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Corrosion Fatigue Crack Growth Modelling for Subsea Pipeline Steels

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Abstract

Based on the recent development of research for environment-assisted cracking (EAC), a crack growth model for the corrosion fatigue (CF) of subsea pipeline steels is proposed in this paper. The model adopts a two-component structure in consistence with the physical fact that CF is a mixed process of stress corrosion (SC) and hydrogen-assisted cracking (HAC). Anodic dissolution (AD) and hydrogen embrittlement (HE) are believed to be responsible for SC and HAC and their models are integrated in the general model within a framework of fracture mechanics. The model is then applied to modelling the CF crack growth of X65 pipeline steel and the results show that the shape of the modelled crack propagation curve can be controlled well using the proposed model, and both the exacerbated cracking rate and behavior features can be captured with appropriate consideration of the effects of mechanical and environmental parameters, such as loading frequency, stress ratio, temperature, and hydrogen concentration, etc.

Keywords: Corrosion fatigue, fracture mechanics, anodic dissolution, hydrogen embrittlement, subsea pipeline

1. Introduction

Subsea pipelines are safety critical structures primarily used to carry oil or gas. They occupy the most extension through space in offshore oil and gas production systems, and often encounter harsh working conditions. As a strategy to guarantee safe operation (DNV-OS-F101, 2013), fracture mechanics based engineering critical assessment (ECA) or fitness-for-service (FFS) has been widely applied in offshore engineering in recent years. However, subsea pipelines are vulnerable to severe environmental assisted crackings (EACs) which lack attention in current mainstream ECA guidelines such as BS 7910 (2013) and API 579 (2009). To acquire high-quality ECAs, it is necessary to develop proper crack growth models for subsea pipelines to account for EAC effects. Depending on the loading profile, there are two major categories of EAC: Corrosion fatigue (CF) and stress corrosion cracking (SCC). CF is the environment enhanced cracking under fatigue loads and SCC represents the cracking induced by the combined influence of a corrosive environment and a static load. Usually both CF and SCC are thought to be able to cause substantial life reduction of subsea pipelines. However, it is doubtful if SCC can be realized in real service conditions since engineering structures are normally exposed to complex operations with varying working stresses that are usually mixture of static and cyclic components. Some researchers even argued that SCC is a special case of CF with the stress ratio being unity (Shipilov, 2002). Although theoretical and experimental studies on SCC are relatively ample, CF is receiving more and more attention in recent years. A number of models have been developed for CF of metals in aqueous environments (Kim et al., 1998). However, most of them are just phenomenological because the material's CF cracking behavior is quite complicated and considerable variables can impose their impacts, while the involved mechanisms are not very clear.

Based on the recent research on the mechanisms and modelling of EACs (Rhodin, 1959; McEvily and Wei, 1972; Gerberich et al., 1988; Parkins, 1992; Rhodes, 2001; Gangloff, 2003; Wang et al., 2013), a two-component CF crack growth model for subsea pipeline steels is proposed in this paper, in which the anodic dissolution (AD) model and the hydrogen embrittlement (HE) model newly established by authors (Cheng and Chen, 2017) are integrated within a framework of fracture mechanics. The model is then applied to modelling the CF crack growth of X65 pipeline steel and the impact of loading frequency, stress ratio, temperature, and hydrogen concentration on the model performance is discussed.

2. Corrosion fatigue mechanisms

Subsea pipelines are exposed to potentially aggressive service environments both the inside and outside. The chemical composition of product fluid may vary from field to field, while the outside environment roughly stays the same. Either tenting of tape or coating at the long-seam welds or under tape wrinkles can create space for seawater contacting and further corroding the bare metal. Also, pre-existing external flaws such as as-built pipeline defects including grooves and weld defects, or dents caused by third-party interference may act as initiating sites for EACs. The research interest of this paper exists in the corrosive effect of seawater on cracks in the external steel surface of subsea pipelines. The influence from environment to the CF crack growth is multi-aspect and complicated. For example, the deep-sea environment may impose huge external hydrostatic pressure and cold temperature. Meanwhile subsea pipelines may also suffer loads from intervention vessels, current and other met-ocean events. Additional attentions should be paid to the possible thermal and pressure expansion and contraction. The related loading frequencies, stress ratios and even temperature may vary in rather big ranges. Besides, as cathodic protection (CP) technique is widely adopted nowadays, intensive hydrogen pick-up may happen due to over protection, which often results in a high hydrogen concentration in the bulk material. All the factors above mentioned may play a role in the CF process of subsea pipelines. Due to the complicated nature of CF, current simplified CF crack growth models often provides predictions that are either over-conservative or under-estimated. For example, as shown in Fig. 1 (a), the over conservatism in predictions by linear models from API 579 (2009) and BS 7910 (2013) is obvious. While in Fig. 1 (b), both bilinear models provided by API 579 (2009) and BS 7910 (2013) fail to predict the CF crack growth rate of X65 pipeline steels in seawater conservatively. The CF data under a loading frequency of 0.01 HZ and a stress ratio of 0.2 go beyond the prediction at relatively high SIF ranges. More attentions should be paid to such problems since the lost CF crack growth rates are fairly high, thus greatly increasing the risk of sudden failure of the component.

No metal is immune from some reduction of its resistance to cyclic loading if the metal is put in a corrosive environment. But different environment-material systems may exhibit distinct CF cracking behaviours. According to McEvily and Wei (1972), those behaviours belong to three types, type A, type B, and the mixed type. Each type is schematically plotted in Fig. 2. Type A describes the behavior where the threshold K_{th} is reduced and crack growth rate is enhanced by the presence of the corrosive environment at all levels of K . Type B represents the behavior typified by the enhanced crack growth beyond the K_{ISCC} and is characterized with a plateau in crack growth rate. The mixed type, where type B behavior happens above K_{ISCC} with type C behavior superimposed on at all K levels below, is

exhibited by a broad range of material-environment systems, and is typical of pipeline steels in seawater.

BS 7910 (2013) defines CF as a type of damage similar to fatigue, except the environment is corrosive instead of inert or dry air. Fatigue failure process of a component or specimen usually begins with the initiation of cracks (stage I), and with continued cyclic loading the cracks grow (stage II), sequentially comes the rapid crack growth leading to the final rupture (stage III). A corrosive environment causes degradation of the material, the most common visible effect being pitting or surface roughness (etching). These notch-like regions act as stress raisers and are generally the sites of crack nucleation (Ellyin, 2012). The corrosive environment thus shortens the crack nucleation stage. Once cracks are initiated, subsequent crack growth may be enhanced by the corrosiveness. Quite a number of fundamental questions regarding the possibility, severity and rate of CF cracking remain unanswered. Foremost among these questions is the problem of the CF mechanism. In contrast to CF, relatively extensive and fruitful research has been performed on the mechanism of SCC (Parkins, 2000; Beavers and Harle, 2001; Woodtli and Kieselbach, 2000; Fang et al., 2003). SCC is the crack growth in a corrosive environment under a sustained load. Crack growth of SCC is a result of the combined and synergistic interaction of corrosion reactions and mechanical stress. In the most general situation, cracks initiate from the bottoms of pits and crevices or other surface blemishes, and propagate into the material either transgranularly, intergranularly or sometimes in a mixed type. For high-pH SCC, the crack often grows along an intergranular path. This is thought to be associated with the strong environmental influence it receives. Local corrosion has been well accepted as its driving mechanism. But for near-neutral pH SCC, it has been suggested that the mechanism is associated with the hydrogen ingress as a byproduct of corrosion, as well as the dissolution at the crack tip. That is to say near-neutral pH SCC is mixture of two damage modes, namely stress-assisted corrosion or stress corrosion (SC) and hydrogen-assisted cracking (HAC). Interestingly, CF usually shows the same transgranular fracture surface as near-neutral pH SCC. The similar morphologies indicate that CF and SCC may have similar cracking mechanisms. Some experimental investigations on the near-neutral pH SCC even used a cyclic loading to initiate cracks from pit bottoms (Parkins et al., 2000). Further studies confirm that the mechanisms, which have generally been proposed to explain SCC, are also responsible for CF (Shipilov, 2002). Thus the main mechanisms for a typical CF are herein considered to be HAC and SC of the metal at the crack tip.

3. Model development

3.1 Fracture mechanics

To be consistent with ECA, the model to be built should be in a framework of fracture mechanics. The principles of linear elastic fracture mechanics (LEFM) were developed in the 1950s by Irwin (1957) to describe the crack growth under sustained loads and introduce the concept of stress intensity factor (SIF) K as

$$K = FS\sqrt{\pi a} \quad (1)$$

where F is the geometry function, S is the stress, and a is the crack depth.

The loading pattern of subsea pipelines is a combination of hydrostatic and cyclic loads. In some extreme cases, the ratio between the hydrostatic load contribution and the cyclic load contribution in subsea pipelines may be greater than unity. This can be accounted for by the mean stress effects in fracture mechanics analysis (Bannantine et al., 1990).

The assumed fatigue crack growth rate at Stage II usually follows of Paris' law (Paris and Erdogan, 1963)

$$\frac{da}{dN} = C(\Delta K)^m \quad (2)$$

where m and C are constants for a specific material; ΔK is the SIF range calculated as

$$\Delta K = (1 - R)K_{\max} \quad (3)$$

with R and $K_{\max}$ being the stress ratio (i.e. $S_{\min}/S_{\max}$) and maximum SIF encountered in a stress cycle, respectively.

In a log-log plot of da/dN vs ΔK , Paris law represents a straight line of a slope m . Behavior of short cracks, i.e. cracks in stage I, departs from a simple Paris' law representation. Below a certain value of ΔK (known as the threshold value, ΔK_{th}) no macrocrack growth is expected. Whereas when $K_{\max}$ approaches K_{IC} the crack growth rate increases rapidly. The fatigue life of a component in an inert environment can be roughly divided into two parts: crack initiation and crack propagation. Fracture mechanics being applied to CF enables the CF life fractions to be defined within the criteria of crack initiation and propagation. For ECAs applied at design stage, prediction of life before crack initiation is decisive, since the initiation process can cost much longer time. However, for ECAs applied in-service for to determine inspection intervals, since cracks already exist, prediction of life after crack initiation becomes decisive. In a typical CF situation, cracks initiate from corrosion sites and grows under the influence from environment, thus the initiation stage is often bypassed. Moreover in practice, cracks

detected on subsea pipelines are mostly at the propagation stage. Therefore this paper focuses on the crack growth in the stage of crack propagation and LEFM is applicable to establish a typical CF model for subsea pipelines at service.

3.2 General model

Several attempts have been made to model environmentally-assisted crack growth under cyclic loading by superposing different processes. The classic model proposed by Landes and Wei (1969), as shown in Eq.(4), simply treats the rate of CF crack growth $(da/dN)_{CF}$, as the algebraic sum of the rate of crack growth in an inert environment $(da/dN)_F$, and that of SCC $(da/dN)_{SCC}$ in the identical aggressive environment.

$$\left(\frac{da}{dN}\right)_{CF} = \left(\frac{da}{dN}\right)_F + \left(\frac{da}{dN}\right)_{SCC} \quad (4)$$

The process competition model of Austen and Walker (1977) is based on the assessment that fatigue and SCC are two mutually competitive processes and that the crack will propagate at a faster rate pertinent to the prevailing SIF. However, to ensure compatibility in terms of crack growth rate per cycle rather than per second, the plateau crack growth rate must be adjusted to the appropriate frequency or to the frequency, stress intensity and stress ratio. And the influences of high frequencies on the crack growth rate plateau are predicted better than those of low frequencies. Other models (Kim et al., 1998) modified these assumptions mostly by adding parameters.

As previously stated, CF crack growth is a joint action of SC and HE. This conclusion is drawn based on the investigation of the mechanisms for SCC and CF. A two-component physical model is proposed herein as

$$\left(\frac{da}{dN}\right)_{CF} = \left(\frac{da}{dN}\right)_{SC} + \left(\frac{da}{dN}\right)_{HAC} \quad (5)$$

where $(da/dN)_{HAC}$ is the crack growth rate by HAC. The proposed model is a linear superposition of two rates. Basically it means that the rate of CF crack growth $(da/dN)_{CF}$ is an aggregate of the rates of SC $(da/dN)_{SC}$ and HAC $(da/dN)_{HAC}$. Anderson (2005) mentioned a similar superposition model, however, it superposed cycle-dependent CF with time-dependent CF, which are defined in a phenomenological way, leading to difficulty in implementing the model. This proposed two-component model proposed is based upon physics and is compatible with fracture mechanics.

3.3 Component model

3.3.1 Anodic dissolution (AD)

Several models have been developed for explaining the crack growth in SCC due to anodic dissolution. A well-accepted one (Parkins, 1979) is the film rupture model, which states while the metal surface is covered by a passive or protective layer, the crack tip is kept free from protective layers continuously or discontinuously by local chemical and mechanical effects, and this allows the accelerated anodic dissolution of local material. To a large extent, AD models for CF are an extension of those proposed for SCC. In such a view, the AD process of CF is described as cyclic plastic strain ruptures a protective film at the crack tip, resulting in transient anodic dissolution, followed by possible repassivation. The amount of environmental crack growth per fatigue cycle depends on the kinetics of the reaction on the clean metal surface as well as the time between film ruptures. All factors imposing influence on CF process seem to work through the film rupture mechanism. Based on the physical facts, a formula is proposed for the crack growth rate from SC over the whole SIF range

$$\left(\frac{da}{dN}\right)_{SC} = \left(\frac{da}{dN}\right)_{AD} = h(\Delta K, R, f) = \begin{cases} h_1(\Delta K, R, f), & K_{max} < K_t \\ h_2(\Delta K, R, f), & K_{max} \geq K_t \end{cases} \quad (6)$$

where $(da/dN)_{AD}$ is the crack growth rate due to anodic dissolution, f is the stress cycle frequency, and K_t is the transition SIF, i.e. where the crack growth rate plateau starts in a abscissa of K_{max} . Endo et al. (1981) solution conducted a series of experiments on carbon steels in NaCl to investigate the SC crack growth rate of steels due to anodic dissolution under fatigue loading conditions. They claimed that the observed CF crack growth followed the sequential model,

$$\left(\frac{da}{dN}\right)_{CF} = \left(\frac{da}{dN}\right)_F + \left(\frac{da}{dN}\right)_{AD} \quad (7)$$

where

$$\left(\frac{da}{dN}\right)_{AD} = Af^b(1-R)^c\Delta K^m \quad (8)$$

A , b and c are the material-environment system coefficients. According to the experimental data, $A = 4.93 \times 10^{-12}$, $b = -0.36$, $c = 2$ and m is the same as that in the expression of crack growth in air, i.e. Eq. (2). And b is consistent with the experimental observation by Bartlett and Hudak (1990) that the higher the frequency, the lower the AD rate, which means b is usually negative. It is suspected that the frequency and stress ratio dependence is determined by the solution conditions. Thus for a medium such as seawater, b may considered the same as that of the NaCl solution. So

$$h_1(\Delta K, R, f) = Af^{-0.36}(1-R)^2\Delta K^m \quad (9)$$

Effects of coefficient A on the shape of the crack propagation curve are shown in Fig. 3.

Aware that CP works through inhibiting the corrosion process. More specifically, CP prevents the occurrence of anodic dissolution at the crack tip. The CF crack growth data of X65 pipeline steel specimen in NaCl water with and without CP are plotted in Fig. 4, as well as those of X70 (Vosikovsky, 1975; Vosikovsky, 1981). As can be seen, the change of crack growth rates after imposing a CP is very small when ΔK goes beyond the plateau, thus roughly

$$h_2(\Delta K, R, f) \approx Af^{-0.36}(1 - R)^2 K_t^m \quad (10)$$

This is consistent with the aforementioned AD mechanism. According to the AD mechanism, as ΔK is growing, there is a critical value of ΔK where repassivation disappears. Below this value, AD occurs discontinuously; above this value, AD effect will be constant. Compared with the increasing HAC rate, this constant rate is very small and its ratio to the HAC rate is still decreasing, evidenced as the pair of crack propagation curves with/without CP approaching each other after the transition point, as shown by Fig. 4.

3.3.2 Hydrogen embrittlement (HE)

As for HAC, crack growth is associated with hydrogen impurity in the material. The hydrogen may be a by-product of corrosion reactions in aqueous solutions or cathodic protection. The absorbed hydrogen atoms diffuse to the region near the crack tip with stresses applied. The presence of hydrogen in metal is known to have an adverse effect on the material's properties, more specifically the concentration of hydrogen ahead of a crack tip can cause a significant reduction in the material's resistance to fracture (Humphries et al., 1989). In such a situation, cracking happens at a lower stress level compared to that of the same material without hydrogen absorption and crack propagates in an enhanced rate. This phenomenon is called hydrogen embrittlement (HE). Although there is still controversy as to the extent to which HAC explains subcritical crack growth in metals stressed in environments that support concurrent crack tip dissolution, passive film formation, and atomic hydrogen production. An agreement has been reached that HE normally prevail for subsea metal structures with cathodic protection (CP) as well as those exposed to gaseous hydrogen (Barnoush, 2011). Studies on the mechanism of HE started several decades ago and are still ongoing. Among the numerous mechanisms that have been raised so far, three have stood critical examinations and been widely accepted, i.e. Hydrogen Enhanced De-cohesion (HEDE), Hydrogen Enhanced Localized Plasticity (HELP), and Adsorption Induced Dislocation Emission (AIDE). Arguments supporting each are not definitive, even not exclusive. For those who are interested, ample information can be found in a recent critical review by Lynch (2012). Not surprisingly, HEDE has been frequently used as the theoretical basis for modelling HE because of its explicit compatibility with fracture mechanics.

HEDE provides the basic notion that hydrogen damage occurs in the crack-tip plastic zone when the local crack tip opening tensile stress exceeds the maximum-local atomic cohesion strength, which has been lowered by the presence of hydrogen (Oriani, 1972). Hence locations of hydrogen damage initial sites are at a distance ahead of the crack tip, i.e. the locations where tensile stresses are maximized. Thus mechanical characteristics of the region near the crack tip should be analyzed, since it can control the precise damage location. A HEDE-based model should involve factors such as the tensile stress distribution in front of a crack tip, the plastic strain and the associated dislocation density profile about the crack tip, the distribution of hydrogen trap sites, and the concentration of environmentally produced hydrogen in the crack-tip area. To establish such a model, the formulated HEDE mechanism by Lee and Unger (1988) was adopted as the basic formula,

$$\sigma_{\max H} = \sigma_{\max 0} - \beta C_H \quad (11)$$

where $\sigma_{\max H}$ and $\sigma_{\max 0}$ are the maximum principal stress at elastic-plastic boundary with and without hydrogen when cracking initiates, respectively. β is a parameter related to loss of critical cohesive stress by hydrogen impurity, and C_H is the local hydrogen concentration. It represents the assumption that fracture occurs when the maximum crack-tip-opening stress exceeds the hydrogen-reduced cohesive strength within an area in front of the crack tip.

C_H can be calculated by Fick's law with respect to the hydrostatic stress (Van Leeuwen, 1974)

$$C_H = C_{H0} \exp\left(\frac{\sigma_h V_H}{k_B T}\right) \quad (12)$$

where C_{H0} is the local hydrogen concentration in unstressed state, σ_h is the hydrostatic stress within plastic zone, V_H is the partial volume of hydrogen, k_B is the Boltzmann constant, and T is the temperature.

With the help of LEFM, expressions of the hydrostatic stress σ_h and the maximum crack-tip-opening stress in front of crack tip $\sigma_{\max}$ under mode I with plane strain condition are (Broek, 1974),

$$\sigma_h = \frac{2}{3}(1 + \nu) \frac{K}{\sqrt{2\pi r}} \quad (13)$$

$$\sigma_{\max} = \sigma_{yy} = \frac{K}{\sqrt{2\pi r}} \quad (14)$$

where r is the radial distance from crack tip, ν is the Poisson's ratio, and σ_{yy} is the stress along the Y direction.

Combining Eqs. (10) - (13), with the assumption that the location of the crack initiation does not change with hydrogen concentration, Wang et al. (2013) obtained the following formula for predicting the hydrogen-degraded fracture toughness,

$$K_{IH} = K_{IC} \left\{ 1 - \left(\frac{\beta C_{H0}}{\omega \sigma_{ys}} \right) \exp \left[\frac{V_H \sigma_{ys}}{k_B T} \cdot \frac{2(1 + \nu)}{3} \cdot \frac{K_{IH}}{K_{IC}} \right] \right\} \quad (15)$$

where ω is a magnification factor of value $3 \sim 5$ accounting for the material's working hardening effect, and K_{IH} is the saturated fracture toughness, i.e. the fracture toughness fully degraded in the environment with a specific hydrogen concentration. This model has been applied to a series of steels to describe the dependence of fracture toughness on hydrogen concentration, and the predicted results match well with the experimental data.

Instead of Paris' law, Forman equation is adopted for its integration of mean stress influence via the stress ratio R and the static failure via K_{IC} in region II and region III,

$$\frac{da}{dN} = \frac{B(\Delta K)^m}{[(1 - R)K_{IC} - \Delta K]} \quad (16)$$

The authors proposed a two-stage Forman equation model for the HE influenced fatigue crack growth in hydrogen gas (Cheng and Chen, 2017). It is based on the corrosion-crack correlation and HEDE theory, which is consistent with Eq. (14). The key assumption in this model is the stress-driven hydrogen diffusion, namely the absorption of the hydrogen ions into the crack tip is through transport up the hydrostatic stress gradients. This process is schematically plotted in Fig. 5. Such an assumption enables the model to calculate several crucial points to capture features of the hydrogen influenced fatigue crack growth. Note that this assumption is not applicable to the fatigue cracking of subsea pipeline steels under CP. CP provides a potential intensifying the hydrogen absorption, which is mainly manifested as the decrease or left shift of K_t in the crack propagation curve, as seen in Fig. 4.

The transition SIF K_t , i.e. the point where the environment affected zone (EAZ) is passed by the plastic zone in front of crack tip and the plateau section of the crack propagation curve starts, can be calculated as

$$K_t = \min \left\{ \max \left\{ K_{tran}, \frac{\Delta K_0}{(1 - R)} \right\}, K_{IH} \right\} \quad (17)$$

where ΔK_0 represents the threshold SIF range at $R = 0$ (usually obtained from experiments), and

$$K_{tran} = \omega \sigma_{ys} \frac{K_{IH}}{K_{IC}} \left(\frac{4\pi^2 D V_H \sigma_{ys}}{k_B T f} \right)^{\frac{1}{4}} \quad (18)$$

The length of SIF range over which the crack growth rate plateau lasts is approximated by

$$K_p = \frac{\Delta K_{th}}{(1 - R)} \quad (19)$$

where ΔK_{th} is obtained by following equation

$$\Delta K_{th} = \Delta K_{th} - 1 \quad (20)$$

The final fracture toughness displayed under a fatigue load, or alternatively the equilibrium fracture toughness K_{IE} , is proposed to be

$$K_{IE} = \begin{cases} \left[1 - \left(\frac{2\pi(\omega\sigma_{ys})^2}{K_{IC}^2} \sqrt{\frac{DV_H\sigma_{ys}}{k_B T f}} \right)^\lambda \right] K_{IN}, & f > f_c \\ K_{IH}, & f \leq f_c \end{cases} \quad (21)$$

where λ is a non-negative parameter used to adjust the rate with which K_{IE} approaches K_{IC} , i.e. the inherent fracture toughness. In consistence with previous work (Cheng and Chen, 2017), λ is considered to be 1.0.

Then K_{IC} in Eq. (16) is substituted by K_{IH} and K_{IE} respectively for the first and second stage Forman equations. Good agreement can be seen from the application of the two-stage Forman equation model to a wide range of carbon pipeline steels (Cheng and Chen, 2017). Thus, the HAC rate of the two-component CF model, is established as

$$\left(\frac{da}{dN} \right)_{HAC} = \left(\frac{da}{dN} \right)_{HE} = g(\Delta K, R, f, K_{IC}, C_{H0}) = \begin{cases} g_1(\Delta K, R, f, K_{IH}, C_{H0}), & K_{max} < K_t \\ g_2(\Delta K, R, f, K_{IE}, C_{H0}), & K_{max} \geq K_t \end{cases} \quad (22)$$

where

$$g_1(\Delta K, R, f, K_{IH}, C_{H0}) = \frac{B_1(\Delta K)^{m_1}}{[(1 - R)K_{IH} - \Delta K]} \quad (23)$$

$$g_2(\Delta K, R, f, K_{IE}, C_{H0}) = \frac{B_2(\Delta K)^{m_2}}{[(1 - R)K_{IE} - \Delta K]} \quad (24)$$

Note that B_1 and m_1 are acquired from the crack growth data measured before ΔK reaches $(1 - R)K_t$, while B_2 and m_2 are acquired from the crack growth data measured after the point $(1 - R)(K_t + K_p)$.

4. Application

In terms of Eqs.(5), (6), and (22), the two-component model can be reformulated as

$$\left(\frac{da}{dN}\right)_{CF} = \left(\frac{da}{dN}\right)_{AD} + \left(\frac{da}{dN}\right)_{HE} = \begin{cases} h_1(\Delta K, R, f) + g_1(\Delta K, R, f, K_{IH}, C_{H0}), & K_{\max} < K_t \\ h_2(\Delta K, R, f) + g_2(\Delta K, R, f, K_{IE}, C_{H0}), & K_{\max} \geq K_t \end{cases} \quad (25)$$

In this section, the proposed model will be applied to X65 pipeline steels to display its general performance. X65, the medium strength pipeline steel, is commonly adopted in offshore oil and gas production systems and is currently one of the highest API grade C-Mn steels for offshore service for its good ductility and toughness. In this application, the influence of wedge effect caused by the accumulation of corrosion product within the crack tip cavity is ruled out by using ΔK_{eff} as the default ΔK .

The two-component model allows the superposition of AD and HE cracking rates. Due to a lack of exact experimental data for determining the AD rate of CF in pipeline steels, the related coefficients will be given approximated values for the demonstrative purpose. On the other hand, CF in seawater shares the same HE nature with the HAC of pipeline steels in hydrogen gas, while the experimental data of HE influenced fatigue cracking in hydrogen gas are plentiful. Ronevich et al. (2016) performed a series of tests on X65 pipeline steels and obtained the HE influenced fatigue crack growth data. Experimental data of mechanical properties for X65 pipeline steels both in dry air and hydrogen gas were collected by Somerday and Marchi (2007). For the gaseous hydrogen environment, C_{H0} is calculated as the product of solubility and the square root of pressure. The values of coefficients for determining the HE rate in the proposed model for X65 pipeline steel are adopted in accordance with those in authors' previous article (Cheng and Chen, 2017).

As for β , as stated in the HEDE notion, β is a parameter related to the loss of critical cohesive stress by hydrogen impurity, which is a material property. Gutierrez-Solana and Elices (1982), and Cialone and Holbrook (1985), conducted a series of tests on the fracture toughness degradation of X42 and X70 pipeline steels in hydrogen gas of different pressures, respectively. Based on the fact that pipeline carbon steels usually share similar chemical composition and exhibit a common ferritic-pearlitic microstructure (Holtam, 2010). It is expected that the sensitivities of C-Mn steels of different strengths to HE should follow some relationship. In this paper, linear interpolation between the β values of X42 and X70 is used with respect to the yield strength, and the β value of X65 pipeline steels is thus determined as $1.56 \times 10^4 \text{ MPa m}^3(\text{mol H}_2)^{-1}$.

The model implemented for X65 pipeline steel is shown in Fig. 6 and the related calculation results are listed in Table 1 ($A = 3.93 \times 10^{-11}$ is adopted in the model calculation for demonstration). In the diagram of remaining fatigue life the effect of HE and CF on remaining fatigue life can be seen

straightforward. The two-component model, by imposing the AD rate to the HE rate, reasonably simulates the shape of the total crack propagation curve. And obviously, the AD process gives a rise of the crack propagation curve before $(1 - R)K_t$. This effect is quite significant at low ΔK regime. An analysis on the influence of the many factors involved in the proposed model is displayed below.

4.1 Loading frequency (f)

For most of commonly used alloys, the effect of frequency on crack growth under constant-amplitude fatigue loads in inert environments is negligible. However for the presence of a corrosive environment the effect of frequency is magnified, since the process of hydrogen transport is a crucial procedure of HE. And the time dependent characteristic of hydrogen transport indicates that the loading frequency f may have an impact to the CF crack growth. More specifically, lowering the loading frequency increases the exposure time of the metal to the corrosive environment, thus within each load cycle more atomic hydrogen may enter the material via the crack tip and diffuse a longer distance. Observations from fatigue tests in corrosive environments have proven the existence of critical frequency f_c , below which the material always shows no frequency dependence. For a CF process at f_c , its propagation curve should be in a shape like that plotted in Fig. 2(a), which is not the typical type of CF behavior for pipeline steels in seawater. Hence this paper puts its focus on the case where $f > f_c$, which is typical for CF of subsea pipeline steels in seawater.

In a typical situation, the influence of loading frequency can be divided into two parts, one is on the AD process before ΔK reaches $K_t(1 - R)$. The decrease of f will cause the increase of AD rates at each level of ΔK , as predicted by Eq. (8). This is because AD rate is affected by the time available for chemisorption at the crack tip during each cycle. The higher growth rate result at lower frequencies occurs because the corrosive environment has more contact time with the crack surfaces, while at higher frequencies, the corrosive environment cannot fully penetrate the crack tip before the crack begins to close again. The experimental observation by Bartlett and Hudak (1990) that a higher frequency resulted in lower fatigue crack growth rates than those at reduced frequencies supports this explanation. But in the log-log plot of da/dN vs ΔK , such increases are attenuated. The other part of the frequency influence is on the HE process, according to Eq. (18), the value of K_t decreases as f increases, manifesting as a left shift of the transition point on the crack propagation curve, as seen in Fig. 7.

4.2 Hydrogen concentration (C_{H0})

The hydrogen concentration C_{H0} is specified as the local hydrogen concentration in the material in front of a crack tip with the unstressed state. Most previous experiments of CF for pipeline steels in

seawater can hardly provide any information of C_{H0} . However, to some extent, the experimental data from tests conducted by Vosikovsky (1975) for API X65 pipeline steels under free corrosion in seawater with a series of loading frequencies can be used to investigate the influence of C_{H0} , since K_{IH} is calculated based on C_{H0} according to Eq. (15). Nevertheless, the K_{IH} for a component is hard to be decided under the free corrosion condition. For a demonstration purpose, a degradation factor δ with a value of $0 \sim 1$, representing the terms except K_{IN} at right hand side of Eq. (15), was used to multiply the inherent fracture toughness to approximate the remaining fracture toughness, namely

$$K_{IH} = \delta K_{IN} \quad (26)$$

Assume $\delta = 0.8$, i.e., 80 percent of the inherent fracture toughness is left providing a saturated hydrogen diffusion condition, then the predicted transitional points K_t on crack propagation curves were calculated and plotted together with the test data in Fig. 7.

Good agreement between model prediction and experimental data is observed in Fig.7. This means that if C_{H0} can be appropriately measured or determined, the CF behaviour of pipeline steels in seawater can be predicted pretty well. Also, if K_t is known or can be measured for a specific material-environment system, it may in turn help investigate the relationship of crack-cavity environment and the C_{H0} .

4.3 Stress ratio (R)

The experimental data of X42 pipeline steels in hydrogen gas provided by Gutierrez-Solana and Elices (1982) are used to investigate the influence of R on the model performance. The modelled crack growth processes and their evolution curves with different stress ratios in the situation of CF, HE, and fatigue, are plotted in Fig. 8.

Fig. 8 (a) and (b) show that the HE model can well capture the features of HAC behavior of X42 pipeline steels under different stress ratios and the influence of AD on cracking process is larger at lower stress ratios (e.g. $R = 0.1$) than that at higher ones (e.g. $R = 0.8$). Fig. 8 (c) - (f) show that both HE and AD's influence on the crack evolution time (cycles) or the remaining fatigue life is closely related to the change of stress ratio R . More specifically, as shown in Fig. 8 (c) and (d), at the low stress ratio $R = 0.1$, HE can cause a much larger reduction of crack evolution time or fatigue life than it at the high stress ratio $R = 0.8$, and thus AD occupies a bigger portion of the life reduction in the latter case. Within each case, HE is more contributive than AD in reducing the fatigue life, while the critical crack size doesn't change much for fatigue, HE influenced fatigue and CF. The little variation of critical crack size is attributed to the fact that when crack growth rate is high, hydrogen supply to the crack-tip region will be insufficient, leading to the recovery of material's fracture resistance, which is consistent

with Eq. (21) for $f > f_c$. On the other hand, Fig. 8 (e) and (f) indicate that the high stress ratio $R = 0.8$ provides a higher crack size where crack growth starts for either fatigue, HE influenced fatigue or CF. Meanwhile the remaining life of either HE influenced fatigue or CF decreases more heavily at the low stress ratio $R = 0.1$ than it at the high stress ratio $R = 0.8$ as the crack size extends. The model incorporates the effects of stress ratio by Eqs. (9) and (16). Eq. (16), known as Forman equation, includes the R effect in its denominator. While Eq. (9) states that for the case of CF, a high stress ratio can lead to a low AD rate at crack tip. It seems that the HE process dominates the influence of R on CF, represented as the shifts of transition points as shown by Vosikovsky (1981). The accompanying changes in crack growth rates can be seen as well in the experimental data by Vosikovsky (1981) for X70 pipeline steel specimens tested in seawater. In conclusion, the increase of stress ratio can cause a left shift of the transition point for the crack propagation curve and a decrease in fatigue life reduction due to CF. Detailed calculation results for X42 pipeline steel are listed in Table 2.

4.4 Temperature (T)

Temperature has impacts on the structural integrity in a few aspects. On the one hand, as indicated by Eqs.(15) and (18), temperature may affect the hydrogen diffusion process and furtherly impose an influence on K_t . On the other hand, the temperature variation may induce a stress/strain variation in the components (i.e. the so-called thermal load). Moreover, the temperature variation in the environment may even change the material's properties. For instance, at low temperatures steels could exhibit severe mechanical property degradation which is well-known as ductile-brittle transition (DBT). The change of mechanical properties are usually obtained by testing and is out of this research's scope. To study the impact that temperature may have on K_t , a temperature range of 20 °C - 300 °C is selected. Within this temperature range, both DBT and creep problems of carbon steels are unlikely to happen (Moura et al., 2009).

The variation trend of K_t of X65 pipeline steels in such a temperature range is plotted in Fig. 9. Fig. 9 shows that K_t increases as the environmental temperature increases, however, the increment is not very significant. When the temperature increases from 20 °C to 300 °C, K_t only increases from 19 MPa $\sqrt{m}$ to a value less than 28 MPa $\sqrt{m}$. However, the increase of K_t doesn't necessarily mean the EAC effects on the material under low temperature is less severe than that under high temperature. First, the environment temperature in this analysis only varies in a limited range where DBT and creep problems are unlikely to happen. Second, K_t is only used to assess the immediate structural integrity state, and can hardly give any information on the whole EAC process, especially if higher cracking rate appears. Lasebikan et al. (2013) concluded from a series of experiments conducted over a range of temperatures that the strain hardening index of steels can be reduced at elevated temperatures. Note

that the strain hardening index has a negative correlation with the crack growth rate in an elastic-plastic way (J-integral). It is very likely that higher crack growth rate occur at elevated environmental temperatures.

5. Conclusions

Subsea pipelines are vulnerable to CF which is a severe type of EACs. And there is a practical need to develop constitutive models for describing the CF process of pipeline steels in seawater. In this paper, the mechanisms of CF for subsea pipeline steels were investigated and it was found that CF is driven by SC and HAC. A two-component model was then proposed for subsea pipeline steels where both the AD and HE models are integrated based on fracture mechanics. The proposed model was applied to simulate the CF crack growth of X65 pipeline steels and the influences of factors of f , C_{H0} , R and T on the model performance were analyzed and discussed. The results show that:

- The proposed two-component model can reasonably simulate the total crack propagation curve. Both the exacerbated cracking rate and behavior features can be well captured with appropriate consideration of the effects of mechanical and environmental parameters.
- The increase of frequency will induce the decrease of AD rates and a left shift of the transition point on the CF crack propagation curve.
- If C_{H0} can be properly measured, the transition behavior of the CF crack propagation curves of pipeline steels in seawater can be well predicted.
- The increase of stress ratio can cause a left shift of the transition point for the crack propagation curve and a decrease in fatigue life reduction due to CF.
- In a limited temperature range where DBT and creep problems are unlikely to happen, K_t increases with the increase of the environmental temperature.

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Abbreviations

AD	Anodic Dissolution
AIDE	Adsorption Induced Dislocation Emission
CF	Corrosion Fatigue
CP	Cathodic Protection
DBT	Ductile-brittle Transition
EAC	Environment Assisted Cracking
EAZ	Environment Affected Zone
ECA	Engineering Critical Assessment
FFS	Fitness For Service
HAC	Hydrogen Assisted Cracking
HE	Hydrogen Embrittlement
HEDE	Hydrogen Enhanced De-cohesion
HELP	Hydrogen Enhanced Localized Plasticity
LEFM	Linear Elastic Fracture Mechanics
SC	Stress Corrosion
SCC	Stress Corrosion Cracking
SIF	Stress Intensity Factor

Table 1. Model calculation results for X65 pipeline steel

X65	R	K_t	Crack evolution time	Life reduction
Fatigue	0.5	-	1.7110×10^6	-
HE	0.5	24	6.5756×10^5	61.568%
CF	0.5	24	3.3674×10^5	80.318%
Unit	-	$\text{MPa}\sqrt{\text{m}}$	Cycle	-

Table 2. Model calculation results for X42 pipeline steels at different stress ratios

X42	R	K_t	Crack evolution time	Life reduction
Fatigue	0.1	-	2.0684×10^6	-
HE	0.1	9.7424	2.2721×10^5	89.016%
CF	0.1	9.7424	7.8647×10^4	96.198%
Fatigue	0.8	-	1.0645×10^7	-
HE	0.8	35	9.1655×10^6	13.902%
CF	0.8	35	8.4309×10^6	20.803%
Unit	-	$\text{MPa}\sqrt{\text{m}}$	Cycle	-

(a)

(b)

Figure1. Experimental CF data in comparison with model predictions of API 579 and BS 7910: (a) Simple linear model; (b) Bilinear model. (X65 and X70 data by Vosikovsky 1975 and Vosikovsky 1981)

Figure 2. Corrosion fatigue behaviour: (a) Type A CF; (b) Type B CF; (c) Mixed type CF. (McEvily and Wei 1972, log-log plot).

(a)

(b)

Figure 3. AD influence on the shape of crack propagation curve: (a) X65; (b) X70

(a)

(b)

Figure 4. Pipeline steel specimens in seawater with and without CP: (a) X65; (b) X70

Figure 5. Model of crack growth for the HE part of CF: (a) Corrosion-crack correlation; (b) Two-stage Forman equation model.

Figure 6. Model application to X65 pipeline steels with R=0.5: (a) crack propagation; (b) crack evolution.

Figure 7. Loading frequency effect on K_t

(a)

(b)

(c)

(d)

(e)

(f)

Figure 8. Stress ratio effect on model performance: (a) crack propagation with $R=0.1$; (b) crack propagation with $R=0.8$; (c) crack evolution with $R=0.1$; (d) crack evolution with $R=0.8$; (e) remaining fatigue life with $R=0.1$; (f) remaining fatigue life with $R=0.8$.

Figure 9. Environmental temperature effect on K_t