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Fatigue Crack Growth Modelling for Pipeline Carbon Steels under Gaseous Hydrogen Conditions

Ankang Cheng, Nian-Zhong Chen*

School of Marine Science and Technology, Newcastle University, United Kingdom

*Correspondence address:

Nian-Zhong Chen (Ph.D.)

Senior Lecturer in Ship and Offshore Structures,

School of Marine Science and Technology,

Newcastle University,

Newcastle upon Tyne,

NE1 7RU, United Kingdom

Tel: +44 (0) 191 208 6964

Fax: +44 (0) 191 208 5491

Email: nianzhong.chen@ncl.ac.uk

Abstract

A corrosion-crack correlation model is proposed for the hydrogen embrittlement (HE) influenced fatigue crack growth modelling of pipeline carbon steels under gaseous hydrogen conditions. The model is developed primarily based on the correlation of environment-affected zone (EAZ) and plastic zone. In the model, fatigue crack growth rate is predicted by Forman equation to take into account the influence from fracture toughness. The critical frequency and the “transition” stress intensity factor (SIF) are derived based on stress-driven hydrogen diffusion and Hydrogen-Enhanced De-cohesion (HEDE) hypothesis, which provides reasonable explanation on the frequency dependence of fatigue crack growth for pipeline carbon steels in hydrogen gas. Furthermore, an approximation formula involving threshold the SIF range and the stress ratio is established to describe the phenomenon of crack growth rate plateau. In addition, a formula is proposed to estimate the equilibrium fracture toughness according to the equilibrium between crack growth and hydrogen delivery rates. A series of experimental data from different grades of carbon pipeline steels (including high-strength grades such as X70 and X80) are utilized for demonstrate the validity of the proposed formulae and model effectiveness. The comparison between model predictions and experimental

data shows that the proposed model is capable of capturing the essence of pipeline carbon steels' fatigue crack growth process under gaseous hydrogen conditions.

Keywords: Fatigue crack growth, fracture mechanics, corrosion, hydrogen embrittlement, pipeline carbon steel

1. Introduction

It has been well known that the cracking process of metals such as carbon steels can be severely aggravated due to aggressive environment. This phenomenon is usually called environment-assisted cracking (EAC). Depending on the loading profile, there are two major categories of environment-assisted damage: stress corrosion cracking (SCC) and corrosion fatigue (CF). Considerable theoretic and experimental studies (Parkins, 2000) have been conducted on SCC. Based on the difference in crack morphology and environment conditions, it has been pointed out that for the two basic SCC modes, i.e. high-pH and near-neutral pH, the latter shares a similar morphology sometimes as well as the environmental conditions with those of CF. On the other hand, engineering structures are normally exposed to complex operations with varying working stresses that are usually a mixture of static and cyclic components (Schütz, 1993). Thus it was suspected that both two phenomena undergo identical EAC mechanisms. Some researchers even claimed SCC is only a special case of CF at the stress ratio equal to unity (Shipilov, 2002). Further studies indicated that the two phenomena are actually both mixtures of two crack-tip damage modes, namely the stress-assisted corrosion/stress corrosion (SC) and hydrogen-assisted cracking (HAC) (Woodtli and Kieselbach, 2000; Beavers and Harle, 2001). In the case of SC, the primary driving force for crack growth comes from the localized chemical corrosion processes occurring at the crack tip, which is usually explained by the theory of anodic slip dissolution (Parkins, 1979; Logan, 1952). Various models for estimating the crack growth rate were proposed based on either theoretical formulae such as Faraday's law or experimental tests or both (Endo, et al., 1981). As for HAC, crack growth is associated with absorbed hydrogen in the material. Those hydrogen may come from corrosion in aqueous solutions, cathodic protection, or high-pressure hydrogen gas, and then diffuse to a pre-existing flaw in the atomic state with stresses applied. Consequently, enhanced crack growth occurs and fracture happens at a lower stress level compared to that of the same material measured in air or inert gas. This phenomenon is called the hydrogen embrittlement (HE). Numerous mechanisms have been raised to account for the degradation of mechanical properties observed in experiments. At present, three of them have been widely accepted: 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. A critical review has been provided by Lynch in 2011. But a consensus is emerging that HEDE is likely to be the dominant mechanism (Gangloff, 2006). There is still controversy centring on 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. Nevertheless, 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).

As nowadays the renewable energy such as wind and solar is booming in its development, the corresponding energy storage technology is attracting more and more attention. In a possible scheme proposed, the fresh energy is converted into gaseous hydrogen by separating water, and then stored and transmitted by pipelines (Leighty et al., 2006). This idea is especially practical for offshore wind and solar farms where water supply is not a problem. Modern pipelines are usually made of medium or low strength steels and often designed by use of defect-tolerant principles, where knowledge of defect size and fatigue crack growth rate can be used to determine the remaining service life of a component. Several comprehensive reviews on the fatigue crack growth behaviour of pipeline carbon steels exposed to gaseous hydrogen were performed by Lam (2006), Nanninga (2010), and Liu (2013).

As seen from above literature review, it is clear that HE has a significant impact on fatigue crack growth behaviour of carbon pipeline steels, especially under high hydrogen pressures and varying loading frequencies. Although both corrosion fatigue and hydrogen influenced fatigue cracking show some frequency dependence, Nanninga (2010) pointed out that the mechanisms and models used to predict hydrogen influenced fatigue cracking in hydrogen may still differ significantly from those used for statically loaded applications or in fatigue situations where the hydrogen derives from aqueous liquids, which usually accompanies corrosion or/and negative potential. Moreover, unlike the situation for sustained load cracking in hydrogen environments, it was found that medium- even low-strength pipeline steels are highly susceptible to HE influenced fatigue cracking (Liu, 2013). However, there is still a lack of such models that can rationally account for the mechanisms of HE influenced fatigue cracking, just as what is mentioned in API RP 579 (2007), equations/models that describe fatigue crack growth behaviour in aggressive environments are only available for limited stress intensity ranges and most often intended for aqueous liquids.

In order to solve this problem, a corrosion-crack correlation model is developed based on the concepts of environment-affected zone (EAZ) and plastic zone. In the model, fatigue crack growth rate is predicted by Forman equation to take into account the influence from fracture toughness. The critical frequency and the “transition” stress intensity factor (SIF) are derived from theoretical basis of stress-driven hydrogen diffusion and HEDE. Furthermore, an approximation formula involving the threshold SIF range and stress ratio is established to describe the phenomenon of crack growth rate plateau. In addition, a formula is proposed to estimate the final fracture toughness determined by the equilibrium between crack growth and hydrogen delivery rates.

2. Hydrogen embrittlement influenced fatigue cracking process

A brief description for the HE influenced fatigue cracking process of pipeline carbon steels is presented in this section with the purpose of providing theoretical basis and clarifying physical meaning to the mathematical equations given in the next section.

As Oriani claimed in 1972, the basic notion of HEDE is that HE cracking occurs when the local opening tensile stress in front of crack tip exceeds the maximum-local atomic cohesion strength, which has been lowered by the presence of hydrogen. Based on this notion, Wang et al. (2013) established a model that can predict the degradation of fracture toughness for alloy steels exposed to gaseous hydrogen. Good agreement was observed between the prediction and experimental data. The success of this model indicates that HEDE works well on explaining the HE of pipeline carbon steels. However, it should be noted that the degraded fracture toughness was measured by testing pre-cracked specimens under sustained loading condition (Gutierrez-Solana, 1982), where there is sufficient time for hydrogen atoms to diffuse to the maximum tensile stress location. Thus fracture resistance is degraded and cracking is enhanced to the same extent corresponding to the hydrogen pressure along the whole test. The fracture toughness measured in such a way is called the saturated fracture toughness for the ambient hydrogen pressure.

However, if crack propagation goes beyond a specific speed (usually around the point when rapid unstable crack propagation begins), diffusing hydrogen cannot keep pace with the growing crack, resulting in an increased resistance against the rapid crack propagation and in turn the rate itself slows its acceleration to establish an equilibrium with the hydrogen delivery rate. Subcritical crack growth may then continue along the equilibrium rate toward an “equilibrium fracture toughness” (API RP 579, 2007). Such a HE influenced fatigue crack process can be observed in fatigue tests performed in hydrogen gas for pipeline steels spanning from low grades such as X42 (Cialone and Holbrook, 1985) to grades as high as X100 (Amaro et al., 2014).

If hydrogen damage is viewed as a special case of corrosion, then the foresaid process is a description on the corrosion-crack correlation mechanism existing in HE influenced fatigue cracking. Based on the typical process description, for a specimen tested in high-pressure hydrogen gas with fatigue loading, three specified types of fracture toughness are defined, namely the inherent fracture toughness, K_{IN} , which is measured in a non-aggressive environment with the same loading conditions; the saturated fracture toughness, K_{IH} , which is obtained under high hydrogen pressure using procedures defined in ASTM E 1820; the equilibrium fracture toughness, K_{IE} , namely the final fracture toughness displayed in the fatigue test.

3. Hydrogen embrittlement influenced fatigue crack growth modelling

For normal fatigue cracking (where aggressive environment is not applicable), in principle, crack propagation starts from the “stage I” (the “initiation” phase), mainly being “short crack”, and continues with the “propagation” phase of stage II and stage III (fast crack propagation), being “long crack” toward final failure (Pugno, 2006), as shown in Fig. 1. However, the process will be changed when HE is introduced. A typical HE influenced fatigue cracking process of pipeline carbon steels in hydrogen gas with constant amplitude loading is shown in Fig. 2 by the solid curve. The process is consisted of three stages. In the beginning, the crack rate grows along the sigmoidal curve dictated by K_{IH} , which is recognized as stage 1. The enhanced fatigue crack growth results from the hydrogen delivered at the crack tip through diffusion. When the crack growth rate goes beyond the hydrogen delivery rate, the crack propagates into the bulk material with little hydrogen, and as a consequence encounters larger resistance. So the growth acceleration decreases, and transition happens, indicating the commencement of stage 2. An equilibrium between the crack growth rate and hydrogen delivery rate is achieved later. Due to the continuity of hydrogen charging and crack growth, the cracking rate will not be dropped. Then, a plateau where the crack growth rate keeps constant at the equilibrium rate, appears in the crack growth curve and it lasts over certain range along the abscissa of K_{max} . Stage 3 starts when the curve merges into the sigmoid oriented by the equilibrium fracture toughness K_{IE} , which is closer to the inherent fracture toughness K_{IN} . The sigmoidal fatigue crack growth curves oriented by K_{IH} and K_{IN} are both plotted using dash lines in the same figure.

At present, linear elastic fracture mechanics (LEFM) analysis has been widely applied to engineering critical assessment (ECA) for steel structures. The main purpose of ECA is to predict the remaining service life of a component with use of Paris’ law (Paris and Erdogan, 1963), which is defined as:

$$\frac{da}{dN} = A(\Delta K)^n \quad (1)$$

where A and n are the Paris coefficients, and ΔK is the SIF range, which can be calculated as

$$\Delta K = F\Delta S\sqrt{\pi a} \quad (2)$$

F is the geometry function, and ΔS is the stress range ($K_{max} - K_{min}$).

It is generally agreed that LEFM can provide reasonable fatigue life estimates for long cracks (Stephens, 2000) and Paris’ Law works well for predicting the fatigue crack growth rate in stage II. For more complicated fatigue crack growth curves that are not straight lines in log-log plots, multi-segment lines are usually constructed in order to model the real curve.

As discussed in Section 2, degradation of fracture toughness of pipeline carbon steels will normally happen in high-pressure hydrogen gas. However, there is no link between crack growth rate and fracture toughness variation in Paris' Law, which is the main manifestation of material degradation in hydrogen gas. In 1967, Foreman et al. proposed an equation as:

$$\frac{da}{dN} = \frac{B\Delta K^m}{[(1-R)K_{IC} - \Delta K]} \quad (3)$$

where K_{IC} represents the fracture toughness in a general sense. The equation covers both stage II and III considering the fracture toughness variation. Therefore, Forman equation is adopted herein as the basic formulation for establishing the fatigue crack growth model of pipeline steels tested in high-pressure hydrogen gas. The model to be constructed is basically a two-stage Foreman equation model, as shown in Fig. 3 in comparison with the HE influenced fatigue crack growth curve and the fatigue crack growth curve in a non-aggressive environment. As seen from the Fig. 3, to schematically describe the HE influenced fatigue crack growth process, four key points should be seized, i.e. the threshold SIF range ΔK_{th} , the Transition stress intensity factor K_{tran} , plateau stress intensity range K_p , and the equilibrium fracture toughness K_{IE} .

3.1 Threshold stress intensity factor range (ΔK_{th})

ΔK_{th} determines when crack starts advancement. Three varying parameters may affect the ΔK_{th} of pipeline carbon steels in high-pressure hydrogen gas, namely the stress ratio R , fracture toughness, K_{IC} , and hydrogen pressure.

Based on the relation proposed by Davenport and Brook in 1979, for fatigue cracking under non-aggressive conditions, its threshold stress intensity factor range ΔK_{th} is given by:

$$\Delta K_{th} = \Delta K_0 \left(\frac{(K_{IC} - K_{max})(1-R)}{(K_{IC} - \Delta K_0)} \right)^{\frac{1}{3}} \quad (4)$$

where ΔK_0 represents the threshold SIF range at $R = 0$, and K_{max} is the maximum SIF. Meanwhile, it should be noted that

$$\Delta K_{th} = K_{maxth}(1-R) \quad (5)$$

and that

$$\Delta K_0 = K_{max0} \quad (6)$$

where $K_{\max th}$ and $K_{\max 0}$ correspond to the maximum SIFs at ΔK_{th} and at $R = 0$, respectively. ΔK_0 is usually obtained from experiments. Based on Eqs. (4) to (6), it follows that,

$$\frac{K_{\max th}}{K_{\max 0}} = (1 - R)^{-\frac{2}{3}} \left(\frac{K_{IC} - K_{\max th}}{K_{IC} - K_{\max 0}} \right)^{\frac{1}{3}} \quad (7)$$

As an example, the relation between R and ΔK_{th} of a 0.15%C-1.5%Mn steel is drawn in Fig. 4 (a).

On the other hand, Fig.4(b) shows that the variation in K_{IC} gives little impact (up to 4%) on the threshold SIF range. This may be because the variation of threshold SIF range of metals exposed to hydrogen mainly comes from the change of threshold SIF range at $R = 0$ under different hydrogen pressures.

In 2007, Somerday et al. did a critical review on hydrogen influence on fatigue crack growth and concluded that at ΔK values near ΔK_{th} , the effect of hydrogen environment on both the crack growth or on the value of ΔK_{th} was not evident. This conclusion was supported by previous experiments conducted by Pendse and Ritchie (1985), and Dauskardt and Ritchie (1986). Therefore, in this paper, the degradation of ΔK_{th} due to hydrogen influence is assumed to be $1 \text{ MPa}\sqrt{m}$ for all hydrogen pressures. The threshold SIF range for pipeline carbon steels tested in high-pressure hydrogen gas is assumed to be:

$$\Delta K_{tH} = \Delta K_{th} - 1 \quad (8)$$

3.2 Transition stress intensity factor (K_{tran})

Transition stress intensity factor, K_{tran} , is determined by establishing the corrosion-crack correlation model, which interprets the fatigue crack growth behaviour of pipeline carbon steel under HE influence as the manifestation of the correlation between the sizes of the environment-affected zone, r_{EAZ} , and plastic zone r_p .

Penetration of chemical agents into a localized crack-tip region would severely damage the bulk of material near the crack tip. The damaged material no longer represents the original bulk material and exhibits accelerated crack growth rates as compared to its baseline behaviour. This damaged zone is called the environment-affected zone (EAZ). In the case of HE influenced fatigue crack growth testing, hydrogen is considered the primary source for creating the damaged zone. According to the HEDE theory, hydrogen damage sites are located at a distance ahead of the crack tip surface where tensile stresses are maximized, namely the plastic-elastic boundary. The elastic-plastic stress field near the crack tip was schematically plotted in Fig. 5

Hill (1998) performed a detailed analysis on the stress field in front of a crack tip with crack tip radius ρ . It was found that the hydrostatic stress within plastic zone σ_h , can be expressed as

$$\sigma_h = \sigma_{ys} \left[\frac{1}{2} + \ln \left(1 + \frac{r}{\rho} \right) \right] \quad (9)$$

where r is the distance in front of the crack tip, and σ_{ys} corresponds to the yield strength.

For hydrogen atoms diffusing into the crack-tip region, their energetic driving force μ , which is created by the chemical potential can be written as

$$\mu = \mu_0 + k_B T \ln \frac{c}{1+c} + \sigma_h V_H \quad (10)$$

where c is the hydrogen concentration at r , V_H is the partial volume of hydrogen, k_B is the Boltzmann constant, T is the temperature and μ_0 is a constant.

Thus the velocity of hydrogen atoms from the crack tip V_r has the expression

$$V_r = - \frac{D}{k_B T} (\nabla \mu)_r \quad (11)$$

where D is the diffusion coefficient, and is calculated from the ratio of hydrogen permeability and solubility for carbon steel (Gadgeel and Johnson, 1979).

Note that the H concentration $c \ll 1$ when the location is away from the immediate crack tip, so the concentration driven part is so small that any explicit concentration dependence of diffusion can be neglected. Also, the aggregation and redistribution process of local hydrogen around the crack tip is quite short and it has no significant influence on the time required for hydrogen atoms to reach the near-crack-tip region (Song and Curtin, 2011). Therefore the dominant force driving hydrogen to diffuse is the hydrostatic pressure in front of the crack tip and the hydrogen concentration at diffusion front line can be viewed as being constant. This analysis yields the following expression for hydrogen diffusion velocity V_r ,

$$V_r = - \frac{D V_H}{k_B T} \left(\sigma_{ys} \frac{\frac{1}{\rho}}{1 + \frac{r}{\rho}} \right) \quad (12)$$

On the other hand, $V_r = - \frac{dr}{dt}$. So, after time t , hydrogen atoms can be spread within a distance of $r = R$ in front of the crack tip, where

$$R = \sqrt{\frac{2 D V_H \sigma_{ys}}{k_B T} t} \quad (13)$$

In a stress cycle, when unloading starts, the residual stress can cause compression in front of the crack tip, since HE effect is limited to tensile stress state, only half time of a cycle contributes to the H accumulation at the near crack-tip region. This limited distance is defined as r_{EAZ} . If this distance in front of the crack tip is not larger than the plastic zone size, it can be calculated by

$$r_{\text{EAZ}} = \sqrt{\frac{DV_{\text{H}}\sigma_{\text{ys}}}{k_{\text{B}}T} \cdot \frac{1}{f}} \quad (14)$$

where f is the loading frequency.

For Mode I loading under the elastic condition, the stress σ_{yy} perpendicular to the crack plane at a distance of r off the crack tip can be expressed in a function of SIF, K_{I} , as

$$\sigma_{yy} = \frac{K_{\text{I}}}{\sqrt{2\pi r}} \quad (15)$$

The maximum principal stress happens at the elastic-plastic boundary r_{c} , where,

$$\sigma_{\text{max}} = \sigma_{yy} = \omega\sigma_{\text{ys}} \quad (16)$$

ω is a magnification factor of value $3 \sim 5$ accounting for the material's working hardening effect.

Accordingly, at fracture initiation point, where $K_{\text{I}} = K_{\text{IN}}$, Eq. (15) can be rewritten as

$$r_{\text{c}} = \frac{1}{2\pi} \frac{K_{\text{IN}}^2}{(\omega\sigma_{\text{ys}})^2} \quad (17)$$

Since the loading frequency may vary when comparing different loading cases, a critical frequency, f_{c} , is determined as where EAZ reaches the elastic-plastic boundary, i.e. $r_{\text{EAZ}} = r_{\text{c}}$. Accordingly,

$$f_{\text{c}} = \frac{4\pi^2 DV_{\text{H}} (\omega\sigma_{\text{ys}})^4}{k_{\text{B}} T K_{\text{IN}}^4} \quad (18)$$

X60 steels used in the experiment conducted by Yu et al. (2015) were taken as an example for illustration and the results were plotted in Fig. 6. In Fig. 6 (a), it can be seen that r_{EAZ} varies as the loading frequency changes. A higher loading frequency corresponds to a smaller size of EAZ, and thus a weaker corrosion effect, which is in agreement with experimental observations (Yu et al., 2015). At the critical frequency f_{c} , the frontline of EAZ draws back to the elastic-plastic boundary, i.e. $r_{\text{EAZ}} = r_{\text{c}}$. The plastic zone size r_{p} varies differently under different loading frequencies. If the loading frequency $f < f_{\text{c}}$, r_{EAZ} will be larger than r_{p} and the relation $r_{\text{EAZ}} > r_{\text{c}} \geq r_{\text{p}}$ holds over the whole fatigue crack growth process as shown in Fig. 6 (b). If $f > f_{\text{c}}$, which represents a typical HE influenced fatigue cracking process of pipeline carbon steels in hydrogen gas shown in Fig. 2, r_{p} may be smaller than

r_{EAZ} at first, corresponding to the stage 1 crack growth. However, as K_{max} increases, the growing r_p comes equal to and then goes beyond r_{EAZ} at some point before K_{max} reaches K_{IC} , corresponding to the stage 1 and 3 crack growth respectively. For the whole fatigue crack growth process, $r_{\text{EAZ}} < r_p \leq r_c$ stays valid. If $f = f_c$, r_p grows and achieves the equilibrium $r_{\text{EAZ}} = r_p = r_c$ at the point where $K_{\text{max}} = K_{\text{IC}}$.

Before having an insight into transition SIF, the assumption that r_c is independent of hydrogen (Wang et al., 2013) is introduced, thus the HEDE mechanism is expressed as (Oriani, 1977)

$$\sigma_{\text{maxH}} = \sigma_{\text{max0}} - \beta C \quad (19)$$

where σ_{maxH} and σ_{max0} are the maximum principal stress at r_c with and without hydrogen when cracking initiates, respectively. β is a parameter related to loss of critical cohesive stress by hydrogen impurity, and C is the local hydrogen concentration.

Insert Eq. (16) into Eq. (19), it follows that

$$\omega_H = \omega - \frac{\beta C}{\sigma_{ys}} \quad (20)$$

ω_H accounts for the material's working hardening effect with hydrogen introduced. Provided that K_{IH} and K_{IN} have been determined, then Eq. (20) can be written as

$$\omega_H = \omega \frac{K_{\text{IH}}}{K_{\text{IN}}} \quad (21)$$

Based on this model, the point where the crack propagation curve starts transition is the very point where r_p grows to an equilibrium with r_{EAZ} , i.e.

$$r_p = r_{\text{EAZ}} \quad (22)$$

Inserting Eq. (14), Eq.(15), and Eq. (17) into Eq. (22), the transition SIF can be obtained as

$$K_{\text{tran}} = \omega \sigma_{ys} \frac{K_{\text{IH}}}{K_{\text{IN}}} \left(\frac{4\pi^2 D V_H \sigma_{ys}}{k_B T f} \right)^{\frac{1}{4}} \quad (23)$$

If the time interval last so long that the hydrogen passes the elastic-plastic boundary, diffusing into the elastic stress field, then the hydrostatic stress afterward holds the following expression

$$\sigma_h = \frac{2(1+v)}{3} \frac{K_I}{\sqrt{2\pi r}} \quad (24)$$

The size of EAZ then should be a sum of hydrostatic stress contribution from both the plastic and elastic zones. But since the summated size is definitely larger than that of the critical plastic zone and maximum stress still occurs at the elastic-plastic boundary, the cracking happens in the same way with that of the situation where $r_{\text{EAZ}} = r_p$.

Note that it was concluded in section 3.1 that hydrogen influence on crack growth rate is not evident at ΔK values near threshold SIF range. This observation results in the lower boundary condition for K_{tran} . Also, the upper boundary condition that K_{tran} should not be larger than K_{IH} . A controlling function for K_{tran} can thus be established as

$$K_{\text{tran}} = \min \left\{ \max \left\{ K_{\text{tran}}, \frac{\Delta K_0}{(1-R)} \right\}, K_{\text{IH}} \right\} \quad (25)$$

The relationship between K_{tran} and f is shown in Fig. 7 where X60 in the experiment conducted by Yu et al. (2015) was taken for illustration. It can be seen that though the loading frequency f can vary in a large range ($10^{-5} - 10$ Hz in the plotting), the transition SIF K_{tran} is restricted to a bounded range by the threshold SIF range and the saturated fracture toughness, as the lower and upper boundaries respectively. The value of K_{tran} is determined by the value of f . If K_{tran} is located at the boundary K_{IH} , it means the crack will grow with saturated hydrogen diffusion along the fatigue crack growth curve dictated by the fully degraded fracture toughness K_{IH} until final fracture happens. In other words, if the K_{tran} value calculated via Eq. (23) is not smaller than K_{IH} , transition behaviours may not be expected, and fractures may happen as K_{max} approaches K_{IH} . This is in accordance with the fact that the lower the loading frequency, the stronger the corrosion effect.

3.3 Plateau stress intensity range (K_p)

In this paper, K_p is defined as

$$K_p = \frac{\Delta K_{\text{tH}}}{(1-R)} \quad (26)$$

K_p can be found in Fig.3 where a nearly constant crack growth rate will be kept before the crack propagation curve catches up and evolve along the new curve that has a fracture toughness locating between the K_{IN} and K_{IH} .

3.4 Fracture toughness (K_{IC})

As mentioned in section 2, three types of fracture toughness should be determined, i.e. the inherent fracture toughness K_{IN} , the saturated fracture toughness K_{IH} , and the equilibrium fracture toughness K_{IE} .

K_{IN} is normally obtained from experiments (Gutierrez-Solana, 1982; Oh et al. 2007; Godefroid et al., 2014; Santos et al., 2010). For most of materials, the inherent fracture toughness K_{IN} can be expressed in terms of any one of the three parameters J_{IN} , CTOD and K_{IN} . If the valid data of K_{IN} are not available, K_{IN} may be approximately estimated by one of the following equations (Broek, 1974):

$$K_{IN}^2 = \frac{J_{IN} \cdot E}{(1 - \nu^2)} \quad (27)$$

$$CTOD = \frac{4}{\pi} \frac{K_{IN}^2}{E \sigma_{ys}} \quad (28)$$

where E and ν are the Young's Modulus and Poisson's ratio of material, respectively.

K_{IH} is calculated herein using Wang's model (Wang et al., 2013):

$$\frac{K_{IH}}{K_{IN}} = 1 - \beta s \frac{\sqrt{P}}{\omega \sigma_{ys}} \exp \left[\frac{2(1 + \nu)}{3} \frac{V_H}{k_B T} \frac{K_{IH}}{K_{IN}} \omega \sigma_{ys} \right] \quad (29)$$

where P is the hydrogen pressure, and s is the solubility. Once all the parameters besides K_{IH} are provided, K_{IH} is then predictable.

Based on Eqs.(14), (17), and (18), a formula is proposed for the equilibrium fracture toughness K_{IE} as:

$$K_{IE} = \begin{cases} \left[1 - \left(\frac{r_{EAZ}}{r_c} \right)^\lambda \right] K_{IN}, & f > f_C \\ K_{IH}, & f \leq f_C \end{cases} \quad (30)$$

In general, K_{IE} is achieved by applying the corrosion-crack correlation theory and Eq. (30) is the correlation function of Eqs.(14), (17), and (18). f_C is calculated from Eq. (18), r_{EAZ} is decided by Eq. (14), and r_c is obtained from Eq. (17). When $r_{EAZ}/r_c \geq 1 \Rightarrow f \leq f_C$, $K_{IE} = K_{IH}$; when $0 < r_{EAZ}/r_p < 1 \Rightarrow 0 < r_{EAZ}/r_c < 1 \Rightarrow f > f_C$, $K_{IE} = \left[1 - \left(\frac{r_{EAZ}}{r_c} \right)^\lambda \right] K_{IN}$; when $r_{EAZ}/r_c \rightarrow 0 \Rightarrow f \gg f_C$, $K_{IE} \rightarrow K_{IN}$. A non-negative parameter λ is used to adjust the rate with which the equilibrium fracture toughness tends toward the inherent fracture toughness. The larger the value of λ , the smaller the difference between K_{IE} and K_{IN} . Material properties and environment conditions can have an impact on the value of λ . In this paper, $\lambda = 1$ is adopted for demonstration purpose.

4. Discussion

The range of loading frequency in practice is large, typically in the range from 10^{-1} to 10^{-6} Hz. For most of the commonly used alloys, the frequency effect of constant-amplitude load on crack growth rate is negligible in dry-air environments. However, because of the presence of hydrogen, the frequency effect on fatigue crack growth behaviour appears for materials under constant-amplitude load. It is generally believed that the fatigue crack growth rate increases with lowering frequency, because lower cyclic loading frequency extends the exposure time of the material to the aggressive environment, which allows more hydrogen atoms to diffuse to a longer distance in front of the crack tip within each loading cycle. Thus the material properties near crack-tip is furtherly degraded. But observations from fatigue tests in aggressive environments have shown that there exists a critical frequency f_C , under which the material properties always show the same extent of degradation. The critical frequency changes as testing material varies (Yu et al., 2015). The proposed Eq. (18) for f_C only depends on material properties and it is thus unique to a particular material, which shows a good agreement with the observations.

To demonstrate the validity of Eq. (18) for predicting the f_C , an experiment conducted by Yu et al.(2015) is utilized for a comparison between model prediction and experimental results. The experiment was performed on X60 pipeline steel and the measured critical frequency was 1.04×10^{-3} Hz. According to the experimental conditions, σ_{ys} of X60 pipeline steel = 414 MPa, $T = 303$ K, $\nu = 0.31$, $V_H = 2 \times 10^{-6}$ m³/mol. Inherent fracture toughness K_{IN} , is assumed as 142 MPa as reported by Guedri et al. (2004). Since the material's working hardening effect ω usually ranges from 3 to 5, f_C is calculated to be in the range of 4.4×10^{-4} and 3.4×10^{-3} Hz. While the measured critical frequency is 1.04×10^{-3} Hz that exactly falls in the predicted range of 4.4×10^{-4} and 3.4×10^{-3} Hz, which shows the critical frequency of the experiment is captured by Eq. (18). It is thus indicated that Eq. (18) may provide reasonable prediction for the critical frequency of a specific material under a given loading frequency if the ω is known.

So far only a few of investigations have been done on EAZ. Kim and Manning (1983) hypothesized that, the size of EAZ r_{EAZ} , is in proportion to the crack-tip plastic zone size (which is a function of the SIF K_I), hydrogen diffusion coefficient for the material, and hold time in the loading profile. However, based on a vast number of experimental data obtained from fatigue tests for Inco 718 alloy under high temperature, Chang (1988) found that r_{EAZ} was a function of hold time and temperature, independent of K_I . As seen from the proposed Eq.(14), the estimate of r_{EAZ} is also only related to the loading frequency and it is independent of K_I , which is well supported by Chang's observation.

Furthermore, the experimental data from tests conducted by Vosikovsky in 1975 for API X65 pipeline steels under free corrosion in seawater with a series of loading frequencies can be used to confirm the validity of Eq.(23), since the hydrogen diffusion of a CF process without CP is also mainly stress-driven. As the saturated fracture toughness is hard to be decided for a component in the free corrosion seawater environment, a degradation factor δ , with a value of $0 \sim 1$, was multiplied by the inherent fracture toughness to approximate the remaining fracture toughness, namely

$$K_{IH} = \delta K_{IN} \quad (31)$$

Assume $\delta = 0.8$, i.e. 80 percent of the inherent fracture toughness is left providing a saturated diffusion condition, then the predicted transitional fracture toughness were calculated and plotted together with the test data in Fig. 8. Good agreement between model prediction (Eq. (23)) and experimental data is observed.

5. Model applications

The effectiveness of the proposed corrosion-crack correlation model is demonstrated by applying it to four types of API pipeline carbon steels and comparing with experimental data, i.e., X42 (Cialone and Holbrook, 1985), X65 (Ronevich et al., 2016), X70 (Drexler et al., 2014) and X80 (Stalheim et al., 2010). X42 were tested in 6.9 MPa hydrogen gas at $f = 1 \text{ Hz}$ and $R = 0.1$ and $R = 0.8$. X65 were tested in 21 MPa hydrogen gas at $R = 0.5$ and $f = 1 \text{ Hz}$. X70 were tested in 5.5 MPa hydrogen gas at $R = 0.5$ and $f = 1 \text{ Hz}$. X80 were tested in 21 MPa hydrogen gas at $R = 0.5$ and $f = 1 \text{ Hz}$.

The model application can be divided into two main steps. The first step is to calculate the parameters f_C and K_{tran} based on the actual experimental conditions. The second step is to achieve the Foreman coefficients using the measured fatigue crack growth data from the fatigue tests. The predicted model parameters are listed in Table 2. The Foreman coefficients obtained by model calibration with experimental data are given in Table 3. The comparison results between model prediction and experimental data are plotted in Figs. 9 - 12.

Fig.9 shows that there is significant difference in crack growth behaviour observed from the comparison between experimental data obtained under hydrogen gas and non-aggressive conditions at $R = 0.1$ and $R = 0.8$. As shown in Table 2, the critical frequency f_C for the X42 specimen under experimental conditions calculated by Eq. (18) is smaller than the specific loading frequency f . Fig.9 also shows that the transitions in the HE influenced fatigue crack growth curves exactly happen at the K_{tran} predicted by Eq. (23), which is listed in Table 2. As shown in Fig.9, K_p determined by Eq. (26)

yields a good approximation to the stress intensity range of the crack growth rate plateau and good agreements can be observed in each stage of the HE influenced fatigue crack growth. This indicates that the model provides good prediction and also successfully captures the essence of the HE influenced fatigue crack growth behaviour of X42 specimens tested in 6.9 MPa hydrogen gas at $f = 1 \text{ Hz}$ and $R = 0.1$ and $R = 0.8$. The results show that the proposed model works well to predict the HE influenced fatigue crack growth behaviour of X42 pipeline carbon steels under both low and high stress ratios. This illustrates that the proposed corrosion-crack correlation model can be applied to predict the HE influenced fatigue crack growth behaviour of low-strength pipeline carbon steels under different stress ratios.

Fig.10 shows that the model curve agrees well with the experimental data of medium-strength carbon steel X65 tested in 21 MPa hydrogen gas at $R = 0.5$ and $f = 1 \text{ Hz}$. The critical frequency f_c calculated from Eq. (18) for the tested X65 specimens, as listed in Table 2, is smaller than the specific loading frequency f . As shown in Fig.10, transition of the fatigue crack growth curve occurred at the point on the abscissa as predicted by Eq. (23) and then a plateau of crack growth rate appeared and lasted over a range of $K_p(1 - R)$. As K_{max} is increased continually, the crack propagates with a rapid crack growth rate approaching the critical size determined by K_{IE} . It is found that in each stage of the HE influenced fatigue crack growth, the model curve agrees well with the experimental data. Again, the results shown in Fig. 10 indicate that the proposed corrosion-crack correlation model provides good prediction for the fatigue crack growth behaviour of medium-strength pipeline carbon steels.

Figs. 11 and 12 shows that the proposed model also works well for prediction of the HE influenced fatigue crack growth behaviour of high-strength pipeline carbon steels X70 and X80. As predicted, Table 2 also shows that the critical frequencies f_c estimated by Eq. (18) under their experimental conditions are smaller than the specific loading frequency f . It can be seen in Figs. 11 and 12 that both transitions of the HE influenced fatigue crack growth curves happen at predicted K_{tran} listed in Table 2. Both figures also show that K_p calculated from Eq. (26) provides a good approximation to the stress intensity range over which the crack growth rate plateau phenomenon occurs. Good agreements are observed between the predicted HE influenced fatigue crack growth curve and the actual experimental data of high-strength pipeline carbon steels.

6. Conclusions

Based on the correlation of environment-affected zone (EAZ) and plastic zone, a corrosion-crack correlation model taking into account the influence from fracture toughness was developed for

fatigue crack growth modelling of pipeline carbon steels under gaseous hydrogen conditions. An experiment performed on X60 pipeline steel and CF tests on X65 pipelines steels were used to demonstrate the validity of the proposed formulae for predicting the critical frequency f_c and Transition stress intensity factor K_{tran} . A series of experiments of four types of API pipeline carbon steels X42 (low-strength), X65 (medium-strength), X70 (high-strength) and X80 (high-strength) were utilized for demonstrating the model effectiveness. The results show that:

- For the HE influenced fatigue crack growth, the derived Eqs.(18) and (23) provides good estimation for the critical frequency and the transition stress intensity factor.
- The proposed corrosion-crack correlation model provides good prediction for the fatigue crack growth behaviour of low-strength, medium-strength, and also high-strength pipeline carbon steels.
- Plateau stress intensity range K_p calculated from the proposed Eq.(26) provides a good approximation to the stress intensity range over which the crack growth rate plateau phenomenon occurs.

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Table 1. Experimental data for API grade pipeline carbon steels

API grade	R	P	f	fatigue test data $da/dN - \Delta K$	Inherent fracture toughness
X42	0.1/0.8	6.9	1	Available	K_{IN}
X42	0.8	6.9	1	Available	K_{IN}
X65	0.5	21	1	Available	J_{IN}
X70	0.5	5.5	1	Available	CTOD
X80	0.5	21	1	Available	CTOD
Unit	-	MPa	Hz	mm/cycle – $MPa\sqrt{m}$	$MPa\sqrt{m}$

Table 2. Model predictions

API grade	R	P	ΔK_{th}	K_{tran}	K_p	f_C	K_{IE}	ω	λ
X42	0.1	6.9	5.7	8.5	6.4	1.6×10^{-4}	145.4	4.2	1
X42	0.8	6.9	2.8	35	14.1	1.6×10^{-4}	145.4	4.2	1
X65	0.5	21	6.8	21.2	13.5	0.0013	158.4	4.8	1
X70	0.5	5.5	6.8	21.4	13.5	0.0020	188.2	4.0	1
X80	0.5	21	6.8	20	13.6	0.0014	188.7	3.2	1
Unit	-	MPa	$MPa\sqrt{m}$	$MPa\sqrt{m}$	$MPa\sqrt{m}$	Hz	$MPa\sqrt{m}$	-	-

Table 3. Model parameters from experimental data

API grade	R	P	B_1	B_2	m_1	m_2
X42	0.1	6.9	6.7×10^{-10}	1.6×10^{-6}	7.0	4.4
X42	0.8	6.9	1.0×10^{-7}	5.2×10^{-6}	4.2	4.5
X65	0.5	21	8.2×10^{-9}	1.2×10^{-5}	4.4	2.7
X70	0.5	5.5	1.7×10^{-10}	1.0×10^{-5}	6.4	3.1
X80	0.5	21	1.0×10^{-7}	2.3×10^{-6}	3.9	3.4
Unit	-	MPa	-	-	-	-

Figure 1. Schematic diagram of a normal fatigue cracking process

Figure 2. Schematic diagram of a typical HE influenced fatigue cracking process

Figure 3. Schematic diagram of the corrosion-crack correlation model

(a)

(b)

Figure 4. (a) ΔK_{th} - R relationship; (b) Influence of K_{IC} on ΔK_{th} (given that ΔK_0 keeps constant)

Figure 5. Schematic diagram of stress distribution in front of crack tip

(a)

(b)

Figure 6. (a) definition of f_c ; (b) relationship between r_{EAZ} and r_p

Figure 7. Relationship between K_{tran} and f

Figure 8. Comparison between predicted K_{tran} and experimental data of X65 steel

(a)

(b)

Figure 9. Comparison between model prediction and experimental data of X42 (a) R = 0.1; (b) R = 0.8

Figure 10. Comparison between model prediction and experimental data of X65

Figure 11. Comparison between model prediction and experimental data of X70

Figure 12. Comparison between model prediction and experimental data of X80