Fatigue crack initiation and growth on a steel in the gigacycle regime with sea water corrosion

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ABSTRACT. *This paper is devoted to the effect of corrosion on the gigacycle fatigue strength of a martensitic-bainitic hot rolled steel used for manufacturing off-shore mooring chains for petroleum platforms. Smooth specimens were tested under fully reversed tension between 10^6 and 10^{10} cycles in three testing conditions and environments: (i) in air, (ii) in air after pre-corrosion, (iii) in air under real time artificial sea water flow. The fatigue strength at greater than 10^8 cycles is reduced by a factor more than 5 compared with non corroded specimens. Fatigue cracks initiate at corrosion pits due to pre-corrosion, if any, or pits resulting from corrosion in real time during the cyclic loading. It is shown that under sea water flow, the fatigue life in the gigacycle regime is mainly governed by the corrosion process. Furthermore, the calculation of the mode I stress intensity factor at hemispherical surface defects (pits) combined with the Paris-Hertzberg-Mc Clintock crack growth rate model shows that fatigue crack initiation regime represents most of the fatigue life.*

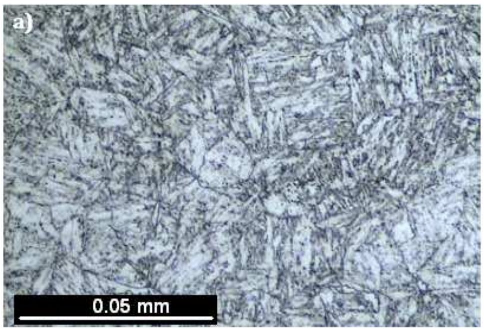
INTRODUCTION

Mooring chains for off-shore petroleum platforms, designed for 30 years, are loaded in fatigue in sea water environment in the gigacycle regime (around 10^9 cycles). The aim of this study is to investigate the gigacycle fatigue strength of a low-alloy steel and the effects on this strength of pre-corrosion and corrosion in sea water environment. Many studies carried out on steel and aluminum alloys in the gigacycle regime have demonstrated that there is not a fatigue limit in such metals after 10^7 cycles as was believed in the past [1, 2]. It has been shown that fatigue cracks initiate mainly at surface defects in the short fatigue life range, but may shift to subsurface in the long life range [3]. Other studies have shown that defects like non-metallic inclusions, pores [4] or pits [5] are the key factors, which control the fatigue properties of metals in very high cycle fatigue (VHCF). Furthermore, in some works it has been proven that crack initiation dominates the total fatigue life of specimens in gigacycle fatigue [6]. In this study the effect of corrosion on the fatigue strength will be quantified and the assessment of the crack propagation period will allow us to investigate the relationship between crack initiation and crack propagation.

MATERIAL AND EXPERIMENTAL CONDITIONS

Material

The investigated material is a non-standard hot rolled low alloy steel grade (named R5) with a typical fine grain microstructure, composed by tempered martensite and bainite, as shown in Figure 1a. Its chemical composition is presented in Table 1. This steel is used after a double quenching in water at 920°C then 880°C and tempering at 650°C with water cooling. After this heat treatment its mechanical properties are as follows: hardness 317 HB, yield strength 970 MPa, UTS=1018 MPa, Young modulus E=211 GPa.



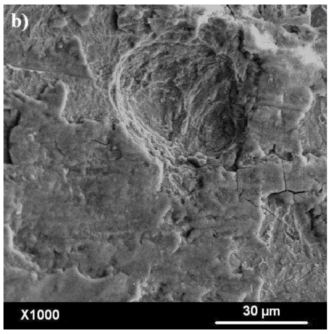


Figure 1: a) R5 steel microstructure; b) typical corrosion pits at the surface of pre-corroded specimen (before any cyclic loading).

Table 1. Chemical composition of the R5 steel (weight %)

C	Mn	Si	P	S	Cr	Ni	Mo	V	Cu	O
0.23	1.22	0.3	0.009	0.003	1.07	1.07	0.5	0.09	0.14	12ppm

Fatigue test conditions and specimen geometry

Testing machine and specimen geometry

All the fatigue tests (crack initiation and crack growth) presented in this paper were carried out with an ultrasonic fatigue testing machine [1] at 20 kHz under fully reversed tension (R= -1) (details can be found in [1]). Since the amplifier and the specimen must work at resonance, the specimen geometry was designed using the elastic wave theory. Figure 2 shows the dimensions of the two types of specimens: (i) for VHCF tests and (ii) fatigue crack growth (FCG) tests. The geometry of these last specimens (FCG) was designed according to the work of Wu and Sun [8, 9], The crack growth was measured with an optical binocular microscope with a maximum magnification x200. The roughness of the tested area of the VHCF specimens was Ra=0.6 µm. The VHCF specimens were tested under three different conditions (i) without any corrosion (virgin state), (ii) after pre-corrosion and (iii) under real time artificial sea water flow. All the VHCF tests were calibrated by using a wide band (100 kHz) strain gauge conditioner and a strain gauge glued on the specimen surface. These tests were carried out until a decrease of the resonance frequency of 0.5 kHz due to the presence of a crack; some times the specimen was broken in two parts. The duration of crack growth period compared to the total life is discussed at the end of this paper.

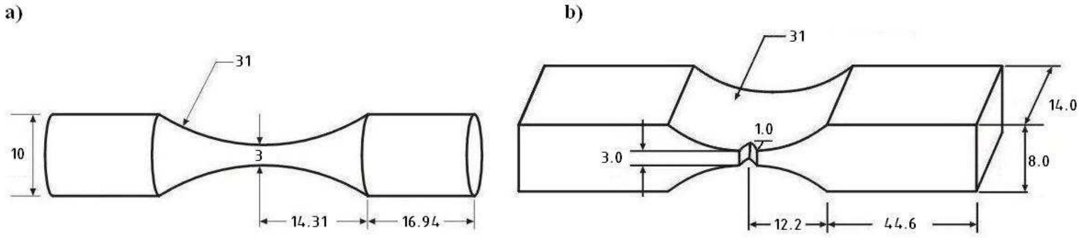


Figure 2: Specimen geometry for (a) VHCF tests, K_t=1.02, (b) crack growth tests (dimensions in mm).

Corrosion of the specimens

The pre-corrosion of the specimens was done according to ASTM G85 standard. The specimens stayed 600 hours in a salt fog corrosion chamber under temperature and humidity control with the following conditions: 35°C with 95% of humidity. The salt solution contains 5% of NaCl, its pH is 6.6 and it is applied in the chamber with a rate flow of 1.52 ml/h. After the pre-corrosion process the specimens were removed from the corrosion chamber, first chemically cleaned and then cleaned with emery paper to remove the oxide layer. Many corrosion pits were created by the salt fog (Figure 1b) their diameter is about 30 to 80 µm.

To carry out VHCF tests in sea water environment a special corrosion cell was designed (Figure 3). To avoid any cavitation it was decided to test the specimens under sea water flow (no immersion[1]). To do that a peristaltic pump creates a flow of sea water (100 ml/mn) on two opposite sides of the specimen surface in the tested area (diameter 3 mm). The sea water used was the A3 standard synthetic sea water; its chemical composition by weight is: 24.53% NaCl, 5.2% MgCl, 4.09% Na2SO4, 1.16% Ca2Cl, 0.695% CaCl and 0.201% NaHCO3. The pH of this solution is 6.6.

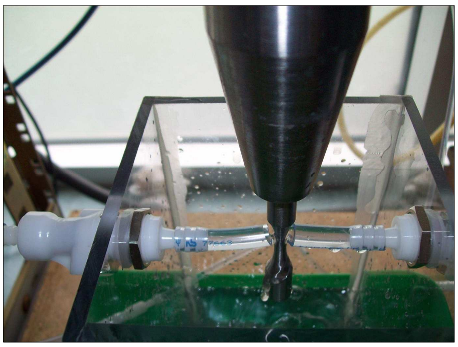


Figure 3: Corrosion cell with peristaltic pump in order to circulate A3 sea water

1 This test condition is representative of the splash zone of mooring chain.

Fatigue crack growth tests

Fatigue crack growth tests, with R=-1, were carried out following a similar methodology as in the ASTM E647 standard. Since with the ultrasonic fatigue testing device the specimen was loaded under displacement control, the range of stress intensity factor ΔK was computed according to the equation proposed in Ref. [8, 9].

RESULTS AND DISCUSION

Fatigue crack initiation tests

Figure 4 shows the SN curves of the crack initiation tests. This figure shows a decreasing (around 50 MPa) of the fatigue strength for the specimens with pre-corrosion compared to the specimens without corrosion. Furthermore, the scatter of the fatigue strength of pre-corroded specimens is larger than for virgin ones. The effect of sea water flow during VHCF fatigue tests is very important. Indeed, for the specimens tested under sea water flow, the fatigue strength at 10^7 cycles is around 300 MPa, not far from the value for pre-corroded specimens (360 MPa), whereas at 3.10^8 cycles the fatigue strength is around 100 MPa only. At this fatigue life the fatigue strength decreasing was -71% compared to pre-corroded specimens and -74% compared to virgin specimens. The fatigue strength decreased significantly in the tests under real time artificial sea water flow due to the detrimental corrosive effect.

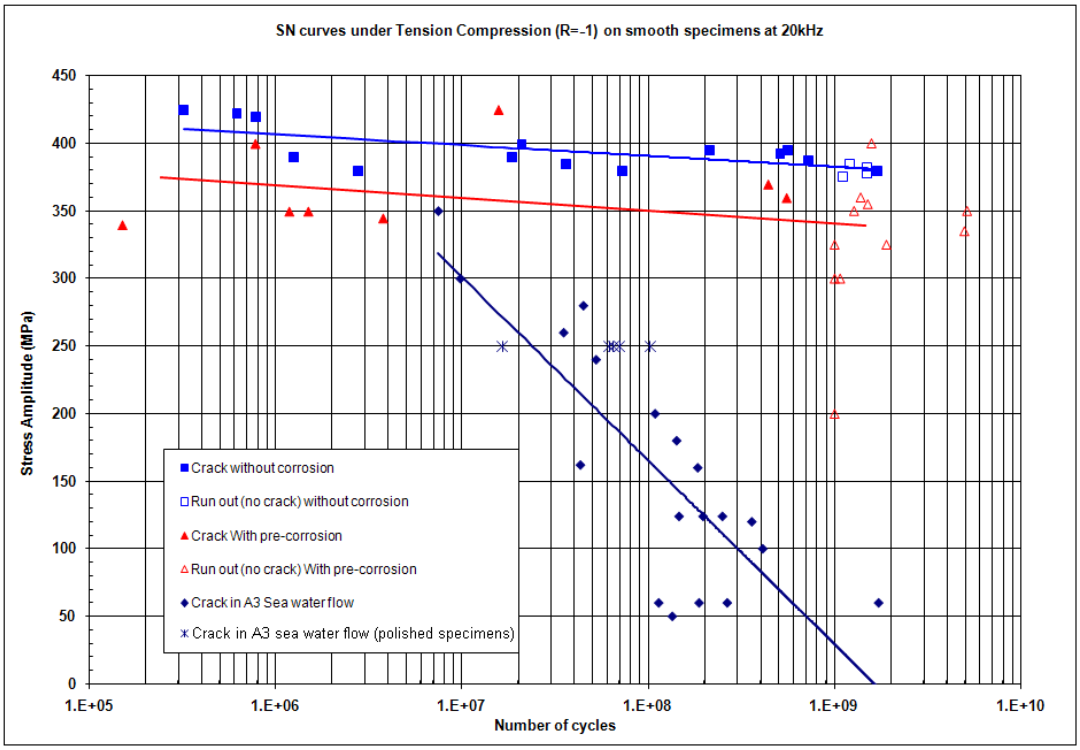


Figure 4: SN curves of the R5 steel under fully reversed tension at 20 kHz

Fractography analysis

Except for some unusual internal crack initiation (Figure 5), fatigue cracks initiated mainly at the specimen surface over the cycle range (10^6–10^9 cycles). Surface defects were the origin of the cracks for non-corroded specimens and corrosion pits for pre-corroded specimens and specimens tested under sea water flow. For the specimens tested under sea water flow the crack initiation areas were all around the specimen surface due to several large corrosion pits. The size of the pits depends on the time (that is to say the number of cycles). At the moment it is difficult to show experimental evidence of the coupling between corrosion and the high strain rate due to the 20 kHz frequency. However the size of corrosion pits is larger under sea water flow (50 to 300 µm) than the pre-corroded specimens (30 to 80 µm). Furthermore, Figure 7 right shows that all the flaws are perpendicular to the loading direction that is to say on the plane of maximum normal stress. This is probably characteristic of a corrosion/cyclic loading interaction. Some complementary fatigue tests were carried out under sea water flow at a stress amplitude of 250 MPa on the same specimens but with a surface polished with emery paper (*Ra*=0.1 µm). There was no evidence of the surface roughness effect under sea water flow (Figure 4). The poor corrosion-fatigue strength was related to the size of pits which are nearly hemispherical surface defects. Due to the major role of the defects, the proportion of the crack propagation period compared to the total life is studied in the following.

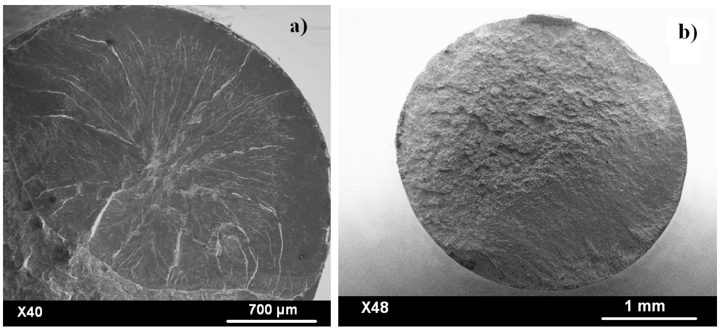


Figure 5. Specimens without corrosion a) Internal crack initiation σ_a=380 MPa, $N_f=2.78\times10^6$ cycles, b) surface crack initiation σ_a=395 MPa, $N_f=5.61\times10^8$ cycles

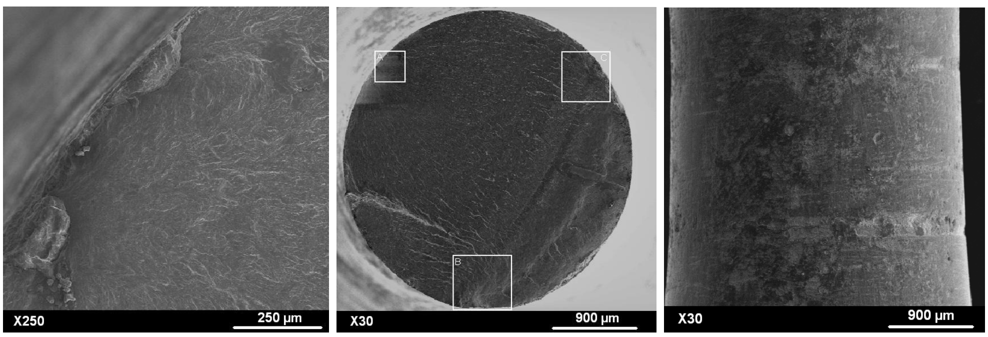


Figure 6. Specimen with pre-corrosion, σ_a=370 MPa, $N_f=4.37\times10^8$ cycles

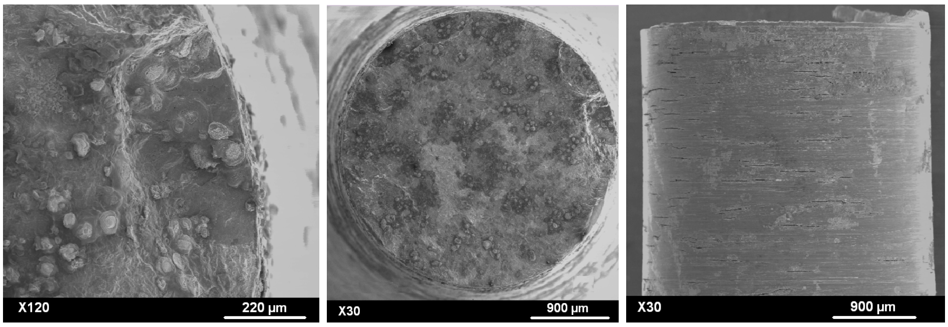


Figure 7: Specimen tested under sea water flow, σ_a=160 MPa, Nf=1. 8×10^8 cycles.

Assessment of the crack initiation and propagation duration

Fatigue crack growth tests gave the *da/dN=f(ΔK)* curve illustrated Figure 8a. This shows that in air the mode I (R=–1) stress intensity threshold for R5 steel is around 3.3 MPa√m. The fatigue crack propagation duration was assessed according to the work of Paris et al. [6, 7], based on the Paris-Hertzberg-Mc Clintock crack growth rate $\frac{da}{dN}=b\left(\frac{\Delta K_{eff}}{E\sqrt{b}}\right)^3$ and $\left(\frac{\Delta K_{eff}}{E\sqrt{b}}\right)=1$ at the corner, where E is the Elastic modulus and b the Burger's vector. Figure 8a shows the agreement of this equation with our experimental data for E=211 GPa and b=0.258 nm. In our experiments at 20 kHz the measurement of K_{op} is not possible, thus in first approximation, ΔK_{eff}~K_{max} and no water interaction with the stress intensity factor was considered.

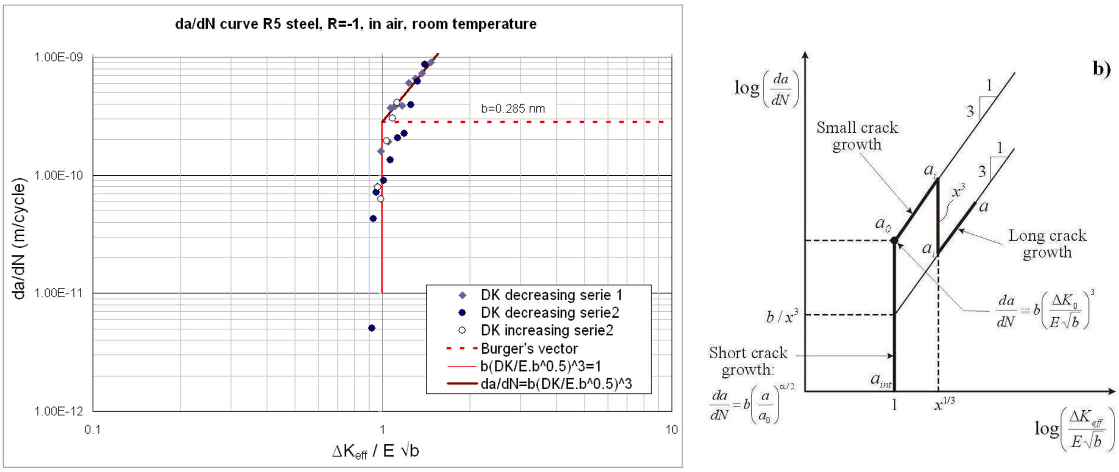


Figure 8. a) Experimental crack growth curve at 20 kHz, b) model of the fatigue growth behaviour for short, small and long cracks according to [7].

To assess the crack propagation phase, a corrosion pit was modelled by an hemispherical surface defect with radius R. A fatigue crack of depth a from the surface of the hemisphere and perpendicular to the loading direction was assumed due to inter-crystalline corrosion cracking. According to the asymptotic approximation proposed by Paris et al. [10] the crack tip stress intensity factor in mode I is: $K_I=\sigma\sqrt{\pi a}\cdot Y(x,\nu)$, where

$x=a/R$, ν is the Poisson ratio,

$$Y(x,\nu)=1.015\left[A(\nu)+B(\nu)\left(\frac{x}{1+x}\right)+C(\nu)\left(\frac{x}{1+x}\right)^2+D(\nu)\left(\frac{x}{1+x}\right)^3\right],\; A(\nu)=1.683+\frac{3.366}{7-5\nu},$$

$$B(\nu)=-1.025-\frac{12.3}{7-5\nu},\; C(\nu)=-1.089+\frac{14.5}{7-5\nu},\; D(\nu)=1.068-\frac{5.568}{7-5\nu}.$$

It was assumed that the propagation of the fatigue crack is divided in three stages: (i) a short crack propagation during $N_{a\mathrm{int}-a0}$ cycles, from initiation a_{int} to the crack size a_0, then (ii) a small crack propagation period of N_{a0-ai} cycles, from a_0 to the crack size a_i, and (iii) a long crack propagation N_{ai-a} cycles, from a_i to the final crack size a. With these three different regimes (illustrated in Figure 8b) the very quick propagation of short crack and the quick propagation of small crack compared to long crack were considered. From [6] the three durations of the three previous stages are :

$$N_{a\mathrm{int}-a0}=\frac{E^2}{Y_0^2\sigma_a^2\pi(\alpha/2-1)}\left[\left(\frac{a_0}{a_{\mathrm{int}}}\right)^{(\alpha/2-1)}-1\right],\qquad N_{a0-ai}=\frac{2E^2}{Y_0^2\sigma_a^2\pi}\left[1-\sqrt{\frac{a_0}{a_i}}\right]\quad \text{and}$$

$$N_{ai-a}=\frac{2E^2Y_0}{Y(a_i/R)^3\sigma_a^2\pi}\left[x^3\sqrt{\frac{a_0}{a_i}}-x^3\sqrt{\frac{a_0}{a}}\right]$$ with $Y_0=Y(a_0/R)$ and $N_{Total}=N_{a\mathrm{int}-a0}+N_{a0-ai}+N_{ai-a}$

The previous equations were applied to our data assuming that the initial crack has a length, a_0, corresponding to the corner: $\frac{da}{dN}=b;\ \frac{\Delta K_{eff}}{E\sqrt{b}}=1$. Two cases were considered in Table 2, a pit of R=48 µm on a pre-corroded specimen with $N_{experimental}=5.5\times10^8$ cycles and a pit with R=300 µm for a specimen under sea water flow with a life of 1.83×10^8 cycles. In the first case 99.1% of the fatigue life is due to initiation. For the second case, 94% of the fatigue life was consumed by the initiation phase. One can notice that with a high α and large a_{int}/a_0 one obtains similar N_{Total} results than with a low α and small a_{int}/a_0, that is because with a higher α the crack does not grow as far due to the slope of the da/dN curve in the threshold region, then a_{int}/a_0 must be larger, and with a lower α, a_{int}/a_0 must be smaller. Furthermore one can notice that an equally appropriate approximation of the crack growth period according to [6]: $N_p=\pi E^2/[2(\Delta\sigma)^2]$ gives results in the same order compared to N_{Total} (Table 2 right).

Table 2.a. Hemispherical surface crack growth vs. experimental fatigue life for a first case, for a_0=9.32 µm, x=3, α=25, 100 and 200, and different a_{int}/a_0 ratios.

Case 1: σ_a=360 MPa, R=48 µm, a= 2.4 mm, $N_{experimental}=5.5\times10^8$ cycles						
α	a_{int}/a_0	$N_{a\mathrm{int}-a0}$	N_{a0-ai}	N_{ai-a}	N_{Total} (cycles)	N_p (cycles)
25	0.9	6,704	44,151	4,507,080	4,557,935	539,600
	0.94	2,947			4,554,178	
	0.97	1,192			4,552,423	
100	0.9	115,801	44,151	4,507,080	4,667,032	
	0.94	13,163			4,564,394	
	0.97	2,300			4,553,531	
200	0.9	11,184,860	44,151	4,507,080	15,736,091	
	0.94	150,676			4,701,907	
	0.97	6404			4,557,635	

Table 2.b. Hemispherical surface crack growth vs. experimental fatigue life for a second case, for a_0=41.5 µm, x=3, α=25, 100 and 200, and different a_{int}/a_0 ratios.

<table>
<tr><td colspan="7">Case 2: σ_a=160 MPa, R=300 µm, a= 2.6 mm, $N_{experimental}$=1.83 10^8 cycles</td></tr>
<tr><td>α</td><td>a_{int}/a_0</td><td>$N_{aint-a0}$</td><td>N_{a0-ai}</td><td>N_{ai-a}</td><td>N_{Total} (cycles)</td><td>N_p (cycles)</td></tr>
<tr><td rowspan="3">25</td><td>0.9</td><td>29,840</td><td rowspan="3">168,014</td><td rowspan="3">10,258,700</td><td>10,456,554</td><td rowspan="9">2,731,700</td></tr>
<tr><td>0.94</td><td>13,120</td><td>10,439,834</td></tr>
<tr><td>0.97</td><td>5,306</td><td>10,432,020</td></tr>
<tr><td rowspan="3">100</td><td>0.9</td><td>515,466</td><td rowspan="3">168,014</td><td rowspan="3">10,258,700</td><td>10,942,180</td></tr>
<tr><td>0.94</td><td>58,593</td><td>10,485,307</td></tr>
<tr><td>0.97</td><td>10,237</td><td>10,436,951</td></tr>
<tr><td rowspan="3">200</td><td>0.9</td><td>49,787,104</td><td rowspan="3">168,014</td><td rowspan="3">10,258,700</td><td>60,213,818</td></tr>
<tr><td>0.94</td><td>670,703</td><td>11,097,417</td></tr>
<tr><td>0.97</td><td>28,504</td><td>10,455,218</td></tr>
</table>

CONCLUSION AND PROSPECTS

Very high cycle fatigue tests were carried out up to 10^9 cycles on smooth specimens in hot rolled martensitic-bainitic steel under three different conditions (i) virgin specimens, (ii) pre-corroded specimens and (iii) under artificial sea water flow during the fatigue test. The fatigue strength at 10^8 cycles is significantly reduced by a factor of 74% compared to the virgin specimens and of 71% compared to the pre-corroded ones. The assessment of the crack growth shows that crack initiation dominates the total fatigue life when N>10^7 cycles. The proposed assessment is reasonable for pre-corroded specimens, but other studies must be done to study the effect of corrosion pits under real time sea water flow. The effect of sea water on the crack growth and the stress intensity factor also needs to be studied. A possible coupling between environment and high frequency cyclic loading should be studied too. However, ultrasonic fatigue test immersed in flowing sea water is the only experimental way to investigate very long life of steel under corrosion conditions.

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REFERENCES

1. Bathias C., Paris P.C. (2005). Gigacycle Fatigue in Mechanical Practice, Marcel Dekker Publisher Co., New York USA.
2. Marines I., X. Bin, C. Bathias (2003). In: *Int J Fatigue* **25**, 1101-1107.
3. Zuo J.H., Wang Z., Han H. (2008). In: *Mat Sci Eng* A **473**, 147-152.
4. Ammar H.R., Samuel A.M., Samuel F.H. (2008). In: *Mat Sci Eng* A **473**, 65-75.
5. Pang H.T., Reed P.A.S. (2007). In: *Mater Sci Eng A* **448**, 67-69.
6. Paris P.C. (2004). In : *Proceedings of VHCF-3*, 1-13.
7. Marines I., Paris P.C., Tada H., Bathias C., (2007). In: *Int J Fat* **29**, 2072-2078.
8. Wu T.Y., Bathias C. (1994). In: *Eng Fract Mech* **47**, 683-690.
9. Sun (2000). Etude du seuil de fissuration à haute fréquence en fatigue et en fretting fatigue. Ph.D. Thesis, CNAM, Paris France.
10. Paris P.C., Palin-Luc T. Tada H. and Saintier N. (2009). Int. Crack Path, 8 pages.