Phase Transformations and Microstructure Development in Low Alloy Steel Welds.

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ABSTRACT

Microstructure development in low alloy steel welds depends on various phase transformations that are a function of weld heating and cooling. The phase changes include non-metallic oxide inclusion formation in the liquid state, weld pool solidification, and solid state transformations. In this paper the mechanism of inclusion formation during low alloy steel welding is considered and the model predictions are compared with published results. The effect of inclusions on the austenite to ferrite transformation kinetics is measured and the mechanisms of transformation are discussed. The austenite grain development is related to the driving force for transformation of δ ferrite to austenite.

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INTRODUCTION

Fundamental understanding of physical processes that occur in fusion welding of low-alloy steels is necessary to design reliable and economical welding processes and consumables (1-2). The physical processes include evaporation of alloying elements in steel, dissolution of gases from arc atmosphere, oxide inclusion formation, solidification, solid state transformation and development of residual stresses (3). Various phase changes that occur as a function of weld cooling, in the weld metal region, are illustrated in a schematic diagram (Figure 1). The weld pool region is usually heated to temperatures as high as 2500 K. As the weld metal cools from this temperature, (I) in the temperature range 2300 to 1800 K, the dissolved oxygen and deoxidizing elements in liquid steel react to form complex oxide inclusions of 0.1-1μm size range; (II) in the temperature range 1800 to 1600 K, solidification to δ ferrite starts and envelops these oxide inclusions and later on (III) transformation to austenite; (IV-VII) in the temperature range 1100 to 500 K, the austenite transforms to different ferrite morphologies such as allotriomorphic ferrite, Widmanstätten ferrite and acicular ferrite.

A typical low alloy steel weld metal microstructure is shown in Figure 1b. The optimum strength and toughness in steel welds are achieved by maximizing the amount of acicular ferrite. It is well known that the inclusions in steel welds promote the formation of acicular ferrite (1,2,4). Hence, there is a great need to understand the inclusion formation in liquid steel and its effect on transformation of austenite to various

Figure 1 – (a) Schematic diagram of continuous cooling transformation diagram showing the development of weld metal microstructure in the low alloy steels. (I) inclusion formation, (II) solidification of liquid to δ ferrite, (III) fully austenitic structure, (IV) nucleation of allotriomorphic ferrite, (V) growth of allotriomorphic ferrite all along the austenite grain boundaries, (VI) Widmanstätten ferrite formation and (VII) acicular ferrite formation. (b) Optical microstructure of typical low alloy steel weld metal microstructure showing (al) allotriomorphic ferrite, (wf) Widmanstätten ferrite and (ac) acicular ferrite.
ferrite morphologies. The present paper pertains to (a) development of a model to describe the oxide inclusion formation in steel welds, (b) effect of inclusions on transformation kinetics of austenite to various ferrite morphologies and (c) theoretical discussion of the austenite grain development from δ ferrite.

INCLUSION FORMATION MODEL

The inclusion formation in low-alloy steel weld is controlled by the concentrations of dissolved gases (such as oxygen and nitrogen) and deoxidizing elements (such as aluminum, titanium, silicon and manganese). The dissolution of gases into the liquid steel is also affected by the arc–atmosphere conditions around the molten steel. For example, it is known that the dissociation of diatomic gases into monatomic species affects the dissolution of gases into liquid steel (3,5). However, in this work the inclusion formation is described for given nominal concentrations of dissolved gases and deoxidizing elements. The inclusion formation will also be controlled by the rate at which the weld cools. Therefore, in this work the inclusion formation is described as a function of weld metal cooling conditions. The important inclusion characteristics are oxidation sequence, inclusion composition, size, number density, and residual amount of deoxidizing elements (4, 6-8). The residual amounts of deoxidizing elements in solid solution, such as Al, Ti, Si and Mn, that remain after the oxidation reaction, control the transformation of austenite to ferrite at lower temperatures. Previous attempts (4, 6, 8) to predict inclusion composition assumed a fixed oxidation sequence of aluminum to Al2O3, titanium to Ti2O3, silicon to SiO2 and manganese to MnO. This assumption was based on the ranking of the oxides Al2O3, Ti2O3 (or to other type of Ti3O7 oxides), SiO2 and MnO as per their standard free energy of formation. However, the oxidation sequence may change as a function of concentrations of reacting elements. Moreover, previous works did not model other inclusion parameters such as inclusion number density, size and residual amounts of deoxidizing elements.

Hence, a comprehensive inclusion model based on fundamental ladle deoxidation reactions, thermodynamic data of oxides, and principles of overall transformation kinetic theories (7) is extended to the continuous cooling conditions of weld. The steps involved in the new inclusion model are (a) calculation of driving force for an oxide formation, (b) calculation of homogeneous nucleation rates, (c) calculation of growth rates, and (d) calculation of overall oxidation kinetics as a function of temperature and composition. The methodologies for each of the above calculations are published elsewhere (7). In this paper, the isothermal oxidation kinetic calculations are applied to the continuous cooling conditions.

Isothermal Overall Oxidation Kinetics

Overall isothermal oxidation kinetics at a particular temperature T1 involving a deoxidizing element (M) and oxygen (Ω) dissolved in liquid steel to produce an oxide inclusion of MxOy is given by following equation (7):

\[ \zeta = 1 - \exp \left\{ - \left( \frac{8\pi}{15} \right) I_v \left( \alpha_i^* \right)^{\frac{3}{2}} \frac{T_1}{\Omega} \right\} \]  

where \( \zeta \) is the extent of reaction, \( I_v \) is the homogeneous nucleation rate of oxide \( M_xO_y \), \( \alpha_i^* \) is the parabolic thickening rate of oxides in liquid steel, \( t \) is the time of reaction at temperature \( T_1 \), and \( \Omega \) is the equilibrium volume fraction of \( M_xO_y \) at \( T \). The volume fraction is given by the following equation:

\[ \Omega = \left( c'_{M} - c''_{M} \right) / \left( c'_{M} - c''_{M} \right) \]  

where \( c'_{M} \) is the nominal concentration of the deoxidizing element, \( c''_{M} \) is the concentration of deoxidizing element in the liquid in equilibrium with the inclusion, and \( c'_{M} \) is the concentration of deoxidizing element in the inclusion. The equilibrium concentration of deoxidizing element, \( c''_{M} \), is obtained by the construction of a tie line to the deoxidation phase diagram (7). In all the above calculations, multi phase equilibria between various oxides and liquid steel are not considered. The expressions for calculating other terms \( \alpha_i^* \) and \( I_v \) and are given in the reference (7). Equation 1 enables one to predict the isothermal oxidation kinetics for various oxides as a function of nominal composition and temperature.
Continuous Cooling Oxidation Kinetics

Isothermal oxidation kinetics is applied to continuous cooling conditions by the following steps: (a) approximate cooling curve equation given by Ion et al. (9) is used for calculating the cooling rates in the weld metal; (b) Kinetic calculations were carried out until the liquidus temperature, \( \sim 1800 \) K; (c) the reactions (i.e., sulfides and nitrides) that occur during the solidification are not considered in this work for simplicity; (d) extent of reaction obtained in equation 1 is extended to the continuous cooling by assuming the additivity of oxide volume fraction that forms as a function of temperature during weld cooling; (e) as the extent of the reaction of an oxide reaches 0.9, the liquid composition is reset to account for the depletion of oxygen and deoxidizing element concentration; (f) integrated number density of first forming oxide is taken as the final number density of inclusions in the steel weld, (This assumption is based on the observation of heterogeneous and layered inclusion structure (6)). The calculation of final inclusion number density is explained below:

\[
Number\ Density = \sum_{i=T_{win}}^{T_{end}} \frac{V_i^{oxide}}{\Delta t_i} \times \Delta t
\]

where \( V_i^{oxide} \) is the nucleation rate of first forming oxide as a function of temperature, \( \Delta t_i \) is the time spent at each temperature from, \( T_{start} \), the temperature at which the first oxide starts to form to, \( T_{end} \), the temperature at which the extent of the first oxide, \( \zeta \), reaches 0.9.

The accumulated volume fraction of oxide inclusions at the liquidus temperature is taken as the final volume fraction of oxides and the residual liquid composition at 1800 K is related to the amount of deoxidizing elements in solid solution. With the knowledge of the final inclusion volume fraction (\( V_f \)) and final number density, the average inclusion radius is calculated (\( V_f = Number\ Density \times (4/3) \times \pi \times r^3 \)). The flow chart illustrating the calculation routine is shown in Figure 2. The inclusion model was then applied to experimental conditions published by Kluken and Grong (6) and the predictions are compared with the experimental results given in the same reference.

EXPERIMENTATION

Effect Of Inclusions On Austenite To Ferrite Transformation Kinetics

The inclusion characteristics are known to influence the microstructure development in steel welds. Therefore, it is desirable to study only the influence of inclusion characteristics on the transformation kinetics of allotriomorphic ferrite, Widmanstätten ferrite, and acicular ferrite. In this work the transformation kinetics of allotriomorphic ferrite and acicular ferrite were measured for two different inclusion characteristics. Two submerged arc welds with two different inclusion characteristics were produced by varying the concentrations of silicon, titanium, aluminum and oxygen. The weld metal compositions and welding process conditions are given in Table I. The inclusion characteristics were measured in a Philips CM-12 transmission electron microscope by the extraction replica technique. The isothermal and continuous cooling heat treatments were performed on the samples extracted from the weld metal region in a Gleeble™ thermomechanical simulator. The weld metal hardenability and the prior austenite grain size were designed to be similar in both welds SW1A and SW2A.

<table>
<thead>
<tr>
<th>Welds</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>Ti</th>
<th>Al</th>
<th>O</th>
</tr>
</thead>
<tbody>
<tr>
<td>SW1A</td>
<td>0.10</td>
<td>0.81</td>
<td>1.61</td>
<td>0.018</td>
<td>0.015</td>
<td>0.084</td>
</tr>
<tr>
<td>SW2A</td>
<td>0.10</td>
<td>0.35</td>
<td>1.67</td>
<td>0.003</td>
<td>0.014</td>
<td>0.029</td>
</tr>
</tbody>
</table>

Theoretical time–temperature transformation diagram calculations (10) indicated the change in silicon levels between welds SW1A and SW2A may not influence the transformation kinetics of allotriomorphic ferrite and acicular ferrite. The austenite grain size was maintained at \( \sim 70 \) \( \mu \)m by austenitizing the samples at 1473 K for 0.6 ks. The continuous cooling transformation kinetics were measured in the 1 to 50 K s\(^{-1}\) range. The details of the isothermal heat treatments are given in Table II. Although the results of isothermal
Create Cooling Curve, Get Composition, Start Temperature Loop

Get temperature and time interval

Calculate Mole fraction of elements
Calculate Activity of various elements
Calculate Free energy of reaction
Calculate the equilibrium concentration
Calculate the growth rate of oxides

Is this the first oxide to form

Yes

Calculate Nucleation Rate

Nucleation Rate of Oxides = Integrated Nucleation Rate of Oxide

No

Calculate Extent of Reaction for the time interval for all oxides:

is $\zeta$ of any oxide $\geq 0.9$

Yes

is this the first oxide

Yes

Set the final inclusion number density = Integrated Nucleation Rate of Oxide

No

Reset the composition

$T < 1800$ K

Yes

Calculate the relative partitioning of elements between inclusions and the liquid

Stop

No

Figure 2. The flow chart illustrating the application of isothermal oxidation kinetic calculation to continuous cooling conditions. The calculation starts with a nominal weld composition at 2300K. The cooling curves are then divided into small isothermal steps. The kinetic calculations are performed at each temperature and time interval and additivity of oxide volume fraction is assumed.

d transformation kinetics of allotriomorphic ferrite and acicular ferrite are published elsewhere (11), the results are discussed again with reference to the results of inclusion model. The kinetic measurements are performed by dilatometric techniques and the details of the procedure are published elsewhere (11). The measured isothermal kinetics are then compared with the theoretical transformation kinetic equations.

Table II. Tabulation of heat treatments and measured austenite grain size.

<table>
<thead>
<tr>
<th>ID</th>
<th>Austenitization treatment</th>
<th>Austenite grain size, $\mu$m</th>
<th>Isothermal transformation</th>
</tr>
</thead>
<tbody>
<tr>
<td>HT1</td>
<td>1473 K, 0.6 ks</td>
<td>71 ± 4</td>
<td>953 K, 1.2 ks</td>
</tr>
<tr>
<td>HT2</td>
<td>1473 K, 0.6 ks</td>
<td>71 ± 4</td>
<td>843 K, 0.6 ks</td>
</tr>
</tbody>
</table>
RESULTS AND DISCUSSIONS

Inclusion Model Evaluation

The inclusion model developed in this work was evaluated with experimental results. The cooling rate data necessary for the inclusion calculation was obtained by using the equation given by Ion et al. (9). However, it is important to note that these cooling rate calculations are approximate and ideally one should use the weld metal cooling rate calculations considering the fluid flow and mass transfer in the weld pool region (12). The calculated inclusion characteristics for welds SW1A and SW2A are compared with experimental data in Table III. The calculations are also applied to the experimental data of Kluken and Grong (6, 13). The comparison of experimental results of inclusion volume fraction, composition, final number density, and inclusion diameter published by Kluken and Grong (6) and the calculations of the present inclusion model are shown in Figure 3.

![Figure 3 - Comparison of experimental inclusion characteristics measured by Kluken and Grong (6) and the calculated inclusion characteristics by the present inclusion model. (a) inclusion composition, (b) inclusion volume fraction, (c) inclusion size and (d) inclusion number density.](image-url)
Table III. Comparison of experimental (denoted as Exp.) and calculated (denoted as Cal.) inclusion characteristics of welds SW1A and SW2A.

<table>
<thead>
<tr>
<th>Weld No.</th>
<th>Size, μm</th>
<th>Num. Density x10^16 m^-3</th>
<th>Inclusion Composition, wt.%</th>
</tr>
</thead>
<tbody>
<tr>
<td>SW1A</td>
<td>0.84</td>
<td>0.70</td>
<td>3.0</td>
</tr>
<tr>
<td>SW2A</td>
<td>0.53</td>
<td>0.60</td>
<td>3.6</td>
</tr>
</tbody>
</table>

The comparisons (Table III and Figure 3) show that the present inclusion model shows a trend similar to that of the experimental data. However, the predictions are not exact. For example, (a) the predicted inclusion volume fractions for weld 5 and 8 of Kluken and Grong (6,13) are quite far from the line of unit slope and (b) the calculated number density is always lower than that of the experimental data. These inaccuracies of the inclusion model could be attributed to the following reasons:

1. Preliminary work (14) with the use of more sophisticated numerical calculations, considering fluid flow and heat transfer conditions, indicated that the cooling rates given by the equation in the reference 9 are lower than actually expected in the weld pool. Therefore, the agreement between calculations and experiments for the number density and size predictions are expected to improve with more accurate cooling rate values.

2. The present model does not consider some other complex oxides like MnO.SiO₂ (Rhodonite) and Al₂O₃SiO₂ (Andalusite) formation during liquid steel cooling.

3. The present inclusion model does not consider the reactions that occur during solidification, e.g., the MnS, TiN formation during the final stages of solidification and in the solid state.

Further work is under progress to remove the above limitations in the present inclusion model (14). In addition to the oxides that form in the liquid state, precipitates like TiN, TiC and other carbo-nitrides form in solid state will be considered. This precipitation will be related to the amount of residual elements that are present after the oxidation reaction. The present inclusion model can estimate the liquid composition at the end of oxidation reaction during weld cooling at 1800 K. The comparison (Figure 4) of experimental measurement and theoretical estimation of aluminum in solid solution shows a fair correlation. However, an excellent agreement is observed with the predicted titanium levels in solid solution.

![Figure 4 - Comparison of experimental aluminum and titanium in solid solution measured by Kluken and Grong (6) and the calculated estimates by the present inclusion model.](image-url)
Figure 5 – Comparison of measured continuous cooling transformation characteristics from welds (a) SW1A and (b) SW2A. The austenitization condition was 0.6 ks at 1473 K. The austenite grain sizes in both the samples were similar at ~70 μm. The measured cooling curves (dotted lines) are also shown. The experimental data points and fitted lines for 1%, 10% and 20% transformation are shown in the plot.

Effect Of Inclusions On Transformation Kinetics
Since the experimental inclusion characteristics of SW1A and SW2A (Table III) are different, a change in the transformation kinetics of ferrite from austenite is also expected. The comparison of CCT behavior is shown in Figure 5. The plots show that the transformation characteristics are almost identical for all the cooling rates in both the weld samples. Although, this result suggests that there is no change in the transformation kinetics due to a change in the inclusion characteristics, it is difficult to differentiate the effect of inclusions on allotriomorphic ferrite and acicular ferrite kinetics. Therefore, further isothermal transformation kinetic studies were performed.

Effect of inclusions on allotriomorphic ferrite transformation kinetics
To study the effect of inclusions, one needs to study the measured transformation kinetics in terms of theoretical kinetic equations in the early stages of allotriomorphic ferrite growth. The transformation kinetics of allotriomorphic ferrite are controlled by nucleation and growth mechanisms. Bhadeshia et al. (15) presented a model based on nucleation and growth. The allotriomorphic ferrite, before site saturation, is modeled as disc with the faces parallel to the austenite grain boundary plane and is assumed to grow in a radial direction. The final expression relating the extent of reaction \( \xi \), defined as measured volume fraction / equilibrium volume fraction, the parabolic thickening rate (\( \alpha_1 \)), the grain boundary nucleation rate per unit area (\( I_B \)), and the reaction time (\( t \)) is given as

\[
-ln\{1-\xi\} = f\{\eta \alpha_1, I_B, t\} \times 2(S_v/\phi) \alpha_1 \sqrt{t}
\]

where \( \eta \) is the aspect ratio of the discs of allotriomorphic ferrite, \( S_v \) is the austenite grain boundary area per unit volume and \( \phi \) is the equilibrium volume fraction expected at a particular temperature and

\[
f\{\eta \alpha_1, I_B, t\} = \int_{\theta_0}^{\theta} \left[ 1 - \exp \left( -0.5 \pi I_s (\eta \alpha_1)^2 \eta^2 (1 - \theta^4) \right) \right] d\theta
\]

where \( \theta = y/(\alpha_1^{0.5}) \) and \( y \) is the distance from the grain boundary plane. The term \( f(\eta \alpha_1, I_B, t) \) approaches unity as the site saturation occurs, after which, the kinetics are controlled by the parabolic thickening rate \( \alpha_1 \) and \( S_v \), until the overlap of diffusion profiles occurs (soft impingement). The equation 3 cannot be applied after soft impingement (15). It is known that the austenite grain boundaries are energetically preferred sites for the allotriomorphic ferrite nucleation compared to the inclusions within the austenite grain. However, the inclusions that lie along a austenite grain boundary plane may modify the inherent nucleation potency of the boundary. As a result, the kinetics of allotriomorphic ferrite may vary with inclusion characteristics, as suggested in reference (1). The allotriomorphic ferrite transformation kinetics measured in the steel welds with two different inclusion characteristics (similar austenite grain size) are analyzed with equation 3.
The measured volume fraction of allotriomorphic ferrite was converted into extent of reaction, $\xi$. The $-\ln(1-\xi)$ term of equation 3 is plotted as a function of the square root of transformation time, $t^{0.5}$. According to equation 3, this term is expected to increase non-linearly until nucleation site saturation occurs, where $f(n\alpha, I_B, t^1) = 1$. Only during this initial stage, the inclusions may influence the nucleation of allotriomorphic ferrite. Hence, for a similar composition and austenite grain size (or $S_v$) the time for site saturation, $\tau$, is expected to be different if the inclusions have an influence on $I_B$. The plots of $-\ln(1-\xi)$ terms against $t^{0.5}$, for SW1A and SW2A in HT1 treatment, are shown in Figure 6. The time for site saturation ($\tau$) is marked in the plots. The plots show only a small difference in the time for site saturation. According to equation 3, after the site saturation, the $-\ln(1-\xi)$ term will vary linearly with a slope of $(2S_v\alpha_1)/\phi$. The slopes after site saturation are identical indicating that both alloys have the same values of $(2S_v\alpha_1)/\phi$.

This result indicates that the transformation kinetics of SW1A and SW2A are approximately the same and there is no significant effect of the inclusion characteristics on the allotriomorphic ferrite transformation kinetics. However, in real welding conditions, the inclusions may indirectly influence the allotriomorphic ferrite kinetics. For example, as observed by Fleck et al. (16) for identical weld compositions, increase in the inclusion volume fraction reduces the austenite grain size. The reduction in the austenite grain size leads to an increase in the value of $S_v$ and the transformation kinetics are enhanced. Previous work (11) has clearly shown that the isothermal allotriomorphic ferrite transformation kinetics are quite sensitive to the austenite grain size.

**Weldmetal austenite grain size**

Since the austenite grain size affects the transformation kinetics to a greater extent, it is important to understand the austenite grain development in low alloy steel welds. The austenite grain development during weld cooling is related to two reactions i.e., (a) transformation of $\delta$ ferrite to austenite during cooling by nucleation and growth at $\delta$ ferrite grain boundaries and (b) austenite grain growth after the completion of the $\delta$ ferrite to austenite transformation during weld cooling. In this work, a theoretical analysis was performed to relate the driving force for transformation of $\delta$ ferrite to austenite ($\Delta G^{\delta\rightarrow\gamma}$) to the experimental austenite grain size measured by Evans (17). Evans has measured the prior austenite grain size as a function weld
Figure 7—The variation of austenite grain size with driving force for transformation of δ ferrite to austenite ($\Delta G^{\delta \rightarrow \gamma}$). The dark line is the fitted line with a relation given by the equations 5 and 6. Experimental data (open circles) are from reference 17.

The experimental austenite grain size (in the units of meters) and the calculated $\Delta G^{\delta \rightarrow \gamma}$ (in the units of J mole$^{-1}$) are fitted to equations 5 and 6. The fitted constants are as following: $A=0.48974$; $B=1.356\times10^8$; $C=4.0869\times10^8$; and $D=3525.4$. The correlation between the predicted and experimental data is shown as a function of $\Delta G^{\delta \rightarrow \gamma}$ in Figure 7. This result shows that austenite grain size estimation for a constant heat input can be made with a knowledge of $\Delta G^{\delta \rightarrow \gamma}$ from thermodynamic calculations and demonstrates the sensitivity of the austenite grain development to the driving force for transformation of δ ferrite to austenite. However, this simple calculation has to be modified to the continuous cooling conditions and to a change in the weld heat input. Unfortunately, there is no detailed analysis of the austenite grain size variation with metal composition for the same welding heat input. The driving force for transformation of δ ferrite to austenite ($\Delta G^{\delta \rightarrow \gamma}$) for the same compositions was calculated from ThermoCalc™ (18) software. In this calculation, the following assumptions are made for simplicity:

(a) Differential thermal analysis (DTA) performed on low alloy steel weld metal compositions (19), similar to that of Evans(17) weld metal compositions, illustrated that solidification to δ ferrite and transformation to austenite completed at 1661 K as the weld metal cooled at 1 Ks$^{-1}$ from the liquidus temperature. It is important to note that the weld metal cooling rates are higher than 1 Ks$^{-1}$ (> 50 Ks$^{-1}$). Since there are no experimental data on the transformation start temperature of austenite from δ ferrite, in this analysis, the transformation from ferrite to austenite is assumed to occur isothermally at a particular undercooling and this temperature is taken as 1673 K. However, in future work the method described below can be easily modified for continuous cooling conditions;

(b) The austenite grain size is assumed to be inversely proportional to nucleation rate of austenite at the δ ferrite grain boundary, where the nucleation rate is a function of $\Delta G^{\delta \rightarrow \gamma}$ as shown below.

$Grain\ Size = 1/(nucleation\ rate)^A$

where $A$ is the constant exponent and the nucleation rate is given by the expression

$nucleation\ rate = B + C \exp\{-D/(\Delta G^{\delta \rightarrow \gamma})^2\}$

where $B$ and $C$ are constants.

(c) In this analysis since the austenite nucleation events occur at δ ferrite grain boundary during rapid weld cooling conditions, the alloying element partitioning during nucleation is not considered. The compositions of austenite and ferrite are assumed to be the same.
Effect of inclusions on acicular ferrite transformation kinetics

It is known that inclusions rich in titanium promote the formation of acicular ferrite. The experimental and calculated inclusion characteristics presented in Table III show that the inclusions in weld SW1A are coarser and richer in titanium than the inclusions in weld SW2A. Therefore a resulting change in the transformation kinetics of acicular ferrite is expected. The measured kinetic plot of acicular ferrite (in SW1A and SW2A during HT2 heat treatment) is shown in Figure 8. The measured volume fraction of acicular ferrite at 843K is converted into the extent of acicular ferrite transformation, $\xi$. The extent of acicular ferrite transformation is given by the ratio of the measured volume fraction to the expected volume fraction of acicular ferrite. The value of expected volume fraction of acicular ferrite is given by the $T_o$ boundary at 843K (i.e., where the austenite and ferrite with similar composition have equal free energy) with a ferrite strain energy of 400 J mole$^{-1}$ (20). The acicular ferrite kinetic results showed that the transformation kinetics are indeed accelerated in weld SW1A compared to SW2A.

According to the above observation, the acicular ferrite microstructures of SW1A and SW2A are expected to be different after the HT2 heat treatment. However, the microstructures in the samples showed no apparent difference. This is because the microstructure after this heat treatment could not be quenched to the room temperature. The residual austenite in both SW1A and SW2A transformed to acicular ferrite during cooling from 843K to room temperature and the microstructure difference at 843K was lost. Hence, the difference in transformation kinetics to acicular ferrite due to a change in inclusion characteristics can not be discerned from the final microstructure in these welds. However, the direct kinetic measurement by dilatometry (11), at a particular temperature, indicated a difference in kinetics does exist.

The accelerated transformation kinetics of austenite to acicular ferrite due to a change in the inclusion characteristics may be related to heterogeneous nucleation of ferrite on inclusions. The heterogeneous nucleation of ferrite on inclusions may be related to the following mechanisms: (a) low lattice misfit between the inclusion and ferrite, (b) inclusions acting as high energy inert substrates, (c) chemical inhomogeneity around the inclusions, and (d) localized thermal strains around inclusions. It is known that Ti-rich (Ti$_x$O$_y$ type) and Galaxite type compounds on the inclusion surface may promote the formation of acicular ferrite (4). The oxidation sequence calculated by the inclusion model is given in Table IV. The calculations show that the proportion of MnOAl$_2$O$_3$ (Galaxite) is more in SW1A than that of SW2A. However, in both welds the formation of Ti$_x$O$_y$ type oxide is predicted. Therefore, the results from inclusion model suggest that the
presence of $\text{Ti}_2\text{O}_y$ type oxide may not be as important as Galaxite in determining the acicular ferrite transformation kinetics in the welds SW1A and SW2A. However, the existence of MnOAl$_2$O$_3$ on the surface of inclusions has to be confirmed by electron diffraction analysis in future work. Moreover, the above inclusion formation model has to be coupled with the theoretical overall transformation kinetic equations of acicular ferrite to estimate transformation kinetics as a function of weld metal composition, inclusion characteristics and weld metal cooling rate.

Table IV – Oxidation sequence calculated by the inclusion model for welds SW1A and SW2A (Table 1). The table indicates the type of oxides that form as a function of temperature, as the welds SW1A and SW2A cool from 2300K.

<table>
<thead>
<tr>
<th>SW1A Weld</th>
<th>SW2A Weld</th>
</tr>
</thead>
<tbody>
<tr>
<td>Temperature, K</td>
<td>Oxide</td>
</tr>
<tr>
<td>2134</td>
<td>Al$_2$O$_3$</td>
</tr>
<tr>
<td>2109</td>
<td>Ti$_3$O$_5$</td>
</tr>
<tr>
<td>2094</td>
<td>SiO$_2$</td>
</tr>
<tr>
<td>2087</td>
<td>MnOAl$_2$O$_3$</td>
</tr>
<tr>
<td>2080</td>
<td>MnOAl$_2$O$_3$</td>
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<td>2038</td>
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<td>2033</td>
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</tr>
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<td>2015</td>
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</tr>
<tr>
<td>1990</td>
<td>MnOAl$_2$O$_3$</td>
</tr>
<tr>
<td>1946</td>
<td>MnOAl$_2$O$_3$</td>
</tr>
<tr>
<td>1836</td>
<td>MnOAl$_2$O$_3$</td>
</tr>
</tbody>
</table>

General Discussions

The work presented in this paper has shown that the microstructure development in low alloy steel weld depends on various reactions that occur as a function of weld cooling. Moreover, it has also demonstrated that by coupling heat and mass transfer models with ladle deoxidation and transformation kinetics one can understand these reactions and theoretically estimate the microstructure development in steel welds. However, the discussions in the above paper have not addressed the external and internal mechanical forces that develop in steel weldments. As pointed out by Kirkaldy (21) the microstructure development and mechanical forces interact with each other during steel processing and service conditions. For example, it is known that acicular ferrite transformation behavior is quite sensitive to external forces. In the work by Babu and Bhadeshia (22), the acicular ferrite microstructure evolution in a steel weld was found to comply with an externally applied elastic stress and the samples exhibited transformation induced plasticity during isothermal transformation. This phenomenon will have due influence on the residual stress development in steel welds. Therefore, future modeling work on the development of microstructure in low alloy steel welds has to consider the effect of external and internal stress conditions.

CONCLUSIONS

A new model for inclusion formation in steel welds has been developed. This model is based on thermodynamic data of liquid steel and oxides, and overall oxidation kinetic principles. The weld metal inclusion characteristics including volume fraction, number density, size, composition, oxidation sequence, and amount of deoxidizing elements in solid solution can be estimated as a function of nominal weld metal composition and weld metal cooling rate. The austenite grain size development in steel welds is related to the driving force for transformation from $\delta$ ferrite to austenite. The transformation kinetics of allotriomorphic ferrite and acicular ferrite were measured in two steel welds with two different inclusion...
characteristics but with similar austenite grain size. The allotriomorphic ferrite transformation kinetics were found to be independent of inclusion characteristics. Acicular ferrite transformation kinetics were found to be dependent on the inclusion characteristics. The accelerated acicular ferrite transformation may be related to the presence of Galaxite oxides in the inclusions.

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