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Fatigue behavior of a high strength steel in vacuum, in air and in 3.5% NaCl solution under cathodic protection

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Fatigue resistance of the high strength steel SE702 (A517 type) in air and in 3.5% NaCl solution under cathodic protection is compared with the fatigue resistance in high vacuum in order to identify the effects of the environment on the fatigue damage processes. Fatigue lifetimes, obtained in air and in vacuum, reveal a strong influence of the gaseous environment and a significant effect of the test frequency in air. Experiments performed in aqueous environment under cathodic protection exhibit an additional reduction of the fatigue resistance relative to results obtained in air. Interrupted fatigue tests show that both ambient air and NaCl solution reduce initiation and propagation stages compared with vacuum. The gaseous environment tends to accelerate transgranular cracking whereas the NaCl environment strongly favors 'intergranular' propagation, enhancing the crack growth rate relative to air. In some conditions, the air appears to be more detrimental than the NaCl solution with cathodic protection on the initiation stage. This work confirms the key role of hydrogen in NaCl solution and the main role of water vapor in air. But it also shows that the role of oxygen can not be neglected in air and especially for the crack initiation stage.

Keywords: Fatigue; Environment; Saline solution; Cathodic protection; High strength steel

1. Introduction

The aim of the present work is to investigate the fatigue resistance of a high strength steel, denoted SE702, in specific environmental conditions defined according to its use in offshore platforms. Legs of such structures, immersed in the sea, are usually protected from general corrosion by the application of a cathodic potential. This potential induces the reduction of water, generating hydrogen at the material surface. Hydrogen can enter the material and lower its resistance to the fatigue loading imposed by the waves and the wind. Moreover, high strength steels are known to exhibit a high susceptibility to hydrogen embrittlement compared with conventional construction steels. Therefore, their fatigue resistance must be assessed, in order to take into account the interactions between mechanical loading and environment.

Most of the works carried out to characterize the effect of a cathodic protection on fatigue resistance deal with long-crack growth behavior [1,2] or welded joints resistance [3]. In contrast, studies on fatigue lifetime very few and are generally devoted to the low-cycle fatigue range [4]. The purpose of the present work is to identify the effects of an aqueous NaCl environment with a cathodic potential on the initiation and growth of microcracks in a high strength steel in the high cycle fatigue range.

The effect of an aqueous environment is generally evaluated by comparison with the fatigue results obtained in air considered as the standard reference environment [3,5]. However, it is well known that fatigue lives of most metals tested in air are considerably reduced when compared with tests performed in vacuum or in inert gases [6]. Therefore, a good comprehension of

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environment effects on fatigue mechanisms requires to refer to the fatigue behavior in an inert atmosphere. The comparison between an active medium and air is justified by the fact that air is the most common environment for the use of structural steels. Moreover, the fatigue processes of this class of steels is better documented in this environment. Nevertheless, it is necessary to keep in mind that the air is an active environment with regard to fatigue.

Consequently, in the present study, fatigue experiments on the SE702 were conducted in a high vacuum, as reference. The effects of air and NaCl solution with cathodic protection were then established by comparison with the results obtained in vacuum for the same mechanical loading conditions. The comparison between the results obtained in these three environments will be used to discuss the effects of the different active species.

2. Material and experimental procedure

2.1. Material

The tested steel, denoted SE702 (comparable to an A517 Grade Q Mod.), was manufactured by Creusot-Loire Industrie. The chemical composition is indicated in Table 1. The mechanical properties of the material were recorded as follows: 0.2% yield stress 780 MPa, ultimate tensile strength 860 MPa, elongation 19%. This steel, quenched and tempered, exhibits a tempered refined bainite/martensite microstructure illustrated in Fig. 1; the prior austenite grain size varies from 10 to 100 μm.

Cylindrical smooth specimens, with a diameter of 5 mm and a gauge length of 13 mm, were cut off in a 160 mm thick sheet, parallel to the rolling direction, at a constant distance of 40 mm from the initial ingot surface. After machining, the specimens were mechanically polished with SiC grinding papers and then with diamond pastes up to 1 μm grade.

2.2. Experimental procedure

Fatigue experiments were performed under stress control at two load ratios: $R = 0.1$ and $R = -1$. Experiments in air (relative humidity $> 40\%$) and in vacuum ($< 2 \times 10^{-4}$ Pa) were performed on a hydraulic machine at 20 Hz, while the experiments in the aqueous environment were conducted at 0.17 Hz on an electro-

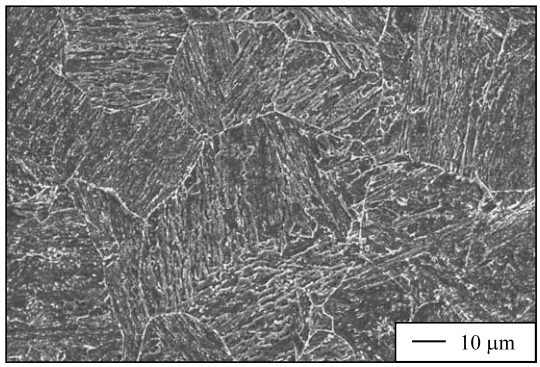


Fig. 1. SEM picture of the microstructure of the SE702 after chemical etching (nital 4%).

mechanical equipment. This frequency corresponds to a representative mean value of the swell frequency in the North Sea. Some experiments were also performed at 0.17 Hz in air in order to estimate the effects of the frequency in this environment and to allow a direct comparison with the NaCl solution data. All the fatigue tests were conducted at room temperature.

For the experiments in the aqueous environment, specimens were polarized with a potentiostat by using a three electrodes electrochemical cell [7] (Fig. 2). The working electrode was the specimen itself, the counter electrode was made of platinized titanium, and the reference electrode was a saturated calomel electrode (SCE). The results presented in this paper were obtained with a cathodic polarization of -1050 mV/SCE corresponding to an overprotection level. The 35 g l^{-1} NaCl solution was aerated and neutralized, the pH being adjusted to 8.1. The specimen is polarized during 20 h in order to stabilize the cathodic current before starting fatigue experiments.

Specimen gauge lengths and fracture surfaces were observed by scanning electron microscope (SEM) after

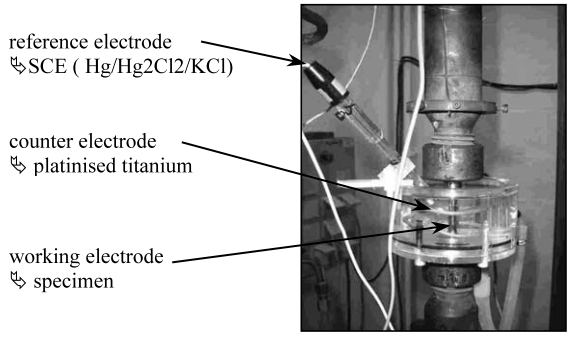


Fig. 2. Experimental device with the three-electrodes electrochemical cell.

Table 1
Chemical composition of the SE702 (wt.%)

Fe	C	Mn	Si	Ni	Cr	Mo	Cu	P	S
Bal.	0.137	1.129	0.213	2.214	0.541	0.526	0.179	0.005	0.0009

failure. Surface deformation, crack initiation and propagation processes were analyzed for the different environments and loading conditions. The relative parts of initiation and propagation were evaluated by performing regularly interrupted experiments associated with SEM observations.

3. Results

3.1. Fatigue lifetimes

3.1.1. Air

Fig. 3 gives the S–N curves (stress vs. number of cycles to failure) obtained at $R=-1$ and $R=0.1$ in vacuum at 20 Hz and in air at 20 Hz and 0.17 Hz. At $R=-1$, a strong detrimental role of the gaseous environment on fatigue lives is clearly revealed. At $\sigma_{max}=450$ MPa, the fatigue life is approximately 25 times lower in air (20 Hz) than in the inert environment. This strong influence of ambient air is still more pronounced at $R=0.1$. In this case, the reduction of fatigue lives in air at 20 Hz, compared to vacum, is greater the one decade even at high stresses.

The results of the experiments conducted in air at 0.17 Hz show that the decrease of frequency in this environment generally causes a reduction in fatigue lifetimes except for the highest stress level investigated at $R=0.1$. This influence of test frequency is all the more important that the maximal stress is low. However, at the lower stress amplitudes, the reduction factor, Nf(air-20 Hz)/Nf(air-0.17 Hz) is approximately 3 whereas the ratio between the fatigue lifetime in vacum, at 20 Hz and the lifetime in air, at 0.17 Hz reaches two decades!

3.1.2. NaCl solution

For the two load ratios at 0.17 Hz, the experiments in the NaCl solution under cathodic protection (−1050 mV/SCE) show that fatigue lives are also considerably reduced with regard to vacuum. Moreover, fatigue lifetimes are shorter or equal than in air, depending on the mechanical conditions (Fig. 4). However, the reduction factor between NaCl and air: Nf(air-0.17 Hz)/Nf(NaCl-0.17 Hz) is relatively limited with regard to the one between air and vacuum. For example, at $R=0.1$, this factor is smaller than 4. At $R=-1$ and for a low stress ($\sigma_{max}=450$ MPa), the reduction factor is approximately 5 (average value) compared to the air and 300 compared to vacuum! For a higher stress ($\sigma_{max}=550$ MPa), there is no significant difference on fatigue lives. Nevertheless, as we will see in the following sections, this does not imply that crack initiation and crack growth kinetics are identical.

3.2. Initiation sites and surface damage

3.2.1. Initiation sites

For the push–pull mode, $R=-1$, the surface deformation induced by cycling is characterized by a localized shearing at martensite lath boundaries leading to the creation of intense slip bands with small extrusions. Whatever the environment, the initiation of fatigue cracks mainly occurred in some of these slip bands, within the prior austenite grains. This transgranular

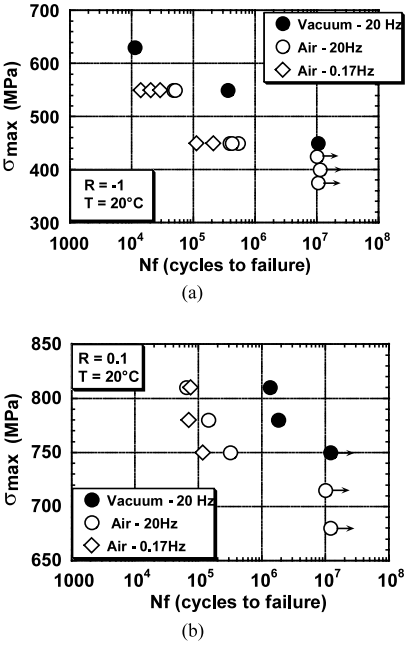


Fig. 3. S–N curves at (a) $R=-1$ and (b) $R=0.1$ in vacuum and in air; influence of the frequency.

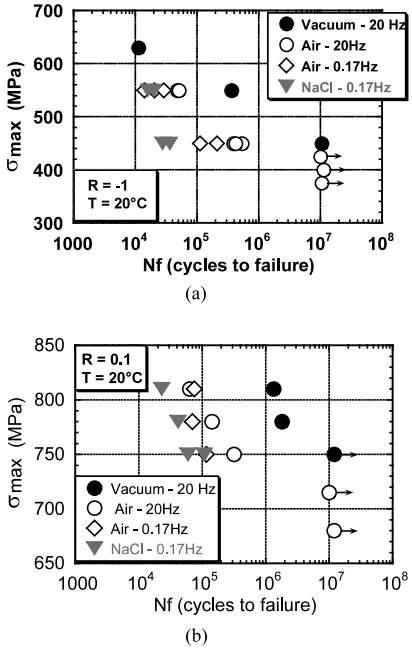
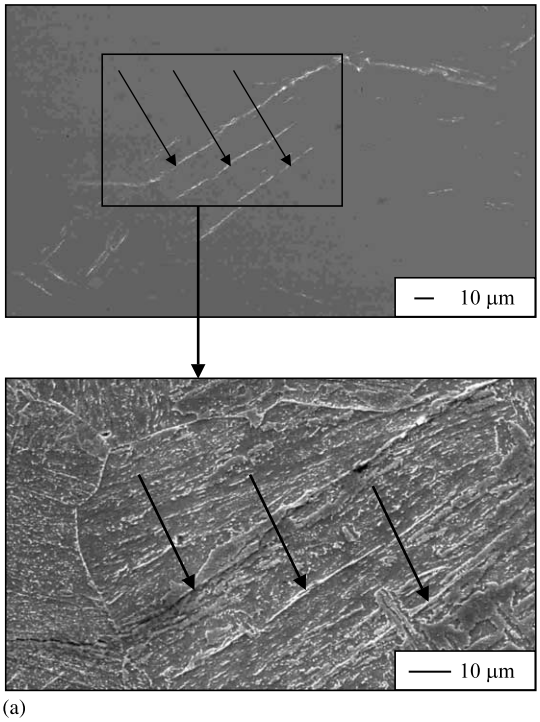
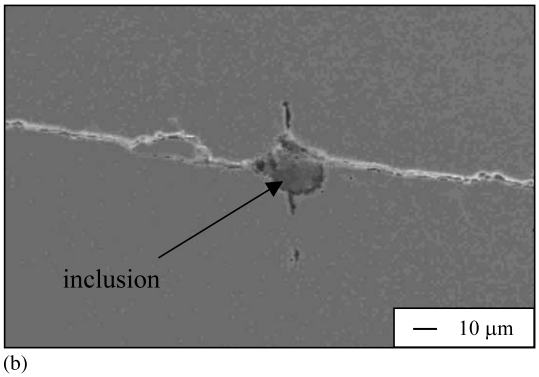


Fig. 4. S–N curves at (a) $R=-1$ and (b) $R=0.1$ in vacuum, in air and in 3.5% NaCl solution with cathodic protection.



(a)



(b)

Fig. 5. Typical initiation sites (a) at $R = -1$, before and after chemical etching, and (b) $R = 0.1$.

initiation which can be called 'interlamellar', corresponds to the stage I cracking period (Fig. 5a).

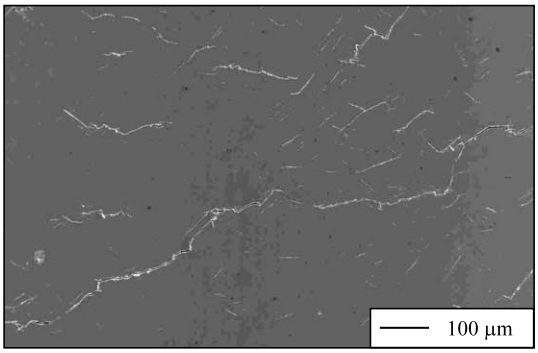
For the positive load ratio, $R = 0.1$, cyclic plasticity level and damage accumulation are lower than in the push–pull mode, due to lower alternate stress amplitudes. Consequently, the slip band density is strongly reduced compared with $R = -1$ tests. In air and in vacuum, the initiation sites are small surface defects, such as porosities or inclusions, that do not seem to play a major role in the push–pull mode (Fig. 5b). In these environments, the accumulation of cyclic damage around these defects increases local deformation favoring the initiation of cracks. In NaCl solution, crack initiation is mainly 'interlamellar' and more rarely intergranular.

3.2.2. Surface damage at $R = -1$

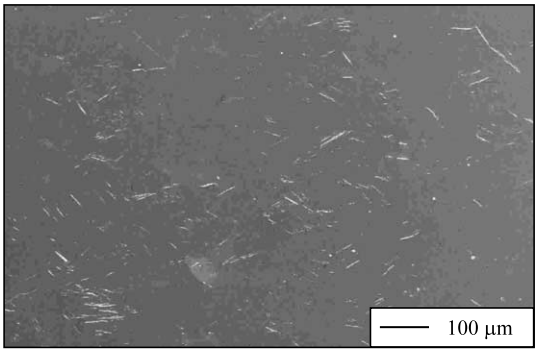
A high slip band density is observed on vacuum failed specimens, due to the important cyclic plastic deformation accumulation, associated with the high number of cycles required to reach failure. However, the specimens cycled in this environment always exhibit a limited number of cracks at failure.

In air, at low stress amplitude ($\sigma_{max} = 450$ MPa), a very low slip band density and very few secondary cracks were detected. In contrast, at a higher stress level ($\sigma_{max} = 550$ MPa), a high density of microcracks was observed. This multicracking process favors the formation of the fatal crack by coalescence (Fig. 6a). At low frequency (0.17 Hz), whatever the stress level, the specimens exhibit a higher density of secondary cracks than at 20 Hz. In spite of such noticeable differences in terms of crack density, the surface crack propagation mechanisms are very similar in air and in vacuum.

In NaCl solution with cathodic protection, at low stress amplitude, the slip band density is very low. Crack damage is limited to the fatal crack so that no secondary crack was observed. Although the slip band density is high for the high stress level, only very few secondary cracks were initiated (Fig. 6b). Failure proceeded from the growth of a single crack without any coalescence. Cyclic damage appears therefore to be much more localized than in air.



(a)



(b)

Fig. 6. Surface damage (a) in air and (b) in 3.5% NaCl solution with cathodic protection at failure after respectively 20.298 and 20.230 cycles at $R = -1$; $\sigma_{max} = 550$ MPa.

Table 2
Number of surface cracks per sample as a function of environment and loading conditions

Environment	$\sigma_{max} = 450$ MPa		$\sigma_{max} = 550$ MPa	
	Number of test samples	Number of cracks	Number of test samples	Number of cracks
Vacuum (20 Hz)	1	1	2	4–9
Air (20 Hz)	3	1–3	1	~ 50
Air (0.17 Hz)	2	9–10	3	~ 120
NaCl (0.17 Hz)	3	1	4	2–3

The most relevant result about damage mechanisms is this multicracking process observed in air. These differences, in terms of crack density, appear in Table 2, where the average number of cracks at failure is reported as a function of testing conditions.

3.3. Relative parts of initiation and propagation stages in total fatigue lifetime at $R = -1$

Interrupted fatigue experiments were performed at $R = -1$, $\sigma_{max} = 450$ MPa and $\sigma_{max} = 550$ MPa, in order to estimate the relative parts of initiation and propagation in each environment. To separate these two stages, a criterion for initiation is required. Cracks being initiated within the prior austenite grains (transgranular cracking mode), the initiation stage is considered to be complete when the fatal crack crosses the boundary of the first prior austenite grain. The specimens gauge lengths were regularly observed using a binocular lens and SEM. This method allows to detect 50 μm long surface cracks, corresponding to the mean value of the grain size.

The relative part of initiation and propagation in the fatigue lifetime has not been precisely established in vacuum. Nevertheless, it is clear that both gaseous and aqueous environments enhance initiation as well as propagation kinetics compared with vacuum. Indeed, for the number of cycles corresponding to the maximal lifetime observed in air, no crack is initiated yet in vacuum, for both stress levels. Moreover, a study of the natural crack growth evolution in this environment has shown that the difference in crack growth kinetics between air and vacuum was larger than one decade for crack depths smaller than 250 μm [8].

Fig. 7 reports an evaluation of Ni, the number of cycles to initiate a crack and Np, the number of cycles to propagate a crack, in air at 20 Hz and 0.17 Hz and in the NaCl solution at 0.17 Hz. The reported data correspond to mean values calculated on several tested specimens.

Whatever the loading conditions, the effect of frequency in air involves a reduction of both initiation and propagation stages.

In the NaCl solution, crack propagation stage is always highly shortened compared with the results in air. In contrast, the effect on crack initiation kinetics

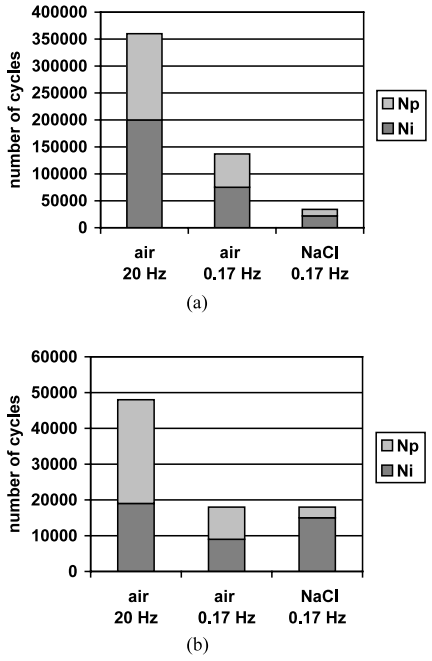


Fig. 7. Relative part of initiation and propagation at $R = -1$; (a) $\sigma_{max} = 450$ MPa and (b) $\sigma_{max} = 550$ MPa.

depends on the stress amplitude level. At $\sigma_{max} = 450$ MPa, the number of cycles necessary to initiate a crack is lower than the one observed in air (much smaller than in vacuum). Therefore, at this stress level, the fatigue life reduction relative to air is due to a reduction of both initiation and propagation stages. At $\sigma_{max} = 550$ MPa, fatigue lives in air and in NaCl solution are comparable. However, in this case the initiation of the very first cracks occurs earlier in air than in NaCl solution. Consequently, for these particular loading conditions, the air has a higher detrimental effects on crack initiation than the NaCl solution with a cathodic potential of -1050 mV/SCE. This faster initiation is associated with multicracking process. Such behavior was observed on several specimens and can be considered as a well-established experimental fact.

3.4. Fractographic analysis

Detailed examinations of the fracture surface morphologies have been made by SEM on specimens

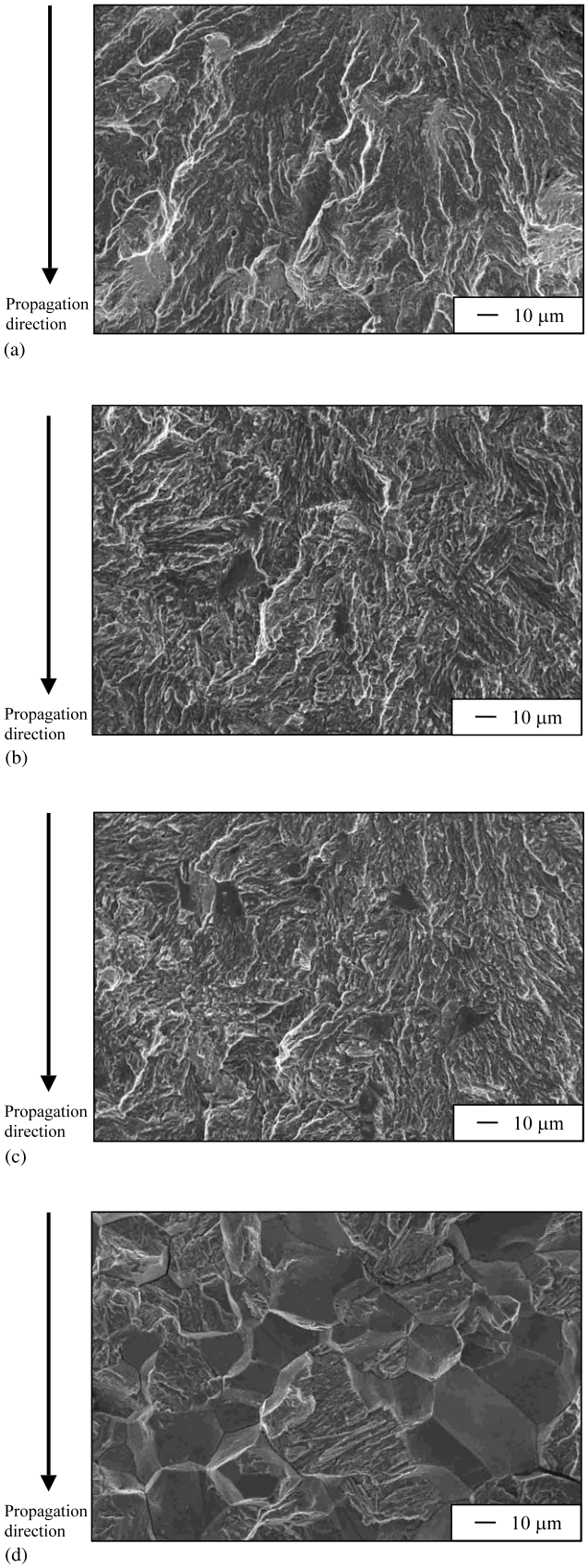


Fig. 8. SEM images of fracture surfaces (a) in vacuum-20 Hz, (b) air-20 Hz, (c) air-0.17 Hz, (d) 3.5% NaCl solution with cathodic protection-0.17 Hz. $R = 0.1$, $\sigma_{max} = 780$ MPa.

tested at $R = 0.1$. These observations have shown that the crack propagation mode is highly dependent on the environment (Fig. 8a–d).

Crack propagation in air and in vacuum is clearly transgranular with respect to the prior austenite grains. The fracture surfaces are nevertheless more brittle in air (Fig. 8b) than in vacuum (Fig. 8a). Some brittle facets, assumed to be intergranular, were also observed in air. It is interesting to notice on Fig. 8c that at low frequency (0.17 Hz), the relative amount of these brittle facets is slightly higher than at 20 Hz. In the NaCl solution (Fig. 8d), the fracture surface morphology is much more brittle than in air with many intergranular separations along prior austenite grain boundaries. The other fraction of the fracture surface mainly consists of brittle transgranular cleavage-like facets. Only few transgranular failures, comparable with those found in air, have been observed. The cleavage-like facets exhibit distinct geometrical markings related to the features of martensite laths.

4. Discussion

The discussion about the influence of the environment on the fatigue behavior of the SE702 is organized as follows:

- First, the comparison of the results obtained in NaCl solution with those obtained in vacuum will allow us to analyze the effect of hydrogen. Indeed, for the overprotection conditions investigated here, the role of oxidation in the aqueous environment can be excluded.
- The role of the various active species present in air, mainly oxygen and water vapor, will be discussed in a second time by comparison with results obtained in vacuum on the one hand and in NaCl solution on the other hand. The results obtained in this saline solution are very helpful to understand the environmental processes acting in air; they allow to separate hydrogen effects from oxygen and water vapor effects.

4.1. Fatigue behavior in the NaCl solution

In saline solutions, cathodic protection conditions lead to a considerable reduction of fatigue lives compared with vacuum. This deleterious effect is due to hydrogen produced by the reduction of water which strongly accelerates both initiation and propagation stages. Hydrogen produced by the cathodic protection is adsorbed at the metal surface and can either produce hydrogen gas (H_2) by recombination or enter the material.

Over the past decades, fatigue crack growth in saline solutions under cathodic protection has been widely studied [2]. There is a general agreement to explain the embrittlement of steels in fatigue by the synergetic action of absorbed hydrogen and cyclically applied stresses. The role of oxygen in this environment can be considered as negligible since in the present polarization conditions no oxide layer is created: the oxide formed before the immersion of the specimen is reduced during the pre-polarization period and the fresh metallic surfaces created by cycling can not react with the dissolved oxygen.

The deleterious effect of hydrogen in this type of environment with an overprotection level can be explained by the 'hydrogen induced decohesion' (HID) model, suggested by Troiano [9] who supposed that hydrogen lowers the interatomic bonding energies. Oriani [10,11] developed this idea and proposed that a critical concentration of hydrogen accumulated into particular sites, leads to an embrittlement of the material by decohesion. The enhancement of crack growth by absorbed hydrogen can also be explained by the theory of 'hydrogen enhanced local plasticity' (HELP) introduced by Beachem [12] and developed by Birnbaum [13]. This model suggests a strong localization of the deformation at the crack tip in relation with the presence of hydrogen at a microscopic scale. In particular, hydrogen can promote both creation and mobility of dislocations [14,15], causing localized high deformation regions. This kind of model does not imply necessarily an embrittlement of fracture surfaces in an active environment.

The strong embrittlement of fracture surfaces found in the present work rather suggests a decohesion mechanism. Prior-austenite grain boundaries and martensite lath interfaces appear to be preferential sites for hydrogen accumulation in the crack propagation stage. In this environment, the role of hydrogen in the embrittlement process is clearly identified and explains the marked crack growth acceleration. Nevertheless, a coupling effect with local plasticity cannot be excluded. For instance, hydrogen transport by dislocations [16] may increase hydrogen local concentrations at interfaces.

Very few studies deal with the effects of a cathodic polarization in NaCl solution on fatigue crack initiation. The present work clearly shows, by comparison with tests performed in vacuum, a strong effect of this environment on crack initiation associated with hydrogen effects. It can be noticed that initiation sites are mainly interlamellar and more rarely intergranular even at $R = 0.1$, at which slip activity at martensite lath interfaces is very weak. This observation suggests that hydrogen also plays a crucial role at these interfaces for microcrack nucleation.

4.2. Fatigue behavior in air

The results obtained in air and in vacuum allow us to conclude that both initiation and propagation kinetics are strongly enhanced in air.

In the case of steels, at room temperature, many authors have reported the harmful influence of the water vapor present in air, especially on the propagation stage. This influence was first mentioned by Dalhberg, who observed a crack growth rate enhancement by a factor of 10 in humid air compared with dry air on the AISI-4340 high strength steel [17]. Other studies carried out on another high strength steel (30NCD16), confirmed this deleterious action of water vapor on crack propagation [18]. The role of oxygen, through oxidation mechanisms, is generally considered to be negligible on crack growth compared with the water vapor effects, although some authors supposed that oxidation accelerates crack growth by reducing slip reversibility at the crack tip [19].

The enhancement of propagation kinetics in air compared with vacuum is usually explained either by the adsorption of water vapor or by a hydrogen assisted process. The theory of 'adsorption assisted cracking' [20,21] suggested that the adsorption of gaseous species could reduce the Griffith's crack surface energy. This idea was complemented by Lynch [22] in his 'adsorbed hydrogen induced plasticity' model. He proposed that hydrogen adsorption at the crack tip would induce a high strain localization by favoring the injection of dislocations, enhancing by this way the contribution of mechanical loading to crack growth. On the other hand, according to Wei et al. [23], the influence of the environment is due to a 'hydrogen embrittlement' process. These authors suggested that the water vapor adsorption at the vicinity of the crack tip is followed by a dissociative chemical absorption and a hydrogen penetration. Then, hydrogen diffuses into the plastic zone, ahead of the crack tip, where hydrogen-metal interactions lead to an embrittlement in the process zone. Petit and Hénaff [18,24], have proposed a model in which fatigue crack growth is assisted either by the adsorption of water vapor or by hydrogen. According to these authors, the crack growth rate enhancement is exclusively due to the adsorption assisted mechanism above a critical rate. For lower crack growth rates, the role of hydrogen resulting from water vapor dissociation, becomes predominant. In the present study, the effects of the gaseous environment on small natural cracks growth are revealed by a slight embrittlement at martensite lath interfaces but above all by the presence of few intergranular facets. This suggests a contribution of hydrogen to damage mechanisms. The reduced amount of these facets in air and the longer propagation stage, compared with the NaCl solution, are evidence that the detrimental environmental effects of the gaseous

environment on crack growth are less severe than in the case of the saline solution.

The frequency effect on damage kinetics in air and especially the increase of the number of intergranular facets at 0.17 Hz tends to confirm the hydrogen contribution. This frequency dependence in gaseous environments containing water vapor has already been shown by Wei and Ressler [25,26]. Because of the high water vapor pressure in air ($>10^3$ Pa), the adsorption processes occurs quasi instantaneously so that the saturated adsorption regime is reached at both studied frequencies. Consequently, the frequency dependence must be mainly linked to the hydrogen assisted process. Indeed, this mechanism depends on the dissociative water vapor absorption, which is time dependent. This could explain why the air has a higher effect at 0.17 Hz than at 20 Hz. Nevertheless, the propagation mode in air is relatively unchanged compared with vacuum: the propagation mode remains transgranular. Thus, absorbed hydrogen has probably also an effect on propagation without a generalized decohesion fracture mode. In air, absorbed hydrogen would act in this case rather through a HELP mechanism favoring slip localization at the crack tip. However, the contribution of adsorption (water vapor and oxygen) remains probably predominant in crack growth acceleration relative to vacuum. Moreover, the contribution of oxidation can not be excluded.

The detrimental influence of the air on crack initiation has been established for several metals and alloys [27,28] but the processes by which environment affect the crack initiation stage are not well documented compared with the one for the crack propagation stage. The present work shows that the air has a strong influence on crack initiation in the case of the SE702, especially at $R=-1$ and $\sigma_{max}=550$ MPa, for which multicracking and coalescence processes are predominant. Adsorption mechanisms as well as hydrogen effects quoted to explain crack growth acceleration could be transposed for the initiation stage. Furthermore, the role of oxygen, generally supposed to be negligible on the propagation stage, must not be neglected in crack initiation. In the literature, oxygen is presented as playing either a detrimental or a protective role on crack initiation. On the one hand, the quasi-instantaneous oxidation of the newly created surfaces causes a decrease of slip reversibility and accelerates crack initiation. On the other hand, an oxide layer may have a beneficial effect by lowering the water vapor adsorption and hydrogen diffusion into the substrate. When compared with the NaCl solution, crack initiation appears to be accelerated by the gaseous environment at relatively high stress amplitude ($R=-1$, $\sigma_{max}=550$ MPa) whereas at lower stress amplitudes ($R=-1$, $\sigma_{max}=450$ MPa and $R=0.1$) the aqueous environment seems to be more deleterious. For the high stress amplitude in air, cyclic shearing at martensite lath boundaries could lead to local fractures of the oxide layer which favor locally water vapor adsorption and dissociation. Such processes could favor hydrogen diffusion and strain localization at martensite lath interfaces. For lower stress amplitudes, the surface damage is very limited, so that the oxide layer at the metal surface could act as a barrier reducing the water vapor effects. On the contrary, it can be assumed that the absence of an oxide layer in the NaCl solution with cathodic protection could favor hydrogen not to be so localized at the martensite interfaces, favoring in this way homogeneous deformation distribution. However, as soon as one crack is created, damage is much more localized in the NaCl solution leading to a strong crack growth acceleration which explains the lack of secondary cracks.

Finally, the greater exposure time at low frequency (0.17 Hz) would allow an enhanced effect of water vapor and hydrogen not only at the crack tip, but also on the initial specimen free surface and especially at martensite lath interfaces. This would explain the earlier crack initiation in air at 0.17 Hz.

5. Conclusions

The resistance to high-cycle fatigue of the high strength steel SE702 was investigated in different environments.

It appears clearly that the fatigue lives in air are much shorter than in vacuum. This gaseous environment leads to a reduction of both crack initiation and propagation stages. This reduction is attributed to water vapor and hydrogen effects with a possible role of oxygen especially on crack initiation stage. These active species have only a slight embrittling effect on fracture surfaces so that the propagation mode remains mainly transgranular, as in vacuum. The enhancement of both initiation and propagation kinetics at low frequency is attributed to a sustained action of water vapor and hydrogen.

The NaCl solution reduces both crack initiation and propagation stages in comparison with vacuum and accelerates crack growth in comparison with air. In this medium, environmental effects are mainly attributed to the hydrogen produced by the cathodic protection. This hydrogen leads to a strong embrittlement of fracture surfaces which are characterized by brittle intergranular and cleavage-like transgranular facets.

The effects of the NaCl solution on the initiation stage, when compared with the air, depend on the loading conditions. At high stress amplitudes, the air can have a more deleterious effect than the NaCl solution on crack initiation, which can be explained by the role of the oxide layer in air.

More generally, our results emphasize the drawback of taking the air as a reference to evaluate the effects of

another active environment and especially to identify the environmental species which participate to damage processes. Moreover, it also appers that fatigue experiments in air should be always performed at the same frequency than the one used in other active environments to allow a suitable comparison of the results.

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