Properties degradation of pipeline steels caused by long-term service in hydrogen enriched environments

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Abstract. Long-term service of steels causes the essential deterioration of their as-received properties, especially the parameters of brittle fracture, and the crack growth resistance first of all. The sensitivity of steels to hydrogen embrittlement increases, therefore, for the evaluation of its in-service degradation it is recommended to test the metal in the hydrogenation conditions, best of all – for the crack growth resistance under action of hydrogenating environment.

The investigations of the properties of different parts of oil and gas pipelines (pipe top and bottom) showed that the metal mechanical properties are lower for the parts with more intensive corrosion damages of the internal surface (pipe bottom) than for the metal of the pipe top part. It indicates the essential effect of hydrogen on the processes of in-service degradation of steels.

Two stages of in-service properties degradation of pipeline steels are considered: deformation aging and dissipated damaging. The role of hydrogen is revealed in the intensification of the dissipated damaging stage. The main regularities of steels properties degradation for the different stages of service are given. So, the strength and hardness increase, plasticity and brittle fracture parameters decrease for the stage of deformation aging according to the conventional point of view. At the same time the brittle fracture resistance parameters and the reduction of area decrease continually but (and it is untypical) the relative elongation increases, the strength and hardness decrease for the stage of dissipated damaging. The last peculiarity demonstrates the development of microdefects and their opening in-bulk material.

The evaluation criteria of pipeline safe service with crack type defects in hydrogen enriched environments are proposed using the approaches of fracture mechanics. The analytical dependence for the description of the cyclic crack growth in steels in hydrogenating environments, which considers hydrogen concentration at the crack tip, is proposed and experimentally confirmed.

Introduction

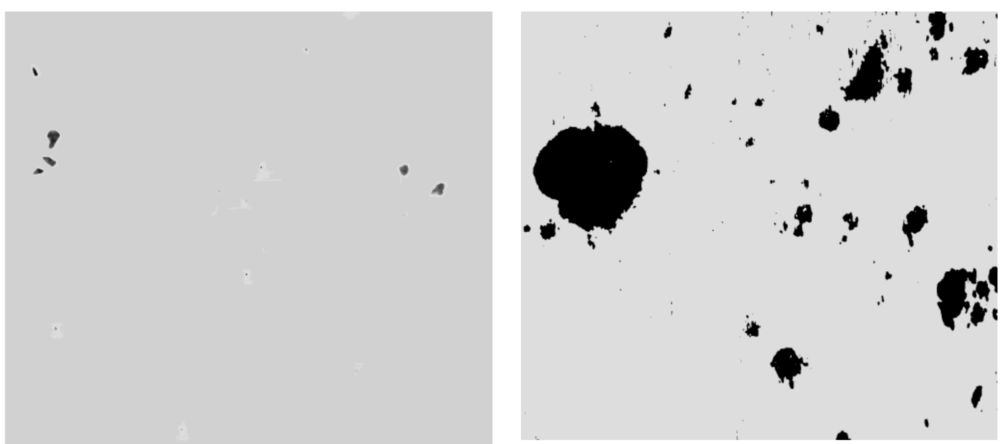
An evaluation of the technical state of long-term exploited structures and calculation of their residual life time is impossible without an account of the change of the material physical and mechanical properties, which appear during service. Really these properties are changed mainly in worse direction, so “in-bulk” material degradation should be considered. It concerns, in particular, pipeline steels exploited at ambient and elevated temperatures. It is evident that diffusion factor can play a dominant role in high temperature degradation of steel properties. It causes changes of the metal microstructure which worsen its resistance to fracture. In particular, the main change for ferrite-pearlite microstructure of power steam pipeline steels consists in dissolution of pearlite and formation of grain boundary carbides [1]. For usual service temperatures change of steel structure is not so pronounced but, for example, for main gas pipeline steels, the changes on the dislocation substructure level, coagulation of grains, grain boundary carbide formation on nanoscale are considered [2].

The peculiarity of pipelines is in their use for transportation of environments which can serve as a source of hydrogen and, correspondingly, to hydrogenate the pipe wall. Then the degradation process occurs during simultaneous action of stresses and hydrogen. The role of hydrogen in the acceleration of macrocrack initiation and propagation is well known but additionally the intensification of in-bulk material degradation (degradation of properties) is considered. In this connection there are two ways of hydrogen effect on nano- and microscale: hydrogen intensifies diffusion [3] and therefore accelerates structure changes [4]; hydrogen facilitates in-bulk dissipated damaging [4,5]. The pipeline steels degradation is considered in the paper just from such positions.

Oil and gas pipelines

The analysis consists of properties comparison in the as-received state (spare pipes) and after 28–40 years of service [4–9].

Main oil pipeline. The 0.10C-1.6Mn-0.30Si steel in as-received state and cut from the upper *up* and lower *bottom* parts of a pipe being in service for 28 years were tested. Residual water deposited from the oil on the pipe bottom was considered as an aggressive environment causing possible in-bulk pipe wall hydrogenation from the internal surface of the pipe. It is confirmed by more intensive internal corrosion damaging the *bottom* part of the pipe (Fig. 1). The tests in simulated residual water showed that the corrosion rate of specimens cut from a middle part of pipe wall was 200% higher for the *bottom* and 170% higher for the *top* parts than for the spare pipe. Thus, the results provided evidence that the changes in the properties of material occur due to long time contact with the residual water as a source of hydrogen absorbed by the exploited metal.



a *b*

Fig. 1. The inner surface of the *up* (*a*) and the *bottom* (*b*) parts of the pipe after 28 years of service.

The results of Charpy tests show that virgin steel exhibits the highest value of *KCV* (180 J/cm^2). The toughness of the material from the *bottom* part is about half of it (95 J/cm^2). Such a dramatic drop in the toughness indicates the degradation of mechanical properties of the steels exploited in oil transport lines. In the case of the *bottom* part of the pipe being in operation, it was not possible to evaluate the impact toughness, because of the formation of cracks parallel to the pipe surface.

Sensitivity of the pipe materials to stress corrosion cracking has been evaluated by slow strain rate (10^{-7} s^{-1}) tensile tests under moderate cathodic polarization (0.5 A/m^2) in water deposited in the oil storage tank. The relative elongation ($K_\varepsilon = \delta^c/\delta$ 100%) and relative reduction in area $K_{RA} = RA^c/RA$ 100%,) of specimens tensile tested in the air (δ, RA) and those tested in corrosion environment (δ^c, RA^c) were compared. The *bottom* part of the material exhibited lower strength and elongation than the virgin part, either tested in air or in the aggressive environment (Table 1).

Table 1. Plasic properties of tensile tested materials

Material	Test environment	δ [%]	RA [%]	K_ε [%]	K_{RA} [%]
As received	air	36	77	39	55
	water	14	42		
Being in service	air	28	56	25	5
	water	7	3		

As it is seen the values of K_{RA} are 55% and 5% respectively for the virgin specimens and cut out from a bottom section of the pipe. The drastic decrease in the K_{RA} and RA values for virgin specimens from the *bottom* part of in-service pipe measured in the air comparing with the virgin ones indicates the possibility of drastic decrease in resistance to brittle fracture of the metal during exploitation.

Oil storage tank. This example is a model for demonstrating the role of hydrogen in degradation of steels. Properties of St3sp (0.2C) steel specimen cut off the high-capacity oil storage tank after 30 years of operation were investigated. Specimens cut off from different areas of the structure were tested: the *upper* part of the wall of the reservoir with air and condensed water during operation (I), areas of the wall that were in permanent contact only with oil (II); areas of the wall near the *bottom* of the reservoir (III), and the *bottom* of the reservoir (IV). Note that areas III and IV were in permanent contact only with residual water.

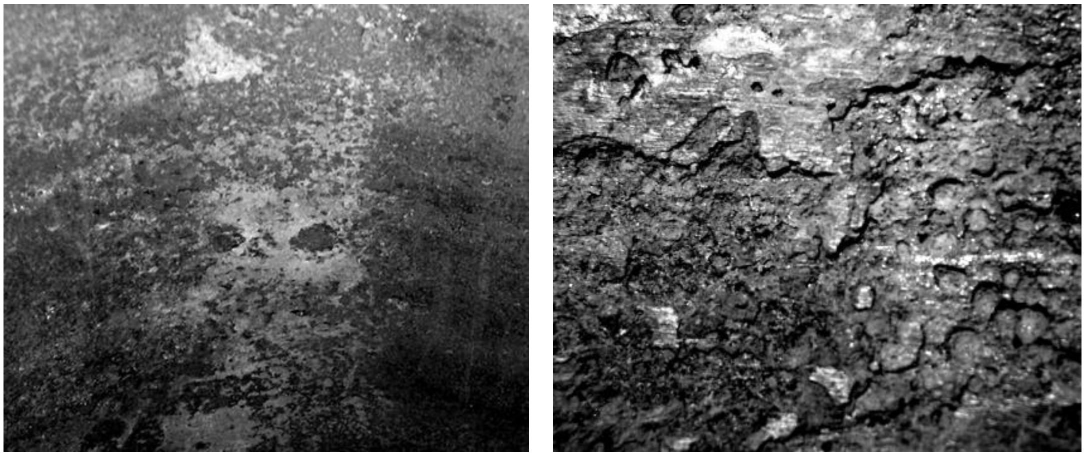
Table 2. Corrosion current density j_{corr} of St3sp steel from different areas of the storage tank (I– IV) in residual water taken from two refineries (conventionally, 1 and 2)

Area / Residual water	j_{corr} 10^6 [A/cm^2]			
	I	II	III	IV
1	3,8	1,7	5,1	4,0
2	6,7	3,3	8,6	7,3

Among the standard mechanical characteristics, the impact toughness appeared to be the most sensitive to the in-service degradation of the steel; the steel from the area II is characterized by the highest mean impact toughness (KCV = 153 J/cm^2), and area III is characterized by its lowest value (62 J/cm^2) due to the joint influence of the aggressive medium and the most intensive loading. For the metal from areas I and IV, which also contacted with the hydrogenating medium, the impact toughness was 72 and 84 J/cm^2 respectively. Note that the metal of the *upper* part of the reservoir is practically not subjected to mechanical loading, and, hence, the low level of KCV is obviously connected only with the action of hydrogen from the condensed water, i.e., the hydrogenation of the metal is accompanied by the initiation of substantial internal stresses. Thus, for the degradation of steels “in the volume,” external load is not mandatory, and only intensive hydrogenation of the metal is sufficient. Correspondingly, corrosion resistance of the metal from different areas of the reservoir determined from the results of electrochemical measurements of the corrosion current (Table 1), also differ, namely, j_{corr} is minimum for area II and maximum for area III. Predicting the corrosion rate of the steels in long operation, this feature should be taken into account.

Main gas pipeline. Steels 17G1S (0.17C-Mn-Si) and API 5L X52 in as-received state (spare pipes) and after 28–40 years of service were studied. The API 5L X52 steel (code X52 for as-received state) was in service for 30 years (code X52-10 and X52-12 for wall thickness 10 and 12 mm respectively). Upper *up* and lower *down* parts of the exploited pipes were distinguished and in some cases specimens were cut out closer to inside *in* and outside *out* parts of the pipe wall. Such

separateness of the different parts of exploited pipe is explained by the fact that in the case of wet gas pipelines the condensed water accumulates at the pipe bottom [10].
As it is seen in Fig. 2 the internal surface of the pipes being in service revealed the deposits of corrosion products. In the case of the *down* surface, non uniform and pitting corrosion was observed (Fig. 2*b*).



a *b*

Fig. 2. The inner surface of pipe X52-10-*up* (*a*) and X52-10-*down* (*b*).

The investigations revealed a number of effects of changes in the physical and mechanical properties after their long operation in comparison with the same steels of spare pipes. In particular, the strength, the hardness (HRB), and the plasticity (RA) of the steels decrease, the plateau of yield appears on tension diagrams, and, in this case, the strain hardening coefficient n increases (Table 3). An abrupt decrease in the yield strength for 17G1S steel and an abrupt increase in the characteristic RA were registered for all steels. However, the influence of operation on the percentage elongation δ is ambiguous: for X52-10 and 17G1S steels its increase was observed. The hardness of the lower areas of the operated steels is smaller than that of the upper regions.

Table 3. Mechanical properties of the studied gas pipeline steels

Steel code	Service time [years]	Pipe part	σ_y [MPa]	σ_{UTS} [MPa]	RA [%]	δ [%]	n	$J_i / J_{0.2}$, [kN/m]
X52	0	-	355	475	72,9	22.7	0.59	86/412
X52-12	30	*bottom*	268	451	64.4	20.8	0.74	50/127
		up	255	460	62.5	22.9		
X52-10		*bottom*	362	536	54.6	29.7	0.82	37/79
		up	335	538	55.0	28.8		
17G1S	0	-	378	595	79.0	20.2	0.58	203/315
	28		403	590	68.2	20.5		
	29		345	547	71.1	19.6	0.76	
	31		419	574	73.8	21.8		87/201
	38		357	520	73.1	25.4	0.97	
	40		302	515	69.2	26.3	0.75	

As a result of the pipes operation, the characteristics of resistance to brittle fracture of the steel, namely the impact toughness KCV and the crack resistance (the critical value of the J -integral $J_{0.2}$

for an increase in the length of the crack of 0.2 mm), also decrease. The resistance to brittle fracture of the metal of a spare pipe correlates well with its hardness: the harder material from the external surface is characterized by the smaller impact toughness. The operated metal has another dependence: *KCV* and $J_{0.2}$ of the material are higher from the external surface of the pipe than those from the internal surface indicating that in the latter case it degrades more severely. This also confirms the determining role of hydrogen here because its content in the metal near the internal wall of the pipe, where it evolves as a result of the corrosion interaction of the steel with the components of transported natural gas, is higher [11].

The impact toughness of X52 steel in the initial state somewhat decreases (by 15–20%) with the decrease of the test temperature due to the energy *A* of propagation of a crack, which enters into the total energy of deformation and fracture (Fig. 3). At the same time, the low-temperature impact toughness of the operated steel is the one-third than that for the nonoperated steel, and the energy of propagation of a crack is the one-fourth than that for the nonoperated steel.

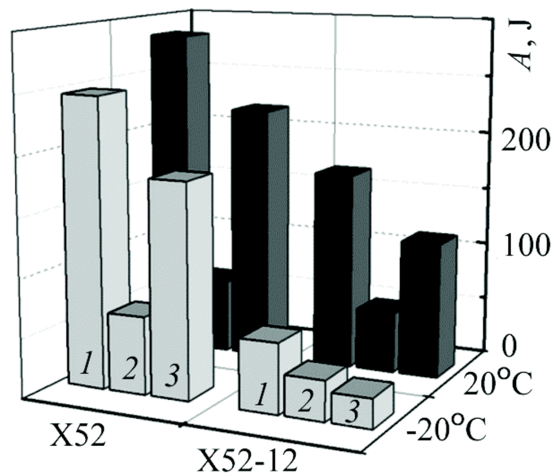


Fig. 3. Total fracture energy (*1*) and its components: energy of crack initiation (*2*) and propagation (*3*) during Charpy tests of API 5L X52 steel.

The stress corrosion cracking in an environment modelled by an aqueous condensate inside the gas pipeline was investigated on smooth specimens and specimens with preliminarily formed cracks. Some experiments were accompanied by moderate cathodic polarization (0.1 A/m^2), which simulated the possible hydrogenation of pipes as a result of cathodic protection. Tests of smooth specimens did not detect the susceptibility to corrosion cracking of the steel in all states at the corrosion potential, but registered it during experiments on cracked specimens (Fig. 4, the threshold of corrosion crack resistance J_{scc} was compared with the value of the *J*-integral J_i of the crack start under intensive loading in the air).

The steel in the initial state is characterized by a maximum resistance to stress corrosion cracking, and cathodic polarization reduced additionally the threshold J_{scc}. This means that the degradation of properties of the pipelines steels manifest itself as an substantial decreases not only in the impact toughness and crack resistance but also in the corrosion resistance and resistance to hydrogen cracking. Moreover, it can be assumed that precisely these parameters of fracture mechanics most sensitive to brittle fracture resistance are also most sensitive to in-service degradation of steel. The embrittlement factors, specifically, decrease of the test temperature and hydrogenation of the metal, enhance this property, and, therefore, the comparison investigations of operated and nonoperated steels under precisely these conditions should be carried out.

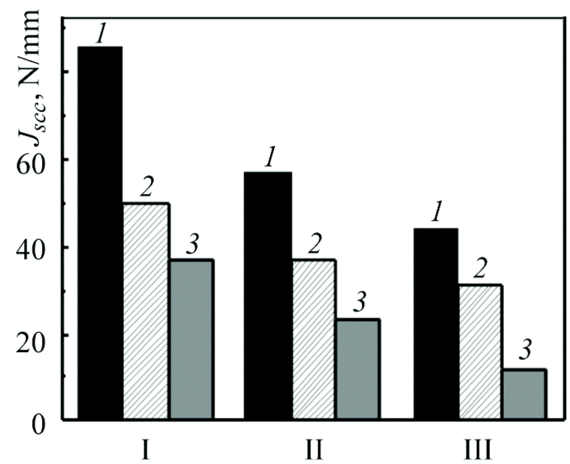


Fig. 4. Crack growth resistance of X52 (I), X52-12 (II) and X52-10 (III) steels in the air (J_i, *1*), in the corrosion environment under corrosion potential (*2*) and cathodic polarisation (*3*).

Moreover, the detected high susceptibility to hydrogen cracking of the metal operated for a long time, despite its low strength, should be taken into account using electrochemical protection of pipelines. Since it is known the electrochemical protection must be used carefully for high-strength steels for the reason that under the metal hydrogenation by cathodic polarization, hydrogen cracking of the pipe is possible. A similar situation is also observed after in-service degradation of a low-strength steel.

Two stages of in-service degradation of steels.
According to the widespread opinion, the main factor of degradation of the main pipelines steels is their deformation aging, which increases the strength and decreases the plasticity and impact toughness (stage I in Fig. 5). However, if the operation time of the metal approaches about 20–30 years, the intensive damage develops, which levels the hardening of the material by deformation aging and simultaneously decreases the hardness and brittle fracture resistance (stage II). Another feature is the change in the plasticity characteristics of the operated steels, namely, the decrease in RA and the increase in δ.

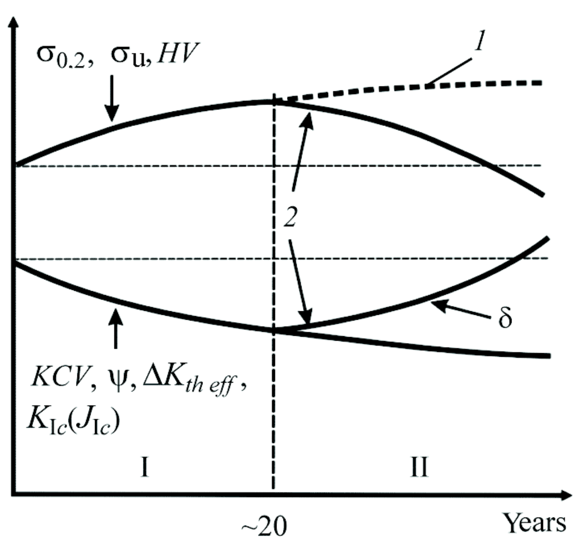


Fig. 5. Changes in steels properties that determine their in-service degradation: (*1*) without hydrogen effect; (*2*) in the presence of hydrogen.

The proposed hypothesis that the microdamage of the operated steel is the main factor of degradation of pipes after their long operation is consistent with the results of investigations of the hydrogen behaviour in the metal obtained by assessing the hydrogen permeability and temperature dependences of the hydrogen extraction from the metal in different states [4–8]. The concentration of hydrogen in the metal and its diffusion coefficient, as a rule, are determined predicting its influence on the decrease of the structural strength because hydrogen may accumulate in the prefracture zone. However, in recent years, known methods of investigating of hydrogen behaviour in metals have been used to assess their possible damage taking into account that hydrogen is predominantly located in defects, which are considered as its traps [12].

During the determination of hydrogen amount in the metal by vacuum extraction, temperature was increased stepwise with some hold in each step. Then, at a relatively low temperature, "low-energy" hydrogen, i.e., hydrogen located in low-energy (taking into account its interaction with defects) traps, leaves the metal. Among such traps are dislocations, for example. At the same time, the "high-energy" hydrogen is located in "deeper" traps (pores, nano-, or microcracks) and, therefore, can leave the metal only at higher temperatures. This is the base of the metal defectiveness analysis [4–8].

The electrochemical method for the determination of hydrogen diffusion coefficient in the metal assumes using a membrane specimen between two electrochemical cells. One side of the membrane (inlet) is polarized cathodically in a potentiostatic mode, and the opposite side is polarized anodically. Hydrogen formed during discharge on the surface of the membrane cathode in the polarization cell partially penetrates into the metal and approaches diffusively the anode side contacting with an alkali solution. Hydrogen atoms from the outlet side of the membrane are almost completely ionized by an applied potential and provide an ionization current proportional to the instantaneous desorption rate of hydrogen. On the whole, the physical (lattice) D and effective D^* (with regard to trapping) diffusion hydrogen coefficients can be determined and, hence, the ratio $D/D^* = 1 + N(k/p)$ can be calculated, where N is the density of traps, k and p are the kinetic constants of trapping and release of hydrogen from traps, and $N(k/p)$ is the efficiency of hydrogen trapping.

Power steam pipeline weld metal

The steam pipeline weld metal posses some peculiarities (Fig. 6) [13]. The weld metal is more sensitive to in-service degradation than the basic. Simultaneous reduction of hardness and strength, on the one hand, and brittle fracture resistance, on the other, contradicts the common rules with the regard to properties of engineering steels. In addition, elongation δ increases and *RA* decreases with the weld metal service, as it was shown above for the exploited steel of transit gas pipelines. It is an evidence of intensive dissipated damaging in exploited metal. From the other hand, the correlation of *RA* with the *KCV*, J_c and $\Delta K_{th\ eff}$ (fatigue threshold) should be considered as a manifestation of the weld metal degradation which testifies its sensitivity to hydrogen cracking and brittle fracture.

Hydrogen effect on fatigue crack growth in gas pipeline steel and fracture risk assessment of the defected pipeline

The study deals with the evaluation of fatigue crack growth rate in pipeline steel under hydrogenation taking into account the local hydrogen concentration near the crack tip the aimed in the fracture risk assessment of defected pipelines [14]. The object of the study was X52 steel. High purity water with pH 7±0.1 and conductivity $\chi < 3$ mS/m served as the basic environment. The formic acid of different concentration C_{HCOOH} was used as an admixture for increasing the hydrogenating ability of the solution. Five series of tests were conducted under different concentrations of formic acid in environment, namely: C_{HCOOH} = 0; 3; 5; 10 and 100 mg/kg. Primary test results were presented as scatter plots of corrosion fatigue crack growth rate *dA/dN* versus stress intensity factor ΔK.

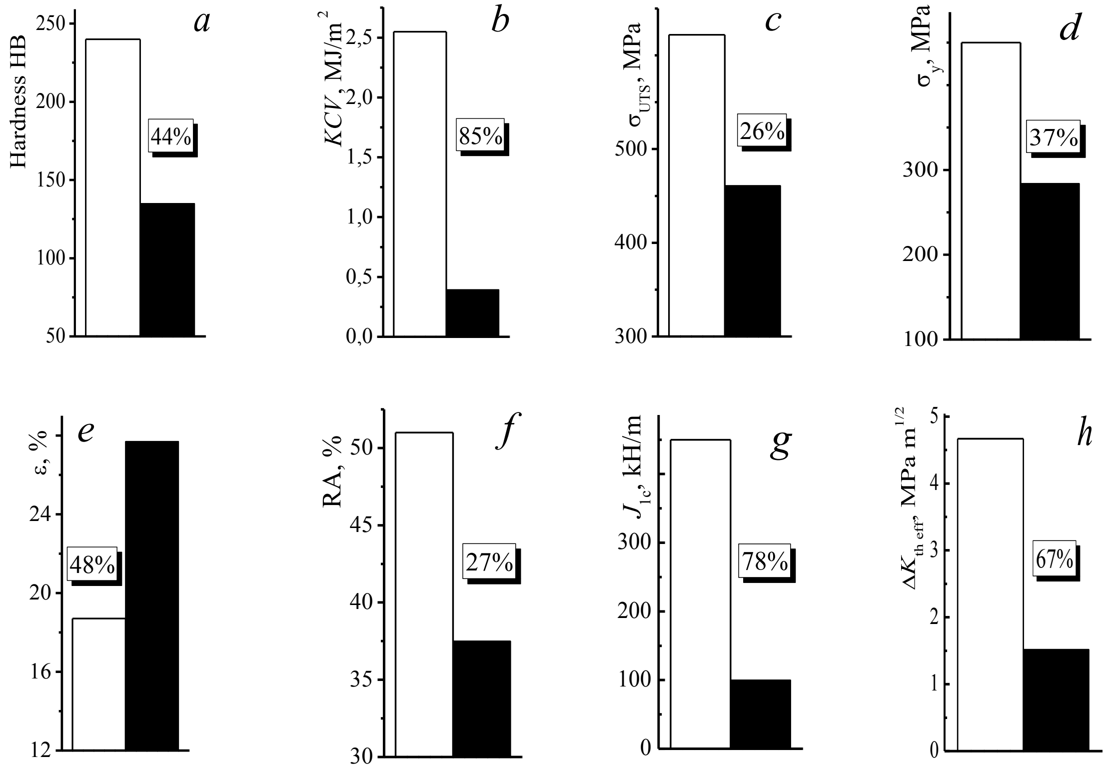


Fig. 6. The mechanical properties of the weld metal (WM) in virgin state (light bars) and after ~$2 \cdot 10^5$ h of service (dark bars). Figures indicate a percentage change in properties as a result of exploitation.

For each series of tests, the measuring of local hydrogen concentration in metal near the fatigue crack tip was made using the special technique of mass-spectroscopy with laser microprobe [15]. The following procedure was applied. After some number of loading cycles N_i and on achieving some crack length a_i, which corresponds to the fatigue crack growth rate $(da/dN)_i$, the test was stopped and specimen was removed from the test rig. Specimen cut-off was subjected to scanning using by laser microprobe. It should be noted that distribution of local hydrogen concentration near the crack tip in perpendicular direction to plain of the crack growth was received under scanning of the specimen surface by laser microprobe at the minimal distance from a crack tip equal 0.1 mm. This point was chosen as a conventional reference point $x = 0$. Within each series of the test the above-mentioned procedure was repeated for cracks of different length a_i. Resulting data were received as the sequence of the following parameters: crack length a_i; fatigue crack growth rate $(da\,/\,dN)_i$; local hydrogen concentration $C_{H(i)}$; stress intensity factor range ΔK_i and number cycles of loading N_i.

The specificity of hydrogen concentration distribution near the crack tip for all tested specimens is the following. Some maximum of local hydrogen concentration was observed at $x = 0$ and the significant decrease of the C_H value under moving away from this point in the both directions $(+x \neq 0, -x \neq 0)$ is also observed. Finally the local hydrogen concentration achieves the hydrogen concentration in the bulk of the metal, $C_{H(v)}$ which depends on the concentration of formic acid in solution. For further analysis the C_H values at $x = 0$ were chosen as conventional maximal value of local hydrogen concentration in the crack tip and which was denoted as $C_{H(t)}$. It has been found that parameter $C_{H(t)}$ strongly depends on the stages of the fatigue crack growth in steel. Thus, it has been accepted as a parameter, which defines the state of materials hydrogenation in the crack tip.

Finally, the received three groups of relations, namely: $da/dN = F_1(\Delta K)$, $C_{H(t)} = F_2(da/dN)$ and $C_{H(v)} = F_3(C_{HCOOH})$ were analysed and processed for all conducted tests. As a result, the generalised diagram in bi-logarithmic coordinates has been built, relating the parameters $(da/dN)/(C_{H(t)}/C_{H(v)})$ and ΔK. This diagram showed the coincidence all data as single scatter plot, which can be described by power function at the mean square deviation $r^2 = 0.98$. Thus, a fatigue crack growth in the given steel under hydrogenation can be presented as a function of local hydrogen concentration in the crack tip $C_{H(t)}$, hydrogen concentration in the bulk of the metal $C_{H(v)}$ and stress intensity factor range ΔK:

$$\frac{da}{dN} = A \cdot \left(\frac{C_{\mathrm{H}(t)}}{C_{\mathrm{H}(v)}} \right) \cdot \left(\Delta K \right)^m , \qquad (1)$$

where A and m are some constants of “material – environment” system. Here it should be noted that this formula is valid under the following physical conditions: $C_{H(v)} \neq 0$ and $C_{H(t)} \geq C_{H(v)}$.

The relation (1) as a generalised result of this study can be used in expert procedure for the evaluation of reliability and strength of pipeline defected components under hydrogenation. For example, it may be done using the statements of the work [16] where the fracture mechanics approach for assessment of workability and fracture risk of pipelines with crack-like defects is proposed.

Let us consider a defected pipe for transportation of hydrogen-containing agents under internal pressure p. Here, all types of detected defects are modelled by semi-elliptical cracks with different ratio of their half-axis a, and c: $(c/a) = 0.01$–0.08. For calculation of a stress intensity factor for longitudinal semi-elliptical crack in pipe wall under internal pressure p the known expressions were used (see ref. [16]).

For corrosion fatigue crack growth behaviour it was assumed that propagating crack saves its semi-elliptical shape, but the ratio of half-axis (c/a) is a variable value [16]:

$$c/a = f\left(C_n, N \right), \qquad (2)$$

where C_n are the constants dependent on “material – environment” system; N is a number of loading cycles.

Here, it was also supposed that the crack growth rate diagram fully defines the resistance to propagation of defects in the pipeline wall in direction of both axis c and a. In our case this diagram was presented in simplified view using the derived formula (1). Such diagram is located between two characteristic values of stress intensity factor range ΔK_{I}, namely: ΔK_{th} – threshold stress intensity factor and ΔK_{fc} – cyclic fracture toughness. The parameter ΔK_{th} defines the limit load, below which the detected defects can be considered as non-propagated and cyclic fracture toughness ΔK_{fc} indicates the critical loading level, above which the detected defects are potentially able to spontaneous (catastrophic) growth. These parameters were chosen as criteria for further consideration of expert assessment of workability and fracture risk of the defected pipeline.

The threshold defects depth c_{th} criterion is based on relationship between the depth of semi-elliptical crack in the tube wall and the value of threshold stress intensity factor ΔK_{th}. Here the threshold defect depth c_{th} defines maximal semi-elliptical crack depth of a given shape (c/a), for which a stress intensity factor equals the threshold value $K_{\mathrm{I}} = K_{th}$.

Thus the criterion of the “safe” crack-like defects is:

$$c \leq c_{th}\left(\Delta K_{th} \right) \text{ under } \left(c/a \right) = \text{const.} \qquad (3)$$

Therefore, all detected crack-like defects by depth $c \leq c_{th}$ can be accepted as “safe”, inasmuch as they don’t have the potential ability to further development into the wall depth.

The critical defects depth c_{fc} criterion is realised according to the well-known criterion of brittle fracture mechanics:

$$\Delta K_{\mathrm{I}} \le \Delta K_{fc}\,, \tag{4}$$

where ΔK_{fc} is a cyclic fracture toughness.

On this basis, the critical defect depth c_{fc} defines the maximal semi-elliptical crack depth of a given shape (c/a), for which the stress intensity factor equals the critical value $K_{\mathrm{I}} = K_{fc}$ and the criterion of the "critical" crack-like defects is:

$$c \ge c_{fc}\left(\Delta K_{fc}\right) \text{ under } \left(c/a\right) = \text{const.} \tag{5}$$

Thus, all detected crack-like defects in the pipelines of the depth about $c \approx c_{fc}$ can be considered as critically dangerous, because the high probability of their spontaneous growth leading to the catastrophic failure of a pipeline exists.

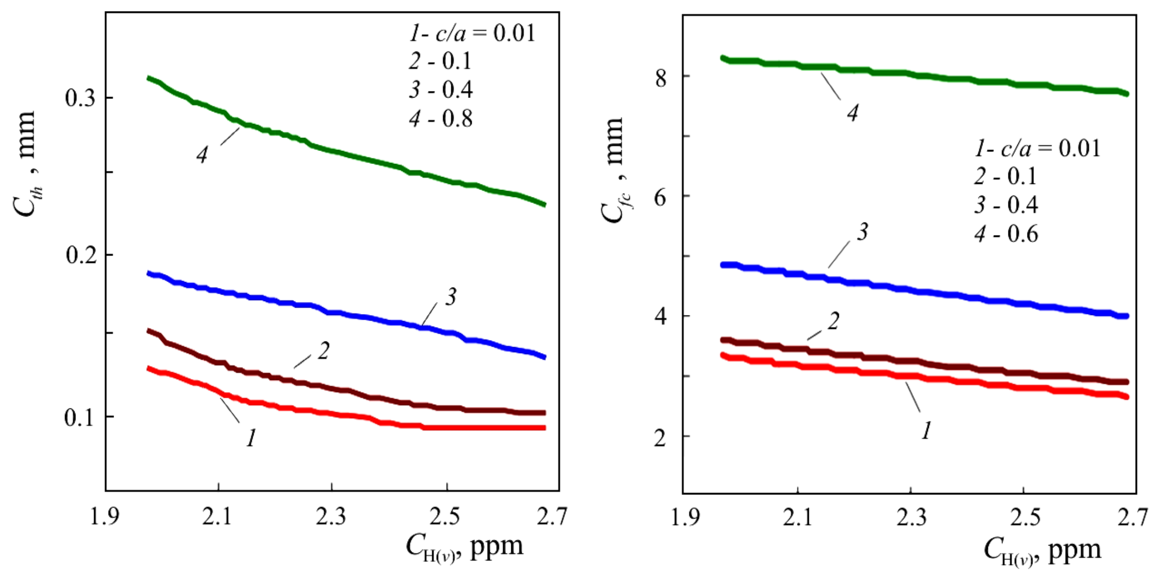


Fig 7. Threshold (*a*) and critical (*b*) depth of defects versus hydrogen concentration in the bulk of the metal for different shapes of crack.

As an example of the above-considered criteria and statements application, the standard gas pipeline with outer diameter $D = 610$ mm and wall thickness $t = 11$ mm subjected to internal pressure of $p = 70$ bars was chosen. The values of threshold depth c_{th} of defects and critical defect depth c_{fc} as a function of hydrogen concentration in the bulk of the metal $C_{\mathrm{H}(v)}$ for different shapes of crack (c/a) were calculated (Fig. 7). These dependencies can be considered as basic diagrams for workability and fracture risk assessment of defected pipeline under given operating conditions and level of metal hydrogenation. Thus, all defects detected by NDT methods under inspection can be compared with these diagrams and as a summary the evaluation of potential risk of each defect can be done.

Conclusions

1. Transport of hydrogen containing environments causes the pipe wall hydrogenation therefore the degradation of the in-bulk material occurs under simultaneous effect of mechanical stresses and hydrogen absorbed by metal.

2. A number of features in the degradation of the mechanical and corrosion-mechanical properties of oil and gas main pipelines and oil storage reservoirs operated for 28–40 years are detected. Thees are: the decrease in the resistance to brittle fracture, expressed in the terms of the impact toughness drop, reduction in area, crack resistance, corrosion resistance, and resistance to hydrogen cracking; anomalies in the mechanical behaviour of the operated metal, namely, the decrease of the resistance to brittle fracture is accompanied with the decrease of the hardness, and the decrease of the reduction in area is accompanied by an increase in elongation; the high sensitivity of the impact toughness to the degradation of steels is caused by the decrease of the propagation energy of cracks, i.e., the parameter of fracture mechanics.

3. The main factor of the degradation of durable pipes is microdamage, which is indirectly substantiated by the intensification of trapping of hydrogen, the increase of the fraction of deep hydrogen traps, the increase of the elongation, and the decrease of the hardness. Two stages of degradation of steels, namely, deformation hardening and development of defectiveness are proposed.

4. The relationship between fatigue crack growth rate and local hydrogen concentration near the crack tip has been established for pipeline steel based on experimental data on crack propagation and results of measurement of the local hydrogen concentration near the crack tip under different hydrogenating conditions. The formula for determination of crack growth rate da/dN as a function of local hydrogen concentration in the crack tip, $C_{\mathrm{H}(t)}$ hydrogen concentration in the bulk of the metal $C_{\mathrm{H}(v)}$ and the stress intensity factor range ΔK has been derived. The engineering application of the received results for workability and fracture risk assessment of defected pipeline under hydrogenation is demonstrated on the base of the conception of fatigue threshold and critical depth of a defect.

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