ASSESSMENT OF HYDROGEN INFLUENCE ON DELAYED FRACTURE OF WELDED JOINTS FROM HIGH-STRENGTH LOW-ALLOYED STEELS

N.N. KHARBIN, O.I. SLEPTSOV, M.N. SIVTSEV and **G.G. VINOKUROV**
V.P. Larionov Institute of Physico-Technical Problems of the North, SD of RAS, Yakutsk, Russia

Hydrogen influence on delayed fracture of welded joints of high-strength low-alloyed steels was studied by the Implant method. A calculation model based on material creep theory was proposed for assessment of hydrogen contribution to delayed fracture process.

Keywords: *arc welding, welded joints, high-strength low-alloyed steels, delayed fracture, calculation model*

Welding of high-strength low-alloyed (HSLA) steels is accompanied by formation of cold cracks which appear at unfavourable combination of structural factors (intensive grain growth, high level of stresses and strains) and increased content of hydrogen [1]. The cause for cold crack initiation is hydrogen-induced cold cracking [2, 3]. Hydrogen influence is manifested in the delayed nature of fracture process [4, 5]. In this connection development of methods of adequate assessment of hydrogen influence on the process of welded joint fracture is urgent.

One of the methods for determination of delayed fracture susceptibility of welded joints is application of implant method, which combines certain features of direct and indirect methods of welded joint testing for cold cracking sensitivity.

The purpose of this work is development on the basis of creep theory of calculated model for assessment of hydrogen contribution into the process of delayed fracture of welded joints of HSLA steels.

Samples-inserts from 14Kh2GMR and T-1 steels (Table) of shape and dimensions as per GOST 26386–84 were tested. Deposition of a bead with sample-insert was performed with 4 mm ANP-2 electrodes in the following mode: I_w = 170 A; U_a = 25 V; v_w = = 10 cm/min. Diffusible hydrogen content in the bead was determined by the glycerin method. Fracture stress in the weakened section σ_{fr} and time to fracture τ_{fr} were recorded. Test results (Figure 1) confirmed the leading role of diffusible hydrogen during delayed fracture of welded joints from low-alloyed steels that is in agreement with the data of work [6]. Delayed fracture susceptibility of samples-inserts made from the above steels turned out to be approximately the same (see Figure 1).

Mechanism of delayed fracture of steel welded joints is characterized by common features of material fracture at creep [2]. In [7] at consideration of crack propagation at stress relaxation as a particular case of fracture under creep conditions, a calculated model based on Yu.N. Rabotnov principle was proposed for assessment of material crack resistance.

In this work the conducted testing procedure allows controlling the force parameters, so that it is rational to replace the value of stress intensity factor (SIF) in [7] by force parameters. The following model was obtained:

$$\frac{\sigma_{fr}}{\sigma_t} = \frac{1}{1 + mq\tilde{\Gamma}(\tau)}, \qquad (1)$$

where

$$\tilde{\Gamma}(\tau) = \int_0^{\tau} \Gamma(\tau - S)dS; \quad m = \frac{1}{1 + p};$$

σ_t is the ultimate tensile strength; σ_{fr} is the stress at which the sample failed; q is the structural parameter related to SIF; $\Gamma(\tau - S)$ is the aftereffect kernel; p = = $p([H])$ is the energy parameter of the system (sample) [7], in this case being a function dependent on hydrogen content.

Let us consider limit cases of model (1).

When the system is open, $p = p([H]) \to \infty$ (m = = 0), i.e. hydrogen content is small enough and system fracture resistance is independent on hydrogen concentration. At hydrogen saturation, when

Composition (wt.%) of 14Kh2GMR and T-1 steels

Steel grade	C	Mn	Si	Cr	Ni	Mo	V	Ti	B	S	P
14Kh2GMR	0.12	1.12	0.32	1.35	0.08	0.38	0.01	0.045	0.0050	0.045	0.023
T-1	0.21	0.95	0.31	0.54	N/D	0.18	0.08	0.028	0.0035	0.017	0.010

$p = p([\mathrm{H}]) \to 0$ ($m = 1$), hydrogen influence is maximum, and system crack resistance also depends on structural parameter q ($0 < q < 1$ [7]). Therefore, parameter m of the proposed model (1) changes in the range of [0–1].

Results of calculations by formula (1) in comparison with experimental data are given in Figure 2. Calculations were conducted at different values of parameter p (q = const) and with the following kinds of integral operator kernel, describing the aftereffect [7]:

1) constant kernel (Maxwell body)

$$\Gamma(\tau - S) = \lambda, \tag{2}$$

where λ is the material constant;

2) exponential kernel (standard linear Kelvin body)

$$\Gamma(\tau - S) = \lambda_1 e^{-\beta(\tau - S)}, \tag{3}$$

where λ_1, β are the rheological constants;

3) power kernel

$$\Gamma(\tau - S) = \frac{\lambda}{(\tau - S)^{1-\beta}}, \tag{4}$$

where $0 < \beta < 1$.

As is seen from these formulas, the first term in the sequence in exponential kernel expansion (3) and the power kernel (4) at $\beta \to 1$ give Maxwell body kernel.

Calculations by model (1) using the above aftereffect kernels showed that experimental data are described best of all by exponential kernel (3) of Kelvin body (Figure 2). Maxwell type kernel (2) leads to somewhat underestimated results at maximum hydrogen content, but at its moderate content, it satisfactorily describes the experimental data. Power kernel (4) gives an unsatisfactory description of the initial sections, but it can be used at longer-term testing (see Figure 2).

Analysis of calculation curves showed that there exists a possibility for assessment of the influence of hydrogen and structural factors on fracture, respectively, by m and q parameters. Limit of $\lim\limits_{\tau \to \infty} \frac{\sigma_{fr}}{\sigma_t}(\tau)$ function when exponential and power aftereffect kernels are used allows assessment of minimum value of fracture stress $\sigma_{fr.min}$ at different hydrogen content. At maximum hydrogen content ($p = 0$) this value, when Kelvin body is used (3), is assessed to be on the level of $\sigma_{fr.min} \approx 231$ MPa ($\sigma_{fr}/\sigma_t \approx 0.33$ at $\tau \to \infty$), and in the case when power aftereffect kernel is used (4) we have $\sigma_{fr.min} = 210$ MPa ($\sigma_{fr}/\sigma_t = 0.3$ at $\tau \to \infty$). As estimates were obtained for the case with maximum possible hydrogen content in the sample, it is anticipated that at real hydrogen content minimum value of fracture stress is close to the experimental result (for 14Kh2GMP steel) $\sigma_{fr.min} = 240$ MPa [2].

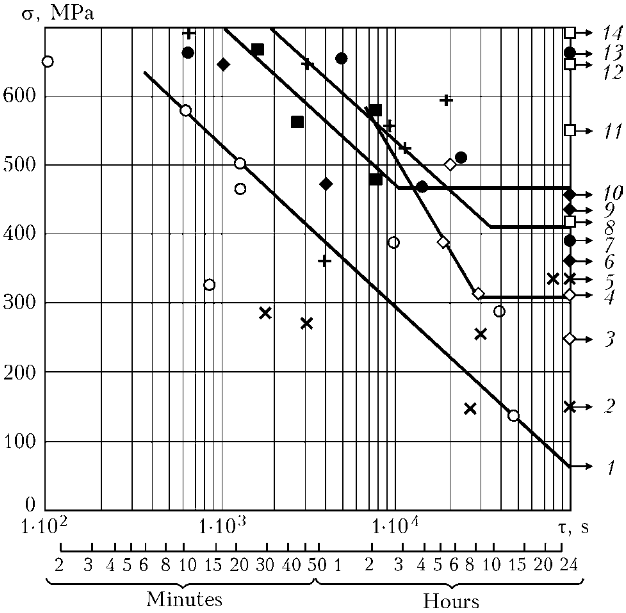

Figure 1. Experimental data on influence of diffusible hydrogen $[\mathrm{H}]_{dif}$ on delayed fracture susceptibility of 14Kh2GMR steel samples-inserts (*1*, *3*, *4*, *6*, *7*, *9*, *11*, *12*, *14*) and T-1 (*2*, *5*, *8*, *10*, *13*): *1* — $[\mathrm{H}]_{dif} = 14.8$ ($T_{av} = -40$ °C); *2* — 15.0 (−25 °C); *3*, *4* — 5.0 (−40 °C); *5* — 7.9 (20 °C); *6*, *9* — 5.0 (20 °C); *7* — 7.9 (20 °C); *8*, *10* — 5.0 (20 °C); *11*, *12*, *14* — 0.6 (−40 °C); *13* — 0.6 cm^3/100 g (20 °C)

Satisfactory description of experimental results by calculated curves allows assessment of the time to fracture, depending on hydrogen content. In the first approximation the simplest Maxwell kernel was used for qualitative assessment of hydrogen influence on time to fracture (2). Substituting (2) into (1), we obtain the time to fracture

$$\tau = (p + 1)\frac{1 - k}{k\lambda q}, \tag{5}$$

where $k = \sigma_{fr}/\sigma_t$.

Based on analysis of empirical data, as well as allowing for exponential nature of thermoactivation dependence describing the energy interactions, parameter p was determined from the relationship

$$p([\mathrm{H}]) = \ln\left(1 + \frac{[\mathrm{H}]_{max} - [\mathrm{H}]_0}{[\mathrm{H}]_0}\right), \tag{6}$$

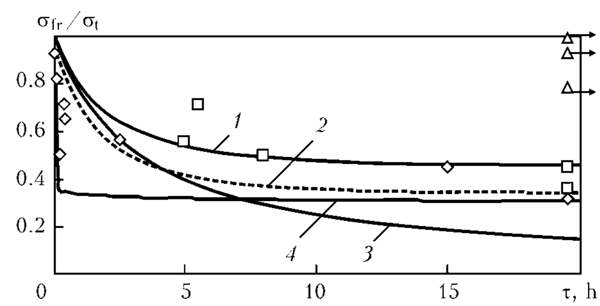

Figure 2. Calculated (curves *1–4*) and experimental data (symbols) with different aftereffect kernels: *1*, *2* — aftereffect Kernel of Kelvin body at $[\mathrm{H}]_0 = 5$ and 15 ml/100 g in the sample, respectively; *3* — Maxwell body at $[\mathrm{H}]_0 = 5$ ml/100 g; *4* — power kernel at $[\mathrm{H}]_0 = 15$ ml/100 g

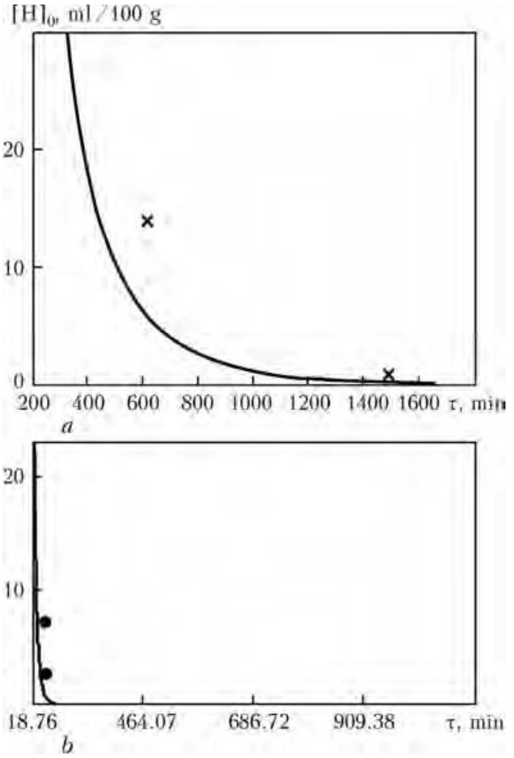

Figure 3. Influence of hydrogen content in samples on delayed fracture at loads σ_{fr} = 300 (×) and 650 (●) MPa (×, ● are experimental points)

where $[H]_{max}$ is the maximum hydrogen content (depends on steel grade and is determined empirically); $[H]_0$ is the hydrogen content in the sample, determined by the glycerin method.

Influence of hydrogen content on delayed fracture of the sample derived by formulas (5), (6) is shown in Figure 3. As is seen from the Figure, the shape of the curves does not change at different loads. However, at lowering of hydrogen content, the curve shifts towards longer time to fracture, that corresponds to increase of delayed fracture resistance of the samples. Results of calculation of the time to sample fracture depending on hydrogen content, based on simplest kernel of Maxwell body, lead to the assumption that when simpler kernels are used, it is possible to more accurately describe this process, but the qualitative picture here will not change. Thus, calculations conducted by the proposed model, showed that at lowering of hydrogen content delayed fracture resistance of welded joints increases. Calculation data are is good agreement with the experimental data.

The proposed experimental model can be used for assessment of hydrogen influence on the technological strength of welded joints of HSLA steels. Comparative simplicity of calculations by this model allows considerably simplifying investigation of the mechanisms of hydrogen-induced delayed fracture of welded joints.

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