ABSTRACT

The formation of acicular ferrite which gives high strength and good impact toughness in C-Mn steel weld metals has been studied using dilatometry in combination with optical microscopy and transmission electron microscopy (TEM) associated with microanalytical technique, i.e. energy dispersive X-ray (EDX) analysis. The results showed that Ti was essential in promoting acicular ferrite and the complexity of the austenite to ferrite transformation was likely to be associated with interactions between Ti, N, O and other alloying elements during welding process.

INTRODUCTION

In modern engineering practice, fusion welding is widely used for applications such as the fabrication of pressure vessels, large offshore constructions and joints in pipelines. Fusion welding is defined as a joining process which is accomplished by melting followed by fusion (Kou, 1987). One of the most common of the fusion welding processes is manual metal arc (MMA) welding also known as shielded metal arc (SMA) welding in the USA and Canada. Up to now, MMA welding is the simplest and perhaps the most versatile process for welding ferrous metals.

In recent years, the demand for more reliable and improved weld metal toughness for high strength low alloy (HSLA) steels has led to a great number of investigations with the objective of achieving such desirable properties by controlling weld metal microstructure.

WELD METAL MICROSTRUCTURE

In steel weld metals, the carbon content is maintained low, typically below 0.05% to approximately 0.1 wt % to maintain good weldability (Lancaster, 1986; Easterling, 1992). As a result, a sequence of transformations occur as they cool from liquid to ambient temperature, i.e. liquid metal $\rightarrow$ δ-ferrite $\rightarrow$ γ (austenite) $\rightarrow$ α (ferrite) as shown in Figure 1.

The austenite to ferrite transformation plays an important role in determining the mechanical properties of weld metal since it controls the final microstructure of weld metals.

It is generally accepted that the weld metal microstructure can be arranged in order of decreasing transformation temperature as follows (Abson and Pargetter, 1986):

1. Allotriomorphic or grain boundary ferrite (GF) forms between 1000 to 650 °C at prior austenite grain boundaries.
2. Widmanstatten ferrite (WF) forms between 750 and 650 °C emanating from the prior austenite grain boundaries.
3. Acicular ferrite (AF) forms below 650 °C within the prior austenite grains.
4. A lath structure which is known as bainite (B), forms at 500 °C.
5. There may be other micro constituents consisting of martensite, retained austenite or carbides (MAC).

Figure 1. Weld metal microstructure (Easterling, 1992)

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Welding was done in the flat position using the stringer bead technique and three beads per layer were deposited. The plate thickness was 20 mm and 27 mm were made. Direct current (electrode positive) was applied to the weld preparation where the current, the voltage and the heat input were 170 A, 25 V and 2.1 kJ/mm respectively. The interpass temperature was standardised at 200 °C. For investigation, all samples were taken from the top central bead of weld metal whereas the chemical compositions of the weld metal under study were made with different levels of Ti and N. Details of their chemical compositions are given in Table 1.

Table 1. Chemistry of weld metals, wt %

<table>
<thead>
<tr>
<th></th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>Cr</th>
<th>N</th>
</tr>
</thead>
<tbody>
<tr>
<td>O</td>
<td>0.074</td>
<td>1.40</td>
<td>0.25</td>
<td>0.0001</td>
<td>0.008</td>
</tr>
<tr>
<td>X</td>
<td>0.069</td>
<td>1.47</td>
<td>0.45</td>
<td>0.0410</td>
<td>0.008</td>
</tr>
<tr>
<td>O2</td>
<td>0.073</td>
<td>1.66</td>
<td>0.27</td>
<td>0.0095</td>
<td>0.0074</td>
</tr>
<tr>
<td>X2</td>
<td>0.068</td>
<td>1.46</td>
<td>0.47</td>
<td>0.0458</td>
<td>0.025</td>
</tr>
</tbody>
</table>

Of note is that each sample contained oxygen approximately 0.63 wt%.

**Optical Microscopy**

Microscopical examination was carried out using a **NIKON** optical microscope with the magnification of X400.

**TEM Studies**

The mechanism of how acicular ferrite to develop was studied using a **PHILLIPS CM 20** transmission electron microscope accompanied with EDX-analysis to identify various phases present in the microstructure.

**Dilatometry**

The transformation kinetics of the weld metal was studied using a Theta high speed dilatometer. Samples for this investigation were made in the form of hollow cylinders with standard specimen dimensions, i.e. 10 mm long by 5 mm outside diameter with a wall thickness of 1 mm. The axis of each specimen was parallel to the weld direction. The samples were continuously cooled at the rate of 13 °C/s since this cooling rate was likely to be near the actual weld cooling rate.

**RESULTS**

**Optical Microscopy**

Figure 3 shows microstructures of weld metal coded O, O2 and X, X2. It can be seen that weld...
metals with no Ti (O and O2) were dominated by Widmanstätten ferrite. The addition of Ti caused the formation of acicular ferrite (X and X2) whereas N tended to refine the weld microstructure.

**Figure 3.** As deposited microstructures of weld metals O, O2, X and X2

**TEM Studies**

The results obtained from TEM showed that acicular was found to nucleate on TiO inclusion as shown in Figure 4.

**Figure 4.** Acicular ferrite nucleated on an inclusion containing TiO

**Dilatometry**

Figure 5. shows an example of dilatometry results obtained during austenite to ferrite transformation.

![Dilatometry Graph](image)

**Figure 5.** An example of dilatation curve

It can be seen that the transformation is accompanied by volume changes. This can be explained based on the fact that austenite is the face centre cubic (fcc) form of iron whereas ferrite is the body centre cubic (bcc) crystal structure of iron so that the fcc crystal structure of austenite is more closely packed than bcc. As a result, the volume expands during the transformation.

The curve shown above can be divided into three regions. Firstly, contraction of austenite occurs linearly with temperature during cooling. This contraction finishes at \( T_1 \) which indicates the beginning of ferrite formation. On further cooling, the curve starts to deviate from linearity and expansion takes place due to transformation until \( T_2 \) is reached suggesting that most austenite has transformed to ferrite. Finally, linear contraction of ferrite occurs starting from \( T_3 \) to ambient temperature. Referring to Figure 5 the fraction of austenite transformed to ferrite can be determined using the following equation:

\[
X = \frac{OA}{OB}
\]  

so that the transformation kinetics of the weld metals can be presented by plotting fraction of austenite transformed to ferrite \( X \) as a function of temperature \( T \) as shown in Fig. 6, whereas the transformation data can be seen in Table 2.
It is generally agreed that transformation behaviour of weld metals can be accounted for in terms of transformation start temperature (\(T_s\)), 50% transformation temperature (\(T_{50}\)) and peak rate transformation temperature (\(T_p\)). The results are given in Table 1 with the error in determining \(T_s\) was about \(\pm 2^\circ C\) whereas the error in \(T_p\) may have been somewhat greater.

### Table 2. Transformation temperatures of the weld metals

<table>
<thead>
<tr>
<th>Metals</th>
<th>(T_s)</th>
<th>(T_{50})</th>
<th>(T_p)</th>
<th>(T_f)</th>
</tr>
</thead>
<tbody>
<tr>
<td>O</td>
<td>760</td>
<td>658</td>
<td>650</td>
<td>554</td>
</tr>
<tr>
<td>X</td>
<td>760</td>
<td>660</td>
<td>630</td>
<td>568</td>
</tr>
<tr>
<td>O2</td>
<td>754</td>
<td>630</td>
<td>606</td>
<td>534</td>
</tr>
<tr>
<td>X2</td>
<td>760</td>
<td>650</td>
<td>638</td>
<td>572</td>
</tr>
</tbody>
</table>

Of note is that \(T_s\) is associated with the formation of grain boundary ferrite whereas \(T_{50}\) and \(T_p\) can be used as an indication of acicular ferrite formation.

**DISCUSSION**

**Kinetics of the Austenite to Ferrite Transformation**

The effects of titanium and nitrogen on the transformation kinetics of C-Mn weld metals can be studied using the dilatometry results as shown in Figure 6. It can be seen that during the early stages of the transformation, typically above 700 °C, the transformation rates of all weld metals were very slow due to the formation of allotriomorphic ferrite. Subsequently, as cooling proceeded, the weld metals studied exhibited variations in the transformation kinetics which could be associated with the presence of inclusions and weld metal compositions.

The transformation reactions of weld metals O and O2 which were dominated by Widmanstätten ferrite occurred at very slow rates. The fraction of austenite transformed to ferrite seen in the two weld metals increased gradually as the temperature decreased and, consequently, the transformation finished at lower temperatures. The 50% transformation temperatures of weld metals O and O2 were achieved at lower temperatures approximately 658 and 630 °C respectively. These results seem to suggest that Widmanstätten ferrite does not grow abruptly from allotriomorphic ferrite but it appears to be the product of a continuously and often irregularly evolving growth process consistent with the previous observation (Aaronson, 1960) in Fe-C alloys. It has been suggested that the development of Widmanstätten ferrite is controlled by the instability of the advancing austenite-ferrite interfaces which may be related to the local carbon diffusion condition at the interface (Trowsend and Kirkaldy, 1968). In the model, carbon is rejected from the ferrite tip and diffuses laterally to the sides of the ferrite plates so that austenite near ferrite-ferrite interfaces is enriched in carbon. As the transformation proceeds, pools of austenite which are trapped between Widmanstätten ferrite plates transform to macroconstituents such as martensite, retained austenite and carbides (MAC).

The transformation kinetics of titanium containing weld metals, namely X and X2, occurred faster rates compared to those observed in weld metals O and O2. It seems that an increase in the transformation rates may be associated with the formation of intragranular acicular ferrite. The results of TEM studies showed that some acicular ferrite nucleated from TiO inclusions leading to interlocking structures, see Figure 4. Therefore, it can be summarised that the major effect of Ti on the decomposition of austenite seen in the present study is that titanium reacts with oxygen to form strong oxides such as TiO which could effectively provide sites for acicular ferrite nucleation hence accelerating the kinetics of the austenite to ferrite transformation.

It seems that oxygen prevents nitrogen from forming TiN since titanium has a strong tendency to form oxides rather than nitrides. Therefore, it is likely that most nitrogen in the weld metals studied is present in ferrite as free nitrogen. Soluble nitrogen well known as an austenite stabiliser, i.e. it expands the γ-phase field in the equilibrium Fe-C phase diagram and lowers the temperature at
Hombury and Bhadeshia, 1995; Sinha, 1989). This effect has been demonstrated by weld metal O2 which transformed at lower temperature compared to weld metal O. However, such a particular effect was less pronounced in weld metals X and X2.

**Mathematical Modelling**

There has been a number of works made with regard to modelling of weld metal transformation in particular acicular ferrite. Tewlis et al. (1997) have recently proposed a model for acicular ferrite formation based on the classical theory of nucleation and growth. According to this model, the rate of heterogeneous nucleation in metals and alloys is commonly given by the following equation:

\[
N_{ac} = N_i \exp \left( - \frac{\Delta G_{ac}}{RT} \right) \exp \left( - \frac{Q}{RT} \right)
\]  
\[(3)\]

where \(N_i\) is a nucleation frequency factor, \(N_i\) is the total number of heterogeneous nucleation sites per unit volume, \(\Delta G_{ac}\) is the energy barrier against nucleation, \(Q\) is the activation energy for atomic migration across the nucleus/matrix interface, \(T\) is temperature and \(R\) is the universal gas constant.

The equation above implies that the nucleation rate depends on the balance between the diffusivity of element at a given temperature and the thermal driving force which acts to overcome the energy barrier to nucleation. At a high temperature, the diffusion rate is high but the driving force is low hence \(N_{ac}\) is small. At a low temperature, the driving force is high but the diffusion rate is low so that \(N_{ac}\) again small. Therefore, there is a certain temperature at which the diffusion rate and driving force are optimised resulting in maximum \(N_{ac}\). However, further work needs to be performed to improve this model since the formation of acicular ferrite is very complex involving many variables such as inclusion chemistry, weld metal composition, prior austenite grain size and cooling rate.

The results obtained from dilatometry and TEM studies in the present investigation have shown that:

1. Acicular ferrite often nucleates heterogeneously on inclusions or on the plates of earlier inclusion-nucleated acicular ferrite known as sympathetic nucleation.
2. The chemistry of inclusions plays a vital role in nucleating acicular ferrite since changes in the inclusion compositions due to Ti additions are accompanied by the microstructural changes.

3. The transformation kinetics of acicular ferrite is faster than that of Widmanstätten ferrite and it seems to be controlled mainly by nucleation within intragranular regions.

Based on the results above and a number of works reported previously (Easterling, 1992 and Tewlis et al., 1997), a model for the transformation kinetics of acicular ferrite formation is proposed. The kinetics of acicular ferrite formation is consistent with the Avrami equation:

\[
X(t) = 1 - \exp(-kt^\alpha)
\]  
\[(4)\]

where \(X(t)\) is volume fraction transformed to ferrite, \(k\) is the rate constant, \(\alpha\) is the Avrami exponent and \(t\) is the time required to transform. The equation above applies to isothermal transformation so that this equation needs to be modified when it is used for continuous cooling transformation. The continuous cooling transformation can be considered to be the sum of a number of a fractional nucleation rate as proposed by Scheil (1935) as shown in Figure 7.

![Figure 7. Fractional nucleation time](image)

\[
\tau(T) = \int_0^{\tau(T)} \frac{1}{\tau(T)} \frac{dt}{d\tau} = \int_0^{T} \frac{1}{\tau(T)} \frac{dT}{dT} = \int_0^{T} \frac{dT}{\tau(T)}
\]  
\[(6)\]

where \(Q(T)\) is the cooling rate during continuous cooling transformation and \(\tau(T)\) is time required to form a certain fraction of ferrite at any given temperature \(T\).

Substituting equation (6) into (4) where \(t = \tau(T)\) results in:
\[ X(T) = 1 - \exp\left(-k \left[ \int \frac{dT}{Q(T)} \right]^n \right) \]  \hspace{1cm} (7)

Of note is that the rate constant \( k \) usually depends on temperature \( T \) so that the equation (7) can be written as follows:

\[ X(T) = 1 - \exp\left(-k \left[ \int \frac{dT}{Q(T)} \right]^n \right) \]  \hspace{1cm} (8)

It can be inferred from the equation (8) that the kinetics of transformation is controlled by \( k \), \( Q \) and \( n \). The rate constant \( k \) gives an indication of nucleation rate (\( N \)) and growth rate (\( G \)). It can be argued that:

1. Growth rate (\( G \)) is not essentially affected by weld metal composition but it is likely to be controlled by surface area/unit volume (\( N \)) and concentration gradient at austenite-ferrite interface.
2. Nucleation rate (\( N \)) is controlled by a number of nucleation sites and their efficiency.

Cooling rate (\( Q \)) is related to the time required by transformation reactions to occur via diffusion process.

It seems that the transformation behaviour of Ti-N type weld metals can be explained based on this model. An increase in the transformation kinetics of Ti- containing weld metals is attributed to TiO inclusions which act as effective sites for nucleation of acicular ferrite and hence cause \( N \) or \( k \) to increase. According to the equation (8), increasing rate constant (\( k \)) will increase the transformation rate. In contrast, soluble N is thought to act as an austenite stabilizer hence it inhibits diffusion process which in turn inhibits nucleation and growth process. As a result, the rate constant (\( k \)) decreases and slows down the transformation kinetics.

CONCLUSIONS

The conclusions that can be drawn from this investigation are as follows:

1. Effect of Ti on the formation of acicular ferrite may be associated with its tendency to form titanium oxides which provide potent sites for nucleation of acicular ferrite.
2. An increase in Ti tends to accelerate the transformation kinetics of C-Mn weld metals with acicular ferrite as the dominant phase.

3. In C-Mn weld metals increasing N content is to lower transformation start temperatures but an effect is less pronounced in Ti-containing metals.

AKNOWLEDGEMENTS

Sincere thanks are extended to Prof. J. Cowhane of the Department of Materials, University of Leeds, UK, for useful discussion with Dr. O.M. Evans of Oerlikon Welding Ltd. for supplying materials used in this investigation.

REFERENCES


