Creep Behavior at the Intercritical HAZ of a 1.25Cr–0.5Mo Steel

Shimpei FUJIBAYASHI and Takao ENDO

1) Faculty of Engineering, Yokohama National University, Tokiwadai, Hodogaya-ku, Yokohama 240-8501 Japan.

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Nowadays the preferential creep damage accumulation at the Intercritical HAZ (ICZ) leading to Type IV cracking has been a great concern for various industries. The ultimate failure of the welded components fabricated from ferritic steels often takes place at this particular region. Type IV cracking has been found in almost all the ferritic steel weldments so far, from a conventional 1.25Cr–0.5Mo steel to a modified 9Cr–1Mo steel. However, the mechanism of Type IV cracking has not yet been understood equivocally. In the present work, cross-weld creep behavior of a service exposed 1.25Cr–0.5Mo steel has been examined in order to clear the feature of Type IV damage. The discussion shall be made on the important role of grain boundaries around small grains, which was transformed into austenite during welding, to promote Type IV cracking. The evident feature of grain boundary facets suggests strongly that Type IV cracking is induced by the grain boundary sliding around small grains. Significant impurity segregation, which is expected to accelerate the damage development by stabilizing cavities, was found at grain boundaries.

KEY WORDS: Type IV cracking; 1.25Cr–0.5Mo steel; grain boundary sliding; multiaxial stress state; impurity.

1. Introduction

It has been commonly recognized that creep damage often appears at the weldment in actual components operated under the creep regime. Types of the creep failure are usually categorized into four groups based upon the location of weldment. The failure location is determined by various factors such as compositions of alloying elements, welding procedure, impurities and operating conditions. Among these four types, Type IV cracking, which is the preferential creep damage at the Intercritical HAZ (ICZ), has been a serious threat to existing high temperature components since the ultimate failure mode of ferritic welds often takes place at this particular region. Type IV cracking has been found in almost all the ferritic steel weldments so far, from a conventional 1.25Cr–0.5Mo steel to a modified 9Cr–1Mo steel. However, the mechanism of Type IV cracking has not yet been understood equivocally. In the present work, cross-weld creep behavior of a service exposed 1.25Cr–0.5Mo steel has been examined in order to clear the feature of Type IV damage. The discussion shall be made on the important role of grain boundaries around small grains, which was transformed into austenite during welding, to promote Type IV cracking. The evident feature of grain boundary facets suggests strongly that Type IV cracking is induced by the grain boundary sliding around small grains. Significant impurity segregation, which is expected to accelerate the damage development by stabilizing cavities, was found at grain boundaries.

2. Experimental Details

2.1. Materials

For the present work, three circumferential butt welds connecting a pipe with a flange were removed from inlet piping to a reactor in a refining plant shown in Fig. 1. The welds had been operated for twenty three years (184 000 h) at 500°C and internal pressure of 1.2–2.5 MPa. An external diameter is 660 mm and nominal thickness is 22 mm. They cracking will be discussed.
were allocated cast references of FJA, FJB and FJC.

A weld preparation was 60°V and Tungsten Inert Gas (TIG) welding was applied to a root run and the remaining was filled with Manual Metal Arc (MMA) welding. Post Welding Heat Treatment (PWHT) was performed at 700°C for 2 h.

2.2. Chemical Compositions and Creep Embrittlement Factor

Chemical compositions of sampled casts are shown in Table 1. Parent materials on the flange side, which exhibited the grain boundary damage at HAZ, contain higher tramp elements such as phosphorus (P), tin (Sn), arsenic (As) and antimony (Sb) than pipe parents which showed no grain boundary damage. Takamatsu et al. 4) correlated the impurity contents using the following factor with creep ductility at Coarse Grained HAZ (CGZ) of a 1.25Cr–0.5Mo steel. This factor, termed Creep Embrittlement Factor (C.E.F), was originally proposed by King 5) to assess the susceptibility to stress relief cracking for a 0.5Cr–0.5Mo–0.25V steel.

\[
\text{C.E.F} = \frac{P + 3.57\text{Sn} + 8.16\text{Sb} + 2.43\text{As}}{1100} \quad (1)
\]

where P, Sn, Sb and As are in wt%.

Although the threshold value to assess the likeliness of premature cracking at CGZ has not been derived, C.E.F values of the cracked reactors fabricated from a 1.25Cr–0.5Mo steel found in relevant papers 6,7) and those of damaged casts in the present work are higher than 0.15. The fact that higher damage at flange CGZs of FJA and FJC is consistent with higher concentrations of impurities in the material on the flange side as discussed in the following section.

2.3. Damage Distribution of a Service Exposed Welds

The grain boundary damage was examined by replication tests although no defect was detected by Wet Fluorecent Magnetic Tests (WFMT) when three welds were sampled. A large difference in damage level was observed among these three casts in spite that they had been closely located during operation. The difference was also observed in the susceptibility to grain boundary cavitation between pipes produced from plate materials and forged flanges. The Heat Affected Zone (HAZ) and a parent material on the pipe side were damage free in all the casts. Most severely damaged weldment in FJA was associated with Transverse Weld Metal Cracking (TWMC), Type III cracking, Type IV damage and isolated cavitation at the parent material which was located adjacent to HAZ on the flange side. The damage observed at the Coarse Grained HAZ (CGZ) and weld metal is shown in Figs. 2 and 3, respectively. Multidirectional grain boundary damage and TWMC, which are the representative feature of stress relief damage, 8) can be seen. From the shape and high hardness of weld metal (240 Hv) in most heavily damaged FJA, it is presumed that the weldment of FJA was repaired at the fabrication with unsuitable PWHT. Higher residual stress and brittle microstructure associated with lower temperature for PWHT should be attributed to TWMC and multidirectional cracking at CGZ rather than service induced stresses. In fact, the damage of weldment for the counter flange of FJB which should have been subjected to the similar condition in terms of stress and temperature was significantly milder. Only isolated cavities were detected at CGZ.

In this paper, a moderately damaged cast FJC that contains micro-cracking at CGZ and isolated cavitation at the ICZ has been mainly examined except the grain boundary facet analyses using the interrupted crept specimen taken from cast FJA. The direction of CGZ damage changes along the depth direction. At the external surface, a main cracking direction is perpendicular to a weld line. But inside the wall thickness, localized cracking which is parallel to a welding direction is observed. These cracks tend to be present at the area termed “cusp region” where the shape
of weld metal discontinues. The extent of creep damage
at the ICZ is diagnosed as "isolated cavities" based upon
Neubauer's classification.9)

Hardness distribution at the weldment measured in the
middle and the bottom of wall thickness for cast FJC is
shown in Fig. 4. It is noteworthy that hardness in the mid-
dle of the wall thickness at the both ICZs of a flange and
pipe is higher than that of parent materials. Hardness at the
bottom (near the root run) where TIG was applied is quite
uniform due to the subsequent heat input during welding,
and little grain boundary damage was detected in this re-

2.4. Specimen Preparation and Testing

For homogenous microstructure of both parent materials
and weld metal, a standardized round bar specimen with
a 6 mm diameter and 30 mm gauge length has been em-
ployed.

Two types of specimens were prepared to examine the
cross-weld creep behavior. One is a 14 mm × 14 mm square
cross-weld specimen to observe the grain boundary damage
at interruptions during creep tests. The other is a round bar
specimen with helical vee notches applied to all the mi-
crostructures constituting the weldment. The geometry of a
spirally notched cross-weld specimen is shown in Fig. 5. It
was referred to the work by McLauglin et al.10) The length
of a screwed part is 40 mm in 50 mm gauge length. The
elastic stress concentration factor, calculated by Finite
Element Analysis, is approximately 2.7. It is a typical value
expected of a discontinuity in a pressure vessel. The area
ratio of a weld metal in gauge length for a square and spi-

rally notched specimen is approximately 0.19 and 0.34, re-

spectively.

Creep and creep rupture tests were conducted with a
constant load technique in air. The damage development
during the test for a cross-weld specimen was observed
using a replication technique. In order to quantify the dam-
age at ICZ, Cavity Density (CD) counted by an optical mi-
roscope with a 450 times magnification was employed.

The cavity nucleus sites at the ICZ was examined using
an interrupted specimen broken in vacuum at the tempera-
ture of −135°C. Grain boundary facets were observed by
Scanning Electron Microscopy (SEM), and particles inside
cavities and segregated elements on the grain boundary
facets and cavity walls were analyzed by Auger Electron
Spectroscopy (AES). The system is a Perkin-Elmer/
Physical Electronics (model PHI595) fitted with a field
emission electron gun. The sample was analyzed using a
3 keV primary electron beam, beam current=8 nA and a
beam size of 400 nm. All the AES analyses were completed
within 4 h from fracturing and 3×10⁻¹⁰ Pa of vacuum was
maintained during observation.

3. Results

3.1. Creep Rupture Properties

Results of creep rupture tests for homogeneous speci-
mens (flange, pipe and weld metal), cross-weld square and
spirally notched specimens taken from cast FJC are plotted
using Manson–Haferd Parameter in Fig. 6.3) The tabulated
values on creep properties were shown in the previous
paper.3) Time to rupture for all types of specimen mostly
lies above the lower boundary of virgin materials (mean
strength −3σ which is based upon NRIM 21B11) for a
Normalized and Tempered 1.25Cr–0.5Mo steel plate) at the
stresses of 30 to 125 MPa. Summaries of testing results for
each type of specimen are described below.

3.1.1. Homogeneous Specimens

The creep lives of a flange material are approximately
three to four times longer than those of a pipe. A flange
parent shows higher creep strength than the mean value of
NRIM data at the stresses lower than 100 MPa, to which the
actual high temperature components are likely to be sub-

jected.

A flange parent produced by forging shows higher level
of bainitic transformation products. It can be presumed that
a pipe parent, which is composed of ferrite and dispersed
carbides, was more equilibrium microstructure associated
with slow cooling at the plate fabrication. Higher creep
strength of a flange parent compared with a pipe should be
attributed to higher bainite ratio in microstructure.

3.1.2. Cross-weld Square Specimens

As noted in Sec. 2.3, the weldment has got service-in-
duced Type IV damage. Cavity Density (CD) before tests
approximately ranged from 200 to 300 (n/mm²). These pre-
existing cavities at the flange ICZ has not significantly
shortened the rupture life in this region. Most failures asso-
ciated with significant grain deformation have taken place
at the pipe parent despite damage development at the flange ICZ during tests. Two specimens failed at the HAZ on the flange side. One failed with the mixed mode, with Type III at the top surface and Type IV in the middle. The other one was broken predominantly with a Type III manner. However, in this specimen, Type IV damage to the level of micro-cracking was found in the middle of wall thickness by sectioning the failed specimen. The ICZ damage at the external surface was significantly less serious. 2) The same tendency was reported by Parker et al. 12) Heavier damage inside the wall thickness suggests that the Type IV damage is accelerated under multiaxial stress state. Thus, the life assessment using the replication technique could lead to non-conservative conclusion.

3.1.3. Spirally Notched Specimens
To increase the triaxiality to all the microstructures constituting the weldment, creep rupture tests were conducted using spirally notched cross-weld specimens. By applying notches, ultimate failures with the Type IV manner were promoted despite relatively short duration of tests. The crack penetration preferentially through the ICZ was observed in these specimens. Time to rupture for all the specimens tested at the nominal stress of 60 MPa is shown in Fig. 7.3) As for homogeneous parent materials, the life prediction by extrapolating the results gained at higher temperatures can be achieved with reasonable accuracy. In the case of cross-weld specimens, the failed position of a cross-weld specimen depends upon the testing temperatures (or testing duration). Therefore, the iso-stress technique is not necessarily available to assess the remnant life of the weldment. At the temperature of 650°C, the final failure of notched specimen is caused by transgranular rupture at a pipe parent associated with significant grain deformation. But the occurrence of cracking at the flange ICZ from the notch root was observed. In this case, time to failure is extended from the lower boundary to the mean value in NRIM 21B due to notch strengthening effect of the ductile pipe parent. At lower temperatures, failure location changes from the pipe parent to the ICZ on the flange side. Time to rupture at 630°C is approximately the same as that of the mean value of virgin materials and it decreases to the lower boundary at the temperature of 610°C.

Fig. 6. Manson–Haferd Parameter–stress correlation for the service-exposed 1.25Cr–0.5Mo steels. The ultimate failure mode of square cross-weld specimens is transgranular failure at pipe parent otherwise denoted. Meanwhile, that for most spirally notched cross-weld specimens is Type IV on the flange side otherwise denoted.

Fig. 7. Iso-stress behavior of homogeneous and cross-weld specimens at 60MPa.
3.2. Damage Development with Life Consumption

A replication technique has been widely utilized to assess the condition of high temperature components in actual plants. Walker et al.\(^{13}\) examined the development of grain boundary damage at ICZ for a 0.5Cr–0.5Mo–0.25V steel weldment. In their work, fairly good correlation was found between life consumption and cavity density.

The Cavity Density (CD) in the present study during and after the creep-rupture tests, is shown in Fig. 8 as a function of the life consumption. CD for a square cross-weld specimen was measured at the position most densely cavitated in the side surface and that for a spirally notched specimen was done by sectioning a specimen. CD at \(t/t_r^1\) must be lower than that shortly before the failure due to the coalescence of cavities.

The large difference can be seen in the value and increase rate of CD between the flange ICZ and pipe ICZ. CD on the flange side increases with the life consumption. In contrast, cavitation is not generated until the final stage of life on the pipe side in a square specimen although creep strength of a pipe parent is significantly lower than that of a flange parent. CD at the both ICZs in spirally notched specimens tends to be higher than that of a square specimen, suggesting that cavitation will be accelerated under the multiaxial stress state. It should be noted that no evident grain boundary damage has been found at a weld metal.

3.3. Cavity Nucleus Sites

To observe the grain boundary damage, cross-weld specimens taken from Cast FJA and FJC were sectioned at interruptions or failures. In Fig. 9, the cross-section adjacent to the final crack path is shown. The grain boundaries at the ICZ on the flange side is damaged heterogeneously. Cavitation takes place preferentially at the triple points of grain boundaries for relatively small grains. All the grain sizes in the area shown in Fig. 9 (7.2 \(\times 10^4\) \(\mu\)m\(^2\)) and the sizes of the smallest grains associated with cavities are shown in Fig. 10. Ninety percent of cavities are generated at the grain boundaries around the grains whose diameters are 10 \(\mu\)m and smaller. The similar tendency is observed at the ICZ of a pipe which is significantly less susceptible to cavitation. The feature of the pipe ICZ is shown in Fig. 11. Colonies of small grains which suffer from intergranular damage are isolated by larger grains with fewer cavities. In the case of the ICZ generated on the bainitic flange, small grains are continuously distributed. The discrete feature of small grain colonies might be the reason for lower sensitivity to Type IV cracking on the pipe side.
3.4. Grain Boundary Facets of the Crept ICZ

It is not easy to break the crept specimen at the ICZ with an intergranular manner. The authors have succeeded in exposing the grain boundary facets only at the flange ICZ probably because the pipe ICZ is less damaged at grain boundaries. The fracture surface of an interrupted creep specimen taken from FJA is shown in Fig. 12, which was obtained by SEM. Most of sub-micron particles appearing on grain boundaries are carbides, and evident damage cannot be seen at the particles and around them. Most cavities are present preferentially at the edges of grains (triple points), and few cavities can be seen at the grain boundaries normal to the tensile axis. In contrast, the cracking is seen along the grain boundaries inclined to the tensile direction. Unlike the Type III damage appearing at the coarse grained HAZ, these cavities are not necessarily associated with particles such as sulfides or carbides, suggesting that cavities are not produced by decohesion between non-wetting particles and matrix. (More than half of cavities are empty inside.) It is to be noted that the shapes of cavities are not spherical but elongated polygonal tubes whose directions are not consistent with that of the maximum stress especially for relatively large cavities. These geometric characteristics are another evidence that cavities are produced by grain boundary sliding.

3.5. Impurity Segregation

Grain boundary facets of the crept FJA, which is shown in Fig. 12, were examined by Auger Electron Spectroscopy. Typical Auger spectrum obtained from a grain boundary is shown in Fig. 13. Significant enrichment of Sb and Sn was found both at grain boundaries without cavities and cavity walls. The maximum extent of segregation reaches ten thousands times for Sb and two hundreds times for Sn by quantifying the AES data using the following equation,

\[
C_i(\text{at}%) = \frac{I_i \times S_i^{-1}}{\sum I_j \times S_j^{-1}} \times 100 \quad \text{.................(2)}
\]

where \(C\) is the concentration of detected element, \(I\) is the peak to peak intensity and \(S\) is the relative sensitivity factor.

4. Discussion

4.1. The Important Role of Grain Boundaries around Small Grains to Promote Type IV Cracking

As mentioned so far, preferential cavity formation was found at the grain boundaries around small grains. It can be presumed that the presence of small grains is due to the transformation into austenite during welding. On cooling from \(\alpha/\gamma\) transformation temperature, these grains become martensite or bainite because of relatively rapid temperature drop as observed by Parker et al.\(^{12}\) They measured the temperature cycle during MMA welding using embedded thermocouples near the weld preparation. The cooling rate from 800 to 500°C ranged from 7 to 17°C/sec. Although cooling rate depends upon various factors such as a size of a component to be welded, heat input and preheat temperatures, hardened grains associated with the transformation followed by rapid cooling could be generated at actual welding. The hardness of small grains measured using a micro-Vickers hardness tester (10 g) was approximately Hv40 higher than that of larger grains. In order to accommodate the strain at the ICZ, which is composed of grains with the different creep resistance, grain boundary sliding will be inevitable at the grain boundary between grains with different creep strength. Parker et al.\(^{14}\) pointed out that the susceptibility to Type IV cracking for a 1.25Cr–0.5Mo steel was largely decreased by raising the temperature of PWHT. It can be interpreted that the effect of PWHT at higher tem-
temperatures is to make creep properties of grains at the ICZ more uniform to prevent the grain boundary sliding around small and strong grains. Furthermore, Kimmins et al.\textsuperscript{11} pointed out that the stress exponent \( n \) for the power law relationship between time to rupture \( t_r \) and stress \( \sigma_t = B \sigma^n \) was 2 to 3 when failed with a Type IV manner. These stress exponents coincide with those for grain boundary sliding of 2 to 3. In addition, significant enrichment of antimony (Sb) and tin (Sn) was revealed both at grain boundaries and cavity walls. It is significantly higher than that expected from the equilibrium segregation during the creep test. Considering the difference in solubility of impurities in austenite, the extent of segregation at the borders of transformed grains could be quite high as described below. Tamaki et al.\textsuperscript{15} proved the impurity segregation at grain boundaries during \( \alpha/\gamma \) transformation associated with welding for a 1.25Cr–0.5Mo steel experimentally and theoretically. The grain boundary phosphorus concentration is significantly higher than an equilibrium value, and this tendency was accelerated by raising a heating rate. These impurities can enhance the cavity formation by reducing the surface energy \( \gamma_s \), and it results in a reduction of the critical cavity radius \( r_c \) given by the following equation,

\[
r_c = 2 \gamma_s / \sigma
\]

where \( \sigma \) is the localized stress perpendicular to grain boundary. When the size of a newly formed cavity is smaller than \( r_c \), it will be sintered.

4.2. Multiaxial Stress State

The effect of multiaxial stress state upon the promotion of Type IV cracking has been proved by experiments using notched specimens and heavier damage inside the wall thickness in the present work. A couple of concepts\textsuperscript{16–18)} has been proposed to describe the creep behavior under the multiaxial loading active on the actual components. In the followings, an equation proposed by Hayhurst\textsuperscript{16} shall be used for further discussion. The equivalent rupture stress \( \sigma_e \), which represents rupture life under the multiaxial condition, is given by the equation described below,

\[
\sigma_e = (\alpha \sigma_1 + (1 - \alpha) \sigma_2)
\]

where \( \sigma_1 \) is the maximum principal stress, \( \sigma_2 \) is equivalent stress (von Mises stress) and \( \alpha \) is the parameter which determines the role of each stress component (\( \sigma_1 \) and \( \sigma_2 \)).

It has been recognized that the maximum principal stress is operative to generate the grain boundary cavities\textsuperscript{19,20} On the other hand, strain is uniquely determined by the equivalent stress. To examine the operative stresses for Type IV failure, these values at the loading \( t = 0 \) h and the steady state \( t = 400 \) h were derived from Finite Element Analysis (FEA). For the sake of simplicity, Norton’s law was employed for the description of inelastic behavior at the condition of 650°C and 40 Mpa. The calculated distribution of the equivalent and maximum principal stress from the notch root \( (r/a = 1) \) to the center of specimen \( (r/a = 0) \) is shown in Figs. 14 and 15, respectively. In these figures, stresses of both parent materials with varying creep exponents \( n = 9.0 \) for a pipe parent and \( n = 6.8 \) for a flange parent) are plotted. At the steady state condition, the equivalent stress is reduced to approximately 80% of the nominal stress. A ductile intragranular fracture mode of a pipe parent suggests that the equivalent stress is operative for this microstructure in the present testing conditions. In the case of square cross-weld specimens, the final failure tends to happen at the pipe parent. Therefore, reduction of equivalent stress by notches would contribute to extending the life of a pipe parent as shown in Fig. 7. Meanwhile, the maximum principal stress at the steady state is approximately 70% higher than the nominal stress at the notch root region where heavy damage was often found. The failure at the ICZ shows a brittle feature with little grain deformation as shown in Fig. 9. From the characteristics discussed above, the conclusion that the maximum principal stress is more operative to develop Type IV cracking will be derived.

However, it might contradict the observation of grain boundary facets discussed in Sec. 3.4 since grain boundary sliding is strain controlled in general. The possible interpretation on this contradiction might be that higher principal stress dominates the damage development after the stress re-distribution associated with grain boundary sliding as proposed by Nix et al.\textsuperscript{21} They have succeeded in predicting the rupture lives for various materials, except aluminum predominantly depending upon equivalent stress and copper

![Fig. 14. Cross-sectional distribution of equivalent stress. \( a \)=radius of testpiece at notch plane, \( r \)=radial coordinate at notch plane.](image-url)
alloys predominantly depending upon the maximum principal stress, under the multiaxial stress state by introducing the concept of principal facet stress. The principal facet stress, which is described by the following equation, is based upon the assumption that grain boundary sliding takes place prior to the onset of cavitation.

$$\sigma_3 = 2.24 \sigma_1 - 0.62(\sigma_2 + \sigma_3) \ldots \ldots \ldots \ldots \ldots (5)$$

where $\sigma_1 > \sigma_2 > \sigma_3$ are principal stresses.

From the significant difference in damage level at the grain boundary between perpendicular grain boundaries and inclined ones against the maximum principal stress, they assumed that the stress should be re-distributed by grain boundary sliding. After the sliding of inclined grain boundaries induced by creep strain, perpendicular grain boundaries will be exposed to higher principal stress. From the relatively dormant increase in CD with life consumption shown in Fig. 8 and heterogeneous feature of grain boundary damage shown in Figs. 9, 11 and 12, it is likely that the abrupt multiplication of cavities will occur at the final stage of life after the grain boundary sliding and resultant increase in the maximum principal stress. Prior to the final stage, equivalent stress will play the major role to cause grain boundary sliding.

4.3. Limitation of the Replication as a Life Assessment Technique

A small increase in CDs at 50% of creep life from those before tests means the occurrence of accelerated cavitation at the final stage of creep. This tendency also suggests that the cavity density measurement in situ should be done very carefully because small cavities found at the early stage of creep damage can be easily smeared by inappropriate preparations. In addition, the stress distribution should be contemplated when applying this methodology. Under the uniformly stressed weldment like the longitudinal seam welds in pressure vessels, higher damage could take place inside the wall thickness at the uniformly stressed welds like a seam weld of a pipe could be the case.

5. Conclusions

(1) Preferential cavitation leading to Type IV cracking takes place at the grain boundaries around small grains. It is presumed that small grains are the products of the $\alpha'/$ transformation during welding and possesses higher creep resistance than original grains. Grain boundary sliding around stronger grains will be inevitable to associate creep strain. The evident feature of grain boundary sliding was found in the crept Intercritical HAZ.

(2) Significant enrichment of tramp elements such as antimony and tin was found at the grain boundaries and inside the cavities. These impurities are expected to accelerate the damage development by stabilizing cavities.

(3) Despite short testing duration, the Type IV failure of a 1.25Cr–0.5Mo steel weldment was generated due to the notch weakening effect of ICZ. The different roles of equivalent stress causing the grain boundary sliding and the maximum principal stress causing the cavity generation are suggested.

(4) The ultimate failure position dose not necessarily depend upon the creep strength of parent materials when loaded with a cross-weld manner.

(5) The life assessment based upon Cavity Density should be done very carefully because of its abruptly increasing tendency at the final stage. In addition, heavier damage inside the wall thickness at the uniformly stressed welds like a seam weld of a pipe could be the case.

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REFERENCES