Near-threshold fatigue crack propagation behavior of austenitic high-Mn steels

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High-Mn austenitic steels utilizing TWIP (twinning induced plasticity) effect have excellent combination of tensile strength and ductility. The near-threshold fatigue crack propagation (FCP) behavior, as represented by the $\Delta K_{th}$ value, of high-Mn steels was examined with the emphasis on the effect of stacking fault energy (SFE), grain size, twinning and tensile properties. Even though no predominant parameter determining the near-threshold FCP behavior of high-Mn steels was found, the SFE showed the most reasonable correlation to the $\Delta K_{th}$ values among the variables examined. It was also suggested that the slip reversibility as determined by SFE could not solely explain the near-threshold FCP behavior of high-Mn steels. The presence of twin boundaries appeared to be not beneficial in improving the resistance to FCP of high-Mn steels in low $\Delta K$ regime. The near-threshold FCP characteristics of high-Mn austenitic steels were discussed and correlated with the well-known parameters presumably affecting FCP.

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1. Introduction

The increasing demand for strong and tough steels for automobile and cryogenic applications leads steel industry to put a strong effort to develop high-Mn austenitic steels utilizing TWIP (twinning induced plasticity) effect [1–7]. Mechanical twins formed during deformation induce high strength and formability of high-Mn austenitic steels, which is called TWIP effect characterized by a high rate of work hardening [5, 8–13]. As the formation of mechanical twins involve the creation of new crystal orientations, twins progressively reduce effective mean free path of dislocations and increase flow stress, leading to high strain hardening [14]. A great deal of studies has been conducted on the deformation behavior of high-Mn steels, but the present knowledge regarding the fatigue behavior is extremely limited. For high-Mn steels under fatigue loading, the formation of mechanical twins has not been reported. It has been argued that the deformation is rather localized and the strain level is not sufficiently high enough to form mechanical twins under cyclic loading [15,16].

Target applications for high-Mn steels include cryogenic tankage, line pipe and parts for automobile chassis that require damage tolerance capability of a structure to sustain defects or cracks. Therefore, the resistance to fatigue crack propagation (FCP), as well as that to high and low cycle fatigue, becomes an important property [17–19]. At present, few literature on this important issue is available, and no systematic studies have been conducted on the FCP behavior of various high-Mn steels with different chemical compositions. Particularly, the FCP characteristics of metals in low stress intensity factor range ($\Delta K$) regime, which are most often represented by a single value called $\Delta K_{th}$ below which no crack growth, have been considered to an important variables to express the damage tolerance ability [20–24]. It has been well established that the FCP behavior of metal is strongly affected by stacking fault energy (SFE), as well as microstructural features. The SFE is the key factor to understand FCP behavior because single phase alloys with low SFE have high slip planarity, enhancing the resistance to FCP by strain localization in the plastic zone. Slip reversibility is expected to be increased with large grain size improving the FCP resistance, as long as the slip characteristics remain constant. High-Mn steels, with Mn content ranging between 15 and 30%, often contain Al, Si, Cr, C and N to obtain stable fully austenitic microstructure with the SFE values ranging from 15 to 30 mJ/m² [4,6–8], and the correlation between SFE and tensile deformation mode has been well established. However, the present understanding is limited on the effect of SFE on the $\Delta K_{th}$ values of high-Mn steels with various chemical compositions.

In this study, the FCP behavior of high-Mn steels with a variety of chemical compositions was investigated to understand the controlling factors determining the near-threshold characteristics of crack propagation. The fracture surfaces of FCP-tested specimens were documented by using a scanning electron microscope (SEM) to identify morphological characteristics in fracture surface. The effects of SFE, grain size and tensile properties on the FCP behavior of high-Mn steels were discussed.

2. Experimental Procedures

A variety of high-Mn steels, including 25Mn, 25Mn0.5Cu, 24Mn4Cr, 22Mn3Cr, 20Mn, 18Mn, 17Mn1.8Al and 16Mn2Al, were supplied by POSCO (Pohang, Korea) to study the FCP behavior of austenitic high-
Mn steels. The specimen designation and chemical composition of high-Mn steels in this study is shown in Table 1. For the micrographic observation, each specimen was polished and etched by using a 2% nitral solution for 50 s. Fig. 1 shows the electron backscatter diffraction (EBSD) figures of (a) 25Mn, (b) 25Mn0.5Cu, (c) 24Mn4Cr, (d) 22Mn3Cr, (e) 20Mn, (f) 18Mn, (g) 17Mn1.8Al and (h) 16Mn2Al specimens. As previously identified for various high-Mn steels with similar chemical compositions [4,21,24], an austenite single phase and annealing twin bands within the grains were observed for each specimen. The grain size and the effective grain size considering the twin boundaries of each specimen were measured by using an image analyzer program. Flat tensile specimens were prepared from each sheet or plate with the tensile axis parallel to the longitudinal direction, and tensile tests were conducted at a nominal strain rate of $10^{-3}$/s at 298 K. The grain size, effective grain size, calculated stacking fault energy and tensile properties are summarized in Table 2.

The FCP tests were performed on the compact tension (CT) specimens at 298 K in lab air at an R ratio of 0.1 with a sinusoidal frequency of 10 Hz in accordance with ASTM E647 [25] using a servo-hydraulic testing machine (Instron model 8516). Load shedding technique was used to obtain the $\Delta K_{th}$ value by decreasing load steps at selected crack size intervals by using the following equation.

$$C = \left( \frac{1}{R} \right) \left( \frac{d \Delta K}{d a} \right)$$

where C is the normalized K-gradient and a is the crack length. During K-decreasing process, C value was controlled to be above $0.08 \text{ mm}^{-1}$. The fatigue crack length was automatically measured by using a DCPD (direct current potential drop) method. The SEM fractographic analyses were conducted on the tested specimens to observe the fracture mode.

### 3. Results and Discussion

Fig. 2 shows the da/dN (FCP rates)-$\Delta K$ curves of high-Mn steels at an R ratio of 0.1 and the testing temperature of 298 K. It was noted that the FCP rates of austenitic single phase, high-Mn steels varied with different chemical compositions, particularly in low $\Delta K$ regime. The $\Delta K_{th}$ value, which represents the critical $\Delta K$ value below which fatigue crack does not propagate, of each specimen is summarized in Table 3, along with the variables, m and C, in Paris’ equation. The $\Delta K_{th}$ values of high-Mn steels were in the range of 9.8 MPa$\sqrt{m}$ for the 18Mn specimen to 5.1 MPa$\sqrt{m}$ for the 16Mn2Al specimen, showing substantial difference between the specimens. The difference in the FCP rates of high-Mn steels became reduced in intermediate and high $\Delta K$ regimes with the similar Paris’ slope, as shown in Table 3 [26]. It has been well established that strain localization tends to increase the resistance to FCP of metals by encouraging planar slip [27,28]. The enhanced slip planarity then increases the slip reversibility under cyclic loading and, eventually, the FCP resistance of metals [27,28]. The slip planarity representing the degree of strain localization for single phase alloy is largely dependent on SFE, the high value of which encourages cross-slip and homogeneous deformation [29,30]. Therefore, single phase alloys with low SFE have high slip planarity, enhancing the resistance to FCP by strain localization in the plastic zone. For metals with planar slip characteristics, the slip length is mainly determined by grain size since the grain boundaries (GBs) act as a major obstacle to slip deformation [28]. In general, the decrease in the distance between slip barriers promotes multiple slip and dislocation-dislocation interaction [31]. Therefore, slip reversibility is expected to be enhanced with large grain size improving the FCP resistance, as long as the slip characteristics remains constant [4,21]. When plastic zone is smaller than or equal to grain size, slip may occur on only one slip system and slip reversibility may not be impeded. Twin boundaries (TBs) can also block the movement of dislocations in the same way as GBs. The TBs are known to interact with dislocations differently as compared to GBs, which is beneficial in improving mechanical properties and even thermal stability of metals [5,32,33]. Under cyclic loading condition, for example, TBs have been reported to behave much stronger in resisting fatigue craking than GBs [34].

The effect of SFE and grain size on the near-threshold FCP behavior of high-Mn steels was investigated. Fig. 3 shows the change in $\Delta K_{th}$ values of high-Mn steels as a function of (a) SFE and (b) grain size. The trend observed in Fig. 3(a) suggested that the $\Delta K_{th}$ values of high-Mn austenitic steels in general increased with decreasing SFE. As mentioned previously, the tendency for increasing FCP resistance with decreasing SFE as a result of enhanced slip reversibility has been well established [21]. However, the correlation between $\Delta K_{th}$ values and SFEs of high-Mn austenitic steels was not completely satisfactory. Fig. 3(b) shows that the grain size of high-Mn austenitic steels in this study did not show any reasonable correlation with the $\Delta K_{th}$ values. Even though a number of studies have demonstrated that grain size affects the near-threshold FCP rates of metals, its effect is somewhat controversial with two opposite trends reported [35–40]. While fine grain size can be beneficial in improving the resistance to high cycle fatigue of metals [35,36], the effect of grain size on the FCP rates in Paris’ regime has been observed to be negligible in most studies [37,38]. Conversely, the improved resistance to near-threshold FCP along with a marked increase in $\Delta K_{th}$ values with increasing ferrite grain size has been also reported in a variety of low-strength steels [39,40].

Fig. 3 strongly suggested that, even though SFE and grain size are important parameters in determining the near-threshold FCP behavior of metals by influencing slip reversibility [21,41], they are by no means the only factors that should be considered. Intrinsically Peierls’ type of stress related to cause dislocation movement on active slip plane may, for example, need to be considered other than slip reversibility [42]. The presence of TBs, as well as GBs, may influence such stress under fatigue loading [34,43]. Well-developed slip bands (SBs) formed in the plastic zone ahead of crack can also affect the FCP behavior of high-Mn austenitic steels [44]. In order to understand the factors affecting the FCP behavior of high-Mn steels, the fracture surfaces of FCP-tested specimens were examined. Fig. 4 shows the SEM fractographs of FCP-tested (a) 25Mn, (b) 25Mn0.5Cu, (c) 24Mn4Cr, (d) 22Mn3Cr, (e) 20Mn, (f) 18Mn, (g) 17Mn1.8Al, (h) 16Mn2Al specimens, documented in the near-threshold $\Delta K$ regime. The fracture mode was basically transgranular ductile cleavage mode for all the specimens, whereas the “degree of ductility” imposed on the cleavage facet differed with different alloy systems. The 18Mn and 24Mn4Cr specimens, for example, showed relatively well-developed cleavage facet planes, while rather ductile cleavage mode was shown on the 16Mn2Al and 22Mn3Cr specimens. Such a subtle variation in the “degree of ductility” in fracture mode was most probably due to the difference in slip planarity. Coincidentally, these 16Mn2Al and 22Mn3Cr specimens have the lowest $\Delta K_{th}$ values among the specimens tested. The tendency for such a ductile cleavage occurred in the 16Mn2Al specimen could be explained by low SFE value, as shown in Fig. 3(a). However, the SFE and the grain size of 22Mn3Cr specimen could not explain the low $\Delta K_{th}$ value. Fig. 5 shows the micrographs of FCP-tested (a) 24Mn4Cr and (b) 22Mn3Cr specimens, documented near the crack path in low $\Delta K$ regime. For the 22Mn3Cr specimen, well-developed slip bands were observed in

<table>
<thead>
<tr>
<th>Specimen</th>
<th>Mn</th>
<th>Cr</th>
<th>Al</th>
<th>Cu</th>
<th>Fe</th>
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<tbody>
<tr>
<td>25Mn</td>
<td>24.5</td>
<td></td>
<td></td>
<td></td>
<td>Bal.</td>
</tr>
<tr>
<td>25Mn0.5Cu</td>
<td>24.4</td>
<td>0.01</td>
<td></td>
<td></td>
<td>Bal.</td>
</tr>
<tr>
<td>24Mn4Cr</td>
<td>24</td>
<td>4</td>
<td></td>
<td></td>
<td>Bal.</td>
</tr>
<tr>
<td>22Mn3Cr</td>
<td>22</td>
<td>3</td>
<td></td>
<td></td>
<td>Bal.</td>
</tr>
<tr>
<td>20Mn</td>
<td>20</td>
<td></td>
<td></td>
<td></td>
<td>Bal.</td>
</tr>
<tr>
<td>18Mn</td>
<td>18</td>
<td></td>
<td></td>
<td></td>
<td>Bal.</td>
</tr>
<tr>
<td>17Mn1.8Al</td>
<td>17</td>
<td>1.8</td>
<td></td>
<td></td>
<td>Bal.</td>
</tr>
<tr>
<td>16Mn2Al</td>
<td>15.5</td>
<td>2</td>
<td></td>
<td></td>
<td>Bal.</td>
</tr>
</tbody>
</table>

**Table 1**

The specimen designation and chemical composition of high-Mn steels in this study.
relatively large area along the crack, indicating a considerable amount of plastic deformation in the plastic zone at the tip of crack. Such a homogeneous deformation in 22Mn3Cr specimen as compared to that of 24Mn4Cr specimen would induce large amount of damage accumulation reducing the resistance to FCP. The enhanced homogeneous deformation along the crack path for the 22Mn3Cr specimen as compared to the 24Mn4Cr specimen was confirmed by the EBSD analysis. The inverse pole figure (IPF) maps and Kernel average misorientation (KAM) maps from the EBSD analysis data in the vicinity of crack path for the FCP-tested (a) 24Mn4Cr and (b) 22Mn3Cr specimens are represented in Fig. 6. The different colors in IPF maps are indicative of the orientation of each point. The KAM analysis was conducted by calculating each measurement point up to its second nearest neighbors where the threshold value was 5°. The yellow and green colors represent higher degree of misorientation which was produced by large amount of deformation, whereas blue color indicates lower degree of misorientation in

Fig. 6. The EBSD figures of (a) 25Mn, (b) 25Mn0.5Cu, (c) 24Mn4Cr, (d) 22Mn3Cr, (e) 20Mn, (f) 18Mn, (g) 17Mn1.8Al and (h) 16Mn2Al specimens. (White arrows indicate annealing twins in the microstructure.)
the less deformed areas. It was noted that the KAM near crack path was higher for the 22Mn3Cr specimen with homogeneous deformation encouraged than the 24Mn4Cr specimen. It was therefore suggested that relatively higher strain would be developed for the 22Mn3Cr specimen during fatigue crack propagating with extensive distribution as compared to the 24Mn4Cr specimen.

Fig. 4(d) clearly indicated that enhanced homogeneous deformation at crack tip obviously promoted the “degree of ductility” on the cleavage facet morphology of 22Mn3Cr specimen. Further studies are required to find the reasons, other than SFE and grain size, for such an enhancement in homogeneous deformation for the 22Mn3Cr specimen during fatigue crack propagating with extensive distribution as compared to the 24Mn4Cr specimen.

Despite a great deal of research efforts on the formation of twins during monotonic plastic deformation and their role in determining the tensile properties of high-Mn steels, the effect of twinning on the FCP behavior, particularly in the near-threshold ΔK regime, has not been well established. When considering the influence of twinning on the FCP behavior, existing twins as a result of pre-deformation and mechanical twins possibly formed during fatigue loading need to be separately considered. At present, it has been generally accepted that no new twins are formed in the plastic zone ahead of crack for TWIP steels under fatigue crack propagating condition [16,45]. Even under low cycle fatigue (LCF) condition when cumulative cyclic plastic strain is large, twin nucleation has not been reported [15], except in rare cases when the grains are particularly large or stresses high [46]. Many researchers have agreed that cyclic plastic strain is not sufficient enough to cause twinning in TWIP steels [15,45]. Under normal circumstances, cyclic straining tends to decrease dislocation density and thickens the existing twins [45]. Cyclic plastic deformation in plastic zone under fatigue loading can then be significantly different from plastic behavior under monotonic loading, so that TWIP effect has no essential role in

**Table 2**
The grain size, effective grain size, calculated stacking fault energy and tensile properties of high-Mn steels.

<table>
<thead>
<tr>
<th>Material</th>
<th>Grain size (μm)</th>
<th>Effective grain size (μm)</th>
<th>Temperature (°C)</th>
<th>YS (MPa)</th>
<th>TS (MPa)</th>
<th>TE (%)</th>
<th>SFE (mJ/m²)</th>
</tr>
</thead>
<tbody>
<tr>
<td>25Mn</td>
<td>13</td>
<td>6</td>
<td>25</td>
<td>436</td>
<td>944</td>
<td>55</td>
<td>29</td>
</tr>
<tr>
<td>25MnCu0.5</td>
<td>16</td>
<td>6</td>
<td>25</td>
<td>439</td>
<td>917</td>
<td>53</td>
<td>34</td>
</tr>
<tr>
<td>24Mn4Cr</td>
<td>38</td>
<td>37</td>
<td>25</td>
<td>499</td>
<td>982</td>
<td>72</td>
<td>27</td>
</tr>
<tr>
<td>22Mn3Cr</td>
<td>29</td>
<td>28</td>
<td>25</td>
<td>462</td>
<td>1016</td>
<td>63</td>
<td>29</td>
</tr>
<tr>
<td>20Mn</td>
<td>24</td>
<td>19</td>
<td>25</td>
<td>526</td>
<td>996</td>
<td>55</td>
<td>29</td>
</tr>
<tr>
<td>18Mn</td>
<td>22</td>
<td>20</td>
<td>25</td>
<td>410</td>
<td>1078</td>
<td>68</td>
<td>26</td>
</tr>
<tr>
<td>17Mn1.8Al</td>
<td>13</td>
<td>12</td>
<td>25</td>
<td>487</td>
<td>943</td>
<td>73</td>
<td>41</td>
</tr>
<tr>
<td>16Mn2Al</td>
<td>14</td>
<td>8</td>
<td>25</td>
<td>494</td>
<td>951</td>
<td>72</td>
<td>44</td>
</tr>
</tbody>
</table>

a: YS: yield strength.
b: TS: tensile strength.
c: TE: tensile elongation.
d: SFE: stacking fault energy [9].

**Table 3**
The ΔKth values and the m and C values in Paris’ equation of high-Mn steels.

<table>
<thead>
<tr>
<th>Material</th>
<th>ΔKth (MPa√m)</th>
<th>m</th>
<th>C</th>
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<tr>
<td>25Mn</td>
<td>8.0</td>
<td>3.46</td>
<td>8.9 × 10⁻¹³</td>
</tr>
<tr>
<td>25MnCu0.5</td>
<td>9.0</td>
<td>3.82</td>
<td>2.4 × 10⁻¹³</td>
</tr>
<tr>
<td>24Mn4Cr</td>
<td>9.0</td>
<td>3.53</td>
<td>3.6 × 10⁻¹³</td>
</tr>
<tr>
<td>22Mn3Cr</td>
<td>5.5</td>
<td>2.57</td>
<td>1.6 × 10⁻¹³</td>
</tr>
<tr>
<td>20Mn</td>
<td>8.0</td>
<td>2.39</td>
<td>1.0 × 10⁻¹³</td>
</tr>
<tr>
<td>18Mn</td>
<td>9.8</td>
<td>3.52</td>
<td>4.0 × 10⁻¹³</td>
</tr>
<tr>
<td>17Mn1.8Al</td>
<td>8.0</td>
<td>3.69</td>
<td>5.8 × 10⁻¹³</td>
</tr>
<tr>
<td>16Mn2Al</td>
<td>5.1</td>
<td>3.47</td>
<td>2.0 × 10⁻¹²</td>
</tr>
</tbody>
</table>

### References

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cyclic deformation of TWIP steels. However, the lack of hardening due to twinning may help maintain ductile nature of TWIP steel during cyclic loading [15,45,47], possibly influencing the resistance to FCP in Paris’ regime beneficially.

Only a limited number of studies have been published on the effect of pre-existing twins on the FCP behavior of metals. Singh et al. have argued that the introduction of nanotwins in ultrafine grained copper can improve the \( \Delta K_{th} \) and decrease the near-threshold FCP rates by increasing the back stress from those twins [48]. However, no \( da/dN-\Delta K \) relationship has been explicitly examined on the long crack. Considering that TBs can hinder dislocation movement more effectively than GBs, the presence of TBs may reduce the slip reversibility, reducing the resistance to near-threshold FCP. To identify the effect of pre-existing twins on the FCP behavior of high-Mn steels, FCP test was conducted on the 35% pre-deformed 17Mn1.8Al specimen. Similar studies on the plastic deformation of high-Mn TWIP steels under monotonic loading condition have suggested that a substantial amount of twins are expected to be formed with 35% tensile-deformation [15].

Fig. 4. The SEM fractographs of FCP-tested (a) 25Mn, (b) 25Mn0.5Cu, (c) 24Mn4Cr, (d) 22Mn3Cr, (e) 20Mn, (f) 18Mn, (g) 17Mn1.8Al and (h) 16Mn2Al specimens, documented in the near-threshold \( \Delta K \) regime.
shows the $da/dN-\Delta K$ curves of 17Mn1.8Al specimen, with and without 35% tensile pre-deformation, at an R ratio of 0.1. It was shown that the FCP resistance of 17Mn1.8Al specimen in near-threshold $\Delta K$ regime decreased with 35% tensile pre-deformation. The $\Delta K_{th}$ value of 17Mn1.8Al specimen, for example, decreased from 8.0 to 6.6 MPa$\sqrt{m}$, clearly indicating that pre-existing twins deteriorated the FCP resistance.

Fig. 9 shows the SEM fractographs of FCP-tested 17Mn1.8Al specimen (a) without pre-deformation and (b) with 35% pre-deformation, along with the schematic illustration of GBs and TBs affecting dislocation movement at the tip of crack. Brittle nature of cleavage facets observed for the 17Mn1.8Al specimen without pre-deformation changed into more-or-less ductile cleavage mode with 35% pre-deformation. The present observation strongly suggested that TBs reduces slip reversibility by hindering dislocation movement, decreasing the resistance to near-threshold FCP of high-Mn steels.

The fatigue strength represents a limit value below which short macro crack does not initiate, and the mechanism for the improved fatigue strength of steel with increasing tensile strength, and arguably yield strength, has been well established[4,16,49–52]. However, the effect of tensile properties on the FCP behavior of steel, particularly in low and intermediate $\Delta K$ regime has not been well understood.

Fig. 10 shows the change in $\Delta K_{th}$ values of high-Mn steels as a function of (a) facet size/grain size.
yield strength, (b) tensile strength and (c) elastic modulus. Early studies have shown that the ΔKth values of medium- to high-strength metals increase with increasing material strength [55], while the inverse trend has been also reported for similar alloy systems [52–54]. One of reasons for such a discrepancy may be since in many cases an increase in grain size is accompanied by a concurrent decrease in strength according to the Hall-Petch relationship [53]. Care must be taken when considering the effect of strength on FCP behavior, particularly in the near-threshold ΔK regime. Early studies based on the dislocation dynamics by Mura and co-workers suggested the direct proportionality between ΔKth and yield strength of ferrous alloys [55]. However, the general consensus is that the effect of material strength on the FCP behavior of metals is not significant, if there is any [56]. For example, increasing yield strength of steels by nearly an order of magnitude has been reported to change the FCP rates in Paris’ regime by only a factor of two or three [57]. Conversely, large dependence of material strength has been reported on the ΔKth and the near-threshold FCP rates [52–54]. The decrease in ΔKth values with increasing strength has been reported for some ferritic-pearlitic steels [53]. Fig. 10 showed a weak correlation between yield strength and the ΔKth values of high-Mn steels, where ΔKth values tended to decrease with increasing yield strength. Moreover, tensile strength and elastic modulus of high-Mn steels showed no notable correlation with ΔKth values. The influence of yield strength on the near-threshold FCP behavior is most probably related to the ease of dislocation movement at the tip of crack. Other than this, the dependence of near-threshold FCP resistance on metal’s strength has often been rationalized in terms of environmental argument particularly for steels [58]. Primary mechanism of environmental attack during crack propagation in moist air is considered as hydrogen embrittlement (HE) [58,59]. Under cyclic straining, hydrogen atoms are transported

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**Fig. 8.** The da/dN-ΔK curves of 17Mn1·8Al specimens, with and without 35% tensile pre-deformed, at an R ratio of 0.1 and 298 K.

**Fig. 9.** The SEM fractographs of FCP-tested 17Mn1·8Al specimen (a) without pre-deformation and (b) with 35% pre-deformation, along with the schematic illustration of GBs and TBs affecting dislocation movement at the tip of crack.
by diffusion or dislocation motion ahead of crack tip. The transported hydrogen atoms then trap at various sites, including GBs, TBs, and precipitates, influencing the fatigue damage process, as well as reducing the intrinsic resistance to fracture [58]. The hydrogen atoms trapped on dislocations could also decrease the slip reversibility, increasing the FCP rates [58]. Since HE can reduce the FCP resistance by various mechanisms and high-strength steels are more vulnerable to HE attack, it can be argued that the $\Delta K_{th}$ may decrease with increasing yield and tensile strengths [58]. Even though this argument is legitimate, the strength contribution, particularly tensile strength, on the $\Delta K_{th}$ value of high-Mn steels appeared to be not significant, as shown in Fig. 10(a) and (b). Early studies on fatigue have argued that the FCP rates in Paris’ regime are proportional to the crack tip opening displacement (CTOD) [58,59]. Since elastic modulus is inversely related to the crack tip opening displacement (CTOD), low elastic modulus induces high CTOD increasing FCP rates [60,61]. However, the issue of elastic modulus on the FCP behavior of steels in low $\Delta K$ regime has been largely unexplored. The present study in Fig. 10(c) indicated that the near-threshold FCP behavior of high-Mn steels was not sensitive to the change in elastic modulus. Fig. 11 show (a) the $\Delta K$ value at $10^{-8}$ m/cycles and (b) the slopes of Paris’ equation, $m$, as a function of elastic modulus of high-Mn steels. This figure also suggested that the effect of elastic modulus was nil in influencing the FCP behavior of high-Mn steels in Paris’ regime. The elastic modulus of steels may affect the magnitude of plasticity-induced crack closure which is considered to arise from the elastic constrain within prior plastic zone [60]. Finite element analysis conducted by Fleck, however, has indicated that the crack closure behavior is insensitive to the change in yield strength and elastic modulus [62]. The FCP behavior of high-Mn steels in low and intermediate. The present observation in Figs. 10 and 11 strongly suggested that the FCP behavior of high-Mn steels in both low and intermediate $\Delta K$ regimes has little relationship with monotonic plastic behavior which is greatly influenced by the TWIP effect.

4. Conclusion

1. The $\Delta K_{th}$ values of high-Mn austenitic steels tended to increase with decreasing SFE, even though the correlation appeared to be weak. Unlike the general belief, no notable relationship was observed between grain size and near-threshold FCP behavior of high-Mn steels.
2. Even though SFE and grain size are known to be important parameters in determining $\Delta K_{th}$, the factors like deformation characteristics at crack tip and intrinsic stress activating dislocation motion were believed to be carefully assessed to understand the near-threshold FCP behavior of high-Mn steels.

Fig. 10. The change in $\Delta K_{th}$ values of high-Mn steels as a function of (a) YS, (b) TS and (c) elastic modulus.

Fig. 11. (a) The relationship between elastic modulus of high-Mn steels and (a) $\Delta K$ value at $10^{-8}$ m/cycles and (b) slopes of Paris’ equation, respectively.
3. The 35% pre-deformation reduced the FCP resistance of high-Mn steel in near-threshold ΔK regime as compared to that of specimen without pre-deformation. The fractographic observation suggested that the pre-deformation encouraged ductile, rather than brittle, cleavage fracture mode with the presence of twin boundaries hindering the dislocation movement.

4. The effect of yield strength, tensile strength and elastic modulus appeared to be negligible on the near-threshold FCP behavior of high-Mn steels, showing no reasonable correlation between the ΔKth values and the tensile properties.

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