HYDROGEN DIFFUSION AND EAC OF PIPELINE STEELS UNDER CATHODIC PROTECTION

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ABSTRACT

The paper deals with the behavior of HSLA steels for pipelines to Environmental Assisted Cracking (EAC) that can take place, in seawater and other similar solutions, through a hydrogen embrittlement mechanism under plastic straining or cyclic fatigue loading, at cathodic potentials.

The behavior is discussed in relation with the hydrogen diffusivities measured through Permeation tests.

The resistance to Stress Corrosion Cracking (SSC) and Corrosion Fatigue (CF) was assessed in artificial seawater under cathodic protection by means of Slow Strain Rate tests and Corrosion Fatigue Crack Propagation tests at low loading frequencies.

Pipeline steels produced by diverse manufacturing process, with different microstructures and tensile yield strength in the range 400 MPa to 640 MPa, were considered.

The results points out the effect of microstructure on the hydrogen diffusion and hydrogen embrittlement resistance of pipeline steels. Martensitic steels with very fine dispersed carbides microstructure show very high hydrogen embrittlement resistance.

Introduction

Steels for pipeline show Environmental Induced Cracking (EAC) phenomena through a Hydrogen Embrittlement mechanism under slow straining and cathodic protection. Hydrogen evolution can take place on the metal surface due to the cathodic protection normally used in order to protect pipelines against general corrosion. The steel is polarized at cathodic potentials in the range –0.8 to –1.1 V vs SCE, but very negative values could be reached in overprotected areas, close to impressed current anodes. The hydrogen ions reduction reaction takes place on the metal surface forming adsorbed hydrogen, which can diffuse into the metal owing its high solubility in the lattice. Afterwards, it concentrates in plastic deformation areas, and promotes brittle crack growth. However, owing to their high resistance, very few field failures caused by hydrogen embrittlement have been found on usual grades of pipeline steels, localized to landslip areas [1,2].

Hydrogen Embrittlement mechanism also induces high crack growth rates on marine structures under cyclic loading at low frequency, especially below 1 Hz, producing Corrosion-Fatigue phenomena. Increase of cathodic polarization enhances hydrogen ions reduction and increases fatigue crack growth for intermediate values of stress intensity factor range (Δ K). A typical plateau on the corrosion fatigue curves (i.e. fatigue crack growth rate vs Δ K curves) can be noted at stress levels above the threshold level. According with the Wei and Landes model, at these levels the crack propagation is dominated from stress corrosion cracking. Hydrogen embrittlement mechanism acts in the cyclic plastic strained region at the cracks tip [3].

The susceptibility to hydrogen embrittlement generally increases with mechanical properties of steels. However previous studies [4] showed that a main factor influencing steel susceptibility is the microstructure. Traditional grades of pipeline steel are produced by means of hot rolling or controlled hot rolling. Quenched and tempered steels are also used.

Hot rolled steels have a banded microstructure of ferrite pearlite, oriented along rolling direction. The adoption of accelerated cooling after controlled rolling for the production of new high-grades steel partially or fully replaces pearlite bands with bainite or martensite.

The microstructure of heat-treated steels depends on tempering, varying from martensite to acicular ferrite and carbide.

Steels with different microstructures show a different resistance to hydrogen embrittlement. Steels with a microstructure characterized by very fine dispersed carbides can have very high hydrogen embrittlement resistance [5].

The paper evaluates the influence of hydrogen diffusivity on behavior of traditional and new grades of pipeline steel to EAC induced by hydrogen embrittlement under slow strain rate and cyclic loading in artificial seawater under cathodic protection.

The EAC behavior has been assessed through SSR tests and CF crack growth tests. Hydrogen diffusion coefficient was measured by means of permeation test along rolling and transverse direction.

Experimental

The tests were performed on HSLA steels with nominal grade from X60 to X100, used for pipelines, risers and flow-line (Table 1). The steels show different microstructures: X60, X65 A and B steels were obtained by means of hot rolling and have ferrite/pearlite bands oriented along the rolling direction. X65 C steel was produced by means of controlled rolling and has a similar microstructure but fine ferritic grain size. The X80 and experimental X100 steels was obtained by means of controlled rolling and has a similar microstructure but fine ferritic grain size. The X80 and experimental X100 steels was obtained by means of controlled rolling and accelerating cooling. X80 shows very fine ferrite bands with bainite islands, X100 has a ferritic/martensitic microstructure. Finally, X65 M and X85 M steels are fully quenched and tempered steels, with a microstructure of tempered martensite. Data relative to X60 and X100 steels was obtained in previous research [6].

Table 1. Microstructure, mechanical properties and chemical composition (by weight %).

Production processing	Microstructure	Steel	TYS (MPa)	UTS (MPa)	с	Mn	Si	Р	S	Cr	Ni
		X60 ¹	430	588	0.22	1.35	0.24	0.012	0.024	-	0.01
Hot rolling	Ferrite pearlite	X65 A	399	518	0.09	1.64	0.24	0.003	0.002	0.031	0.017
		X65 B	485	567	0.08	1.60	0.31	0.009	0.003	0.053	0.034
Controlled rolling	Ferrite pearlite	X65 C	507	579	0.05	1.55	0.16	0.002	0.003	0.031	0.005
Controlled rolling and accelerated cooling	Ferrite/ pearlite/bainite	X80	547	658	0.07	1.89	0.19	0.017	0.006	-	0.28
	Ferrite/martensite	X100 ¹	663	750	0.07	1.96	0.34	0.035	0.007	0.03	0.31
Oil quenching and	Tompored martaneita	X65M	552	619	0.10	1.12	0.30	0.010	0.002	0.142	0.418
tempering	rempered martensite	X85M	637 ²	738	0.10	1.11	0.29	0.015	0.002	0.17	0.42

¹ by Cabrini et al. [6]

²TYS_{0.5 %}

SSR tests

Slow Strain Rate (SSR) tests were performed according to ISO 7539-7 on 3 mm diameter cylindrical tensile specimens at strain rates of 10^{-5} s⁻¹. The experimental details were described elsewhere [7]. The hydrogen embrittlement effects were evaluated by normalizing the Reduction in Area to the value obtained after the test in air. Fracture analysis was also performed in order to establish brittle areas on fracture surface and secondary cracks.

Corrosion-Fatigue tests

Fatigue tests were performed on Single-Edged-Notched-Three Points Bending specimens according to an experimental procedure adopted in previous researches [8]. The tests were conducted according to the requirements of ASTM E647 standard.

Crack size (a_N) was evaluated through compliance (C_N) by using the relation (ASTM E399-90 Standard)

$$a_{N}/W = 0.9997 - 3.95 \cdot U + 2.98 \cdot U^{2} - 3.214 \cdot U^{3} + 51.52 \cdot U^{4} - 113.0 \cdot U^{5}$$
(1)

$$U = \{1 + [4 \cdot E' \cdot B \cdot C_N \cdot (W/S)]^{1/2} \} - 1$$
(2)

where B is the specimen thickness, W the specimen width, S the span length, E' the Young's Modulus (E) for plane stress or $E/(1-v^2)$ for plane strain and v is the Poisson's Ratio.

The data of crack length (a_N) measured as a function of elapsed cycle number (N) were fitted through an incremental secondorder polynomial method [9] on 25 successive points in order to compute crack growth rates (da/dN). Stress intensity factor range (ΔK) was evaluated according to the relation of ASTM E399-90 Standard (ΔP_N is force range at N cycles):

$$\Delta \mathbf{K} = \left[\left(\Delta \mathsf{P}_{\mathsf{N}} \cdot \mathsf{S} \right) / \left(\mathsf{B} \cdot \mathsf{W}^{3/2} \right) \right] \cdot \mathbf{f}(\mathsf{a}_{\mathsf{N}} / \mathsf{W}) \tag{3}$$

$$f(a_{N}/W) = \{3 \cdot (a_{N}/W)^{1/2} \cdot [1.99 - (a_{N}/W) \cdot (1 - a_{N}/W) \cdot (2.15 - 3.93 \cdot a_{N}/W + 2.7 \cdot a_{N}^{2}/W^{2})]\} / [2 \cdot (1 + 2 \cdot a_{N}/W) \cdot (1 - a_{N}/W)^{3/2}]$$
(4)

A sinusoidal loading was imposed with 0.6 minimum to maximum stress ratio (R) at 20 Hz and 0.2 Hz frequencies, for the tests in air and in seawater respectively. Two types of test were performed:

1. Constant-Force-Amplitude fatigue tests with constant force range (ΔP) and increasing stress intensity factor range during crack propagation;

2. K decreasing fatigue test with decreasing force range in order to produce decreasing stress intensity factor range during crack propagation.

Permeation tests

Hydrogen permeation tests were carried out according to the electrochemical methods proposed by Devanathan-Stachurski [10,11]. A metal membrane was settled in a permeation cell between two compartments and acted as a bi-electrode. One side (hydrogen entry side) was in contact with 3.5% NaCl solution and polarized at -1050 mV vs Saturated Calomel Electrode (SCE) through a potentiostat and a platinum counter electrode. The opposite side (hydrogen exit side) was in contact with a 0.1M NaOH solution and polarized at +200 mV vs SCE. The reference electrode was a double junction Saturated Calomel Electrode. The electrode surface was 1 cm² wide. The temperature was regulated at 25°C by means of a thermostat. The solution in the cathodic compartment (hydrogen entry side) was re-circulated in order to avoid concentration variations and formation of hydrogen bubbles on the metal membrane. Anodic currents on hydrogen exit side was monitored during the permeation test up to a steady state of hydrogen permeation flux was reached. The diffusion coefficient was evaluated along the three principal directions: rolling direction (R), transverse direction across thickness (T_s) and transverse direction perpendicular to the longitudinal section (T_L). The tests were also carried out on specimens with the anodic side electroplated by palladium.

Environment

The slow strain rate tests and the corrosion-fatigue tests were carried out in aerated substitute ocean water (ASTM D1141-75 Standard Specification) under cathodic protection at –900 and –1050 mV vs SCE.

Results and discussion

All the steels showed similar stress vs strain curves in SSR tests (figure 1). The effect of hydrogen embrittlement on SSR curves is clearly evident at potentials of -1050 mV vs SCE, although the examination of fracture surface shows the presence of brittle areas and secondary cracks in the necking zone, on the specimens tested at potentials lower than -850 mV vs SCE. Thus, the reduction of area on the specimens tested in artificial seawater decreases at level much lower then air tests at very negative cathodic potentials, due to evident hydrogen embrittlement (figure 2). Table 2 summarizes the effect of the potential on the normalized reduction of area.



Figure 1. SSR curves obtained on X65B steel, in artificial seawater under cathodic protection, at 10⁻⁵ s⁻¹ strain rate.



Figure 2. Effect of potential on the morphology of fracture surface of X65B specimens, after SSR tests.

Table 2. Normalized reduction of area after SSR tests in artificial seawater under cathodic protection, at 10⁻⁵ strain rate.

Е	Steel								
(V vs SCE)	X60 ¹	X65 A	X65 B	X65 C	X80	X65 M	X85 M	X100 ¹	
-0.80	-	-	-	-	1	-	1	-	
-0.85	1	0.8	1	0.9	0.9	1 ²	0.9	1	
-0.90	1	0.7	0.8	0.7	1 ² 0.9	1 ²	1 ² 0.9	1	
-0.95	-	-	-	-	0.8	-	0.7	-	
-1.05	0.5	0.7	0.6	0.6	0.6 ²	0.7 ²	0.9 ²	0.7	
1	by Cabrin	i at al.[6]	•	•		•			

² SSR Tests in 35 g/L NaCl solution



Figure 3. Fatigue curves in air and in artificial seawater under cathodic protection at different potentials for the X65C steel

Figure 3 shows fatigue and corrosion-fatigue crack growth as a function of ΔK . The fatigue curves in air are in agreement with the Paris' law (Table 3).

$$\left(\frac{da}{dN}\right)_{F} = C \cdot \Delta K^{n} \qquad ; \qquad \Delta K_{\text{th}} \leq \Delta K \leq K_{C} (1-R) \qquad (6)$$

Paris coefficient	Steel								
	X65 A	X65 B	X65 C	X80	X65 M	X85 M			
С	6.2·10 ⁻⁹	5.3·10 ⁻⁹	2.6·10 ⁻⁹	4.2·10 ⁻⁹	4.3·10 ⁻⁹	4.6·10 ⁻⁹			
n	3.0	3.1	3.2	3.1	3.1	3.1			

Table 3. Paris coefficient by fatigue curves in air in the range 4 to 40 MPa m^{1/2} ΔK .



Figure 4: Morphology of fracture surface of the steel X65 specimen after fatigue tests at 0,2 Hz in sea water at –1050 mV vs SCE: a) air pre-cracking zones; b) C-F propagation zones during K decreasing test; c) C-F propagation zones during ∆K increasing test.

In synthetic seawater under cathodic protection, the crack growth rate becomes much higher than the rate measured in air, at intermediate values of stress intensity factor range and low loading frequencies. The crack growth morphology changes owing to the environmental effects as shown in figure 4. The effect is more pronounced as the cathodic polarization increases from -900 to -1050 mV vs SCE. At high stress intensity factor range, the curves in air and in synthetic seawater approach together. Mechanical factors prevail under crack instability conditions, as stress intensity factor approaches material fracture toughness. The crack growth rate becomes higher than the correspondent value in air above a threshold (ΔK_{eth}) that can be related with a critical value of the maximum stress intensity factor. According to the Model of Wei and Landes [3], such a value can be assumed as an estimation of critical stress intensity factor for SCC (K_{SCC}) under cyclic loading

$$K_{\rm SSC} = \Delta K_{\rm eth} / (1-R) \tag{7}$$

Table 4 describes the estimated values of ΔK_{eth} and K_{SCC} . ΔK^* is also illustrated. It is the value of ΔK for which K is always over the value of K_{SCC} during the loading cycle

$$\Delta \dot{\mathbf{K}} = \Delta \mathbf{K}_{\text{eth}} / \mathbf{R}$$
(8)

The steels show K_{SSC} of about 30-40 MPa m^{1/2}, slightly decreasing as the cathodic polarization increases from -900 mV to -1050 mV vs SCE.

Steel	Potential (mV vs SCE)	∆K _{eth} (MPa⋅m ^½)	K _{scc} (MPa⋅m ^½)	∆K [*] (MPa⋅m ^½)
X65 A	-900	14	35	23.3
703 A	-1050	13	32.5	21.7
	-900	15	37.5	25
702 B	-1050	14	35	23.3
VGE C	-900	17	42.5	28.3
X65 C	-1050	15	37.5	25
X80	-1050	15	37.5	25
X85 M	-900	14	35	23.3
	-1050	14	35	23.3

Table 4. Values of ΔK_{eth} , K_{SCC} and ΔK^* estimated from corrosion fatigue curves.



Figure 5. Corrosion contribute to fatigue crack growth rate in artificial seawater at -1050 mV vs SCE, 0.6 R, 0.2 Hz.

According to the model proposed by Wei and Landes, the crack growth rate can be considered the sum of a pure fatigue contribute $(da/dN)_F$ evaluated by means of Paris' law, and a corrosion contribute $(da/dN)_C$. Thus the contribute of corrosion can be evaluated by relation

$$\left(\frac{da}{dN}\right)_{c} = \left(\frac{da}{dN}\right)_{CF} - \left(\frac{da}{dN}\right)_{F}$$
(9)

Figure 5 shows the corrosion contribute to fatigue crack growth rate calculated by the experimental fatigue and corrosion fatigue curves at -1050 mV vs SCE, as a function of ΔK . As ΔK increases above 20 MPa m^{1/2}, the contribute due to corrosion approaches a constant value illustrated in the table 5.

It is well known that crack growth rate due to SCC is related to stress intensity factor (K). Crack growth occurs at almost a constant rate $(da/dt)_{scc}$ above K_{SCC} up to stress intensity factor approaches material fracture toughness. During fatigue cycle, SCC phenomena can only take place during the period of load increasing because during the decreasing period the crack tip is under compression. Thus, in order to estimate $(da/dt)_{scc}$ by the experimental corrosion-fatigue curves the following relation

can be adopted, valid for K always above K_{scc} during the loading cycle (that is $\Delta K > \Delta K^*$), by considering the frequency (f)

$$\left(\frac{da}{dt}\right)_{SCC} = \left(\frac{da}{dN}\right)_{C} \cdot 2 \cdot f \tag{10}$$

Steel	-900 n	nV vs SCE	-1050 mV vs SCE			
	(mm/cycle)	Standard deviation	(mm/cycle)	Standard deviation		
X65 A	2.4·10 ⁻⁴	3.8·10 ⁻⁵	3.1.10-4	2.1·10 ⁻⁵		
X65 B	2.1·10 ⁻⁴	5.8·10 ⁻⁵	3.8·10 ⁻⁴	2.9·10 ⁻⁵		
X65 C	2.9·10 ⁻⁴	2.3·10 ⁻⁵	4.0·10 ⁻⁴	3.9·10 ⁻⁵		
X65 M	2.3·10 ⁻⁴	3.6·10 ⁻⁵	3.8·10 ⁻⁴	2.0·10 ⁻⁵		
X80	3.5·10 ⁻⁴	5.1·10 ⁻⁵	6.5·10 ⁻⁴	3.1·10 ⁻⁵		
X85 M	2.8·10 ⁻⁴	2.6 [.] 10 ⁻⁵	4.6.10-4	3.6·10 ⁻⁵		

Table 5. Average values and standard deviation of stationary corrosion contribute (mm/cycle) to fatigue crack growth rate in artificial seawater at 0.6 R, 0.2 Hz.

Figure 6 shows the estimated $(da/dt)_{scc}$ as a function of the mechanical properties of the steel. Values of about 10⁻⁴ mm/s are found, slightly increasing with the cathodic polarization.

At -900 mV vs SCE potential, the differences between the steels are comparable to the experimental scattering. At -1050 mV vs SCE, the steels with microstructures of banded ferrite with pearlite or bainite or martensite inside the bands presented increasing of the cracks rate with the mechanical properties. A similar behaviour, but less pronounced differences, was also observed for the martensitic steels.



Figure 6. SCC crack growth rate estimated through Wei and Landes Model from Corrosion Fatigue data in artificial seawater, under cathodic protection.

The hydrogen embrittlement phenomena take place when the atomic hydrogen reaches a critical concentration inside the metal lattice. The hydrogen permeates the metal lattice through a diffusion mechanism. Thus, the hydrogen transport depends on hydrogen concentration on the surface exposed to the environment and the hydrogen diffusion coefficient. Table 6 summarizes the hydrogen diffusion coefficient. A number of test were obtained on specimens with the hydrogen exit

side electroplated with palladium but no systematic difference with respect to the specimens without electroplating was noted. The results on the martensite steels were obtained by Zucchi et al. [12] during a coordinated research project performed on these materials. The X60 steel was drawn from a pipeline placed in the '60s which showed field cracking under cathodic protection and very slow straining induced by settling of soil.

Because hydrogen mainly diffuses through pre-existent paths as ferritic grain boundary [13], the diffusivity increases for the control rolled steel with respect to hot rolled steel owing to their fine ferritic grain. In the tempered martensite steel, the effect of very fine structures is contrasted by very fine and dispersed precipitates that can act as traps subtracting hydrogen to diffusion. Owing to yield strength depends on ferritic grain dimension and formation of hardening precipitates, a correlation between yield strength and diffusivity can be expected. However, figure 7 also shows important differences in behavior between steel with similar microstructures. The hot rolled steels by recent production have much lower diffusivities than the X60 steel produced in '60s, that has rather high content of carbon and sulfur and coarse microstructure with very pronounced pearlite band. In order to explain such a difference, less efficient hydrogen trapping could be supposed.

	Steel									
Diffusion direction	Ferrite/pearlite				Ferrite/bainite	Ferrite/martensite	Tempered martens			
	X60	X65 A	X65 B	X65 C	X80	X100	X65 M	X85 N		
Rolling	8.5 ^b	0.9 1.9 ^{Pd}	2.2 2.0 ^{Pd}	4.4 4.1 ^{Pd}	-	3.9 ^b	-	-		
Transverse T _s	5.6 ^b 11.5 ^a	2.6 2.3 ^{Pd}	2.7 2.2 1.7 1.7 ^{Pd}	4.2 2.4 ^{Pd}	4.7 ^a	3.9 ^b	4.2 ^a	4.0 ^a		
Transverse T _L	-	2.0	1.4	4.6	-	_	-	-		

Table 6. Hydrogen diffusion coefficient $(10^{-7} \text{ cm}^2/\text{s})$ measured by means of permeation test.

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^a by Zucchi et al. [12] ^b by Cabrini et al. [6]

 2.0^{Pd}

Pd specimens with the anodic surface electrochemically covered with palladium

3.1 ^{Pd}

 2.0^{Pd}



Figure 7. Hydrogen diffusion coefficient as a function of mechanical properties of the steel



Figure 8. Normalized Reduction of Area after SSR tests in artificial seawater under cathodic protection at -1050 mV vs SCE as function of hydrogen diffusion coefficient



Figure 9. Crack growth rate for SCC evaluated from corrosion fatigue curves in artificial seawater under cathodic protection by means of Wei and Landes model, as function of hydrogen diffusion coefficient

Figures 8 and 9 relate the hydrogen diffusion coefficient with the results of SSR test and the crack growth rates for stress corrosion cracking estimated by corrosion fatigue tests. As hydrogen diffusivity increase, the effect of hydrogen embrittlement in SSR test and crack propagation become more evident. Doubling in crack growth rates can be noted as the diffusivity increase from $2.7 \cdot 10^{-7}$ cm²/s to $5 \cdot 10^{-7}$ cm²/s at the potential of -1050 mV vs SCE.

It usually assumed that Hydrogen embrittlement occurs through mechanisms involving hydrogen transport to the plastic strain zone at the crack tip and a critical concentration for failure initiation.

An increase of hydrogen transportation to the plastic zone of a propagating crack is expected to arise crack growth rates, because the critical content of hydrogen can be reached in a short time. However, a complex situation has to be hypothesized as far as the relation between material susceptibility and hydrogen diffusivity.

Litterature data [14,15] confirm an increase of 4 times in crack growth rates of C-Mn steels during corrosion-fatigue tests in seawater under cathodic protection, at 20 MPa.m^{1/2} ΔK in the range from 0°C a 25°C. This is in accordance with the increasing effect of temperature on diffusion coefficient [16].

On the other hand, it is also well known that with an increase in temperature from room temperature the hydrogen deleterious effect minimizes, in despite of a considerable increase of diffusivity. Actually, hydrogen transport also depends on trapping into the metal and surface concentration of hydrogen adsorbed on steel. The surface concentration decreases with temperature because recombination reaction to produce hydrogen molecule is favored. Decreasing of probability of being trapped by reversible and irreversible traps, and increasing of hydrogen solubility in the lattice also occur.

Under the environmental conditions adopted for testing, the surface concentration of adsorbed hydrogen mainly depends on cathodic polarization, in absence of any hydrogen recombination poison like sulfide. Thus, in order to interpret the experimental data a constant surface concentration could be roughly assumed. The data show an effect of microstructure that cannot only be related to diffusivity. Tempered martensite steels show low susceptibility to environmental effects during both SSR test and corrosion-fatigue test, owning to their fine microstructure with very fine and dispersed precipitate. These fine microstructures are able to distribute hydrogen over spread irreversible traps and reduce probability of reaching local critic hydrogen concentration for cracking initiation.

Finally, the diffusivity measured in this research has been obtained on steel without any deformation during test. Hydrogen diffusion is dramatically influenced by the presence of deformation as found in previous research [17]. Thus, the correlation between steel susceptibility and hydrogen diffusivity shown in figures 8 and 9 could be related, at least partially, to the influence of microstructure both on diffusivity and critical condition for the initiation of environmental assisted cracking.

Conclusions

The results confirm that the hydrogen embrittlement in artificial seawater under cathodic protection can rise on steel with tensile yield strength in the range of 400-660 MPa in presence of cathodic polarization and dynamic loading conditions.

Hydrogen Embrittlement effects induced a decreasing of Reduction of Area after Slow Strain Rate tests. The results pointed out the best resistance to hydrogen embrittlement for steels with fine microstructure of tempered martensite.

Due to corrosion-fatigue, the crack growth rate increases in synthetic seawater from 10 to 25 times the rate measured in air at stress intensity factor range exceeding 18 MPa m^{1/2}, at 0.2 Hz loading frequency. Increasing cathodic polarization from –900 mV to –1050 mV vs SCE produced a slight increase in corrosion fatigue propagation. The rising contribute due to hydrogen embrittlement stress corrosion cracking to fatigue crack propagation was estimated through Wei and Landes model. A correlation between diffusion coefficient and the results of SSR test and Corrosion-Fatigue tests was found.

References

- 1. Cabrini M., Pistone V., Sinigaglia E., Tarenzi T., "Unique HSC scenario leads to gas line failure", Oil & Gas Journal, 6, 61-65, (2000)
- Carter C.S.and Hyat M.V.t, Review of Stress Corrosion Cracking in Low Alloy Steels with Yield Strength Below 150 ksi, in: R.W. Staehle, J.Hochman, R.D. McCrigth, J.E. Slater (Eds.), Stress Corrosion Cracking and Hydrogen Embrittlement of Iron Base Alloys, NACE, Houston TX, pp. 524-600, 1977
- 3. Wei R.P., Landes J.D., "Correlation between sustained-load and fatigue crack growth in high strength steel", Materials Research and Standards 9, 7, (1969)
- Barsanti L., Cabrini M., Pastore T., Spinelli C., "Effect of microstructure on the behaviour of HSLA steels to hydrogen embrittlement under cathodic protection", Proceedings of EICM04, Banff (Canada), 20th – 24th September 2004, in press
- 5. Tau L., Chan S.L.I., Shin C.S., "Hydrogen enhanced fatigue crack propagation of bainitic and tempered martensitic steels", Corrosion Science, **38** 2049-2060 (1996).
- 6. Cabrini M., Cogliati O., Maffi S., "Effetto della microstruttura sulla diffusione dell'idrogeno in acciai al carbonio per pipeline", La metallurgia italiana, **3**, 13-20, (2003)
- Cabrini M., Migliardi L., Pastore T., Spinelli C., Synergic Effect Of Cathodic Potential And Strain Rate On Hydrogen Embrittlement Of Different Microstructure Carbon Steels, H-CDI: Hydrogen – Corrosion Deformation Interaction, Jackson Lake Lodge, Wyoming, (2001), pp.16-21.
- 8. Cigada A., Mazza B., Pastore T., Pedeferri P., "Fatigue Crack Growth of HSLA Steel in Seawater", Metals/Materials Technology Series, 8520-011, America Society for Metals, Metals Park, Ohio, pp. 1–7, (1985)
- 9. Standard Test Measurement of Fatigue Crack Growth Rates; ASTM, E 647-00.
- 10. Devanathan M.A.V., Stachurski Z., Proc. Roy. Soc. London A 270 90, (1962)
- 11. Cabrini M., Pastore T., "Effect of microstructure on hydrogen diffusion in high strength low alloyed steels", 16th International Corrosion Congress, Proceedings on CD Rom, Beijing, September 2005
- 12. Zucchi F., Grassi V., Frignani A., Monticelli C., Trabanelli G., in Atti del Convegno Nazionale AIM, Associazione Italiana di Metallurgia, Vicenza, November 17-19, (2004)
- 13. McBreen J., Nanis L., Beck W., J. Electrochemical Soc. 113, 1218, (1966)
- 14. P.M.Scott "Effects of environmental on crack propagation", in Developments in fracture mechanics 1, London Applied Science Publisher LTD, pp.222-253 (1979).
- 15. Jaske et al. "Corrosion fatigue of metals in marine environments" Heidelberg, Spring-Verlag, (1981)

- Bollinghaus Th., Hoffmeister H., Middel C., Welding in the Word, **37**, 1, 16-23 (1996)
 Cabrini M., Razzini G., Maffi S., "Evaluation of Hydrogen embrittlement behaviour by means permeation current measure in slow strain rate conditions of a micro-alloyed steel", Materials Science Forum Vols. 286-292 (1998) pp 1245-1256, Trans Tech Publications, Switzerland
- 18. Bolzoni F., Cabrini M., Spinelli C., "Hydrogen diffusion and hydrogen embrittlement behaviour of two high strength pipeline steels", Eurocorr 2001, Riva del Garda, Proceedings on CD-ROM, AIM, Milano, pag. 1-10