EFFECT OF PURITY AND THERMOMECHANICAL TREATMENT ON FRACTURE BEHAVIOUR OF LOW-ALLOY MARTENSITIC STEELS

J. Eickemeyer and G. Zouhar

Akademie der Wissenschaften der DDR. Zentralinstitut für Festkörperphysik und Werkstoffforschung, Dresden, German Democratic Republic

ABSTRACT

Strength, fracture toughness and stress corrosion cracking threshold of tempered martensitic silicon-manganese steels were analysed in relation to local failure mechanisms after high temperature thermomechanical treatment (HTMT) and conventional treatment (CT). In order to investigate the role of minor impurities two material qualities containing different amounts of minor impurities are compared. The results indicate that the fracture toughness \( K_{IC} \) at constant strength level is markedly enhanced by the HTMT applied and also by decreasing the level of minor impurities. In comparison with these effects on \( K_{IC} \), the influence of HTMT on the stress corrosion cracking threshold \( K_{ISCC} \) is far less pronounced. An influence of the decreased impurity content on \( K_{ISCC} \) did not appear. It is concluded that the HTMT suppresses the enrichment of detrimental impurities at the austenite grain boundaries with the consequence of transcrystalline plastic fracture at unstable crack growth in the materials of commercial purity. However, this purification of grain boundaries by HTMT produces no marked effect on the stress corrosion cracking behaviour because of the cooperative action of hydrogen and detrimental impurities, which yield intercrystalline brittle fracture.

KEYWORDS

Martensitic steels; minor impurities; segregations; conventional treatment; thermomechanical treatment; fracture toughness; stress corrosion cracking; hydrogen embrittlement.

INTRODUCTION

Primary recrystallization of prior austenite in low tempered low alloy martensitic steels during HTMT may yield intercrystalline brittle fracture during unstable crack growth with relatively low fracture toughness \( K_{IC} \). This was observed on a 0.1%Cr-0.7%Mn- and a 0.5%Cr-0.1%V-
steel (Zoubar Jr., 1979, 1984). It was suggested, that critical enrichments of detrimental impurities at the austenitic grain boundaries appear by nonequilibrium segregation during primary recrystallization (Kieting, 1976, 1979; Beyer, 1981). If this enrichment is transmitted to the martensite during quenching it decreases the cohesion of the prior austenitic grain boundaries. Furthermore from the results on the 0.95Cr-0.2Fe-0.7V-steel it was concluded, that lattice defects introduced in the stable austenite by comparatively small amounts of plastic deformation during the last pass at HVTM without initiation of primary recrystallization increase the matrix solubility, and some purification of the grain boundary results. This is connected with transcrystalline plastic fracture and results in an increased fracture toughness, $K_{IC}$, as well as an increased stress corrosion cracking threshold, $K_{ISC}$ (Zoubar Jr. and co-workers, 1981). Investigations as to what extent this purification mechanism by HVTM may be applied also to low alloy Si-Mn-steel are not known. In this paper results about the influence of HVTM and CT on strength, toughness and stress corrosion cracking behavior of high strength martensitic Si-Mn-steel are presented. In order to elucidate effects of minor impurities on the fracture toughness and the stress corrosion cracking threshold two steel qualities with different purity levels are examined.

**EXPERIMENTS**

The chemical composition of the Si-Mn-steels used in this investigation is shown in Table 1. The abbreviation CP stands for the arc-melted material of commercial purity.  

<table>
<thead>
<tr>
<th>No.</th>
<th>Material</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Cr</th>
<th>Cu</th>
<th>Ni</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>65Si3Mn7 (CP)</td>
<td>0.39</td>
<td>1.62</td>
<td>0.73</td>
<td>0.074</td>
<td>0.325</td>
<td>0.18</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>2</td>
<td>65Si3Mn5 (CP)</td>
<td>0.50</td>
<td>1.10</td>
<td>1.11</td>
<td>0.061</td>
<td>0.036</td>
<td>0.21</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>3</td>
<td>55Si3Mn5 (CP)</td>
<td>0.47</td>
<td>1.44</td>
<td>0.63</td>
<td>0.020</td>
<td>0.034</td>
<td>0.42</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>4</td>
<td>&quot; &quot;</td>
<td>0.50</td>
<td>1.54</td>
<td>0.68</td>
<td>0.026</td>
<td>0.040</td>
<td>0.13</td>
<td>0.17</td>
<td>0.08</td>
</tr>
<tr>
<td>5</td>
<td>&quot; (HP)</td>
<td>0.42</td>
<td>1.44</td>
<td>0.71</td>
<td>0.035</td>
<td>0.036</td>
<td>0.03</td>
<td>0.02</td>
<td>0.04</td>
</tr>
</tbody>
</table>
|- not determined

The HP-quality was melted in a 10kg-vacuum induction furnace from high-purity base materials.

Rolled billets of 38 mm diameter (materials No. 1, 2, 3) were rolled down on a HVTM-experimental rolling mill (Schmitt and co-workers, 1976) in thirteen passes (austenitizing temperature 1250°C, finish-rolling temperature 850°C) to rods with an oval sectional area of about 50 mm² for prestressed concrete. The reduction of cross sectional area in the last two passes amounted to about 20%. Immediately after rolling the material was quenched in water, followed by an isochronous shock tempering treatment at different temperatures between 450 to 610°C (time of tempering about 2 s) to different strength levels. Further details on the deformation-temperature-time-regime of HVTM applied to materials No. 1, 2 and 3 are given elsewhere (Zoubar and co-workers, 1984). The conventional heat treatment was carried out by austenitizing at about 950°C, followed by oil quenching and bath annealing at 300, 400 and 450°C to obtain different strength levels. In total of the HVTM- and CT-conditions applied to materials No. 4 and 5 are described in another paper (Zoubar Jr. and co-workers, 1979). The HVTM and CT were carried out on this room. These rods were tempered at 200°C for 1 h, after quenching from finish-rolling temperature (HVTM) and austenitizing temperature (CT).

The specimens for tensile tests were prepared with their axes parallel to the rolling direction. The orientation of the fatigue precrack for the fracture mechanics SERR-bending specimens was at right angles both to rolling direction and plane. The threshold for stress corrosion cracking $K_T$, was measured in a saturated Ca(OH)₂ + Pao₅-solution ($pH = 4.5$). Details of the test method are published elsewhere (Sickkerseyel and co-workers, 1974). All tests were conducted at room temperature.

**RESULTS AND DISCUSSION**

Figure 1 illustrates the fracture toughness-yield stress and the stress corrosion cracking threshold-yield stress relations for the CP-materials 55Si3Mn7, 65Si3Mn5 and 65Si3Mn7, resulting from HVTM and CT.

**Fig. 1. Fracture toughness $K_{IC}$ and stress corrosion cracking threshold $K_{ISC}$ dependence on the yield stress $R_p$ for different Si-Mn steels after HVTM and CT. Numbers 2, 4 and 5 are related to scanning electron micrographs in Fig. 2.**

**Fig. 3. Fracture toughness $K_{IC}$ and stress corrosion cracking threshold $K_{ISC}$ dependence for different purity after HVTM and CT. Dashed symbols from Fig. 1. HP = high purity, CP = commercial purity.**
after HTMT (Fig. 25.), a pronounced intercrystalline brittle fracture with very small plastic areas predominates after CT (Fig. 24.).

Figure 3 shows the fracture toughness $K_{IC}$ and stress corrosion cracking threshold $K_{WSCC}$ in relation to the yield stress $R_{p0.2}$ for the steel 50SiMn2 with high and commercial purity after $P_{0.2}, HMT$ and CT. After CT the HP-material indicates an increased fracture toughness in comparison with the $K_{IC}$ values of the CP-material. The HP-material shows transcrysalline plastic fracture. The fracture toughness yield stress relations of the HP-material after HTMT fit in the corresponding $K_{IC}$ of the CP-materials (Fig. 4). Independant of $P_{0.2}$, ine of the CP-materials (Fig. 4). Independant of $P_{0.2}$, ine of the CP-materials (Fig. 4).

The above results reveal that the HTMT controls the cohesion of the prior austenite grain boundaries in high strength martensitic low-alloy steels. Grain refinement and/or polygonized dislocation substructures developed in steels austenite during HTMT control the kinetics of the martensitic transformation and the accommodation processes during the transformation in such a way that results a fine dispersed martensite with homogeneously distributed and increased local stresses and stresses (Zöhrer and co-workers, 1986; Zöhrer and co-workers, 1985; Zöhrer, 1985). These factors are not very likely as a reason for the observed increase in fracture toughness and stress corrosion cracking threshold by the used HTMT. Polygonized dislocation substructures are not expected in the hot rolled condition of the steel for precracked concrete (Zöhrer and co-workers, 1985), and nearly the same prior austenite grain size is observed after HTMT and CT. Therefore it has to be suggested that the different fracture behaviour of the Si-alloy steels with commercial purity after HTMT and CT is caused by different enrichments of detrimental minor impurities at the grain boundaries in the austenite.

It is well known that critical enrichments of minor impurities may be produced at austenite grain boundaries by equilibrum and nonequilibrium segregation. Figure 4 shows these mechanisms schematically. If such critical enrichments are transmitted to the martensite during quenching they decrease the cohesion of the prior austenite grain boundaries, the comparatively low temperature of martensitization at CT supports the development of critical enrichments at the austenite grain boundaries of the CP-materials by equilibrum segregation with the consequence of intercrystalline brittle fracture at unstable crack growth (Fig. 5.).

It is expected, that for conditions examined grain boundary segregations of substitutional elements like phosphorus (Schart and co-workers, 1983) and their interaction with copper (Wieting, 1971) are of concern. Sulphur should be bound to manganese. In decreasing the content of minor impurities the $K_{IC}$ values are increased by about 50% in consequence of HTMT compared to the $K_{IC}$-values after CT at yield stresses $R_{p0.2} = 1000$ to 1500 MPa. This result corresponds with the different fracture behaviour after HTMT and CT during subcritical crack growth, whereas a mixed fracture mode prevails.
way to control detrimental effects of minor impurities on the fracture resistance of high-strength low-alloy Si-Mn steels during unstable crack growth.

From the absence of a mutiny effect on the stress corrosion cracking threshold, it is concluded that a reduction of the phosphorus and sulphur content to 50 ppm is not sufficiently to influence $K_{frc}$. By variation of the phosphorus and sulphur contents, respectively, between 10 and 200 ppm in an AISI 4340 steel an influence on $K_{frc}$ also did not appear (Sandoz, 1971). In comparison with phosphorus and sulphur, the alloying elements manganese and silicon should have an overwhelming effect on $K_{frc}$ (Banerji and coworkers, 1970). These elements cooperate also, whereby a comparatively strong interaction between manganese and phosphorus has to be taken into account (Courtman, 1970). In addition silicon increased the solubility of hydrogen in steel (Szczepanski, 1963). In this way manganese and silicon may support the hydrogen-induced fracture by increasing the critical contents of minor impurities to lower levels with respect to $K_{frc}$. This was proved by removal of manganese and silicon from a low-alloy steel (Banerji, 1978). Thereby the stress corrosion cracking threshold $K_{frc}$ was increased five- to sixfold, although the steel contained 100 ppm phosphorus and 200 ppm sulphur. Therefore an increase of the $K_{frc}$ values of the used Si-Mn steels probably requires such decreases contents of minor impurities, which are scarcely of interest on technical conditions.

It is suggested, that the small increase of the $K_{frc}$ values by TMT is caused by the distribution of minor impurities and by a small anisotropy effect, such that longitudinal cracks parallel to rolling plane appear after $K_{frc}$ testing.

CONCLUSIONS

1. The fracture toughness of low-alloy martensitic Si-Mn steels may be increased by about 100% at nearly the same strength level by TMT with respect to fracture toughness after CF. A reduction of the level of minor impurities also enhances the $K_{frc}$ value.

2. In comparison with these effects on $K_{frc}$ the influence of TMT on the stress corrosion cracking threshold is far less pronounced. An influence of the increased impurity content on $K_{frc}$ did not appear.

3. It is concluded that the TMT applied suppresses the enrichment of detrimental minor impurities at the austenite grain boundaries with the consequence of transcrystalline plastic fracture during unstable crack growth in the Si-Mn steels with commercial purity. This purification mechanism by TMT appears as a beneficial way to control detrimental effects of minor impurities on the fracture resistance of high strength low-alloy Si-Mn steels during unstable crack growth.

4. However, this purification of grain boundaries by TMT produces no marked effect on the condition of stress corrosion cracking because of the cooperative action of hydrogen and detrimental impurities, which yield intercrystalline brittle fracture. Manganese and silicon should enhance the sensitivity of this type of steel to hydrogen induced fracture by decreasing the critical contents of minor impurities to very low levels with respect to the initiation of stable crack growth ($K_{frc}$). Therefore an increase of the $K_{frc}$ values of the Si-Mn steels probably requires such low contents of minor impurities, which are scarcely of interest on technical conditions.

REFERENCES


Metallurg [Prentzel], 1, 193-203.


