The Austenite/Ferrite Front Migration Rate during Heating of IF Steel

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The austenitization of interstitial free (IF) steel was investigated through real time imaging of evolving surface relief structures through a hot-stage confocal scanning laser microscope. Nucleation was observed at grain edges but site saturation did not occur. As expected the, the migration rate was found to be controlled by an interface reaction and the rate was, within the experimental errors, nearly indistinguishable between IF steels, pure Fe and previously reported rates for pure Fe with activation energies between 174–180 kJ/mol. There was no apparent effect of solute drag caused by the trace solute elements in the IF steel, but the non-metallic oxide and nitride particles present in the IF steel appeared to impede the front migration and caused it to become jagged in appearance.

KEY WORDS: confocal scanning laser microscopy (CSLM); austenitization; interstitial free (IF) steel; interface migration.

1. Introduction

The rate and mechanism of solid-state phase transformations in steels is of fundamental importance to designing heat treating procedures, consisting of various steps of heating, annealing, cooling, quenching, and tempering. The austenitization of steels is one of these steps, but may proceed too rapidly in many cases for classical investigative methods which involve soaking a sample for various times and subsequently evaluating the resulting microstructure through sectioning, polishing, and optical or electron microscopy. In many cases, the austenite cannot be retained for ambient temperature characterization, as it decomposes upon cooling into a variety of phases and microstructures. Significant improvement to current understanding of austenite formation is possible using in situ observational techniques to study the behavior of migrating interfaces, in terms of rate, evolving morphology, and effect of initial structure, during the phase change. One of these techniques, more recently developed by Emt1 for high temperature observation of metals, is the hot stage Confocal Scanning Laser Microscope (CSLM), which will be utilized in the current investigation.

Speich and Szirmae2 conducted a conventional study of austenite formation in iron and iron-carbon alloys, by heat treating samples and subsequently analyzing the resulting microstructures, and developed a model for the interface migration rate in pure iron, where the rate is controlled by the interface reaction (uncorrelated jumping of atoms across the interface) only:

\[ v = \left( \frac{\delta v}{kT} \right) \exp \left( \frac{\Delta S^*}{k} \right) \exp \left( \frac{\Delta H^*}{kT} \right) \Delta g^{α→γ} \] ........(1)

where \( v \) is the velocity of the interface, \( \delta \) is the boundary width, assumed be 5 Å, \( v \) is the jump frequency, \( \Delta S^* \) and \( \Delta H^* \) are the activation entropy and enthalpy for migration of an atom across the interface, and \( \Delta g^{α→γ} \) is the driving force for the reaction (per atom). The activation enthalpy for atoms across the boundary was approximated by the activation enthalpy for grain boundary diffusion in pure iron, with a value of 2.76×10^19 J/atom (166 kJ/mol). This value is also in the same order as, though somewhat smaller than, more recent data for activation energy for migration of high angle grain boundaries in α-iron (200–400 kJ/mol). Because the jump frequency and activation entropy are not well known, the term \( v \cdot \exp \left( \frac{\Delta S^*}{k} \right) \) was determined experimentally from data for pure iron; a maximum interface velocity of 1.6 cm/s was reported for 950°C, where the driving force is 4.18×10^10 J/atom, and so this term is approximated to be 1.65×10^17 s⁻¹. This model was compared to a dilatometric study of fully ferritic, ultra low alloy steels, during constant heating, by Caballero et al.3 In this study, the interface control model appears to overestimate the rate of the transformation for pure iron. One possible reason for the disagreement between calculated and experimental data is that it required Speich and Szirmae’s model to be combined with assumed microstructural data that may not necessarily be valid; in extrapolating a volume fraction it was necessary to include certain assumptions about the microstructure and nucleation phenomena (sites, kinetics). Observation of the evolving microstructure during the transformation may be able to resolve this disagreement.

In addition to pure iron, the study of ultra low alloy steels like the interstitial free (IF) alloys may provide additional insight into the mechanism of this transformation at a fundamental level. Because the processing of IF steel is limited to strengthening through grain refinement, knowledge of the mechanism of the ferrite to austenite phase transforma-
tion may also improve the performance of these alloys. Ultra Low Alloy IF steels contain less than 0.1 wt% of any alloying element, and less than 0.01 wt% of the interstitial elements like carbon and nitrogen. It can then be determined if impurities at this level have a measurable effect on the transformation mechanism, to see if the even more limited impurities in laboratory pure iron samples could cause the previously noted differences between experiment and theory. There are a number of effects that these impurities could have. If there is free energy required for diffusion of solute atoms (there are small amounts of impurity elements even in a high purity iron) to and from the interface during transformation, and this energy is significant relative to the energy required for the interface reaction, the transformation could be controlled by long-range diffusion. This kind of mechanism is the basis of interface migration models developed by Zener, Hillert, and Trivedi. Another possibility is that atoms which may sit preferentially at the interface (i.e. grain boundary segregation) may retard the motion of the interface, since these atoms also require the dissipation of free energy for diffusion in order to move along with it. This phenomenon, often referred to as ‘impurity drag’ or ‘solute drag’, was characterized by John Cahn in 1962. Other effects, such as the requirement for a certain interface structure or migration of defects at the interface, have been considered more recently. In the particular case of IF steel, the addition of specific elements such as aluminium and titanium result in small inclusion particles. The effects of these inclusion particles (e.g. oxides, nitrides, sulfides) on migrating phase boundaries have been recognized as a useful method for refining the microstructure of steels. Therefore, an understanding of the effect of non-metallic inclusions on the ferrite to austenite transformation will further advance processing techniques.

The objectives of this study are to compare migration rates of austenite measured through evolving surface relief patterns in the CSLM with Eq. (1), developed by Speich and Szirmae, in pure Fe and IF steels. Any observed differences may be used to assess the role of solutes present in IF steels in light of the effects described above.

2. Materials and Experimental Approach

2.1. Materials

Two interstitial free steels, IF-A and IF-B (chemistries shown in Table 1), and 99.995% pure iron (Alfa-Aesar Puratronic® brand) were the focus of this study. SEM images of the three sample types are shown in Figs. 1(a)–1(c). Both IF steels contain less than 0.006% C, and were chosen to significantly reduce or eliminate the effect of carbon diffusion on the rate of any phase changes. The high-purity iron samples were used as a comparison to see if the small residual levels of carbon or any of the small alloying additions in the IF steels had any significant effect on the austenite front migration rate. Industrially obtained ‘lollipop’ samples of the IF steels were sectioned into ~2 mm thick slices, and then cut into ~4×4 mm squares. The pure iron was obtained from a 4 mm diameter rod, cut into ~2 mm thick slices. After removing any burrs from cutting, one face on each of the samples was ground with SiC and polished with 9, 3, and finally 1 micron diamond particles. Polished samples were cleaned thoroughly in an ultrasonic cleaner, using first a residue-free and corrosion-inhibited detergent solution, and then ethanol. Just before an experiment was to be run, each sample was etched with a 1% Nital solution (fully concentrated nitric acid dissolved in 200-proof ethanol) for up to one minute to reveal grain boundaries. Also revealed by this process are very small second phase particles, shown in Figs. 2(a)–2(b). These non-metallic particles were observed in the SEM, and EDX analysis (Tables 2(a)–2(b)) has indicated that the particles are aluminum-rich oxides and titanium nitrides. The oxides also contained small amounts of magnesium, calcium, titanium, and/or manganese.

Table 1. Composition of interstitial free (IF) steels in current investigation

<table>
<thead>
<tr>
<th>Element</th>
<th>C</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Si</th>
<th>Cr</th>
<th>Ni</th>
<th>Cu</th>
<th>Mo</th>
</tr>
</thead>
<tbody>
<tr>
<td>IF-A</td>
<td>0.015</td>
<td>0.05</td>
<td>0.001</td>
<td>0.001</td>
<td>0.002</td>
<td>0.05</td>
<td>0.05</td>
<td>0.005</td>
<td></td>
</tr>
<tr>
<td>IF-B</td>
<td>0.006</td>
<td>0.06</td>
<td>0.01</td>
<td>0.006</td>
<td>0.02</td>
<td>0.08</td>
<td>0.05</td>
<td>0.006</td>
<td></td>
</tr>
<tr>
<td>Element</td>
<td>Sn</td>
<td>Al</td>
<td>Nb</td>
<td>V</td>
<td>Ca</td>
<td>Pb</td>
<td>Ti</td>
<td>Cr</td>
<td>N</td>
</tr>
<tr>
<td>---------</td>
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<td>----</td>
<td>----</td>
<td>----</td>
<td>----</td>
<td>----</td>
</tr>
<tr>
<td>IF-A</td>
<td>0.004</td>
<td>0.05</td>
<td>0.001</td>
<td>0.002</td>
<td>0.002</td>
<td>0.0003</td>
<td>0.06</td>
<td>0.0005</td>
<td>0.006</td>
</tr>
<tr>
<td>IF-B</td>
<td>0.005</td>
<td>0.04</td>
<td>0.001</td>
<td>0.002</td>
<td>0.002</td>
<td>0.001</td>
<td>0.06</td>
<td>0.00001</td>
<td>0.006</td>
</tr>
</tbody>
</table>

Fig. 1. Starting microstructure of (a) IF-A steel, (b) IF-B Steel, (c) 99.995% pure Fe. Samples polished to 1/4 μm diamond and etched with 1% nital to expose grain boundaries.
2.2. High Temperature Confocal Scanning Laser Microscopy

Migrating interfaces were observed experimentally with a hot stage CSLM. The system used in the present investigation utilizes a gold-plated, ellipsoid shaped furnace (shown in Fig. 3(a)). Radiation from a halogen lamp at the lower focus is concentrated on the sample, which is located at the upper focus of the ellipsoid. Because the laser radiation used to form the image is much more intense than thermal radiation from the sample, and because the pin hole mentioned earlier is also able to deflect out of phase radiation that does not originate from the laser, a clear image of the surface can be obtained. The temperature is measured by a thermocouple placed near the surface of the sample (Fig. 3(b)). The response of this thermocouple is also used to control the intensity of the heating lamp, and the temperature profile during an experiment can be programmed as desired. Confocal optics, shown in Fig. 3(c), enable the detection of a strong signal from the focal plane while decreasing the intensity of signals not in the focal plane. A screen with a pin hole deflects any signals that originate at locations above or below than the focal point, or on any surface which is not perpendicular to the laser, and thus an image is created, with a ~5–10μm depth of field. The depth of field results from the variation of signal strength with elevation, roughness, or inclination, and is based primarily on the topography of the surface being viewed. Because this depth of field is fairly narrow, very small changes in elevation, inclination, or roughness will cause noticeable changes in signal strength, and thus the migrating fronts can be observed as long as some disruption of the surface occurs in the transformed region.

A vacuum pump and gas delivery system are used in tandem to evacuate the furnace of undesired gases before an experiment and provide a constant atmosphere of desired composition during an experiment. The gas used in this study is grade 5.0 Argon ($P_{O_2} \sim 10^{-5}$ atm), and has been passed through a gas cleaning system as follows: one steel tube filled with silica desiccant to remove any moisture and three steel tubes wrapped with heating tape and ceramic fiber insulation; one filled with copper turnings ($T \sim 700^\circ C$) and the next two filled with magnesium chips ($T \sim 600^\circ C$). The furnace chamber is repeatedly evacuated and refilled with the purified Ar gas from the gas cleaning system, which was subsequently allowed to flow for 1 h. The oxygen potential in the outlet gas stream was measured periodically during the study by a ceramic oxygen sensor in order to ensure that appreciable oxygen leaks were not present. Generally a $P_{O_2}$ of less than $10^{-10}$ atm is maintained. The gas flow rate during an experiment is on the order of 1 L/min.

Contrast changes observed by the CSLM during solid-state phase transformations are a result of topographical changes caused by surface deformation due to the displacive\textsuperscript{18–20} and/or dilatometric nature of the phase transformations, surface diffusion leading to smoothing of the surface,\textsuperscript{21} or the formation of non-metallic phases (liquid or solid oxides, sulfides, etc.) on the surface. The relationship between these two processes generally means that the faster the migration rate of the interface, the less likely it is to be smoothed out before it can be observed. In previous studies of CSLM observations, interface migration has been observed for a range of transformations in steels: Widmanstätten ferrite precipitation\textsuperscript{4,5}, austenite to $\delta$-ferrite transformation\textsuperscript{1,6}, austenite to pearlite\textsuperscript{4} and ferrite plus pearlite to austenite.\textsuperscript{5} In many of these cases, the surface structure is not preserved after the transformation, due to either rapid surface diffusion and high temperatures or sub-

\begin{table}[h]
\centering
\caption{Inclusion particle identification data}
\begin{tabular}{|c|c|c|c|c|}
\hline
Particle & Type & Diameter (μm) & Sample & Location & Figure \\
\hline
1 & Oxide & 3.75 & IF-A & Matrix & - \\
2 & Oxide & 2 & IF-A & Matrix & - \\
3 & Oxide & 3.5 & IF-A & Matrix & - \\
4 & Nitride & 1.25 & IF-A & Matrix & - \\
5 & Oxide & 1 & IF-B & Matrix & - \\
6 & Oxide & 2.5 & IF-B & Matrix & 3a \\
7 & Oxide & 1.75 & IF-B & G8 & 3b \\
8 & Nitride & 1.25 & IF-B & Matrix & - \\
9 & Nitride & 1 & IF-B & Matrix & 3a \\
10 & Nitride & 1.75 & IF-B & On Surface & - \\
\hline
\end{tabular}
\end{table}

\begin{table}[h]
\centering
\caption{Estimated composition (atomic percent) of inclusion particles as measured by Energy Dispersive X-Ray Spectroscopy (EDX)}
\begin{tabular}{|c|c|c|c|c|c|c|c|c|c|}
\hline
Particle & Ti & Ca & Al & Mg & Fe & Mo & O & N & N \\
\hline
1 & 0.28 & 1.17 & 29.78 & 5.30 & 3.19 & - & 60.38 & - & - \\
2 & 0.51 & 2.36 & 24.47 & 2.00 & 18.63 & - & 52.63 & - & - \\
3 & 0.34 & 0.21 & 30.25 & 5.96 & 0.99 & - & 62.27 & - & - \\
4 & 41.33 & - & - & - & - & - & 58.67 & - & - \\
5 & 1.07 & 2.18 & 21.29 & 3.94 & ignored & 0.69 & 70.86 & - & - \\
7 & 2.76 & 2.81 & 16.29 & 3.04 & ignored & - & 75.09 & - & - \\
8 & 53.06 & - & - & - & - & - & 46.91 & - & - \\
9 & 57.65 & - & - & - & - & - & 42.35 & - & - \\
10 & 41.46 & - & - & - & - & - & 58.54 & - & - \\
\hline
\end{tabular}
\end{table}

Fig. 2. Typical inclusion particles found in the IF steels investigated. Composition of particles are given in Table 2(b). The particle in (b) is located at a ferrite/ferrite grain boundary.
sequent transformations upon cooling. Therefore, the exact nature of the relief can not be quantified and a general mechanism has not yet been proposed. Keeping this in mind, it appears to be appropriate to relate the surface relief fronts to the migration of interface boundaries.\textsuperscript{19,20}

2.3. Experimental Procedure

The surface of each sample was observed in the CSLM during heating. Samples were heated at 10 K/s to 700°C, held for 1 min to ensure homogeneity, and then heated at rates which varied from 3 K/s to 10 K/s through the temperature range 700 to 1000°C. Multiple identical experiments were run for some of the heating rate/sample combinations, in order to ensure reproducibility of the results. The measured temperatures were compared to the known melting temperatures of pure silver and copper, in order to calibrate the temperature measurements. The surface was continuously recorded by a SVHS video recorder at a rate of 30 frames per second. Tapes were then digitized using MGI VideoWave\textsuperscript{TM} software. The position of each migrating front was measured as the furthest point on the front from its origin. If the origin was not visible on the screen, the front was tracked from its first appearance in the field of view.

The rates were measured by using a ruler to measure distances directly on a flat-screen monitor while toggling the digital video back and forth. The growth distance of a particular front was consistently measured as the furthest distance on an observed austenite/ferrite interface from where it originated. If the origin was not visible on the screen, the front was tracked from its first appearance in the field of view.

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2.4. Temperature Control and Calibration

Because the interface reaction control model is very strongly dependent on temperature (a few degrees increase may increase the migration rate by an order of magnitude), a careful consideration of transformation temperatures is essential. Speich and Szirmae\textsuperscript{8} assumes no nucleation barrier for the initial formation of incoherent austenite precipitates in ferrite grains. Therefore, the initial assumption is that the transformation will begin almost immediately at the thermodynamically expected transformation temperature. Because of the nature of the temperature profile in the confocal microscope, where infrared radiation is concentrated at a point by reflection from the walls of the elliptical furnace chamber, it is possible for a significant difference in incident radiation on the sample and the thermocouple tip just above it. It also seems reasonable, since the furnace walls are water cooled and the flowing atmosphere is not preheated, that this thermal gradient will increase as the hot zone in the furnace increases in temperature. This thermal profile was observed in the context of peritectic solidification in iron alloys (in the current experimental setup) by McDonald,\textsuperscript{7} although this data is limited to one dimensional horizontal displacement of the sample within the chamber. The thermocouple and sample are displaced vertically in the current study, and this method for measuring the gradients can not be used. Thus a different method for calibration was carried out, in which very small flakes of pure metals (whose melting temperatures are close to the austenite formation temperature, and are accurately known) were placed on the surface during an experiment, and their melting temperatures were measured and compared to the Fe\textsubscript{\(\alpha\rightarrow\gamma\)} transformation temperature. Fe\textsubscript{\(\alpha\rightarrow\gamma\)} has been assumed to be that of pure iron, i.e. 1187 K, as there is no evidence to suggest that the small amounts of alloying/impurity elements in IF-A and IF-B have any significant effect on this temperature. This is further supported by the fact that the pure Fe samples did not transform at temperatures outside the experimental scatter for the IF steels. The heating rate during these trials was 20 K/min, which is slow enough to ensure that melting is observed nearly isothermally, as the process of melting may require significant time due to inertial and cohesive force restrictions of the metal to visibly spread out onto the surface as it melts. Thus accurate meas-
measurements of the melting temperature can be made. An example of such an experiment is provided in Figs. 4(a)–4(g). Since the austenite formation, and melting of the silver and then copper pieces are observed within a few hundred micrometers, it is assumed that there is no significant temperature gradient between them. This procedure was carried out several times, and the average measured temperatures for each transformation were compared to the thermodynamically expected temperatures. The departure from expected temperature was plotted vs. the expected temperature (Fig. 5) in order to see what kind of dependence existed. The average experimental error in measured temperature for each phase change was approximately 12 degrees, and it can be seen from this plot that the 3 points can be assumed to be linear within experimental error. Because metals do not experience superheating during melting, and because the difference between expected and measured temperatures varies consistently with increasing temperature, the assumption is made that the austenite formation also begins at the thermodynamically expected temperature and experiences no superheat at any of the heating rates.

This assumption is required in order to carry out a comparison with the model, because experimentally measured temperatures for the onset of austenite formation in the current study varied significantly (~860°C ±20°C) and the temperatures are not consistent from experiment to experiment. Accordingly, only the time and position measurement have been considered when comparing experimental data to the model. This also means that even if the actual thermodynamic transformation temperature varies slightly from that of pure iron, it will not affect the ability to compare experimental results to the model since the model is much more dependent on the relative effect of temperature on the free energy change rather than the absolute effect of tem-

Fig. 4. Series of confocal stills from a calibration experiment in which (a, b) austenite forms from ferrite, (c, d) silver melts, and (e–g) copper softens and melts.

Fig. 5. Difference between measured and expected temperatures for each phase transformation during calibration experiments. A linear relationship, to within experimental error, suggests that the austenite formation begins at the thermodynamically expected temperature, with no superheat required.
3. Results and Discussion

A series of confocal still images of austenite formation in the IF-A steel is shown in Figs. 6(a)–6(e). These images show a migrating austenite/ferrite interface from just after the moment that it first appears, until the entire viewable surface has been completely austenitized. Even at very early stages, the migrating interface is noticeably jagged in some places, and smooth in others. Examples of what is meant by jagged and smooth areas of a migrating boundary have been labeled in Figs. 6(c)–6(d) with the letters “j” and “s”, respectively. This becomes more pronounced as the austenite regions grow in size and the interface becomes longer. This appearance of a planar, though somewhat jagged, growth interface suggests that it is likely to be an incoherent boundary.\textsuperscript{22,23} The front that begins to grow is able to move through grain boundaries visible at the surface, and only stops by impingement with other migrating fronts. In this particular series of images, it is difficult to determine exactly whether the austenite nucleated at a second phase particle or at a grain boundary (both seem possible given the appearance of a small inclusion seen at the top of Fig. 4(a)) although it is quite clear that fronts do not begin growing at or near every grain boundary observable on the surface. Figure 7 shows the nucleation behavior of austenite precipitates; in Figs. 7(a), 7(b) nucleation occurs at a grain corner or edge (observable as the intersection of three grains at the surface) while Fig. 7(b) shows the nucle-
ation of a new precipitate at a previously existing austenite/ferrite interface. A significant implication of these observations is that there is no site saturation of nucleation (i.e. at grain corners) as has been suggested previously.\(^{8,10}\)

Nucleation does seem to occur at grain corners as expected, but not every such location results in a new precipitate. Because far fewer fronts nucleate and grow, and each front will migrate much further than if site-saturation existed, and the transformation would occur over a longer time period. In addition, the nucleation of fronts at existing interfaces means that nucleation does not occur all at the same time. This may explain, to some extent, why these previous models have overestimated the transformation rates measured by dilatometry for fully ferritic low-alloy steels (or pure iron) with a similar initial microstructure.\(^{10}\)

The location of new austenite boundaries is also shown in Fig. 6(e). It would appear that as growing austenite fronts impinge, the newly formed boundaries appear to contain some of the small inclusion particles. Accordingly, the behavior of interfaces as they interact with these particles may also be of interest; such an interaction is shown in Figs. 8(a)–8(d). While complete pinning of the migrating interface does not occur, there is definitely an observable hindrance to growth. The interface becomes stalled momentarily in the immediate vicinity of the particle, but eventually the growing austenite engulfs the particle and continues to grow unimpeded. While this is occurring, however, it is necessary for the boundary to deviate from a smooth, planar appearance, and it must also increase in length. These interactions may thus result in the jagged appearance of some regions of a migrating boundary seen at a lower magnification (i.e. in Fig. 6). In addition, a migrating interface that is stalled at a particle may be impinged upon by the rapidly moving interface of another precipitate, explaining why some inclusion particles seem to sit at the newly formed austenite/austenite boundaries at the completion of austenitization.

Figures 9(a)–9(c) present the front position vs. time for heating rates of 3, 5, and 10 K/s. From these plots it appears

![Fig. 8. Series of confocal stills showing an austenite/ferrite front that is momentarily “pinned” by an inclusion particle. Whited dotted lines indicate the approximate location of the migrating front.](image)

![Fig. 9(a). Comparison of experimental front migration data to the model for interface reaction controlled migration velocity, modified to calculate position vs. time, at a constant 3 K/s heating rate.](image)

![Fig. 9(b). Comparison of experimental front migration data to the Speich model for interface reaction controlled migration velocity, modified to calculate position vs. time, at a constant 5 K/s heating rate.](image)
that there is no appreciable difference in the extent of migration with time between these alloys. These results confirm that the minimal alloying/impurity elements in the IF do not affect the transformation kinetics and that no free energy is dissipated by long-range diffusion of these additional elements. The nature of this transformation would make it impossible to analyze the region near an interface for partitioning or segregation, because it cannot be preserved. However, even if the other elements are accounted for, the two-phase field at such dilute compositions is of negligible width in a phase diagram, and the $T_g$ temperature (thermodynamic requirement for partitionless transformation$^{24}$) would be nearly identical to the transformation temperature for pure iron. The presence of small non-metallic inclusion particles (e.g. oxides and nitrides) also means that significant amounts of several of the alloying/impurity elements in the IF steels and impurity elements in the pure iron are not dissolved in the matrix during austenitization. The fact that the elements that do remain in solution (e.g. copper, nickel, manganese, chromium) do not result in any significant difference between 99.995% pure iron and IF steels also suggests that, at the amounts contained in these alloys, there is not any appreciable solute drag at a migrating ferrite/austenite interface.

The calculated curves included in these graphs are based on the velocity in Eq. (1). In order to compare with position vs. temperature curves produced by the CSLM data, this relationship has been integrated numerically, with respect to time, using the midpoint rule with each time step equal to 0.002 of the total time. Because time is directly proportional to temperature ($T = K_t t$, where $K$ is a constant equal to the heating rate), this result can be used to produce the desired relationship between position and temperature. A comparison between experimental data and the interface control model is presented in Fig. 10 for several heating rates. The apparent agreement of the experimental data with the interface reaction controlled migration model, especially at earlier stages of growth, is consistent with a partitionless transformation at an incoherent boundary. It is also consistent with a lack of solute drag, since this model does not account for such effects. In previous work$^{25}$ on grain boundary migration in aluminum alloys it was found that the activation energy required for boundary migration will only “break away” from that of the pure material when the solute concentration reaches some minimum value. Thus it is reasonable to conclude that the solute concentrations in the IF alloys have not reached these values and that the activation energy and mechanism of austenite formation in IF steels is essentially identical to that of pure iron under the experimental conditions. Some important factors have been considered when comparing the experimental data to the model calculations. Two parameters have been allowed to vary slightly, in order to more closely follow the data. First, the activation enthalpy was adjusted slightly in each case relative to the value (166 kJ/mol) used by Speich and Szirmae.$^8$ The values required to fit the entire range of data at each heating rate only varied by 5–8% (174–180 kJ/mol), and the same range of activation enthalpies was valid at every heating rate. Given that such a wide range of activation enthalpies have been observed for the migration of a variety of high angle grain boundaries$^9$ it seems reasonable that some variation in this value from front to front would exist due to slight differences in the orientation of the austenite and ferrite at an interface. The curves which have been presented in Figs. 5 and 6 only show the lower bound (174 kJ/mol) of the activation enthalpy, the higher value (180 kJ/mol) would cause the curves to be shifted to the other side of the data spread. Second, it is unlikely that an austenite precipitate will be visible immediately when it nucleates and begins to grow, because they are very small and the resolution of the microscope limits the ability to see a precipitate less than one micron in diameter. It has been observed that there is not site saturation of nucleation at grain corners and so there is not a large density of precipitates, thus many of the observed precipitates probably nucleated somewhere below the surface and are not visible until they have been growing for some time. The lack of site saturation also might suggest that there may be a slight delay in nucleation due to a small but significant nucleation barrier. As a result of these factors, it has been approximated that, on average, a precipitate will not be visible until it has had the opportunity to nucleate and grow during the first 0.5 s. Thus the temperature at which growth is first observed will be somewhat higher than the thermodynamic transforma-

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**Fig. 9(c).** Comparison of experimental front migration data to the Speich model for interface reaction controlled migration velocity, modified to calculate position vs. time, at a constant 10 K/s heating rate.

**Fig. 10.** Comparison of experimental front migration data to the interface reaction control model of migration velocity, modified to calculate position vs. time, with representative data for all three constant heating rates.
transformation, and mobility, where determined from the following relationship:

\[ \nu = \frac{M\Delta G}{V_m} \]  

\( M \) is the mobility, \( \Delta G \) is the free energy change associated with the transformation, and \( V_m \) is the molar volume. This is essentially a simplified version of Eq. (1). The mobility, \( M \), can then be expressed as a function of temperature in the form:

\[ M = M_0 \exp \left( -\frac{Q}{RT} \right) \]  

in which \( M_0 \) is a pre-exponential constant and \( Q \) is the activation energy. In terms of Eq. (1), activation energy is equivalent to activation enthalpy, \( \Delta H^* \), and the pre-exponential constant would be simply

\[ M_0 = \left( \frac{\partial \nu}{RT} \right) \exp \left( \frac{\Delta S^*}{R} \right) \cdot V_m \quad (\text{m}^4 \cdot \text{J}^{-1} \cdot \text{s}^{-1}) \]

The results of these studies, along with the present results, are contained in Fig. 11. As very little data is available for such mobilities, studies of three different kinds of migrating interfaces have been included: ferrite/ferrite during grain growth and recrystallization, \(^{26}\) ferrite/austenite during heating, \(^{81}\) and austenite/ferrite during cooling. \(^{27,28}\) The mobilities are plotted versus temperature, with the lines indicating the different values of \( M_0 \) and \( Q \) in Eq. (3) that are suggested by each study. The present data are indicated by 2 lines, indicating the upper and lower bounds of activation enthalpy reported presently, and the other ferrite to austenite line is simply Eq. (1) as determined by Speich and Szirmae. \(^{81}\) It should be noted that the \( M_0 \) suggested by Refs. (26), (27) and (28) are not allowed to vary with temperature, though this is not expected to make a significant difference. The current investigation is the only one contained in this plot in which direct observation of a migrating interface has been analyzed; the other studies rely on dilatometry, calorimetry, and/or \textit{a posteriori} microscopy and indirect analysis of those results. In addition, the two on-cooling transformation studies used iron samples that were alloyed with 1 or 2 atomic percent of a variety (Mn, Co, Al, Cu, or Cr) of substitutional elements. It is quite apparent upon examination of this figure that the on-cooling and grain growth/recrystallization mobilities vary significantly (orders of magnitude) from the present results, although this seems to result primarily from discrepancies between the pre-exponential factors as the slopes (activation energies) are all relatively similar. As the pre-exponential factor contains several unknown quantities which could vary significantly with temperature, composition, and interface type, it follows that the current results are consistent with previously reported mobilities. As a rough attempt to consolidate all of the ferrite to austenite transformation data, the dotted line in Fig. 11 could be fitted to the present and to the earlier\(^{81}\) results. The activation energy and pre-exponential values for this

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**Fig. 11.** Comparison of front mobilities for recrystallization of ferrite, ferrite to austenite, and austenite to ferrite in iron and substitutional iron alloys. Thick regions of the lines indicate regions for which experimental data was collected in determining the relationship between mobility and temperature. The dotted line is a preliminary attempt to consolidate the present results with earlier ferrite to austenite interface migration data.
relationship are 287 kJ and $6.95 \times 10^6 / T \cdot m \cdot mol \cdot J^{-1} \cdot s^{-1}$, respectively. These values do, in fact, fit the entirety of the present results and come quite close to the only previously reported ferrite to austenite interface mobility. It should be noted, however, that because of the relatively narrow range of temperatures over which this transformation occurs, the required values of mobility can be obtained by an almost endless number of combinations of $M_0$ and $Q$ used in Eq. (3). Thus, until there is concrete data for all of the constants contained within the pre-exponential—which does not seem very likely given the extreme difficulty in experimentally isolating and measuring these values—the authors are hesitant to suggest, at least for the ferrite to austenite front mobility, such a drastic departure (factor of $10^3$) in $M_0$ from that which was carefully established in Speich and Szirmae’s study. What is most important to conclude from Fig. 11 is that the mobilities measured in the current study are in good agreement with a mobility reported in a previous study of the same type of interface under the similar conditions, and that the activation energy range reported presently is in reasonable agreement with those reported for a wide variety of migrating interfaces in iron and in dilute substitutional iron alloys.

4. Conclusions

(1) Austenite formation in ultra low alloy, interstitial free steels can be observed in a confocal scanning laser microscope as moving fronts of surface relief due to dilatometric changes and roughening as a result of the phase transformation. Nucleation of austenite occurs at grain corners, but does not achieve site saturation and may also occur at interphase boundaries at a later stage of the transformation. The fronts are generally planar, though occasionally jagged, and appear to move through grain boundaries, indicating that moving interfaces are incoherent.

(2) No significant difference in rate is observed, under identical experimental conditions, among the 3 samples (pure iron and two IF alloys) in this study. Experimental data is reproducible and in good agreement with an interface reaction control model for austenite front migration. One important condition for this agreement is that the transformation begins at very little superheat with respect to the expected ferrite to austenite transformation temperature. This assumption has been carefully tested and shown to be reasonable through calibration with pure metals that melt just above the transformation temperature.

(3) Austenite front migration in the IF steels has been compared to 99.995% pure iron, with an insignificant difference in rate for a given thermal profile. This suggests that there is little or no effect of solute drag due to the alloying/impurity elements in the IF steels. A lack of solute drag further supports the agreement between experimental and calculated results, as the model does not incorporate any impurity drag effects at the interface.

(4) The strong likelihood of incoherency at the observed interfaces, good agreement with an interface reaction control, and the lack of any solute drag effect suggested by the results, signifies that austenite formation in ultra-low alloy steels occurs by a partitionless transformation which requires no long-range diffusion of solute elements.

(5) Small oxide and nitride inclusions in the IF steels have been shown to have an impeding effect on migrating interfaces, but not complete pinning. This does result in some of the inclusions being trapped at new austenite grain boundaries, and may explain the deviation of measured growth kinetics from the calculated values at longer growth distances.

(6) The austenite/ferrite interface mobility measured presently is in good agreement with that reported in a previous study of the ferrite to austenite transformation, and the experimentally determined activation enthalpy is in reasonable agreement with those reported for a wide variety of migrating interfaces in iron and in dilute substitutional iron alloys.

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