On the Toughness and Micro-Structure in Low Carbon Steels Subjected to Weld Thermal Cycles

By Hiroshi Mimura*, Makio Iino*, Hiroyo Haga*, Nobuchika Nomura*, Kazuo Aoki* and Koichi Aoki**

Abstract

With a view to making clear some of the relations between toughness and metallurgical structure in a weld joint subjected to a large heat input, toughness was studied through careful examination of micro-structure, of two steels A and B (A: 0.1C, 0.25Si, 1.15Mn; B: 0.1C, 0.25Si, 1.15Mn and 0.30Mo) subjected to varied weld thermal cycles. Test results are summed up as follows:

1. There appear high carbon martensite, M*, between the needles of an acicular ferrite in the steel containing 0.3%Mn, if cooled at medium cooling rates from 1350°C.

2. If the structure of a specimen contains M*, toughness deteriorates seriously.

3. If a specimen is free from M*, variation in toughness is rather gradual with respect to the cooling rate. This is due to the fact that an austenitic grain size or cleavage facet size gradually changes with the cooling rate.

I Introduction

Nowadays, there is a tendency to adopt a large heat input in automatic welding. A weld joint, however, is apt to be more brittle as heat input is increased.

Therefore, it is desirable to make clear the relationship between the toughness of a weld joint subjected to a large heat input and the factors such as chemical composition, microstructure and welding method. As the first step to do this, the toughness of low carbon, low manganese steels subjected to weld thermal cycles was investigated through careful examination of micro-structure.

II Experimental procedures

Two alloys, steel A (0.1C, 0.25Si, 1.15Mn) and steel B (0.1C, 0.25Si, 1.15Mn and 0.30Mo), were vacuum melted, cast to an ingot of about 20 kg weight and hot rolled to 14 mm in thickness. Square rods of 12×12×65 mm³ were cut from the hot rolled plates. The rods were quenched from 930°C, tempered at 600°C for 40 min. and subjected to weld thermal cycles.

Induction heating was applied to give a weld thermal cycle to the central part of the rod with a homogeneous temperature zone of about 5 mm length. The maximum temperature of the cycle was set at 1350°C. From the rod a 2 mm V-notch Charpy test piece was machined.

The chemical compositions of specimens are shown in Tab. 1 and cooling rates of the thermal cycles in Fig. 1.

Table 1. Chemical compositions of specimens (wt %)

<table>
<thead>
<tr>
<th>Spec. No.</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>Mo</th>
</tr>
</thead>
<tbody>
<tr>
<td>V 816</td>
<td>0.11</td>
<td>0.24</td>
<td>1.13</td>
<td></td>
</tr>
<tr>
<td>V 853</td>
<td>0.11</td>
<td>0.26</td>
<td>1.23</td>
<td></td>
</tr>
<tr>
<td>V 892</td>
<td>0.13</td>
<td>0.25</td>
<td>1.19</td>
<td></td>
</tr>
<tr>
<td>V 1056</td>
<td>0.09</td>
<td>0.26</td>
<td>1.16</td>
<td></td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Spec. No.</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>Mo</th>
</tr>
</thead>
<tbody>
<tr>
<td>S 775</td>
<td>0.10</td>
<td>0.27</td>
<td>1.16</td>
<td>0.29</td>
</tr>
<tr>
<td>V 1008</td>
<td>0.11</td>
<td>0.25</td>
<td>1.14</td>
<td>0.29</td>
</tr>
<tr>
<td>V 1122</td>
<td>0.09</td>
<td>0.23</td>
<td>1.17</td>
<td>0.29</td>
</tr>
<tr>
<td>V 866</td>
<td>0.10</td>
<td>0.26</td>
<td>1.17</td>
<td>0.28</td>
</tr>
</tbody>
</table>

Fig. 1. Weld thermal cycles used in the present work. The cooling rate of cycle W, which appears in Fig. 3 and Fig. 3', was prescribed to 4°C/min. throughout.

III Experimental Results

1. Transformation characteristics

1-a C.C.T. diagrams.

The C.C.T. diagrams for steels A and B are shown in Fig. 2. In a medium cooling rate range, an acicular ferrite, Fₐ, precipitated at austenite grain boundaries, while there coexisted a polygonal structure etched gray under an optical microscope, which seemed to be a kind of sorbite, S. Austenite between acicular ferrite needles was decomposed into a ferrite-cementite mixture (pearlite + ferrite, or Zwischenstufe, Zw) or transformed into a high carbon martensite, M*, in part in steel B. And even if the main structure was similar, when observed under an optical microscope of low magnification, there existed a wide variety of

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1-b Metallurgical structure of steels subjected to weld thermal cycles.

As seen from the C.C.T. diagram, the structure of steel B subjected to cycles X, Y or Z, was composed mainly of an acicular ferrite containing $M^*$ and free from $F$. The width of $F_a$ was decreased with an increasing cooling rate.

On the other hand, the structure of steel A varied from an acicular ferrite and cementite to a mixture of an acicular and polygonal ferrite and pearlite owing to the cooling rate change from cycle $Z$ to $X$. Fraction of So structure diminished with an increasing cooling rate or alternatively by addition of 0.3% Mo.

In effect, addition of molybdenum of 0.3%, retarded appearance of allotriomorphs and made the structure stable to the effect that an upper bainitic structure is obtained and $M^*$ appears in a wide cooling rate range. In Photo. 2 are shown metallurgical structures of steels A and B subjected to various weld thermal cycles.

2. Notch toughness

Steel A had a stable toughness, i.e. an energy transition temperature in Charpy test $\psi_{TE}$, fell in the narrow range from $-60^\circ$C to $-30^\circ$C, for the whole cooling rate range of the present thermal cycle, while toughness in steel B deteriorated considerably as the cooling rate was decreased, but recovered at further slower rates. These results are shown in Fig. 3.

3. Response to artificial thermal cycles.

In order to make clear the influence upon toughness of a high carbon martensite $M^*$ found in steel B, specimens were subjected to an artificial thermal cycle $N$, or $L$ which was intended to replace $M^*$ by $Z_{w}$.

Fig. 2. The C.C.T. diagrams for Steels A and B. Maximum heating temperature is 1350°C.

Fig. 3. Energy transition temperature ($\psi_{TE}$) of specimens subjected to weld thermal cycles as a function of the cooling rate. Values of cooling rates represented on abscissa are the averaged ones between 800°C and 500°C.

Fig. 3'. Energy transition temperature vs. cooling rate for Steel B subjected to various thermal cycles. O. Q.; oil quenched from 1350°C. The dotted line in the figure shows the supposed relation between $\psi_{TRE}$ and cooling rate in the absence of $M^*$. 

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Photo 1. Micro-structure of So and $M^*$ by optical microscopy ($\times$500), replication ($\times$3000) and direct observation ($\times$45000, 30000)
Photo. 2. Micro-structure of steels A and B by optical microscopy (×200)

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or vice versa without changing the other metallurgical factors. Thermal cycle N consists of the following heat treatment: a specimen is cooled from 1350°C at the rate of cycle X down to 450°C, kept at this temperature for 20 min and cooled to R.T.

Steel B subjected to cycle N could be, unintentionally, divided into two definitely distinguishable groups, as shown in Fig. 4. In the first group, austenite between needles of an acicular ferrite was decomposed into Zn completely. The other metallurgical characters were quite similar to those of one subjected to cycle X except that the edges of cementite rods in specimens cycled N were slightly rounded. In the second group M* did not disappear yet. This is explained if one supposes that the keeping time of 20 min. was just critical for completion of Zn transformation. The impact transition temperature of the first group was quite low as compared with that of the second group, subjected to normal cycles.

In steel A which was free from M* originally, toughness depended little on whether cycle N or X was applied.

Thermal cycle L corresponds to the following heat treatment: a specimen was cooled from 1350°C at the rate of cycle Z down to 550°C, kept at this temperature for 1/2 min and quenched to R.T. Steel A subjected to this cycle had a structure composed of Zn and M*, and the morphology of Zn was quite similar to that of one subjected to cycle Z. As seen from Fig. 5, T_{RE} of specimens subjected to cycle L was quite high as compared with that subjected to cycle Z. Then it will be understood that existence of M* causes T_{RE} to rise by a large amount.

4. Properties of the high carbon martensite M*.

Direct observation by electron microscopy showed that M* was of a twinned type characteristic of a high carbon martensite. In this regard, it would be interesting to measure the M* point of M*. The volume fraction of M* is, however, too small to be detected by dilatation due to martensitic transformation. Then, the specimen was cooled from 1350°C at a proper rate down to temperature T and kept at this temperature for time t, sufficient for completion of transformation γ→Zn. If T is below the M* point, tempered high carbon martensite islands must be found. A tempered martensite was distinguished from a Zn structure under an electron microscope. The M* point thus obtained is quite low, especially on slow cooling, as is shown in Table 2. This may explain why steel B deteriorated when slow-cooled.

<table>
<thead>
<tr>
<th>Cooling rate</th>
<th>M* point (°C)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Z</td>
<td>370 ~ 400</td>
</tr>
<tr>
<td>Y</td>
<td>270 ~ 300</td>
</tr>
<tr>
<td>X</td>
<td>250 ~ 280</td>
</tr>
</tbody>
</table>

IV Discussion

What may be taken into account as factors controlling the toughness are, firstly main metallurgical characteristics as stated in III—1—b, secondly an austenitic grain size, and thirdly co-existence of another brittle phase such as high carbon martensite islands M* (G).

It will be seen from comparison between γT_{RE} and structure that the main structure is not a controlling factor in the present case, for steel A has a stable toughness without regard to a considerable variation in structure or in the cooling rate, while γT_{RE} of steel B shows a wide variation in the cooling rate range in which a structure is almost acicular ferrite. Of course, the width of a needle of an acicular ferrite is increased as the cooling rate is decreased, but widening of the needle does not seem to be detrimental, as will be discussed in appendix I.

Variation in the fraction of So has no correlation, either, with that in toughness, for So, which is more often found in steel A than in Steel B, is rather a brittle structure, as confirmed by fractographical observation.
An austenitic grain size is slightly increased as the cooling rate is decreased, presumably because of longer exposure to high temperature for slower-cooled specimens. Change in the austenitic grain size is gradual, which explains the slow rise in \( T_{\text{TE}} \) at very slow cooling rates in steel A. But the variation in toughness in steel B cannot be explained in terms of that in austenitic grain size.

On the other hand, the deterioration observed in steel B cooled at a medium cooling rate has good correlation with appearance of \( M^* \). For instance, the structure of steel B subjected to a very slow cooling cycle \( W \) is very coarse to all appearance, consisting of a polygonal ferrite and a small fraction of pearlite, but free from \( M^* \). And its toughness is far better. Improvement in toughness by an artificial thermal cycle manifests the detrimental effect due to \( M^* \) more clearly. Only in the case in which \( M^* \) was replaced by \( Z \) by application of cycle \( N \), toughness was improved. An inverse case was seen in steel A subjected to cycle \( L \).

The degree of deterioration due to \( M^* \) may be given by the amount of improvement in \( T_{\text{TE}} \) by cycle \( N \). If \( M^* \) does not appear in steel B, \( T_{\text{TE}} \) seems to vary gradually with the cooling rate, as shown in Fig. 3' by a dotted line. Remembering that steel A has a stable toughness with respect to the cooling rate, it will be concluded that the toughness of steel subjected to a weld thermal cycle is rather stable, if free from \( M^* \), and deteriorates seriously, when \( M^* \) appears.

V Conclusions

Toughness was studied through careful observation of microstructure in two low carbon, low manganese steels (A: 0.1C, 0.25Si, 1.15Mn; B: 0.1C, 0.25Si, 1.15Mn, 0.3Mo) subjected to weld thermal cycles. It was found that:

1. There appear high carbon martensite islands, \( M^* \), between the needles of an acicular ferrite in the steel containing 0.3\%Mo, if cooled at medium cooling rates from 1350°C.

2. If the structure of a specimen contains \( M^* \), toughness deteriorates seriously.
3. If a specimen is free from \( M^* \), variation in toughness is rather gradual with respect to the cooling rate. This is due to the fact that an austenitic grain size or cleavage facet size gradually changes with the cooling rate.

Acknowledgment

The authors wish to thank Dr. T. Ohtake, Director of Tokyo Research Institute, Yawata Iron and Steel Co., Ltd. for his interest and encouragement.

Appendix 1 Relation between \( \tau \) grain size and cleavage facet size.

It is usually expected that steel with a fine \( \tau \) grain size has high toughness. This may be due to the fact that steel with fine \( \alpha \) grains is obtained by reduction of the \( \tau \) grain size.

However, it is not very clear that reduction of the \( \tau \) grain size favours toughness in steels of an acicular ferritic structure in a similar way to the case of a polygonal ferrite. In order to make this point clear, the size of a cleavage facet was examined on Charpy test pieces.

It was found that a cleavage path changes its direction about 3–4 times in one prior austenite grain on the average, without regard to whether the structure is polygonal or acicular ferritic, or to whether an acicular ferrite needle is broad or narrow. Then the \( \tau \) grain size can be a good measure of the grain size effect upon toughness in the present case.

Appendix 2 Micro-cracks associated with martensite islands

The high carbon martensite island is expected to initiate a crack by acting as a stress concentrator or by cracking itself. In fact, microscopic observation of martensite islands near a fracture path showed that there were microcracks associated with the martensite islands, as is seen in Photo. 1A.

Photo. 1A. Observed microcracks associated with \( M^* \). (\( \times 3000 \))
References


2) P.S. Trozzo and G.E. Pellissier: "Application of Fracture Toughness Parameters to Structural Metals" (1960),