# EFFECT OF METALLURGY ON STRESS CORROSION CRACKING AND HYDROGEN EMBRITTLEMENT OF ULTRA/HIGH STRENGTH STEELS

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## INTRODUCTION

The scope of application of ultra-high strength steels tends to be limited by their brittle fracture resistance, which apart from the fracture toughness [1,2] is also substantially dependent on their susceptibility to premature and delayed fractures [3]. A detailed analysis of their causes has shown that delayed fractures in ultra-high strength steels are related not only to the specific effects of the martensitic phase transformation, but also to the influence of metallurgical factors [3,4,5]. Furthermore, the resistance to delayed fracture is also dependent on the environment, since e.g. the presence of hydrogen greatly reduces the threshold stress level [6].

This paper reports on an investigation into the susceptibility of ultra-high strength steels, with a structure of low-tempered martensite, to brittle failure in an aggressive environment. In this study, the authors also examined the way this susceptibility is affected by various modes of heat treatment, and by various metallurgical parameters, especially by various oxygen contents of the steel.

## EXPERIMENTAL WORK

The research was conducted on five different steels: a Fe-Ni-C steel, further referred to as steel A; two multiple alloyed steels of the MnSiMo and MnSiCrNiMo types, designated B and C respectively; and two different melts of MnSiNiMo steel, marked $D_1$ and $D_2$ respectively. The chemical analyses of all these materials are listed in Table 1. Melts $D_1$ and $D_2$ differed mainly by their oxygen contents, which were 0.0040% and 0.0078% respectively. Steel A was produced in a laboratory furnace, all the other grades were taken from routine production melts of open-hearth steels.

The steel A specimens were heat treated by two successive anneals at 1050°C followed by water cooling, which left a fully austenitic structure. They were then quenched by 30 minutes at -196°C, which produced an average of 94% of martensite in the structure, and tested after room-temperature ageing.

The specimens of steels B, $D_1$ and $D_2$ were taken from plates 14 mm thick which had been heat treated at 920°C with subsequent water quenching and tempering at 200°C for 4 hours followed by cooling in air. The steel C specimens were taken from plates 16 mm thick which had undergone either of two alternative heat treatments. The conventional heat treatment, further denoted CT, consisted of water quenching from 900°C and tempering

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at 200°C for 4 hours. The alternative was high-temperature thermo-mechanical treatment, further designated HTMT; this involved a 70% deformation at about 900°C, followed by water quenching in a quenching press within 15 seconds after final pass [1,7], followed by tempering at 200°C for 4 hours with subsequent air cooling. The final structures of the C, $D_1$ and $D_2$ specimens then consisted of low-tempered martensite. Only steel B was after its quenching found to contain a small proportion of lower bainite in its structure.

The stress corrosion cracking was studied on flat single edge notch tensile specimens measuring $((4\div10)x25x250).10^{-3}$m. A fatigue crack was induced at the tip of the notch at approximately 0.5 $K_{Ic}$, and the initial value of the stress intensity factor $K_{Ii}$ was plotted against the time to fracture $\tau$. Moreover, in selected cases a compliance technique was applied to investigate the kinetics of stable crack growth, in terms of the dependence of the stable crack growth rate v $(\Delta a/\Delta t)$ upon the instantaneous magnitude of $K_I$ [8,9]. The $K_{ISCC}$ values were established in a 3.5% aqueous solution of NaCl, the $K_{ISH}$ testing was performed while the specimens were electrolytically hydrogen charged in a 1 M HCl + 0.1 M $N_2H_4$ solution at a current density of $2.10^2 Am^{-2}$ [10].

## RESULTS

Table 2 lists the basic mechanical properties of the steels under investigation, along with their $K_{Ic}$, $K_{ISCC}$ and $K_{ISH}$ values.

Figure 1 shows the $K_{Ii}$ versus $\tau$ dependences established in steel A during the stress corrosion cracking tests and during the investigation of the effects of hydrogen embrittlement, including the relevant threshold levels of $K_{ISCC}$ and $K_{ISH}$. Figures 2 and 3 present the $K_{Ii}$ versus $\tau$ plots ascertained in the corrosive and the hydrogen charging solutions on steel B, and on the CT and HTMT specimens of steel C, respectively. It is worth noting that in the HTMT specimens, the threshold level of $K_{ISCC}$ is about 30% higher, and that of $K_{ISH}$ almost twice as high, as in the CT material (Table 2).

Figure 4 shows the $K_{Ii}$ versus $\tau$ dependences for the $D_1$ and $D_2$ specimens in the 3.5% aqueous solution of NaCl. Figure 5 indicates how, in the $D_1$ and $D_2$ specimens, the stable crack growth rate v varies with the instantaneous value of $K_I$. At $K_I$ = 35 MPa.m$^{1/2}$, this rate is greater roughly by one order of magnitude in the $D_2$ than in the $D_1$ material.

Fractographic analyses revealed that after exposure in the 3.5% NaCl solution, the fractures had been prevalently intercrystalline, but that transcrystalline cleavage predominated in the specimens which had fractured in the course of their hydrogen charging [4]. These fracture surfaces were much finer on the HTMT than on the CT specimens.

## DISCUSSION

Figures 1 and 2 prove that both the environments to which the specimens were exposed had caused a pronounced drop in the values of the stress intensity factor $K_{Ic}$ of the investigated steels. In steel A, the threshold level of $K_{ISCC}$ is on average 60% lower, and that of $K_{ISH}$ some 70% lower, than the relevant $K_{Ic}$ value. In steel B, the $K_{ISCC}$ threshold is on average 73% lower and the $K_{ISH}$ level approximately 80% lower than $K_{Ic}$.





Similar relationships were also ascertained in both the CT and the HTMT specimens of steel C, although it is clear from Table 2 that in the HTMT material the $K_{ISCC}$ and $K_{ISH}$ threshold levels diminished less, in relation to the $K_{Ic}$ value, than in the CT material.

Figure 3 demonstrates that HTMT produced higher $K_{ISCC}$ and $K_{ISH}$ values than CT, even though the strength properties of the HTMT material were superior by about 15% to those of the material CT. This finding is vitally important for the practical application of ultra-high strength steels, because it confirms the beneficial effect of HTMT even in the presence of a detrimental hydrogen-containing environment. It has already been shown [1,2,3,8] that the favourable effect of HTMT is closely connected with the refinement of the prior austenitic structure, and the consequent refinement of the martensitic platelets, although we may assume that many other factors must also contribute to the overall rise in the mechanical metallurgy parameters conferred by this treatment [1,2,3,4,7].

The results in Figure 4, and the corresponding data on the kinetics of crack growth in the 3.5% NaCl solution in Figure 5, show that the higher-oxygen $D_2$ steel is more liable to brittle failure than the $D_1$ steel. The $K_{Ii}$ versus $\tau$ curve of this steel is shifted towards the region of lower times to fracture, and the threshold level of $K_{ISCC}$ is also lower. This suggests that the higher oxygen content probably impairs the cohesive strength of the prior austenitic grain boundaries in conjunction with the specific characteristics of the martensitic transformation, this reduction of the cohesive strength could account for the changes set out above [3,4].

To supplement the findings on the mechanical metallurgy characteristics of steels with various strength levels, Figure 6 confronts the newly obtained data with some previously ascertained threshold values of $K_{ISCC}$ and $K_{ISH}$, including the appropriate $K_{Ic}$ values [1,2,3,8,9,11], plotted against the ultimate tensile strengths of the materials. The relationships summarized on this diagram indicate that $K_{ISH}$ represents the limiting criterion for assessing the susceptibility of steels to brittle failure in an aggressive environment. However, no generalized conclusions can be drawn from this evidence until a substantially larger body of data has been assembled and processed.

## CONCLUSION

The results gained in this work have corroborated that, no matter whether we base our evaluation on the $K_{ISCC}$ or the $K_{ISH}$ levels, HTMT always produces a material more resistant to brittle failure than CT does. A further finding is that an increased oxygen content both reduces the $K_{ISCC}$ level, and increases the stable crack growth rate v at a comparable magnitude of the $K_I$ values. These observations supplement our existing knowledge of the properties of ultra-high strength steels, and can contribute to the process of extending their potential scope of application.

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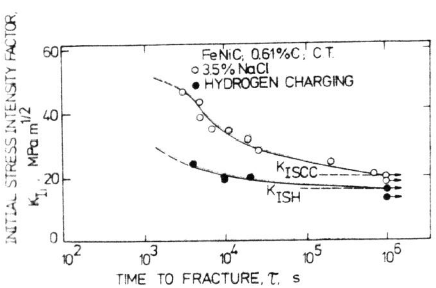


Figure 1 $K_{Ii}$ versus τ for Steel A

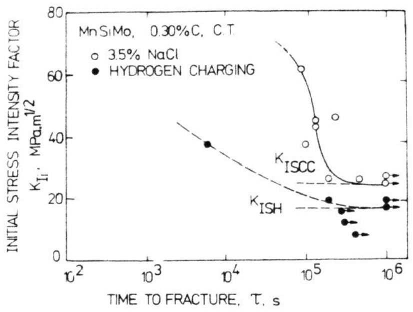


Figure 2 $K_{Ii}$ versus τ for Steel B

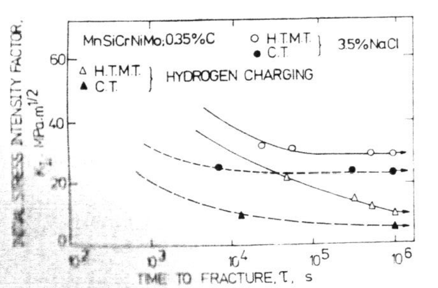


Figure 3 $K_{Ii}$ versus τ for CT and HTMT Specimens of Steel C

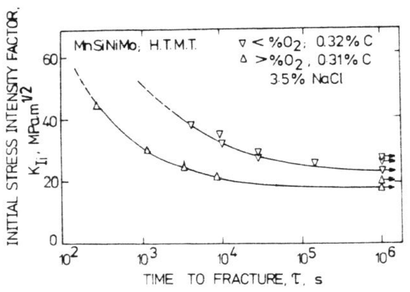


Figure 4 $K_{Ii}$ versus τ for Steels $D_1$ and $D_2$





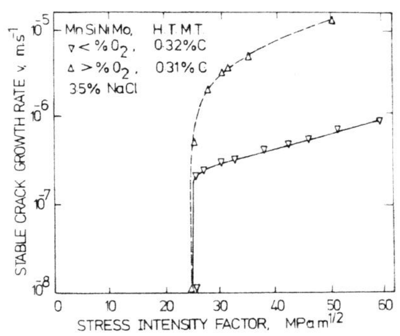


Figure 5 Stable Crack Growth Rate versus $K_I$ Dependence in Steels $D_1$ and $D_2$

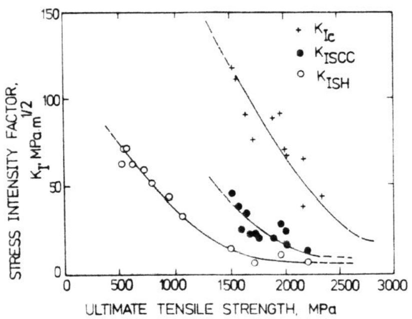


Figure 6 The Dependence of $K_{Ic}$, $K_{ISCC}$ and $K_{ISH}$ Upon the Ultimate Tensile Strength

Table 1 Chemical Composition of the Investigated Steels (wt - %)

| Steel | C | Mn | Si | P | S | Cr | Ni | Mo |
|---|---|---|---|---|---|---|---|---|
| A | 0,61 | 0,54 | 0,62 | 0,020 | 0,018 | - | 20,40 | - |
| B | 0,30 | 1,39 | 1,45 | 0,015 | 0,013 | 0,46 | 0,42 | 0,30 |
| C | 0,35 | 1,00 | 1,23 | 0,015 | 0,012 | 1,30 | 1,40 | 0,36 |
| $D_1$ | 0,32 | 1,10 | 1,19 | 0,016 | 0,013 | 0,38 | 0,95 | 0,35 |
| $D_2$ | 0,31 | 1,06 | 1,25 | 0,014 | 0,013 | 0,40 | 0,91 | 0,37 |

Table 2 Mechanical Properties of the Investigated Steels

| Steel | $\sigma_y$ | $\sigma_{UTS}$ | Elongation | $K_{Ic}$ | $K_{ISCC}$ | $K_{ISH}$ | Note |
|---|---|---|---|---|---|---|---|
| | MPa | | % | MPa.m$^{1/2}$ | | | |
| A | 1420 | 1680 | 9,8 | 51,5 | 20,2 | 16,1 | -- |
| B | 1520 | 1700 | 10,2 | 90,5 | 24,2 | 16,7 | -- |
| C | 1450 | 1650 | 8,2 | 76,5 | 21,4 | 5,3 | CT |
| C | 1680 | 1870 | 9,8 | 91,4 | 27,9 | 10,2 | HTMT |
| $D_1$ | 1565 | 1720 | 10,1 | 88,5 | 25,8 | - | -- |
| $D_2$ | 1540 | 1710 | 9,8 | 87,2 | 19,0 | - | -- |