A multi-scale brittle fracture modelling for irradiated RPV materials

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Abstract. Further development of local approach for irradiated RPV materials is considered through (i) the analysis of applicability of available local cleavage fracture criteria for radiation embrittlement modelling; (ii) the analysis of the effect of radiation-induced damage on the processes of cleavage microcrack nucleation and propagation; (iii) the discussion of the effect of the material hardening caused by cold work and neutron irradiation on brittle fracture.

Introduction

Intensive development of local approach models in fracture mechanics over the past 25 years was stimulated, to a great extent, by the needs of assessing the structural integrity of reactor pressure vessels (RPV). A key input to calculation of the RPV structural integrity is the fracture toughness versus temperature curve, $K_{JC}(T)$. However, classical (global) fracture mechanics requires full-sized cracked specimen testing for adequate determination of $K_{JC}(T)$ over wide temperature range that is practically impossible for irradiated RPV materials. This explains the reason why local approach models are now intensively developed. The application of a local approach allows the prediction of the $K_{JC}(T)$ curve for an irradiated steel on the basis of small-sized specimen testing. Further development of local approach for irradiated RPV materials has to include in the consideration radiation-induced defects so that a multi-scale fracture modelling may be achieved.

Thus, a multi-scale approach in brittle regime should be understood as a chain from radiation-induced defects to local cleavage fracture criterion and up to fracture on a macro-scale.

At present there is a gap in this chain as there is no link of radiation-induced defects with fracture criteria, more precisely, with the critical parameters in criteria. From the physical viewpoint it means that there is no link of radiation defects with the processes of cleavage microcrack nucleation and propagation.

The main objective of the present paper is to try to bridge this gap. The present paper reviews briefly local cleavage fracture criteria, considers their application for prediction of brittle fracture of RPV materials and develops a multi-scale modelling for irradiated RPV materials through the consideration of the mechanisms of the radiation damage effect on cleavage microcrack behaviour.

Local Cleavage Fracture Criteria

Formulations and Physical Background. Currently two local cleavage fracture criteria are mainly used in brittle fracture modelling. Traditional formulation is written as [1-3]

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

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

where σ_{eq} is the equivalent stress, σ_Y, the yield stress, σ_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 [4-8]. This formulation is based on the following considerations summarizing the main physical and mechanical features of cleavage fracture [8].

1. Three processes have to occur for cleavage fracture: (i) microcrack nucleation, (ii) microcrack start (unstable growth of a nucleus microcrack up to the nearest barrier), (iii) microcrack propagation (unstable growth of microcrack through various barriers). Cleavage fracture on a macro-scale may be controlled by each of these processes.
2. Continuous nucleation of cleavage microcracks occurs when the nucleation condition has been satisfied. For ferritic materials, microcrack nucleation happens on carbides or at other particles.
3. A cleavage microcrack nucleus is a sharp microcrack with the tip radius equal to lattice parameter, microcrack start is determined by Griffith's condition.
4. If $\sigma_1 \geq S_0$ (S_0 – microcrack start stress) at the moment of sharp microcrack nucleation, this microcrack starts. If $\sigma_1 < S_0$, the nucleus microcrack is blunted by dislocation emission from its tip and transforms into a void. Such a microcrack-void can not be an initiator of cleavage fracture: to initiate cleavage fracture it is necessary to nucleate a new sharp microcrack.
5. A propagating cleavage microcrack may be arrested by various barriers such as grain boundaries, microstresses, slip bands, and boundaries of dislocation substructure arising under plastic deformation. The parameter S_C is interpreted as the stress required for microcrack propagation through various barriers.

The proposed formulation is written in the form [4-8]

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

$$\sigma_1 \geq S_C(æ)\,, \tag{2b}$$

where the effective stress is $\sigma_{eff} = \sigma_{eq}-\sigma_Y$, i.e. σ_{eff} is the strain hardening controlled by plastic strain, $æ = \int d\varepsilon^p_{eq}$ is the accumulated plastic strain, $d\varepsilon^p_{eq}$ is the equivalent plastic strain increment, σ_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 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 σ_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 [4-8]

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

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

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

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 σ_{Ys} is the temperature-dependent component of the yield stress.

From the physical viewpoint, Eqs. (1a) and (2a) are the condition for cleavage microcracks nucleation, and Eqs. (1b) and (2b) – the condition 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 σ_d. It is important that the plastic strain when microcrack is nucleated may exceed 0.2% and increases with the temperature growth [6, 8].

The important consequences follow from this difference between (1a) and (2a). In criterion (2) two critical parameters - S_C and σ_d may control cleavage fracture and this depends on material properties and loading conditions, mainly, on the ratio S_C/σ_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) [7, 8]. 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 Fracture Properties on a Macro-scale. Prediction of brittle fracture on a macro-scale in a stochastic manner may be performed on the basis of criterion (1) with the Beremin model [9] and on the basis of criterion (2) with the Prometey model [10]. Both models use the Weibull statistics for stochastic parameters and the weakest link model to predict the brittle fracture on a macro-scale. The Beremin model takes the critical stress S_C as stochastic parameter, and the Prometey model uses two stochastic parameters - σ_d and S_C.

Applicability of the Beremin model was widely studied for various materials [4-7, 11]. It was found that there are difficulties with the use of this model for medium and high strength steels, in particular, for RPV steel [4-8]. It was shown [12, 13] 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, the parameter K_{JC} does not practically depend on temperature that is in contradiction with test results.

Some attempts (for example, [14]) were undertaken to reform the Beremin model by introduction of the temperature dependence for the parameter S_C (or the parameter σ_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 [4-7] 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).

It should be noted that there is no problem with prediction of $K_{JC}(T)$ curve with the Prometey model 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. The Prometey model was verified by application to RPV steels in various conditions (initial, irradiated and highly embrittled) [7, 10, 13, 15].

Applicability of Local Fracture Criteria for Radiation Embrittlement Modelling

From the viewpoint of modelling of the effect of irradiation on cleavage fracture it is important, first of all, to analyse the applicability of criteria (1) and (2) in terms of having an adequate link with the physical mechanisms of radiation embrittlement.

At present, for RPV steels three types of radiation-induced damage are found: *matrix damage* caused by radiation-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 [16-20]. These radiation damages and mechanical properties of irradiated RPV materials are linked as follows. The matrix damage and precipitates result in an increase of σ_Y as they affect the dislocation mobility. An increase of σ_Y is caused by an increase of the athermal component σ_{YG} of the yield stress. Segregation of impurities, as a rule, is not associated with changes in σ_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. The role of non-hardening mechanism in embrittlement of RPV steels is clearly seen also from the data on post-irradiation annealing [16] which show that after post-irradiation annealing of 2Cr-Ni-Mo-V RPV steel at temperature 475^{o}C the full recovery of σ_Y occurs but DBTT recovers not fully.

When analysing the applicability of the above criteria for radiation embrittlement modelling it should be taken into account that the critical stress S_C does not practically depend on neutron fluence (at least, for transcrystalline fracture) [13, 18-20]. This follows from experimental and theoretical results. From a physical viewpoint, radiation-induced lattice defects and precipitates result in not decreasing the critical stress S_C. It follows from the model [6] 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 σ_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 σ_1 due to increase of σ_Y.

Criterion (2) contains two parameters - σ_d and σ_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 σ_d. This process is 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 [17] and result in decreasing the interface strength (i.e. in decreasing σ_d). This means that the nucleation of cleavage microcracks becomes easier compared with unirradiated steel.

It seems that an application of criterion (1) may be justified from viewpoint of empirical correlations for irradiated materials. Indeed, the correlation ΔT_{tr} vs $\Delta\sigma_Y$ is known to follow qualitatively from the Yoffe scheme [2] based on criterion (1). However, if we estimate on the basis of criterion (1) for steels with various σ_Y the response in ΔT_{tr} caused by the same value of $\Delta\sigma_Y$ we obtain the predicted curve 1 shown in Fig. 1 that differs essentially from curve 2 constructed according to empirical correlation. This correlation is used in the form $\Delta T_{tr} = 1070\left(\dfrac{\Delta\sigma_Y/\sigma_Y}{4.53+\Delta\sigma_Y/\sigma_Y}\right)^{0.857}$ [21] as obtained by treatment of large experimental data set for RPV materials. This correlation shows that ΔT_{tr} for $\Delta\sigma_Y$=const decreases when σ_Y increases as schematically shown by curve 2 in Fig. 1. These quite different trends for curves 1 and 2 mean that criterion (1) cannot be used for radiation embrittlement modelling for RPV steels.

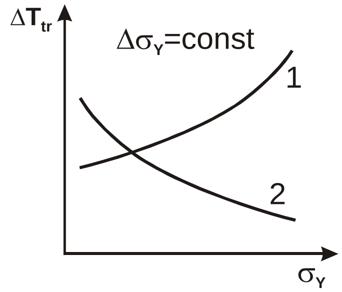


Fig. 1. ΔT_{tr} vs. σ_Y for $\Delta\sigma_Y$=const according to criterion (1) (curve 1) and to test results (curve 2) (scheme).

Thus, the above analysis shows that radiation embrittlement of RPV materials cannot be considered as a result of the material hardening only. Criterion (2) in principle allows the description of radiation embrittlement caused by both material hardening and the initiator weakening. The physical models of the effect of radiation defects on cleavage microcrack nucleation provide the dependence of the critical stress σ_d on neutron fluence F [10, 22-24].

The Effect of Radiation Defects on Cleavage Microcrack Nucleation

The Effect on the Critical Stress σ_d. For RPV steels in unirradiated condition the cleavage microcrack initiators are known to be mainly globular carbides [25]. In [10, 24] it has been concluded on the basis of microstructural data [16, 17] that 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. Therefore, it is necessary to answer a question, how the radiation-induced defects may affect cleavage microcrack nucleation on carbides.

The possible mechanisms of the effect of radiation-induced defects on cleavage microcrack nucleation may be divided into two groups. The first group includes the mechanisms [23, 24] connected with decreasing the strength of carbide-matrix interface. One mechanism of decreasing σ_d is the impurity (P) *segregation* on ferrite-carbide interfaces [17, 19, 20] caused by the P diffusion accelerated by irradiation. A simple physical model for impurity segregation on carbide-matrix interface under irradiation [23] provides the dependence $\sigma_d(F)$ in the form

$$\sigma_d(F) = \sigma_d^0 \exp\left[-\alpha_1\left(\frac{F}{F_0}\right)^m\right] \quad (6)$$

$\sigma_d^0 = \sigma_d(F=0)$; α_1, m - constants for a given condition of irradiation; F_0 - the normalizing factor.

Another mechanism for decreasing σ_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 σ_{nuc} being less than σ_{nuc} for unirradiated material. By another words, dislocation loop "works" as a wedge between carbide and matrix. The dependence $\sigma_d(F)$ for this mechanisms is similar to Eq. (6) [24].

The second group includes the mechanism connected with easier formation of dislocation pile-ups near microcrack initiators due to radiation defects [10, 22, 24]. This process may be described by increasing the probability of dislocation pile-up formation in material with high concentration of barriers for dislocations. As known, for most dislocation models of microcrack nucleation, the influence of concentration of barriers is not taken into account. In particular, the condition (2a) does not contain any parameter characterizing concentration or distribution of barriers. Then for material with barriers of the same strength and different concentration, nucleation of a microcrack happens for the same σ_{nuc}. It is clear that this result is valid for the plane barriers only. It may is expected that for compact barriers (for example, globular carbides) microcrack nucleation is easier for material with high concentration of barriers as the probability of formation of dislocation pile-up increases. This probability increases for irradiated material as radiation defects are barriers for dislocations and their concentration increases with increasing the neutron fluence. The detailed consideration is given in [10, 22, 24] 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. It has been shown that for this mechanism the dependence $\sigma_d(F)$ may be represented as exponential one.

Thus, the type of the dependence $\sigma_d(F)$ is similar for all the considered mechanisms. This conclusion is important for engineering application when predicting the transformation of $K_{JC}(T)$ curve as a function of neutron fluence with the Unified Curve method [23] based on the Prometey model. The Unified Curve for the fracture probability P_f=0.5 and specimens thickness B=25 mm is described by equation $K_{JC(med)}(T) = K_{JC}^{shelf} + \Omega \cdot \left(1 + \tanh\left(\frac{T-130}{105}\right)\right)$ (K_{JC}^{shelf} =26 MPa√m, units of K_{JC}: MPa√m, and T: ^oC). From the exponential function for $\sigma_d(F)$ a functional type of the only calibrated parameter $\Omega(F)$ is found as exponential [23].

The Effect of Radiation Defects on "Driving Force" σ_{nuc}. Evidently, except for the effect of radiation defects on σ_d these defects result also in an increase of "driving force" $\sigma_{nuc} \equiv \sigma_1 + m_{T\varepsilon} \cdot \sigma_{eff}$ in condition (2a) that is caused by increasing σ_1 as a result of increasing σ_Y. In principal, it is possible that the coefficient $m_{T\varepsilon}= m_0\sigma_{Ys}(T)m_\varepsilon(æ)$ also increase due to increase of the parameter m_0. (Other parameters, σ_{Ys} and σ_{eff}, do not practically vary under irradiation.) 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. However, at present, the experimental data are too few to allow the analysis of this effect and do not allow the determination of the dependence $m_0(F)$.

Calculation of $K_{JC}(T)$ curves with the Prometey model for increasing function of $m_0(F)$ has shown (Fig. 2a) that for this case the transformation of $K_{JC}(T)$ curve may be practically exactly described as a lateral shift. The $K_{JC}(T)$ curves calculated with the Prometey model for decreasing σ_d are represented in Fig. 2b. (Here the decrease in σ_d models increasing neutron fluence.). Transformation of $K_{JC}(T)$ curves shown in Fig. 2b is in good agreement with test results for irradiated (embrittled) RPV materials: a lateral temperature shift for small degree of embrittlement and a variation in the $K_{JC}(T)$ curve shape for high degree [8, 13, 15, 23].

It is appropriate to mention here 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 these models by the $\sigma_Y(T)$ dependence. Thus, it should be concluded that at present the Prometey model based on criterion (2) is the only model that allows the prediction both of pure lateral shift and variation in shape for $K_{JC}(T)$ curves.

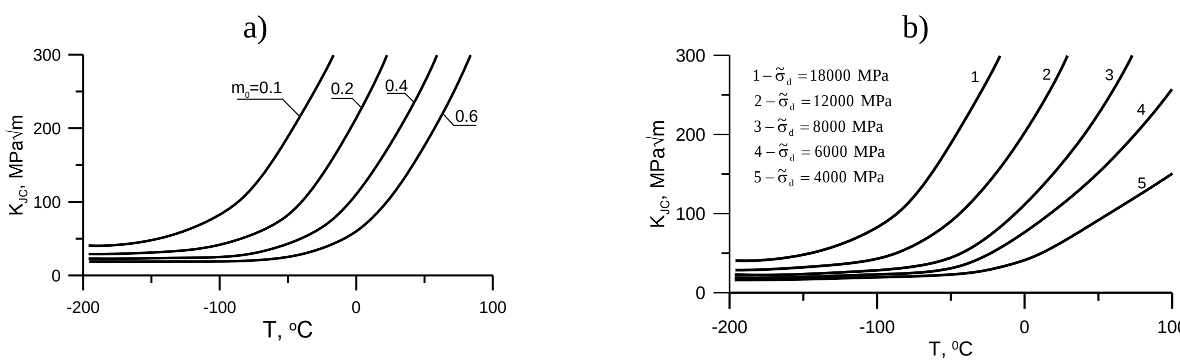


Fig. 2. The $K_{JC}(T)$ curves for RPV steel calculated with the Prometey model for increasing m_0 and $\tilde{\sigma}_d$ =const (a) and for decreasing $\tilde{\sigma}_d$ and m_0=const (b). (P_f=0.5 and B=25 mm.)

Some Consequences

Radiation Embrittlement of Low-strength and High-strength Steels. The performed study has shown that radiation embrittlement is caused by two factors: (1) the increase of σ_1 as a result of the σ_Y increase (the mechanical factor) and (2) the decrease of the critical stress σ_d (the physical factor). The contributions of the mechanical and physical factors in irradiation embrittlement may depend on material properties, mainly, on the ratio S_C/σ_Y. This ratio determines which physical process – microcrack nucleation or propagation - controls brittle fracture of cracked specimen. For steels with large value of S_C/σ_Y, for example, for low-strength steel ($S_C/\sigma_Y^{20} \geq 4$), condition $\sigma_1=S_C$ is satisfied only when σ_1 increases significantly due to strain hardening, i.e. after large plastic strain. By this the microcrack nucleation condition (2a) has been satisfied earlier. It means that brittle fracture of cracked specimens from these steels is mainly controlled by condition (2b) and hence, the mechanical factor predominates in radiation embrittlement for this steel.

For medium-strength steel with the ratio $S_C/\sigma_Y^{20} \leq 2.5$, in particular, for RPV steel, condition $\sigma_1=S_C$ near the crack tip is met earlier than condition (2a), therefore the microcrack nucleation mainly controls brittle fracture of cracked specimens. When fracture is considered in the probabilistic statement both conditions control brittle fracture. Therefore both the mechanical and physical factors determine radiation embrittlement of RPV steel.

The physical and mechanical factors of radiation embrittlement have been analyzed in detail in [22]. In Fig. 3 the calculation results obtained with the Prometey model are shown for medium-strength 2Cr-Ni-Mo-V RPV steel ($S_C/\sigma_Y^{20} \approx 2.3$) and for low-strength M16C steel ($S_C/\sigma_Y^{20} \approx 4$). Here the relative contribution of the mechanical factor is shown in terms of the ratio $(\Delta T_{tr})_\sigma/\Delta\sigma_Y$. The value $(\Delta T_{tr})_\sigma$ is the temperature shift determined for K_{JC}=100 MPa√m from the $K_{JC}(T)$ curves calculated for $\Delta\sigma_Y$=var. As seen from Fig. 6 for M16C steel the ratio $(\Delta T_{tr})_\sigma/\Delta\sigma_Y \approx (0.4\div0.6)^{o}$C/MPa that is in good agreement with test results for the irradiated steel [22]. It means that for low strength steel the contribution of the material hardening in embrittlement is dominant. Here decreasing σ_d does not change brittle fracture resistance controlled by microcrack propagation.

At the same time for RPV steel the mechanical factor provides only 20% of total temperature shift as $(\Delta T_{tr})_\sigma/\Delta\sigma_Y \approx 0.1^{o}$C/MPa (see Fig. 3) and for RPV steels with low impurity content experimental values of $\Delta T_{tr}/\Delta\sigma_Y$ are from 0.4 to 0.5 ^{o}C/MPa [18].

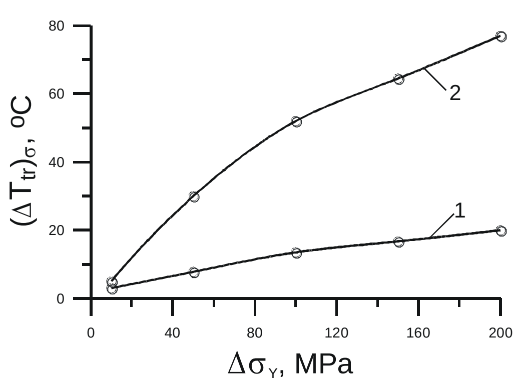


Fig. 3. The temperature shift $(\Delta T_{tr})_\sigma$ caused by the material hardening $\Delta\sigma_Y$ for medium-strength 2Cr-Ni-Mo-V RPV steel (curve 1) and for low-strength M16C steel (curve 2).

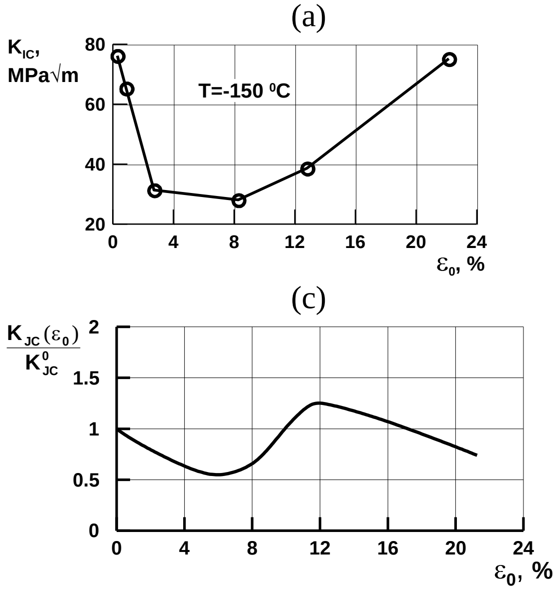
The Effect of Cold Work on Brittle Fracture. Analysis of this issue shows clearly a link of local criterion with the embrittlement mechanisms as cold work is known to result in the yield stress increase as well as neutron irradiation. However, the test results [26] show that material hardening caused by cold work may result in both decreasing and increasing fracture toughness (Fig. 4a) as distinct from the hardening caused by irradiation always resulting in material embrittlement.

Analysis of applicability of criteria (1) and (2) for prediction of the cold work effect on brittle fracture shows the following [24]. From viewpoint of criterion (1) (that is represented by the Yoffe scheme) the degree of embrittlement is the same for the same increment of σ_Y both for irradiated and cold worked materials. (This is valid for plastic pre-strain ε_0 <10% that does not result in S_C growth.) At the same time experimental data show that the ratios $\Delta T_{tr}/\Delta\sigma_Y$ are different for irradiated and cold-worked materials. For example, for medium-strength cold-worked A533B steel $\Delta T_{tr}/\Delta\sigma_Y$ =(0.1÷0.15)°C/MPa for ε_0 <10% [27] and for irradiated steel $\Delta T_{tr}/\Delta\sigma_Y \cong$0.5°C/MPa [18]. It should be noted that this ratio $\Delta T_{tr}/\Delta\sigma_Y$ obtained experimentally for A533B steel when σ_Y increases due to cold work, is close to $(\Delta T_{tr})_\sigma/\Delta\sigma_Y$ calculated on the basis of criterion (2).

Calculations have shown also that according to criterion (1) the dependence $K_{JC}(\varepsilon_0)$ is always decreasing (Fig. 4b) as the ratio S_C/σ_Y decreases when ε_0 increases. Thus, criterion (1) does not describe a difference in the behaviour of irradiated and cold-worked material. It is connected with that criterion (1) takes into account only one physical process of brittle fracture – cleavage microcrack propagation and does not describe adequately microcrack nucleation.

Stress-and-strain controlled criterion (2) shows why the material hardening caused by cold work may result in both decreasing and increasing fracture toughness. The reason consists in the opposite influence of cold work on microcrack nucleation. Cold work may result in easier microcrack nucleation as it decreases the critical stress σ_d due to arising of the internal stresses on carbide-matrix interface and increasing the probability of dislocation pile-up formation. Both mechanisms are caused by increasing the dislocation density and similar those proposed for radiation defects. On the other side, cold work may make more difficult microcrack nucleation that is caused by decreasing a number of possible initiators. It is because cleavage microcracks are nucleated on weak initiators during pre-strain but not propagate. Afterwards, these initiators cannot nucleate new microcracks and, hence, the probability of microcrack nucleation decreases.

(a) (c)



(b)

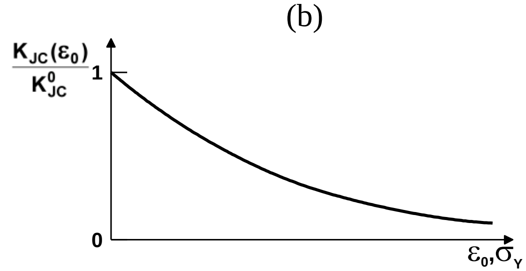


Fig. 4. The effect of cold work on fracture toughness: (a) test results for RPV steel [26]; (b) the relative dependence $K_{JC}(\varepsilon_0)/K_{JC}^0$ predicted with criterion (1); (c) the relative dependence $K_{JC}(\varepsilon_0)/K_{JC}^0$ predicted with the Prometey model based on criterion (2).

This opposite influence of cold work leads to non-monotonic variation of σ_d. As a result, the dependence $K_{JC}(\varepsilon_0)$ may be also non-monotonic. In Fig. 4c the relative dependence $K_{JC}(\varepsilon_0)/K_{JC}^0$

(where K_{JC}^{0} is fracture toughness for initial condition) is shown as resulted from the Prometey model for RPV steel [24]. It is seen that the dependence $K_{JC}(\varepsilon_0)$ may be both decreasing and increasing function depending on the plastic pre-strain value.

On the Intergranular Fracture Mode. Brittle fracture of RPV steels after irradiation and post-irradiated annealing may occur by mixed trans- and inter-granular mechanism. The most difficult for interpretation observation is that a fraction of intercrystalline fracture in the fracture surface is not always correlated with mechanical properties. For example, for irradiated steels the mechanical parameters (T_{tr} and σ_Y) speak about significant embrittlement, however a fraction of intercrystalline fracture is usually small (never larger than 20% of fracture surface) [16]. On the contrary, after annealing of the irradiated materials, significant recovery of mechanical properties is observed, at the same time, a fraction of intercrystalline fracture may increase [16].

These findings may be explained with criterion (2) and the proposed mechanisms of the effect of radiation-induced defects on the critical stress σ_d. In deterministic statement, σ_d is determined by minimum value of the two values σ_d^{grain} and $\sigma_d^{boundary}$ - the strengths of carbide-matrix interfaces for carbides locating in a grain and on grain boundaries. The fracture mode may be transcrystalline or intercrystalline that depends on which value is less. In probabilistic statement, mixed transcrystalline and intercrystalline fracture may be expected if the difference of σ_d^{grain} and $\sigma_d^{boundary}$ is not large.

The interpretation for variation of the parameters σ_d^{grain} and $\sigma_d^{boundary}$ is schematically shown in Fig. 5, 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 (steels for WWER-440 and WWER-1000 RPV respectively). For unirradiated steels (Fig. 5a) σ_d^{grain} is less than $\sigma_d^{boundary}$ as brittle fracture occurs mainly by the transcrystalline mechanism. The difference of σ_d^{grain} and $\sigma_d^{boundary}$ is less for 2Cr-Ni-Mo-V steel as some fraction of intercrystalline fracture is observed. This difference for two steels is connected with their different sensitivity to temper embrittlement: 2.5Cr-Mo-V steel is not sensitive and 2Cr-Ni-Mo-V steel shows a tendency to temper embrittlement which is caused by P diffusion, intensified by Ni, to phase and grain boundaries.

For irradiated steels (Fig. 5b) σ_d^{grain} and $\sigma_d^{boundary}$ decrease. For carbides located on grain boundary, segregation of impurities and arising of internal stresses caused by dislocation loops occur more intensively. Nevertheless, for irradiated 2.5Cr-Mo-V steel the condition $\sigma_d^{grain} < \sigma_d^{boundary}$ is more possible, so that transcrystalline brittle fracture is more typical. For irradiated 2Cr-Ni-Mo-V steel the situation is possible when $\sigma_d^{grain} \approx \sigma_d^{boundary}$ as this steel is more sensitive to the P segregation so that mixed trans- and intercrystalline brittle fracture is observed.

After post-irradiation annealing at T_{ann}=475^{o}C (Fig. 5c) phosphorus segregations dissociate in a grain only, and do not dissociate on grain boundary [16, 20]. Radiation-induced dislocation loops and precipitates may be dissociated practically completely both in a grain and on grain boundary. Therefore the σ_d^{grain} value increases up to the value for the unirradiated condition but the $\sigma_d^{boundary}$ value remains 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 σ_d^{grain} and $\sigma_d^{boundary}$ become close, although this annealing results in full recovery of the mechanical properties (T_{tr} and σ_Y). For 2Cr-Ni-Mo-V steel the situation is possible when intergranular carbides become "weakest link" and intercrystalline fracture predominates, and although σ_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 the value $\sigma_d^{boundary}$ that does not recover fully.

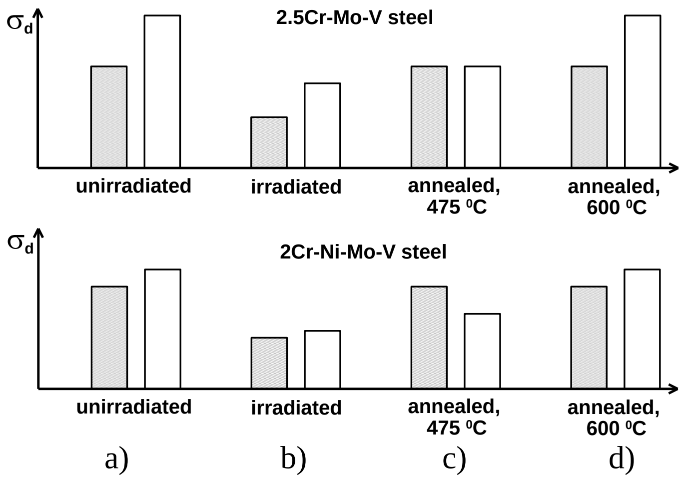


Fig. 5. Variation of σ_d^{grain} (shaded bar) and $\sigma_d^{boundary}$ (open bar) for RPV steels in various conditions: unirradiated (a), irradiated (b), after post-irradiation annealing (c) and (d).

Annealing at $T_{ann}\approx 600^oC$ (Fig. 5d) results in full recovery of $\sigma_d^{boundary}$ as dissociation of grain boundary P segregations occurs at this temperature [16, 28] and the situation become close to the unirradiated condition (Fig. 5a). Full recovery of the mechanical properties and the fracture modes is observed. New SEM study results in [29] confirm this conclusion.

Conclusions

1. Applicability of available cleavage fracture criteria has been analyzed for material embrittlement modelling. Stress-controlled fracture criterion used in the RKR and Beremin models and stress-and-strain controlled fracture criterion used in the Prometey model have been considered. It has been shown that stress-and-strain controlled criterion provides adequate prediction for all known properties of embrittlement of a material.

2. It has been found that radiation embrittlement of steels is caused both by the material hardening (mechanical factor) and the microcrack initiator weakening (the physical factor). The contributions of the mechanical and physical factors in irradiation embrittlement depend on material properties, mainly, on the ratio S_C/σ_Y. This ratio determines which physical process – microcrack nucleation or microcrack propagation - controls brittle fracture of cracked specimen.

3. Irradiation-induced defects are shown to result in more easy cleavage microcrack nucleation. The proposed physical-and-mechanical models of the effect of radiation defects (dislocation loops, precipitates and segregation) on cleavage microcrack nucleation are reviewed.

4. As distinct from stress-controlled fracture criterion, stress-and-strain controlled criterion explains why the material hardening caused by cold work may result in both decreasing and increasing fracture toughness. The reason consists in the opposite effect of cold work on cleavage microcrack nucleation that may become easier or more difficult depending on pre-strain value.

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