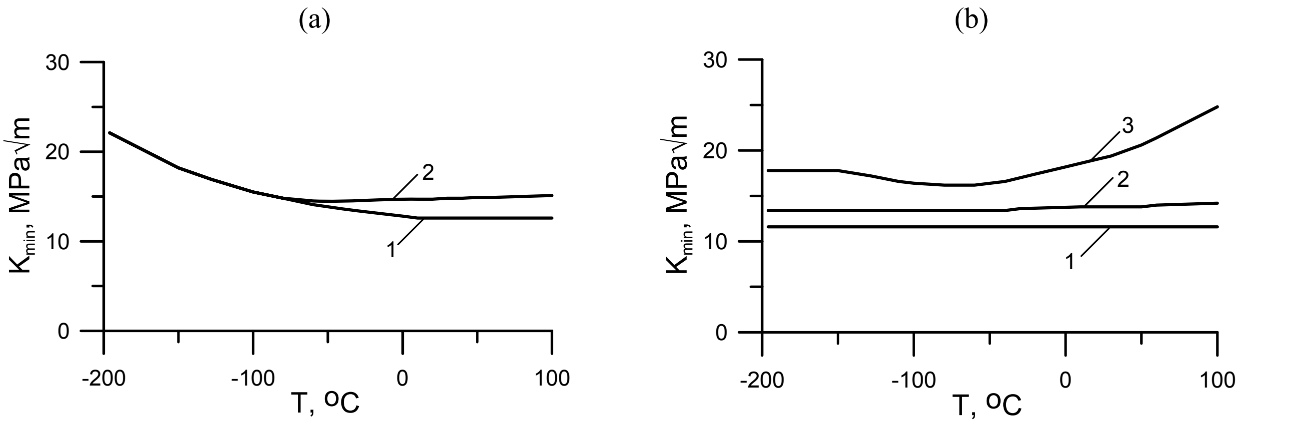


Figure 4 The $K_{min}(T)$ curves calculated with the Beremin (a) and Prometey (b) models model for RPV steel: (a): 1 - $S_0$=1300; 2 - $S_0$=1500 MPa; (b): 1 - $\sigma_{d0}$ = 1300; 2 - $\sigma_{d0}$ =1500; 3 - $\sigma_{d0}$ =2000 MPa.

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