MODELLING OF FRACTURE FOR AUSTENITIC MATERIALS WITH FERRITIC PHASE ON THE BASIS OF LOCAL APPROACH

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ABSTRACT
Main regularities of fracture of two-phase materials are considered. An approach is proposed for description and prediction of fracture of two-phase materials. Modelling of fracture processes is performed on the basis of local approach as applied to austenitic cladding for reactor pressure vessel.

1 INTRODUCTION

When welding austenitic and ferritic steels by austenitic materials, two-phase microstructure appears to be typical for cladding and weld metals. Percent fractions of two phases for cladding and weld metals may be usually 92…98% of austenitic phase and 8…2% of ferritic phase.

Fracture of such two-phase materials differs from fracture of both austenitic and ferritic steels and may be characterised by a series of particularities. First of all, over some temperature range, critical fracture parameters such as fracture strain for tensile specimens and critical value of J-integral, J_C, for cracked specimens decrease as temperature decreases. This behaviour is known to be typical for brittle fracture of ferritic steels. At the same time, when testing cracked specimens over the same temperature range, stable growth of crack is observed and J_R-curves may be determined. This type of behaviour tells about ductile fracture of a material. In the present report, an attempt is undertaken to explain and describe this phenomenon by using local approach methods.

2 MAIN REGULARITIES OF FRACTURE OF AUSTENITIC CLADDING AND THEIR RELATION WITH MICROSTRUCTURE

Fracture of austenitic materials with ferritic phase differs from fracture of FCC metals in the temperature dependencies of critical fracture parameters. The critical fracture parameters of FCC metals are known to be weakly sensitive to temperature. When testing specimens from austenitic cladding over wide temperature range, two temperature ranges are revealed. Over low temperature range, critical fracture parameters such as critical strain for tensile specimens and critical value of J-integral, J_C, for cracked specimens decrease strongly as temperature decreases. This behaviour is typical for brittle fracture of BCC metals. Over higher temperature range, the critical fracture parameters for austenitic cladding become weakly sensitive to temperature that is usually observed for ductile fracture. The temperature dependence of critical fracture strain for austenitic cladding is schematically shown in Fig. 1.

At the same time, it should be noted that fracture of austenitic cladding over low temperature range ($T < T_{tr}$) differs from brittle fracture of BCC metals. When testing cracked specimens from austenitic cladding over this temperature range, stable growth of crack is observed and J_R-curves may be determined. This stability of fracture process is usually observed for ductile fracture.

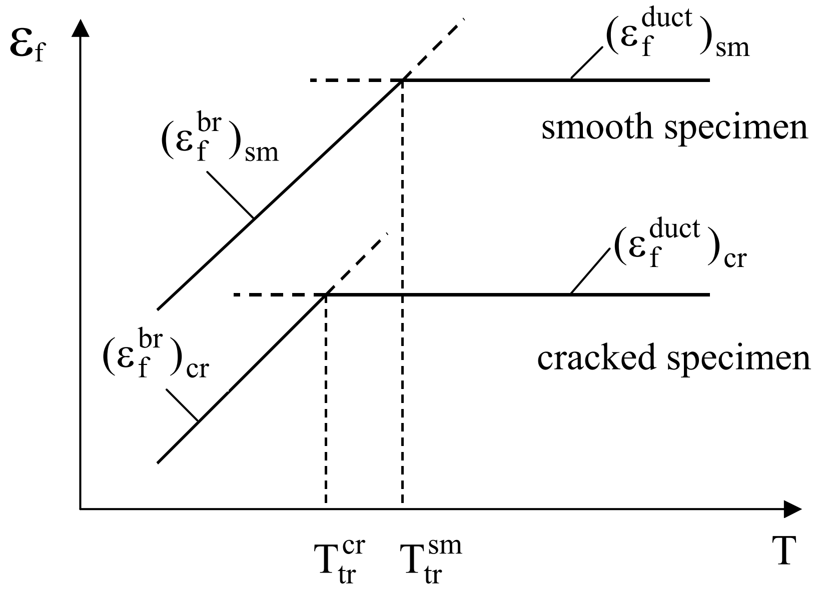

Figure 1: The temperature dependencies of fracture strain for smooth tensile and cracked specimens from austenitic cladding (scheme).

The above particularities of fracture of austenitic cladding may be related with microstructure. Austenitic cladding being spatially heterogeneous material consists of austenitic matrix with ferritic phase located mainly on grain boundaries, and may contain some fraction of σ–phase that depends on tempering condition.

Fractographical studies of fracture surfaces show that fracture of cladding may occur by two mechanisms (Nikolaev [1]): (i) transcrystalline ductile fracture caused by nucleation, growth and coalescence of voids (classical ductile fracture which is typical for both BCC and FCC metals); (ii) intercrystalline quasi-brittle fracture caused by nucleation and propagation of cleavage microcrack in ferritic phases located on austenitic grain boundaries. The ductile fracture mechanism is realised over temperature range $T > T_{tr}$ and intercrystalline quasi-brittle fracture is observed over temperature range $T < T_{tr}$ (Fig. 1). Ductile fracture of cladding and other two-phase metals may be described by known ductile fracture models.

In the present report, an approach is proposed which allows the description and prediction of intercrystalline quasi-brittle fracture of austenitic cladding. This approach may be also used for other two-phase metals.

3 LOCAL CRITERION OF INTERCRYSTALLINE QUASI-BRITTLE FRACTURE FOR TWO-PHASE METALS

Intercrystalline fracture of two-phase metals is caused by nucleation and propagation of cleavage microcrack in BCC phases (or in other brittle phases) located on FCC grain boundaries. These cleavage microcracks are arrested by FCC phases and unstable growth of cleavage microcracks on a macro-scale does not occur. FCC phase ligaments are ruptured on ductile mechanism.

Criterion of intercrystalline quasi-brittle fracture for two-phase metals may be formulated on the basis of modification of local criterion of brittle fracture for BCC metals as proposed in Margolin [2, 3]. This criterion consists of two conditions

$$\sigma_1 + m_T \cdot m_\varepsilon \cdot \sigma_{eff} \geq \sigma_d \tag{1a}$$

$$\sigma_1 \geq S_C(æ) \tag{1b}$$

where σ_1 is the maximum principal stress, the effective stress is $\sigma_{eff} = \sigma_{eq} - \sigma_Y$, σ_{eq} is the equivalent stress, σ_Y is the yield stress, $æ = \int d\varepsilon^p_{eq}$ is Odqvist's parameter, $d\varepsilon^p_{eq}$ is the equivalent plastic strain increment, σ_d is the strength of some particles on which cleavage microcracks are nucleated, $S_C(æ)$ is the critical brittle fracture stress, the parameters m_T and m_ε are the known functions of temperature and plastic strain respectively. The parameter m_T characterises the dislocation pile-up blunting (the width of dislocation pile-up) and may be written as

$$m_T(T) = m_0 \sigma_{Ys}(T) \tag{2}$$

where m_0 is a constant which may be experimentally determined and σ_{Ys} is the temperature-dependent component of the yield stress. The parameter m_ε is determined by dislocation pile-up length arrested by cleavage-nucleating particles. As a common case, this length is equal to grain size for initial state of a material and decreases for deformed state due to deformation substructure formation.

Condition (1a) is the nucleation condition for cleavage microcracks. Condition (1b) is the propagation condition for cleavage microcracks. It was shown in Margolin [2, 3] that for RPV steels, brittle fracture on a macro-scale may be controlled by condition (1a) or condition (1b) that depends on stress triaxiality. For example, brittle fracture of tensile smooth specimens is controlled by condition (1b) and brittle fracture of cracked specimens – condition (1a).

The above criterion may be modified for two-phase metals by using the following considerations.

1. It is assumed that dislocation pile-ups which may generate cleavage microcracks in BCC phase, locate in FCC grains. Length of these pile-ups may be taken not to depend on plastic strain as formation of dislocation substructure in FCC grain occurs for sufficiently large plastic strain. Thus, the parameter m_ε in eqn (1a) may be taken as constant m_ε = const.

2. The parameter $\sigma_{eff} = \sigma_{eq} - \sigma_Y$ for FCC metals is approximated by equation $\sigma_{eff} = A_0\sqrt{æ}$, where A_0 is material constant.

3. BCC phases are not plastically deformed as compared with FCC grains. It means that dislocation substructure in BCC phases is not formed over the considered temperature range. Taking into account that the critical brittle fracture stress S_C is stress for propagation of cleavage microcrack in BCC phases through dislocation substructure and the increase in $S_C(æ)$ is caused by plastic deformation, we may assume that $S_C = S_0$ = const.

Thus, criterion of intercrystalline quasi-brittle fracture for two-phase metals may be formulated in the form

$$\sigma_{nuc} \equiv \sigma_1 + m \cdot \sigma_{Ys} \cdot A_0\sqrt{æ} \geq \sigma_d \tag{4a}$$

$$\sigma_1 \geq S_0 \tag{4b}$$

where m and S_0 are constants ($m = m_0 \cdot m_\varepsilon$).

This formulation allows the explanation of the described particularities of fracture of two-phase metals and modelling of fracture process for cracked specimens from austenitic cladding that is presented hereafter. It is important to emphasise that for the considered two-phase materials, intercrystalline quasi-brittle fracture on a macro-scale is controlled by condition (4a) for both smooth tensile and cracked specimens. This consideration is drawn from test results of smooth tensile specimens at $T<T_{tr}$. These results show that $\max\sigma_1$ (which for brittle fracture is usually taken as the fracture stress) is a function of temperature. It means that at $\sigma_1 = S_0$ condition (4a) is not satisfied and fracture of specimen does not happen. It is clear that for cracked specimens for which stress triaxiality increases, condition (4a) also controls intercrystalline fracture.

It may be shown from eqn (4a) that for intercrystalline quasi-brittle fracture (at $T < T_{tr}$, see Fig. 1) the critical strain ε_f for tensile specimens increases as temperature increases. Indeed, the parameters σ_{Ys} and A_0 decrease as temperature increases. To satisfy the fracture condition (4a), plastic strain should be increased. In other words, the critical strain ε_f^{br} increases over temperature range of intercrystalline fracture. Over temperature range of ductile fracture (at $T > T_{tr}$, see Fig. 1), the critical strain ε_f^{duct} does not depend practically on T. Some temperature $T=T_{tr}$ may be found for which $\varepsilon_f^{br} = \varepsilon_f^{duct}$ and the transition from intercrystalline quasi-brittle to transcrystalline ductile fracture occurs (Fig. 1). One more interesting result is predicted from condition (4a). When testing notched or cracked specimens, the transition temperature T_{tr}^{cr} is lower than T_{tr}^{sm} for smooth specimens. Indeed, stress triaxiality σ_m/σ_{eq} affects ε_f^{br} and ε_f^{duct} by different manner. As seen from condition (4a), ε_f^{br} decreases weakly when σ_m/σ_{eq} and, hence, σ_1 increases. On the other hand, according to Hancock [4] $\varepsilon_f^{duct} \sim \exp\left(-1.5\dfrac{\sigma_m}{\sigma_{eq}}\right)$. Then for the critical strain near the crack tip $(\varepsilon_f)_{cr}$, we have $(\varepsilon_f^{duct})_{cr} = (\varepsilon_f^{duct})_{sm}/k_1$ and $(\varepsilon_f^{br})_{cr} = (\varepsilon_f^{br})_{sm}/k_2$ and $k_1>k_2$ (here $(\varepsilon_f)_{sm}$ is the critical strain for tensile smooth specimen). Illustration for this phenomenon is also shown in Fig. 1 and confirmed by test results.

4 MODELLING OF CRACK GROWTH AND PREDICTION OF J_R-CURVES FOR INTERCRYSTALLINE QUASI-BRITTLE FRACTURE

On the basis of the proposed local criterion for intercrystalline quasi-brittle fracture, modelling of crack growth may be performed by using a procedure in Margolin [5]. According to this procedure, ductile crack growth is simulated as the consecutive fracture of some unit cells located near the crack tip on the crack extension line. For the considered intercrystalline fracture, unit cell with size ρ_{uc} has to be defined in the following way (Fig. 2.). Unit cell is taken to consist of a region of FCC phase that is ruptured on ductile mechanism and a region of BCC phase in which cleavage microcracks are nucleated and propagate. Thus, intercrystalline crack grows through periodical ductile and brittle regions. This periodical structure reflects schematically structure of fracture surfaces that was studied by SEM for cracked specimens from austenitic cladding.

Intercrystalline crack growth may be represented as follows. When crack is arrested by ductile region, J-integral increases up to some value for which condition (4a) is satisfied over some distance r_f (Fig. 2) $\sigma_{nuc}\big|_{r=r_f} = \sigma_d$. When this condition has been satisfied, cleavage microcrack is nucleated and propagates up to ductile regions 1 and 2. Ductile fracture of ligaments 1 occurs practically without increases of load. As a result, crack extents on value ρ_{uc}. Under subsequent loading this fracture process repeats.

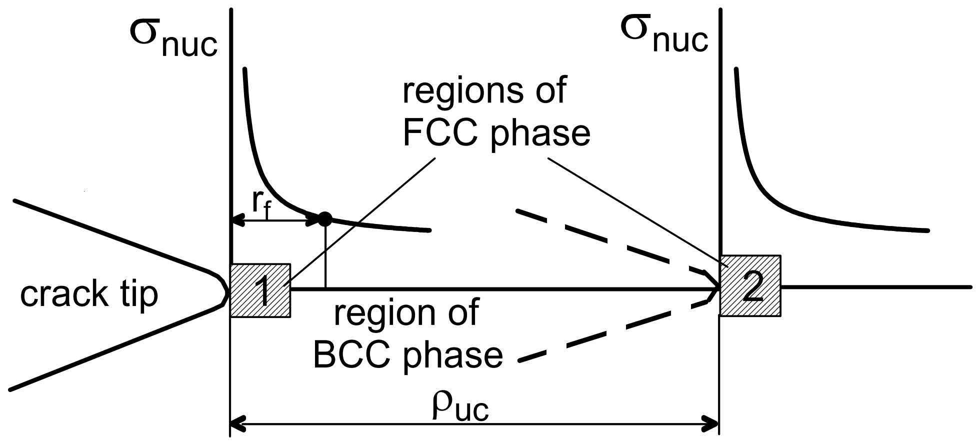

Figure 2: Simulation of crack growth for two-phase materials.

Two unknown parameters m and σ_d in eqn (4a) may be calibrated from test results of smooth and notched tensile specimens over temperature range $T<T_{tr}$ (assuming that the dependencies $\sigma_{Ys}(T)$ and $A_0(T)$ are known).

Simulation of crack growth was performed by FEM as applied to SEB specimen from austenitic cladding for PRV. For each crack extension on value ρ_{uc} a value of J-integral was calculated for which condition $\sigma_{nuc}\big|_{r=r_f} = \sigma_d$ was satisfied and, as a result, J_R-curve was predicted. Size of unit cell was taken to be equal to ρ_{uc}=0.1 mm and the parameter r_f=0.01 mm. These values were determined on the basis of measurement of sizes of structural ductile and brittle elements that was performed by SEM analysis of fracture surfaces for cracked specimens. In Fig. 3 the calculated J_R-curves are presented at various temperatures. It is seen from this figure that values of J_C decrease as temperature decreases. The calculated results are in good agreement with experimental data.

Space does not permit a detailed consideration and discussion of the used procedures and obtained results. Two remarks should be noted here. First of all, it is of interest to note that calculations predict decreasing of the parameters J_C and $\frac{dJ_R}{da}$ with temperature decreasing (this is connected with decreasing of the plastic strain æ near the crack tip for which eqn (4a) is satisfied). Some temperature T_{unst} may be found such that for $T \leq T_{unst}$ $\frac{\partial J}{\partial a} \geq \frac{dJ_R}{da}$. It means unstable crack growth through the whole specimen. This explains the experimentally observed large jumps of crack that result in specimen fracture. Value of T_{unst} depends on type and sizes of cracked specimens. The performed calculations showed that for the investigated cladding, such a type of fracture is predicted at $T_{unst}\approx$-200…-180°C for SEB specimens with sizes of 10x10x55 mm. This conclusion is confirmed by test results in Nikolaev [1], where unstable fracture of specimens from austenitic cladding is observed.

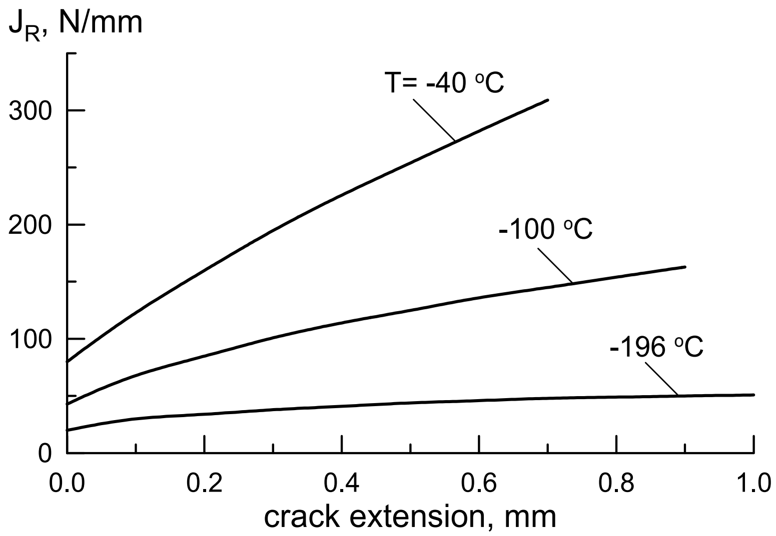

Figure 3: The calculated J_R-curves for austenitic cladding over temperature range of intercrystalline quasi-brittle fracture.

Second remark concerns the effect of crack front length on the critical fracture parameters. As known, brittle fracture of BCC metals is obeyed the weakest link theory. For the considered intercrystalline quasi-brittle fracture, the weakest link theory can not be applied as brittle fracture of one unit cell does not result in fracture of the whole specimen. It follows from this consideration that for the considered two-phase materials, the effect of crack front length on fracture toughness is practically absent.

In conclusion, it should be noted that the proposed approach may be successfully used for austenitic cladding in irradiated condition, for that the parameter σ_d in eqn (4a) should be decreased.

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