ABSTRACT

To investigate the failure mechanism of pipelines subjected to mechanical damage, Charpy impact, crack-tip-opening displacement (CTOD) and fatigue-crack growth tests were carried out for six series of line pipe steels with uniaxial plastic prestrain, $\varepsilon_{pr}$. The Charpy absorbed energy and critical CTOD ($\delta_c$) decreased with increasing $|\varepsilon_{pr}|$. In $\delta_c = a \varepsilon_{pr} + \beta$. The derivative, $d\delta_c/d\varepsilon_{pr}$, was dependent on the ductile-to-brittle transition temperature of the steels. In the CTOD tests, the prestrain caused ductile-to-brittle transition for the steels with a higher transition temperature. The effects of the compressive $\varepsilon_{pr}$ on both the reduction of $\delta_c$ and ductile-to-brittle transition were larger than those of the tensile $\varepsilon_{pr}$. The compressive $\varepsilon_{pr}$ accelerated both the fatigue-crack initiation and growth.

INTRODUCTION

Pipelines can suffer mechanical damage caused by outside forces. Previous studies showed that a gouge in a dent is the most typical and also the most severe form in pipelines with respect to burst behavior (Mayfield et al., 1979, Maxey, 1986).

In addition to the burst behavior, the fatigue behavior has also become of great interest because of internal pressure fluctuations. Therefore, Hagiwara et al. (1999) and Hagiwara and Oguchi (1999) studied the fatigue behavior of line pipes having a gouge-in-dent type of mechanical damage. Hagiwara and Oguchi (1999) reported that the fatigue behavior of pipelines depended on the defect geometry and fracture toughness of the steels; when pipes with a lower fracture toughness were deeply dented, low-cycle fatigue occurred associated with initial ductile-crack growth. On the contrary, high-cycle fatigue occurred for pipes with a lower fracture toughness having a shallow damage and also for pipes with a higher fracture toughness. In these cases, the high-cycle fatigue life depended on the defect depth and stress amplitude (Hagiwara et al., 1999).

These results suggested that plastic deformation due to denting would locally reduce fracture toughness at mechanical damage, in particular, for pipes with a lower fracture toughness. It was assumed that the reduction in fracture toughness caused by plastic deformation would depend on the fracture toughness level of the pipes. In addition, the fatigue properties could also be affected by the plastic deformation. Thus the effects of prestrain on the fracture toughness and fatigue properties of line pipes should be clarified to investigate the failure mechanism of damaged pipelines.

Oguchi et al. (2000) showed that denting applied compressive prestrain to pipes at the bottom of the dent, therefore, the effects of both tensile and compressive...
prestrain need to be investigated. If the prestrain dependence of the fracture toughness and fatigue properties for line pipes is available, data necessary for pipeline maintenance, such as fatigue life and the maximum allowable defect size, would be estimated.

The effects of plastic prestrain on the mechanical properties of steels are also of importance for various steel structures which will have been exposed to plastic deformation. Therefore, research on steels has been conducted with respect to the relationship between fracture toughness and compressive prestrain by Inoue et al. (1986), that between low-cycle fatigue properties and prestrain by Iida (1988), and that between crack-tip-opening displacement (CTOD) and prestrain by Hanxing et al. (1992), Homma et al. (1998) and Miki et al. (2000). Although valuable findings have been reported, the effects of the prestrain ranging from tension to compression on fracture toughness and fatigue properties of line pipes are not yet fully understood.

The effects of tensile and compressive prestrain on the fracture toughness and fatigue properties of various line pipes were investigated in the present study. Charpy impact, CTOD and fatigue-crack growth tests were conducted for specimens with prestrain up to 20 percent.

**EXPERIMENTAL**

**Materials**

Base metals of five series of DSAW types (Steels A, B, C, D and E) and a series of ERW type (Steel F) of line pipes were evaluated. Steel C was a thermo-mechanical controlled process (TMCP) type of steel. The remaining ones were the controlled-rolled types. Table 1 summarizes the dimensions and mechanical properties of the pipes. Table 2 shows the chemical compositions of the pipes.

**Application of Prestrain**

**Compressive Prestrain.** Figures 1(a) and (b) illustrate the geometry of test samples for Steels A and B and that for Steels C, D and E, respectively. Prior to compression, a constraint jig was fastened to the samples to prevent its buckling. Strain gages were mounted on the center of the parallel portion of each sample.

Using a universal-type, 245 kN testing machine, a compressive load was applied to each sample along the longitudinal direction of the pipes at room temperature. The crosshead speed of the testing machine was 0.2 mm/min. Each sample was compressed until a predetermined residual prestrain was obtained. After unloading, measurement of the gage length at the center of the parallel portion of each sample allowed us to determine the compressive prestrain by solving Equation 1:

\[
\varepsilon_{pr} = \ln \left( \frac{L}{L_0} \right)
\]

where \(\varepsilon_{pr}\) is the uniaxial residual prestrain, \(L_0\) and \(L\) are the gage lengths before the compression and after removing the load used for the compression, respectively.

The maximum target \(\varepsilon_{pr}\) was -20%. As a -5% prestrain manifested a marked barrel-type deformation peculiar to the compressive tests, the parallel portion of the samples that have undergone the prestrain of \(\varepsilon_{pr}<-5\%\) was smooth grained, and the compressive load was applied again to have a target \(\varepsilon_{pr}\).

Fracture toughness of pretrained steel generally dependent on aging time due to the strain aging effect. Horikawa (1980) reported that pretrained steels heat-treated at 250 °C for 1 hour had almost the same Charpy absorbed energy as those aged at room temperature for 180 days. After the compression, the samples were therefore heat-treated at 250 °C for 1 hour in an electric furnace.

**Tensile Prestrain.** Figure 1(b) illustrates the geometry of the samples for all the steels. Strain gages were mounted on the center of the parallel portion of each sample. Using the universal testing machine, a tensile load was applied to each sample along the longitudinal direction of the pipes at room temperature. The crosshead speed of the testing machine was 0.2 mm/min. After removing the load, \(\varepsilon_{pr}\) was calculated from Equation 1 in the same manner as for the computation of the compressive \(\varepsilon_{pr}\). After tension, the samples were heat-treated at 250 °C for 1 hour.
Charpy Impact Tests

V-notch Charpy impact test specimens (10×10×55 mm, notch depth: 2 mm) were extracted from the center of the parallel portion of the prestrained samples [Figures 1(a) and (b)] of Steels A and B with an ε_p of ±10% and ±20%. Test specimens without prestrain (ε_p=0%) were directly extracted from the pipes longitudinally. A notch was machined in each specimen along the through-wall direction of the pipes which was perpendicular to the direction of the prestrain.

Charpy impact tests were performed at 0 °C and also at room temperature. The absorbed energy and percent brittle fracture were determined for each sample.

CTOD Tests

To perform the CTOD tests in compliance with BS 7448-1 (1991), three-point bending specimens (10×10×50 mm) were extracted from the center of the parallel portion of the prestrained samples [Figure 1(b)] of Steels C, D, E and F. Test specimens without prestrain (ε_p=0%) were directly extracted from the pipes longitudinally. A notch was machined in each specimen along the through-wall direction of the pipes. A fatigue precrack was introduced for each specimen.

A universal-type, 196 kN testing machine was used for the CTOD tests which crosshead speed was 0.25 mm/min. The tests were performed at 0 °C for the prestrain specimens and at temperatures between 0 °C and -100 °C for the specimens without prestrain (ε_p=0%) using ethyl alcohol and dry ice or using isopentane and liquefied nitrogen. The opening displacement of the notch was measured with a clip gauge. The fatigue-crack length, a (mm), was determined with Equation 3 employing the change in the elastic component of the displacement versus the load, P (MPa), that is, compliance (Tada et al., 1985):

\[ V_g/P = [65a + W^2 - B] [0.76 - 2.08X + 3.87X^2 - 2.04Z + 0.66/(1 - X)^2] \] (3)

where W and B are the width and thickness of the specimens (mm), X = a/W, E = 2.1x10^5 MPa, and S = 80 mm is the distance between the fulcrums of the load.

The fatigue-crack growth rate, da/dN (mm/cycle), was obtained from the differentiation of the polynomial approximation of a versus the number of cycles, N. It was preliminarily confirmed that the error from the polynomial approximation was within 0.008 mm of its standard deviation.

RESULTS AND DISCUSSION

Charpy Impact Tests

Figures 2 and 3 show the relationship of ε_p with the yield stress (MPa) of each material without prestrain (ε_p = 0%) at room temperature (Table 1). Note that σ_y was also used for the prestrained samples.

Fatigue-Crack Growth Tests

Three-point bending type specimens were extracted from the center of the parallel portion of the prestrained samples of Steels A and B with an ε_p of ±10% and ±20% in accordance with Figure 1(c). Test specimens without prestrain (ε_p=0%) were directly extracted from the pipes longitudinally. A notch was machined for each specimen along the through-wall direction of the pipes.

Using a 11.1 kN fatigue testing machine, fatigue-crack growth tests were preformed for the specimens at room temperature. Cyclic loads having sinusoidal waves of 6 Hz were applied to each specimen. The maximum and minimum values of the cyclic loads, P_max and P_min, were 568 N and 392 N or 451 N and 392 N, respectively. The opening displacement, V_g (mm), of the notch was measured with a clip gauge. The fatigue-crack length, a (mm), was determined with Equation 3 employing the change in the elastic component of the displacement versus the load, P (MPa), that is, compliance (Tada et al., 1985):

\[ V_g/P = [65a + W^2 - B] [0.76 - 2.08X + 3.87X^2 - 2.04Z + 0.66/(1 - X)^2] \] (3)

where W and B are the width and thickness of the specimens (mm), X = a/W, E = 2.1x10^5 MPa, and S = 80 mm is the distance between the fulcrums of the load.

The fatigue-crack growth rate, da/dN (mm/cycle), was obtained from the differentiation of the polynomial approximation of a versus the number of cycles, N. It was preliminarily confirmed that the error from the polynomial approximation was within 0.008 mm of its standard deviation.
Charpy absorbed energy and percent brittle fracture for Steels A and B, respectively. A large difference was observed in the results between both steels. Figure 2 shows that the Charpy absorbed energy in Steel A with a higher fracture toughness slightly decreased with the increasing tensile $\epsilon_{pr}$, while it steeply decreased with the increasing compressive $\epsilon_{pr}$; thresholds of the compressive $\epsilon_{pr}$ for the reduction of the absorbed energy were observed between $-10\%<\epsilon_{pr}<0\%$ at 0 °C and between $-20\%<\epsilon_{pr}<10\%$ at room temperature.

In response to the thresholds of the compressive $\epsilon_{pr}$, Figure 2 shows a sharp rise in the percent brittle fracture versus $\epsilon_{pr}$ curve for Steel A, which implies a ductile-brittle transition. The critical $\epsilon_{pr}$ for the ductile-brittle transition where the percent brittle fracture, that is, crystallinity, equals 50% was observed at $\epsilon_{pr}=-5\%$ at 0 °C and $\epsilon_{pr}=18\%$ at room temperature. The critical $\epsilon_{pr}$ was largely dependent on the testing temperature; the critical $\epsilon_{pr}$ decreased with the lowering the temperature. This result consisted with that the curves of the absorbed energy and percent brittle fracture versus $\epsilon_{pr}$ at room temperature were both shifted to the compressive side compared with those at 0 °C; $\epsilon_{pr}$ had larger influence on the absorbed energy and percent brittle fracture when the temperature lowered.

On the contrary, Figure 3 shows that $\epsilon_{pr}$ significantly reduced the absorbed energy for Steel B with a lower fracture toughness, irrespective of tension or compression. A peak was observed at $\epsilon_{pr}=0\%$ in both the absorbed energy versus $\epsilon_{pr}$ curves and the percent brittle fracture versus $\epsilon_{pr}$ curves. The thresholds of $\epsilon_{pr}$ were observed for the Charpy absorbed energy between $-10\%<\epsilon_{pr}<0\%$ and between $0\%<\epsilon_{pr}<+10\%$ both at 0 °C and room temperature.

The difference between the results of Steels A and B indicated that the effect of $\epsilon_{pr}$ on the Charpy impact properties depended on the fracture toughness level of the steels; the compressive $\epsilon_{pr}$ reduced the absorbed energy and caused the ductile-to-brittle transition irrespective of the steels. On the contrary, the tensile $\epsilon_{pr}$ reduced the absorbed energy and caused the ductile-to-brittle transition only for Steel B with a lower fracture toughness.

**CTOD Tests**

**Reduction of CTOD by prestrain.** Figure 4 shows the temperature dependence of $\delta_{CTOD}$ for Steels C, D, E and F. Here, $\delta_{CTOD}$ represents $\delta_C$ for the specimens without prestrain ($\epsilon_{pr}=0\%$). When $\epsilon_{pr}=0\%$, a ductile-to-brittle transition observed at $-90$ °C (the ductile-to-brittle transition temperature, $T_{DBT}$) for Steel D; $-50$ °C for Steel E and $-40$ °C for Steel F. On the other hand, no ductile-to-brittle transition was observed for Steel C within the range of the experiments.

Figure 5 shows $\delta_C$ versus $\epsilon_{pr}$ for Steels C, D, E and F. In this figure, the open and solid symbols represent the ductile and brittle (cleavage) fractures, respectively. $\delta_C$ of each steel had a peak at $\epsilon_{pr}=0\%$ and decreased with both increasing tensile and compressive $\epsilon_{pr}$ irrespective of the fracture toughness properties. In particular, $\delta_C$ steeply decreased with the increasing compressive $\epsilon_{pr}$.

Figure 5 indicates that $\ln \delta_C$ decreased almost linearly with increasing $|\epsilon_{pr}|$ in ductile fracture regions for both tension and compression;

$$\ln \delta_C = \alpha \cdot |\epsilon_{pr}| + \beta$$ (4)

where $\alpha$ and $\beta$ are coefficients.

In order to quantitatively evaluate the effect of $\epsilon_{pr}$ on $\delta_C$, $\alpha$ and $\beta$ for each steel were determined. Table 3 summarizes the results of the least square fitting for Equation 4 in the ductile fracture regions of Figure 5.

Table 3 shows that Steels C and D had a relatively smaller $|\alpha|$ ($\approx 0.60 \sim 0.12$). On the other hand, Steels E and F had a relatively larger $|\alpha|$ ($\approx 0.30 \sim 0.99$). This indicated that the tested steels could be roughly classified into two groups: one (Group 1) includes Steels C and D with a lower $T_{DBT}$ and a relatively smaller $|\alpha|$. The other (Group 2) includes Steels E and F with a higher $T_{DBT}$ and a relatively larger $|\alpha|$. Figure 6 shows relationship between $|\alpha|$ and $T_{DBT}$ for Steels C, D, E and F. $|\alpha|$ increased almost linearly with the increasing $T_{DBT}$. Therefore, it was concluded that the reduction of $\delta_C$ by prestrain was dependent on the $T_{DBT}$ level of the steels; the steels with a higher $T_{DBT}$ deteriorated more easily due to prestrain than
the steels with a lower $T_{DBT}$.

Figure 6 shows that $|\alpha|$ for the compressive $\varepsilon_p$ was greater than $|\alpha|$ for the tensile $\varepsilon_p$ for each steel, in particular, for Steel C with a higher $T_{DBT}$. It was therefore concluded that the compressive $\varepsilon_p$ had a larger effect on the reduction of $\partial_c$ than the tensile $\varepsilon_p$ and that difference between the effects of the tensile and compressive $\varepsilon_p$ was significant for steel with a higher $T_{DBT}$.

Figure 7 shows the relationship between $\varepsilon_p$ and the normalized CTOD, that is, $\partial_c/\partial_{c_0}$. The normalized CTOD decreased with the increasing $\varepsilon_p$. In Figure 7, the derivatives of $(\partial_c/\partial_{c_0})/\partial\varepsilon_p$ in the ductile fracture regions due to tension were approximately 0.05 and 0.5 for Groups 1 and 2, respectively. This result showed that the steels with a higher $T_{DBT}$ were 10 times more susceptible to tensile prestrain than the steels with a lower $T_{DBT}$ with respect to deterioration of $\partial_c$.

**Ductile-to-Brittle Transition.** Figure 5 shows that all the specimens of Steels C and D underwent ductile fracture at 0 °C, irrespective of the prestrain level. Concerning Steel E, on the contrary, ductile fracture was observed only when $0 \leq \varepsilon_p \leq +2\%$ and brittle fracture was observed when $+3\% \leq \varepsilon_p \leq +1\%$ at 0 °C; a critical $\varepsilon_p$ for the ductile-to-brittle transition ($\varepsilon_{DBT}$) was identified at $+2\% < \varepsilon_p < +3\%$ and $-1\% < \varepsilon_p < 0\%$. This result indicated that the compressive $\varepsilon_p$ had a greater effect on the ductile-to-brittle transition than the tensile $\varepsilon_p$.

Concerning Steel F, ductile fracture was observed only when $0 \leq \varepsilon_p \leq +2\%$ and brittle fracture was observed when $+2\% \leq \varepsilon_p \leq +$$5\%$ at 0 °C; $\varepsilon_{DBT}$ was identified at $\varepsilon_p = +2\%$ for tension which was in good agreement with the results of Steel E ($+2\% < \varepsilon_p < +3\%$).

Concerning Steel D, the ductile-to-brittle transition was caused only by a $\Delta T$ of 90K but not by the prestrain between $-10.5\% < \varepsilon_p < +18.2\%$. It is inferred from the results that the ductile-to-brittle transition would be caused by prestrain of $\varepsilon_p < -10.5\%$ and $+18.2\% < \varepsilon_p$ for Steel D.

In summary, the steels with a lower $T_{DBT}$ (Group 1) showed no ductile-to-brittle transition by the applied prestrain within the range of the experiments. On the other hand, the steels with a higher $T_{DBT}$ (Group 2) underwent ductile-to-brittle transition due to a few percent of the applied prestrain.

Table 3 and Figure 8 show $\varepsilon_{DBT}$ at 0 °C and $T_{DBT}$ when $\varepsilon_p=0\%$. Figure 8 indicated that prestrain and cooling had a similar effect on the ductile-to-brittle transition; $\varepsilon_{DBT}$ of $+2\%$ and $-0.9\%$ corresponded to a temperature shift ($\Delta T$) of 50K of Steel E and $\varepsilon_{DBT}$ of $+1.9\%$ corresponded to a $\Delta T$ of 40K for Steel F. Figure 8 also indicated that the ductile-to-brittle transition behavior of the steels caused by prestrain was dependent on their $T_{DBT}$ level; $|\varepsilon_{DBT}|$ increased with the decreasing $T_{DBT}$.

**Failure at Mechanical Damage in Pipelines.** Figures 5 and 6 showed that the effects of the plastic prestrain on $\partial_c$ was relatively small for the pipes with a lower $T_{DBT}$ such as Group A. This result consists of the previous result that no initial ductile-crack growth was observed for this type of pipes subjected to severe mechanical damage (Hagiwara and Oguchi, 1999). Therefore, only high-cycle fatigue should be considered for maintenance of this type of pipelines. In addition, the $T_{DBT}$ of this type of steels remained low enough (less than 0°C) for the prestrained samples. Therefore, brittle fracture should not occur in these pipes at service temperatures in Japan even when severe damage is introduced.

On the contrary, the effects of the plastic prestrain on $\partial_c$ was large for the pipes with a higher $T_{DBT}$ such as Group B. This result is also consistent with the previous result that low-cycle fatigue associated with the initialductile-crack growth was observed in this type of pipes subjected to severe mechanical damage (Hagiwara and Oguchi, 1999). The results of the CTOD tests (Figure 5) indicated that the initial ductile-crack growth observed was caused by a reduction of $\partial_c$ due to denting. This suggested that not only a high-cycle but also a low-cycle fatigue associated with initial ductile-crack growth should be taken into account for maintenance of this type of pipelines.

Furthermore, only a few percent of prestrain increased $T_{DBT}$ to 0 °C or more for the pipes with a higher $T_{DBT}$.
(Group 2). This indicated that brittle fracture might occur in region of mechanical damage of these pipes after initial ductile and/or fatigue-crack growth. However, the brittle fracture in region of mechanical damage has never been reported for pipelines in the field. The possibility of brittle fracture after ductile and fatigue-crack growth in region of mechanical damage should therefore be investigated for these pipes using internal pressure fatigue tests.

**Fatigue-Crack Growth Tests**

**Fatigue-Crack Initiation.** Figure 9 shows the relationship between \(\varepsilon_p\) and \(N\) under two types of loading conditions as the fatigue crack grew to 0.1 mm from the notch tip. These \(N\) would nearly equal to the \(N\) for fatigue-crack initiation, that is, fatigue strength, because most of these \(N\) were necessary for the fatigue-crack initiation.

Figure 9 indicated that \(N\) for fatigue-crack initiation of Steel A was the largest at \(\varepsilon_p=0\%\) and decreased almost proportionally to the compressive \(\varepsilon_p\), in the order of 60 to 70% at \(\varepsilon_p=-22\%\). The reduction of the fatigue strength by the tensile \(\varepsilon_p\) was smaller than that by the compressive \(\varepsilon_p\). Furthermore, an almost equal level of \(N\) for fatigue-crack initiation is seen at \(+10\% \leq \varepsilon_p \leq +21\%\); the effect of the tensile \(\varepsilon_p\) on the fatigue strength was saturated at \(\varepsilon_p=+10\%\).

Although a similar trend was also seen for the compressive \(\varepsilon_p\) for both steels, Steel B showed a different result from that of Steel A such that the \(N\) for fatigue-crack initiation at \(\varepsilon_p=+20\%\) increased to the level of \(\varepsilon_p=0\%\).

Oguchi et al. (2000) showed that the mechanical force would apply a maximum compressive \(\varepsilon_p\) of -10% at the tip of a gouge in a dent in pipelines. Considering the present result that \(\varepsilon_p=-10\%\) decreased \(N\) for fatigue-crack initiation by 20% or more irrespective of the steels, a reduction in fatigue strength due to denting should be taken into consideration for a conservative evaluation of the mechanical damage in pipelines.

**Fatigue-Crack Growth.** Figures 10 and 11 illustrate the relationship between \(a\) and \(da/dN\) for Steels A and B, respectively. These figures indicated that \(da/dN\) was the smallest at \(\varepsilon_p=0\%\) and the largest at \(\varepsilon_p=-22\%\) for both steels.

For both steels, the effect of \(\varepsilon_p\) on \(da/dN\) was clearly observed when \(a<9\) mm. On the other hand, the effect of \(\varepsilon_p\) was hardly observed when \(a>9\) mm. Therefore, the difference in the curves in Figures 10 and 11 would result from the difference in \(da/dN\) while \(a<9\) mm.

In order to quantitatively evaluate the effect of \(\varepsilon_p\) on the fatigue-crack growth, the relationship between \(\varepsilon_p\) and the parameters involved in a fatigue-crack growth formula was examined. Here, the following relation was applied which was derived from the fatigue-crack growth formula proposed by Mogami et al. (1990):

\[
da/dN = C (\Delta K (B - K_{max}))^n\]

where \(B\) (MPa \(\sqrt{m}\)), \(C\) (mm/cycle) and \(n\) are coefficients and \(\Delta K = K_{max} - K_{min}\) (MPa \(\sqrt{m}\)). Here, \(K_{max}\) and \(K_{min}\) are stress intensity factors under the maximum and minimum loads, respectively.

A non-linear optimization of the coefficients (\(B\), \(C\), and \(n\)) in Equation 5 for fitting with the experimental results revealed that \(B\) remained almost constant irrespective of \(\varepsilon_p\). \(B\) was thus fixed to its average value of \(B = 53.5\) MPa \(\sqrt{m}\).

Figures 12 and 13 show the relationship between log \(da/dN\) and log \(\Delta K/B - K_{min}\) for Steels A and B, respectively. A straight line was obtained for each \(\varepsilon_p\) level in these figures. \(C\) and \(n\) were then determined from the respective straight lines by the least square fitting. Figures 14 and 15 illustrate the relationship of \(\varepsilon_p\) with \(C\) and \(n\), respectively. Referring to Figure 14, \(C\) was almost constant when \(-22\% \leq \varepsilon_p \leq -10\%\) and decreased at \(\varepsilon_p=-22\%\) for both steels. Figure 15 indicated that \(n\) decreased in proportion to the compressive \(\varepsilon_p\) for both steels. There was a proportional reduction, while Steel B decreased only at \(\varepsilon_p=-22\%\).

These results are applicable to numerical simulations for high-cycle fatigue life predictions. Using the simulation technique proposed by Oguchi et al. (2000) and the present data, the fatigue life of pipelines having properties believed to be representative of the region of a.
gouge in a dent are estimable with high precision.

CONCLUSIONS

1. The effect of ε_pr on the Charpy impact properties depended on the fracture toughness level of the steels. The compressive ε_pr markedly reduced the absorbed energy and caused the ductile-to-brittle transition irrespective of the steels. On the contrary, the tensile ε_pr reduced the absorbed energy and caused the ductile-to-brittle transition only for Steel B with a lower fracture toughness.

2. In δ_c decreased almost linearly with increasing |ε_pr| in the ductile fracture regions for both the tension and compression; In δ_c = α · ε_pr + β. The reduction of δ_c by prestrain (α) was dependent on the T_DBT level of the steels; |α| increased almost linearly with the increasing T_DBT of the steels. The compressive ε_pr had a larger effect on the reduction of δ_c than the tensile ε_pr, in particular, for steel with a higher T_DBT.

3. In the CTOD tests, the ductile-to-brittle transition temperature shift caused by the prestrain was dependent on the T_DBT level of the steels without prestrain; the steels with a lower T_DBT showed no ductile-to-brittle transition at 0°C within the range of the experimental data. On the contrary, the steels with a higher T_DBT underwent a ductile-to-brittle transition at 0°C with a few percent of the applied prestrain. The compressive ε_pr had a greater effect on the ductile-to-brittle transition than the tensile ε_pr.

4. When a fatigue crack was initiated, the fatigue strength decreased proportionally with the increasing compressive ε_pr. On the other hand, the tensile ε_pr lowered the fatigue strength when ε_pr ≥+10%, but was saturated or increased the fatigue strength when ε_pr ≥+10%.

5. The compressive ε_pr of -20% accelerated the fatigue-crack growth rate, in particular, when the crack was relatively short. The dependency on ε_pr of the fatigue-crack growth rate, which was a function of the steels, was quantitatively evaluated by introducing a fatigue-crack growth formula.

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### Table 1 Dimensions and Mechanical Properties of the Line Pipes

<table>
<thead>
<tr>
<th>Pipe</th>
<th>Outer Diameter (mm)</th>
<th>Wall Thickness (mm)</th>
<th>0.2% Yield Stress (MPa)</th>
<th>Tensile Strength (MPa)</th>
<th>Elongation (%)</th>
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<tbody>
<tr>
<td>A</td>
<td>609.0</td>
<td>12.7</td>
<td>356</td>
<td>578</td>
<td>36.9</td>
</tr>
<tr>
<td>B</td>
<td>609.0</td>
<td>12.7</td>
<td>338</td>
<td>581</td>
<td>39.9</td>
</tr>
<tr>
<td>C</td>
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<td>15.1</td>
<td>512</td>
<td>635</td>
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</tr>
<tr>
<td>D</td>
<td>610.0</td>
<td>12.7</td>
<td>396</td>
<td>546</td>
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<tr>
<td>E</td>
<td>762.0</td>
<td>12.9</td>
<td>294</td>
<td>414</td>
<td>38.0</td>
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<tr>
<td>F</td>
<td>267.4</td>
<td>7.8</td>
<td>446</td>
<td>519</td>
<td>29.2</td>
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### Table 2 Chemical Composition of the Line Pipes (wt%)

<table>
<thead>
<tr>
<th>Pipe</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
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<tbody>
<tr>
<td>A</td>
<td>0.10</td>
<td>0.25</td>
<td>1.39</td>
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<td>0.004</td>
</tr>
<tr>
<td>B</td>
<td>0.09</td>
<td>0.20</td>
<td>1.46</td>
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</tr>
<tr>
<td>C</td>
<td>0.07</td>
<td>0.24</td>
<td>1.48</td>
<td>0.008</td>
<td>0.001</td>
</tr>
<tr>
<td>D</td>
<td>0.12</td>
<td>0.33</td>
<td>1.30</td>
<td>0.015</td>
<td>0.010</td>
</tr>
<tr>
<td>E</td>
<td>0.15</td>
<td>0.01</td>
<td>0.46</td>
<td>0.011</td>
<td>0.009</td>
</tr>
<tr>
<td>F</td>
<td>0.14</td>
<td>0.21</td>
<td>0.68</td>
<td>0.019</td>
<td>0.018</td>
</tr>
</tbody>
</table>

### Table 3 Results of the CTOD Tests

<table>
<thead>
<tr>
<th>Pipe</th>
<th>( T_{\text{ini}} ) for ( \varepsilon_p = 0 % ) (°C)</th>
<th>Direction of ( \varepsilon_p )</th>
<th>( \varepsilon_{\text{ini}} ) at 0 °C (%)</th>
<th>( \ln \delta = \alpha \varepsilon_p + \beta )</th>
</tr>
</thead>
<tbody>
<tr>
<td>C</td>
<td>-100</td>
<td>Tension</td>
<td>&gt;+18.1</td>
<td>(-0.061)</td>
</tr>
<tr>
<td>D</td>
<td>-90</td>
<td>Compression</td>
<td>&lt;-10.0</td>
<td>(0.116)</td>
</tr>
<tr>
<td>E</td>
<td>-50</td>
<td>Compression</td>
<td>&lt;-10.5</td>
<td>(0.118)</td>
</tr>
<tr>
<td>F</td>
<td>-40</td>
<td>Tension</td>
<td>+2.0</td>
<td>(-0.299)</td>
</tr>
</tbody>
</table>

\( \alpha = -0.321 \), \( \beta = -1.351 \)
Fig. 1  Dimensions of test specimens

Fig. 2  Effect of prestrain on Charpy Impact properties (Steel A)

Fig. 3  Effect of prestrain on Charpy Impact properties (Steel B)
Fig. 4 Temperature dependence of $\delta_c$

Fig. 5 Prestrain dependence of $\delta_c$

Fig. 6 Transition Temperature versus $|\alpha|$

Fig. 7 Prestrain dependence of dimensionless $\delta_c$

Fig. 8 Transition temperature versus critical prestrain

Fig. 9 Effect of prestrain on the number of cycles to 0.1 mm of fatigue-crack growth
Fig. 10 Relationship between crack length and crack growth rate

Fig. 11 Relationship between crack length and crack growth rate

Fig. 12 Relationship between $\frac{\Delta K}{(B-K_{\text{max}})}$ and crack growth rate

Fig. 13 Relationship between $\frac{\Delta K}{(B-K_{\text{max}})}$ and crack growth rate

Fig. 14 Effect of prestrain on the coefficient $C$

Fig. 15 Effect of prestrain on the coefficient $n$