The Flux Effect On Radiation Embrittlement Of WWER RPV Materials

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Abstract. Investigation of the influence of neutron flux on a radiation embrittlement of reactor pressure vessel (RPV) materials of WWER by means of comparison of the test results of surveillance specimens (SS) and the test results of the specimens irradiated by high neutron flux in the frame of research programs (SRP) is presented. The analysis of flux effect is carried out at various mechanisms of embrittlement of WWER RPV materials.

Introduction

In many cases it is necessary to estimate RPV material embrittlement on the basis of accelerated irradiation with high neutron flux. Therefore a question arises if it is possible to use the test results of specimens irradiated by high flux for prediction of embrittlement of RPV material during the operation.

In literature there is a contradictory enough information on the neutron flux effects on radiation embrittlement of steels with different chemical compositions and concentration levels of impurity and alloy elements [1-7]. For example, in some papers it is shown that the flux effect is absent [1] or that the sign of this effect depends on a flux value [2]. In other papers a positive flux effect is discussed [3], i.e. with increasing neutron flux a ductile to brittle transition shift obtained due to neutron irradiation ΔT_F increases. In still other papers [4-6] the authors discuss a negative flux effect when an increase of neutron flux results in a decrease of ΔT_F. In the review presentation [7] it is shown that a neutron flux either results in a decrease or does not affect on ΔT_F.

The aim of this paper is the investigation of a neutron flux as applied to WWER-1000 and WWER-440 RPV materials, as well as the analysis of the radiation embrittlement mechanisms resulting in demonstration of this effect.

To perform this aim experimental data on radiation embrittlement obtained by Research Centre “Kurchatovsky institute” in frame of the surveillance specimens programs and the data base obtained in the frame of the research programs of CRISM “Prometey” are compared.

Main mechanisms of radiation embrittlement

Radiation-induced defects result in embrittlement of a material by two basic mechanisms: hardening mechanism and non-hardening mechanism [8]. The feature of hardening mechanism is that material embrittlement is accompanied with its hardening, i.e. with yield strength growth. This mechanism is connected with radiation-induced dislocation loops and precipitates. For non-hardening mechanism, material embrittlement is not accompanied with its hardening. This mechanism is mainly connected with impurity segregation.

Material embrittlement by hardening mechanism is caused by the mechanical and physical factors. The mechanical factor consists in increasing of stresses near macro-crack tip (postulated flaw in RPV) and as a result, start and propagation of cleavage microcrack (Griffith’s crack) occur at lower value of K_J. The physical factor consists in arising of inner self-balancing stresses that makes cleavage microcrack nucleation easier.

Material embrittlement by non-hardening mechanism is caused mainly by the physical factor. Impurity segregations locate on any interfacial surfaces (for example, on carbide-matrix interfaces) and grain boundaries. Microcracks are nucleated usually on such boundaries. It is clear that the phosphorus segregation results in decreasing the interface strength and, hence, the nucleation of cleavage microcracks becomes easier compared with unirradiated steel.

The effect of radiation defects on material embrittlement is schematically shown in Fig. 1.

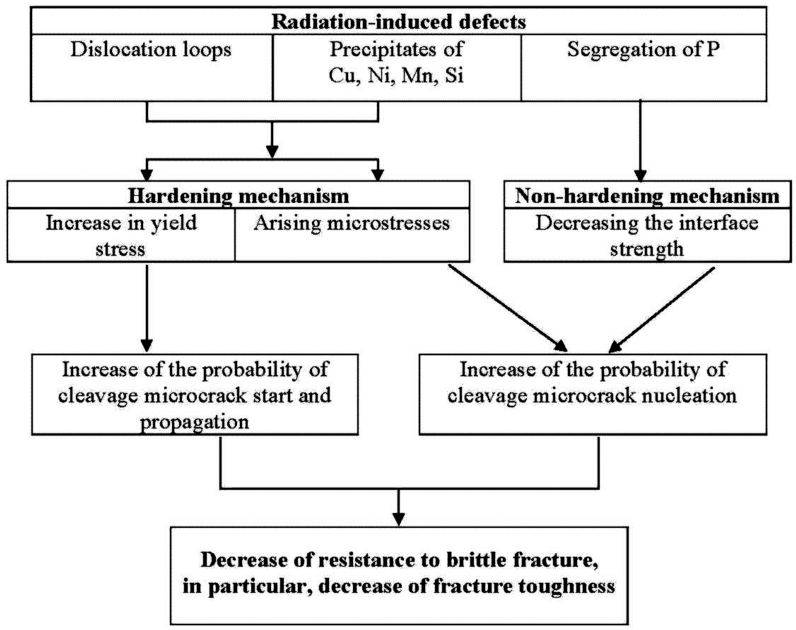

Fig. 1. The effect of radiation defects on material embrittlement (scheme).

The flax effect is observed when the material embrittlement ΔT_k depends not only on the neutron dose but on time. Time factor can affect through the thermal aging mechanisms (carbon thermal aging and segregations) and kinetics of radiation-induced defects (dislocation loops, precipitates and also segregations).

We investigate two RPV steels: 2.5Cr-Mo-V for WWER-440 and 2Cr-Ni-Mo-V for WWER-1000. 2.5Cr-Mo-V steel and its weld metal which practically don’t contain Ni and contains higher concentration of Cr as compared with 2Cr-Ni-Mo-V steel. That’s why these materials are not sensitive to carbide thermal aging. As these materials contain Mo and don’t contain Ni they are not practically sensitive to thermal aging due to phosphorous segregation at least for $T \leq 350^{o}C$. Old WWER-440 RPV’s have a high phosphorus and copper content, but second generation of WWER-400 RPV’s are clean enough of these impurities. Hence, for WWER-440 RPV materials the flux effect is connected with kinetic of radiation-induced defects (mainly dislocation loops and phosphorous segregations and cooper precipitations).

2Cr-Ni-Mo-V steel and its welds metal (with nickel content) are sensitive to carbide thermal aging. These materials are clean enough of phosphorus and copper content. Hence, for WWER-1000 RPV materials the flux effect is connected with thermal aging or/and kinetic of radiation damage (mainly dislocation loops and precipitates of Ni, Mn, Si).

Thus, WWER-440RPV’s materials with low Cu content and high P content can be used for analysis of flux effect when dislocation loops and segregations control embrittlement of material. WWER-1000 RPV’s materials (having low Cu and P content) can be used for flux effect analysis when dislocation loops and precipitations (in the first place Ni, Mn, Si) control material embrittlement. Let us consider the above material with point of view flux effect analysis.

Trend curve for prediction of embrittlement of WWER-1000 RPV materials and irradiation conditions

According to [9, 10] trend curve of material embrittlement of WWER-1000 is presented in the form:

$$\Delta T_{к}(F\ ,t) = \Delta T_{t}(t) + \Delta T_{F}(F\ )\ ,\ °C \tag{1}$$

where ΔT_t is a transition temperature shift resulted from the effect of irradiation temperature (thermal ageing); ΔT_F is a transition temperature shift depending only on of neutron irradiation.

The value ΔT_F is described by the formula:

$$\Delta T_F = A_F \cdot \left(\frac{F}{F_0}\right)^n,\ °C \tag{2}$$

where F is neutron fluence, F_0=1.0·10^{22} n/m^2, A_F is a coefficient of radiation embrittlement, n is an exponent.

The value ΔT_t is described by the formula:

$$\Delta T_t\ (t) = \left(\Delta T_t^{inf} + b_T \exp\left(\frac{t_T - t}{t_{OT}}\right)\right) \cdot th\left(\frac{t}{t_{OT}}\right),\ °C \tag{3}$$

where t is an ageing time, ΔT_t^{inf} is a transition temperature shift at t→∞; t_{OT}, t_T and b_T are material constants dependent on a temperature of ageing.

The parameters of equation (3) for a given class of materials at an irradiation temperature T_{irr}=290-300°C are given in paper [9, 10].

The experimental data on radiation embrittlement in the frame of research program were obtained as a result of accelerated irradiation, i.e. at small time of exposure. The time of irradiation in the WMR-M reactor did not mainly exceed ~1000 hours. Therefore in the case of accelerated irradiation the contribution of temperature ageing ΔT_t in ΔT_k, according to formula (3), is insignificant and it can be neglected ($\Delta T_t \approx 0$°C).

For the data obtained in the frame of the SS program the value ΔT_t calculated by formula (3) cannot be neglected in view of the ageing at the temperature of RPV operation.

The value of the transition temperature shift obtained as a result of neutron irradiation ΔT_F is described by formula (2).

According to [9, 10] the values *n* and A_F in dependence (2) can be calculated at T_{irr}=290-300°C using the following formula:

for base metal: n = 0.8; A_F = 1.45, °C, (4)

for weld metal: n = 0.8; $A_F = \alpha_1 \exp(\alpha_2 \cdot C_{eq})$, °C, (5)

where

$$C_{eq} = \begin{cases} C_{Ni} + C_{Mn} - \alpha_3 C_{Si},\ \text{если}\ C_{Ni} + C_{Mn} - \alpha_3 C_{Si} \geq 0 \\ 0, \qquad\qquad\qquad \text{если}\ C_{Ni} + C_{Mn} - \alpha_3 C_{Si} < 0 \end{cases}, \tag{6}$$

where α_1 = 0.703; α_2 = 0.883; α_3 = 3.885, C_{Ni}, C_{Mn}, C_{Si} are contents of nickel, manganese and silicon in weight %.

Irradiation of SS was carried out at neutron flux ≈10^{15} n/m^2·s and temperature 295± 5°C. Irradiation of specimens in the frame of research programs was carried out at neutron flux ≈10^{17} n/m^2·s and temperature 180÷310°C. To analyse the flux effect we should have at least two data sets for different fluxes at identical irradiation temperature. That’s why it is necessary to recalculate the values ΔT_F from the actual irradiation temperature to a temperature of 300°C for materials irradiated in the frame of research program. The temperature equal to the upper bound (namely 300°C) of the SS irradiation temperature was chosen in order to estimate the maximum possible negative influence of the neutron flux effect.

According to papers [9, 10] for base metal of WWER-1000 RPV the coefficient A_F can be taken as independent on the content of alloying elements. At the same time it should be noted that formulas (5) and (6) for calculation of the coefficient of irradiation embrittlement are valid over a wide range of changes in a parameter C_{eq}. In particular, with the content of the considered elements that are characteristic of 2Cr-Ni-Mo-V steel ($C_{Ni} \approx 1.3\%$, $C_{Mn} \approx 0.4\%$, $C_{Si} \approx 0.25\%$) calculation by formulas (5) and (6) gives the value $A_F = 1.34$°C, which approaches the value $A_F = 1.45$°C obtained for this steel on a direct processing of the experimental data. Therefore formulas (5) and (6) can be used both for weld metal and base metal.

The coefficient of radiation embrittlement with regard for irradiation temperature effect (for weld metal and base metal) can be presented in the form:

$$A_F = A_F^{temp} \cdot A_F^{chem}, \text{°C} \tag{7}$$

where A_F^{temp} is the temperature part of the coefficient of radiation embrittlement, depending only on an irradiation temperature, A_F^{chem} is the part of the coefficient of radiation embrittlement, depending only on nickel, manganese and silicon content.

Comparing formulas (7) and (5) and processing the experimental data we obtain:

$$A_F^{temp} = 31.83 \cdot \exp(-0.01307 \cdot T_{irr}), \text{°C} \tag{8}$$

$$A_F^{chem} = \exp(\alpha_2 \cdot C_{eq}) \tag{9}$$

Each experimental point was normalized in the form: $A_F^{temp} = \Delta T_F / (A_F^{chem} \cdot (F/F_0)^n)$

Comparison of $A_F^{temp}(T_{irr})$ and experimental data is presented in Fig. 2.

For materials irradiated in the frame of research programs using the obtained dependence (8) recalculation of the values ΔT_F from the actual irradiation temperature to a temperature of 300°C was made in the following way:

$$\Delta T_F^{300} = \Delta T_F \cdot \frac{A_F^{temp}(300°)}{A_F^{temp}(T_{irr})}, \text{°C} \tag{10}$$

where ΔT_F is a transition temperature shift at $T_{irr} < 300$°C.

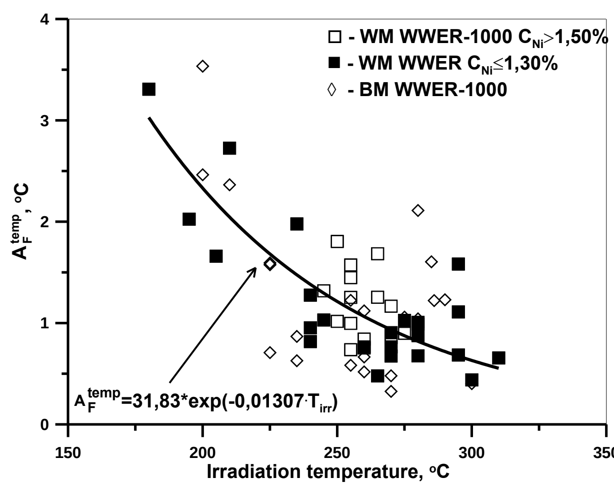

Fig. 2. Dependence of A_F^{temp} on an irradiation temperature for the combined set that includes base and weld metals: ■, □, ◊ are the experimental data obtained in the frame of research programs; line is the approximation of the experimental data by dependence (8).

Trend curve for prediction of embrittlement of WWER-440 RPV materials and irradiation conditions

According to [11, 12] for WWER-440 RPV materials the value $\Delta T_t = 0°C$ and the value ΔT_F is described by the formula (2). Authors of this paper in [11] proposed the following trend curve for base metal in form (2), where the value $n = 0.483$, while a coefficient of radiation embrittlement A_F is calculated by formula:

$$A_F = 0.651 + 358 \cdot (0.046 \cdot C_{Cu} + (C_P - 0.002)),\ °C \qquad (11)$$

where C_P and C_{Cu} are phosphorus and copper content in weight %.

According to [11, 12] for weld metal of 2.5Cr-Mo-V steel in dependence (2) the values n = 1/3, while a coefficient of radiation embrittlement A_F is calculated by formula:

$$A_F = \begin{cases} 6.4 + 610 \cdot (C_P + 0.07 \cdot C_{Cu} - 0.01), \text{ if } C_P + 0.07 \cdot C_{Cu} \geq 0.01 \\ 6.4 \qquad\qquad\qquad\qquad\qquad\quad , \text{ if } C_P + 0.07 \cdot C_{Cu} < 0.01 \end{cases},\ °C \qquad (12)$$

Irradiation of SS was carried out at neutron flux $\approx 10^{15}$ n/m^2·s. Irradiation of specimens in the frame of research program was carried out at neutron flux $\approx 10^{17}$ n/m^2·s. Irradiation temperature is equal to 270°C.

Analysis of neutron flux effect on material embrittlement on different mechanisms

In the general case radiation embrittlement of metal proceeds by the following mechanisms (see Fig. 1).

1. Hardening of material at the expense of dislocation loop formation. We will call it a mechanism “A”.
2. Hardening of material at the expense of formation of barriers in the form of precipitates (clusters) enriched with Cu, Ni, Mn and Si – it is a mechanism “B”.
3. Segregation of elements at any interphase boundaries or grain boundaries, which results in a decrease of strength of these boundaries. It is a mechanism “C”. In most practically important cases a segregating element is phosphorus.

For detection of flux effect each point of an array belonging to the same flux was normalized as $\Delta T_F / A_F$. Then, new array is treated by function in the form:

$$\frac{\Delta T_F}{A_F} = \eta \cdot \left(\frac{F}{F_0} \right)^n \qquad (13)$$

where η is the coefficient indexing flux effect: if $\eta^l \approx \eta^h$ then flux effect is absent, if $\eta^l < \eta^h$ then flux effect is positive, if $\eta^l > \eta^h$ then flux effect is negative. η^l, η^h correspond to low and high neutron flux respectively.

Let us consider the influence of neutron flux on the mechanism “A”. Under the influence of irradiation new dislocation loops are generated and annihilated. In the absence of the annihilation process of dislocation loops density ρ_d would be proportional to neutron fluence F. Based on this assumption, i.e. on the assumption of a linear connection of ρ_d with F and taking into consideration that $\Delta\sigma_{0.2} \sim \sqrt{\rho_d}$ [13], as well as taking $\Delta T_F \sim \Delta\sigma_{0.2}$ we can obtain a trend curve in the form $\Delta T_F \sim \sqrt{F}$. Such form of the dependence was experimentally established in one of the pioneer works [14].

At the same time, the rate of dislocation loops annihilation depends in the general case on the current ρ_d and time [13, 15]. As time decreases, the number of annihilated dislocation loops decreases. Consequently, an increase of neutron flux at a given fluence results in a growth of ρ_d and hence in an increase of hardening and embrittlement of material. Thus, the accelerated irradiation under the research programs should result in a higher value of ΔT_F compared to the transition temperature shift obtained on SS.

It should be noted that the mechanism "A" will exert a dominating influence on radiation embrittlement of material, if the mechanisms "B" and "C" do not happen. Such situation is typical for the steels in which elements causing formation of precipitates and segregations are absent. The 2.5Cr-Mo-V steel with a low content of impurity elements – copper and phosphorus - can be considered as the indicated steels.

Comparison of the test results of SS (obtained by Research Centre "Kurchatovsky institute") [11, 16, 17] and SRP (obtained by CRISM "Prometey") shows that with a low content of copper (C_{Cu}≤0.11%) and low content of phosphorus (C_P≤0.013%) the influence of flux has not been revealed (see Fig. 3). The dependences $\Delta T_F(F)$ with different neutron fluxes (the fluxes differed by ≈ 100 times) practically coincide (for SS η^l=1.00; for SRP η^h=0.98). Thus, the following conclusion can be drawn: if a dominating mechanism of radiation embrittlement is hardening at the expense of dislocation loops formation (mechanism "A"), then the flux effect can be neglected.

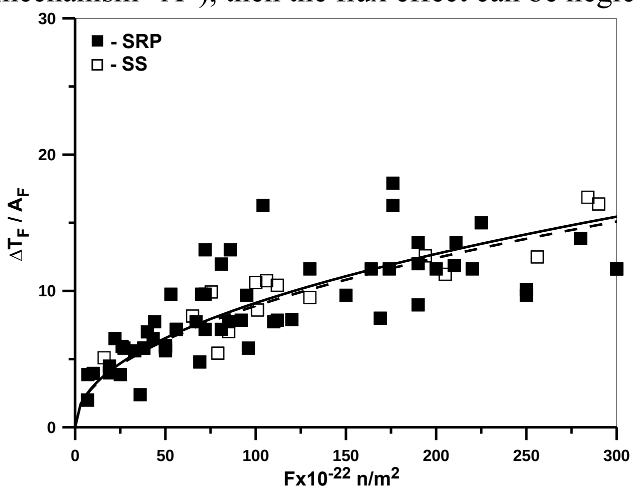

Fig. 3. Normalized dose dependences for the 2.5Cr-Mo-V steel with contents C_{Cu}≤0.11% and C_P≤0.013%: □ are SS data ($\varphi \approx 10^{15}$n/m^2·s); ■ are SRP data ($\varphi \approx 10^{17}$n/m^2·s); solid line is the dependence (13) for SS (η^l=1.00); dashed line is the dependence (13) for SRP (η^h=0.98).

Now let us consider the influence of neutron flux on the mechanism "C". In the general case the segregation of some elements on sinks, those are interphase boundaries and grain boundaries, proceeds at the expense of a diffusion process. Diffusion, as a typical thermoactivated process, depends on temperature and time. A typical segregation process resulting in embrittlement of metal and depending on temperature and time is temper brittleness [18].

It is necessary to note that, there is representative enough experimental data that demonstrates a fundamental difference of the conditions of segregation processes realization under thermal ageing and neutron irradiation.

Paper [19] presents the experimental data demonstrating embrittlement of the 2Cr-Ni-Mo-V steel at the expense of nonhardening mechanism "C". In [19] it was shown that, at T_{irr}=50-80^{o}C a strong influence of phosphorus on ΔT_F is observed, while $\Delta\sigma_{0,2}$ is not changed with an increase of phosphorus content.

The fact of a strong influence of phosphorus on ΔT_F at T_{irr}=50-80°C cannot be explained from the point of view of the classical diffusion processes, since at such irradiation temperature the diffusion processes proceed undoubtedly slower by several orders than at T_{irr}=290-300°C. Hence, phosphorus is translated to interphase boundaries and (or) grain boundaries by some other mechanism that differs from the classical thermoactivated diffusion process. And this other process is directly related not to time, but to a neutron dose or, as a first approximation, to neutron fluence. Most probably translation of phosphorus to interphase boundaries proceeds through primary cascades that are formed on collision of a neutron with metal. Translation of phosphorus inside a cascade proceeds considerably faster than the one by the classical diffusion mechanism. Cascades are an initial cause of different types of crystalline lattice flaws on irradiation. Since lattice flaw formation at the expense of cascades proceeds preferably at interphase boundaries or grain boundaries [20] (where the bond between atoms of a matrix is weakened), concentration of phosphorus at these boundaries is connected with the parameter that controls the number of cascades, that is, it is connected with neutron fluence. A similar mechanism of an accelerated translation of phosphorus at the expense of a set of vacancies formed on neutron irradiation is considered in papers [21, 22].

Thus, under neutron irradiation the concentration of phosphorus at interphase boundaries is mainly controlled by neutron fluence and depends slightly on the time of exposure, that is, on neutron flux.

The experimental proof of the absence of neutron flux influence on ΔT_F at the expense of realization of the mechanism "C" is the data on radiation embrittlement of weld metals of WWER-440 RPV. The considered data base on ΔT_F for weld metals of the 2.5Cr-Mo-V steel includes the data with a high content of phosphorus over the range of $0.007\% \le C_P \le 0.032\%$.

The performed calculations show that for SS $\eta^l=0.97$, while for SRP $\eta^h=0.99$ (see Fig. 4); therefore even with high content of phosphorus ($C_P \le 0.032\%$) in weld metal of the 2.5Cr-Mo-V steel an effect of the flux has not been revealed.

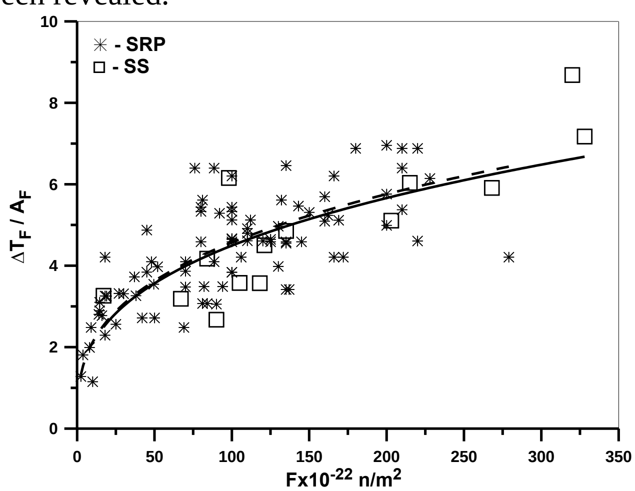

Fig. 4. Normalized dose dependences for weld metals of the 2.5Cr-Mo-V steel with contents $0.02 \le C_{Cu} \le 0.23\%$ and $0.007 \le C_P \le 0.032\%$. □ are SS data ($\varphi \approx 10^{15}n/m^2\cdot$s); * are SRP data ($\varphi \approx 10^{17}n/m^2\cdot$s); solid line is dependence (13) for SS ($\eta^l=0.97$); dashed line is dependence (13) for SRP ($\eta^h=0.99$).

Let us consider the influence of neutron flux when mechanism "B" predominates.

Treatment of experimental data for SS (obtained by Research Centre "Kurchatovsky institute") and SRP (obtained by CRISM "Prometey") by formula (13) gives: $\eta^l \approx \eta^h$ as for both base metal and weld metal. From obtained results the following conclusion can be drown: for base metal flux effect is practically absent; for weld metal flux effect is remarkable. Similar tendencies on the influence of a flux effect were obtained with the other investigation data base (accelerated irradiation) in paper [5].

Neutron flux for base metal does not practically exert any influence on the change of ΔT_F. Evidently, a different sensitivity of weld metal and base metal to the flux is connected with different total contents of nickel and manganese ($C_{Ni}+C_{Mn}$). Indeed for base metal the total content $C_{Ni}+C_{Mn}$ = 1.6%, for weld metal with low nickel content ($C_{Ni}\leq1.3\%$) this value $C_{Ni}+C_{Mn}$ =1.9%, for weld metal with high nickel content ($C_{Ni}>1.5\%$) $C_{Ni}+C_{Mn}$ =2.6%.

Paper [6] presents a large data base IVAR on reactor pressure vessel steels of different chemical compositions.

Form processing of the experimental data of paper [6] it was obtained that a content increase of such elements as Ni and Cu in reactor pressure vessel materials results in an increase of the difference between yield strength increments obtained under irradiation by low ($\phi1$) and high ($\phi2$) fluxes. This fact undoubtedly indicates that there is an influence of these elements on the flux effect. An increase of manganese content in the considered materials results in a less significant increase of the difference between yield strength increments. It should be noted that it is difficult to judge the influence of C_{Mn} on an increase of a flux effect shown in paper [6], since the main data base corresponds to the variations of C_{Mn} over a very narrow range. The analysis of experimental data in [6] showed that an increase of phosphorus content does not result in the flux effect. This fact agrees well with our conclusion that follows from the analysis of the mechanism “C”.

As it was shown above degree of neutron flux effect on material embrittlement depends on sum ($C_{Ni}+C_{Mn}$). When $C_{Ni}+C_{Mn}$ >1.9%, flux effect begins to be remarkable. Let us compare this conclusion with data in [6].

Fig. 5 shows the dependence of the difference of yield strength increments by low and high flux levels on the total content of manganese and nickel for the experimental data of paper [6]. Let us assume that the flux begins to affect hardening when the difference of yield strength increments exceeds the error in determination of this difference – 10 MPa. Then from Fig. 2 it is seen that the flux effect starts manifesting itself at least with $C_{Ni}+C_{Mn}\geq1.79$ %, which corresponds to the level of 10 MPa.

With $C_{Ni}+C_{Mn}$=2.6 % typical for weld metal of WWER-1000 RPV (with $C_{Ni}\geq1.5$ %) the value of the difference of yield strength increments is equal to 22.1 MПa, which indicates that there is a pronounced enough influence of the flux effect. Thus the presented investigation results of the flux effect on radiation embrittlement of WWER-1000 RPV materials correspond to the tendencies that follow from the experimental data [6].

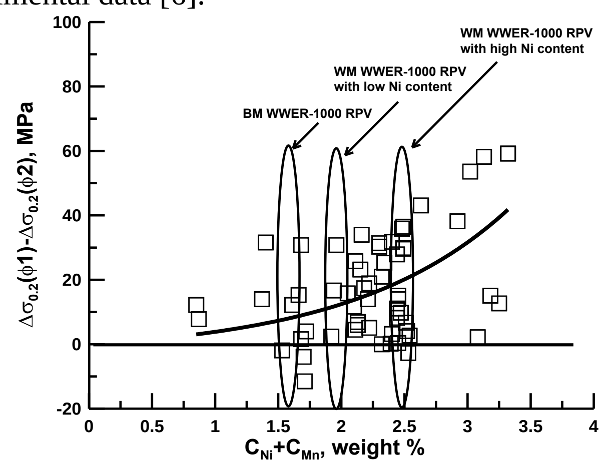

Fig. 5. Dependence of the difference of yield strength increments by low and high flux on ($C_{Ni}+C_{Mn}$): □ are the experimental data; solid line is approximation of the experimental data by the dependence $\Delta\sigma_{0.2}(\phi1)-\Delta\sigma_{0.2}(\phi2)=3.128\cdot[\exp(0.802\cdot(C_{Ni}+C_{Mn}))-1]$.

Let us consider the influence of a neutron flux from point of microstructure changes. Paper [23] presents the data on irradiation of steel SA533B by the same neutron fluence and different neutron fluxes. In [23] the chemical composition of the steel and data on the sizes and compositions of the precipitates that are formed in the steel under neutron irradiation with different fluxes are presented.

Let us estimate the contribution of the mechanism of precipitate formation to material hardening. You will recall that a considerable contribution to hardening is also made by formation of dislocation loops. The value of material hardening at the expense of precipitates $\Delta\sigma_{0,2}^{prec}$ can be estimated based on the Orowan stresses. Let us take that $\Delta\sigma_{0,2}^{prec} \approx \tau_{orov}$. As it is known, the Orowan stress [24] is calculated by the formula:

$$\tau_{orov} = \alpha \cdot \frac{G \cdot b}{\lambda}\text{, MPa} \tag{14}$$

where α is a constant depending on a barrier type, G is a shear module, b is the Burgers vector, λ is an average distance between barriers. The parameter λ can be calculated by formula:

$$\lambda = \left(\sum \rho_i \cdot d_i\right)^{-1/2}\text{, m}^{-2} \tag{15}$$

where ρ_i is a concentration of the i^{th} clusters of size d_i.

Our calculations by formulas (14) and (15) show that there is the practically same function λ(F) for different level of fluxes. As for considered steel C_{Ni}+C_{Mn} =2.0 % we can conclude that for this sum flux effect is practically absent.

Conclusions

1. Investigation of the flux effect on embrittlement of WWER RPV materials by comparing the test data of surveillance specimens and the test data of specimens irradiated by high neutron flux in frame of the research programs was carried out.
2. The analysis of the neutron flux effect under different mechanisms of embrittlement of RPV materials was made.
3. It was shown that when the radiation embrittlement controlled by the mechanisms “A” (a hardening mechanism at the expense of dislocation loops) and “C” (a non-hardening, segregation mechanism) the flux effect is negligibly small. On dominance of the hardening mechanism at the expense of barrier formation in the form of precipitates or clusters (mechanism “B”) the flux can affect the radiation embrittlement of reactor pressure vessel materials, this influence is negative: with a flux increase ΔT_F decreases. The main elements forming precipitates are Ni, Mn и Cu. Based on the experimental data for WWER RPV materials, PWR RPV materials (steels A508, SA533B), as well as model melts it was established that the flux effect starts manifesting itself when total content of nickel and manganese C_{Ni}+$C_{Mn} \approx 1.8$ %. Such conclusion corresponds to the materials with low content of copper $C_{Cu} < 0.12$ %. As copper has smaller solubility in α-iron than nickel and manganese, it exerts a stronger influence on radiation embrittlement than nickel or manganese, the flux effect starts manifesting itself even with its relatively low content $C_{Cu} \approx (0.12 \div 0.14)$ %. This conclusion follows from papers [4, 6].

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