Comparison of Alloy Element Partition Behavior and Growth Kinetics of Proeutectoid Ferrite in Fe-C-X Alloys with Diffusion Growth Theory*

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Synopsis

Experimental data in the literature on the partition of alloying element during the growth of proeutectoid ferrite allotriomorphs in 9 Fe-C-X systems ($X = \text{Al, Si, V, Cr, Mn, Co, Ni, Cu}$ and Mo) are compared with the theory of diffusional growth which assumes that all component species are in local equilibrium (LE) at the advancing interface and also with the theory which assumes that only local equilibrium with respect to carbon is maintained at the interface (paraequilibrium (PE)). The temperature ranges in which the formation of ferrite occurs with or without alloy partition are explained in a qualitative manner by the former theory. The partition behavior in each alloy is discussed in terms of thermodynamic properties of alloy element. A major deviation of partition from the LE theory occurs at lower reaction temperatures in alloys with higher Mn or Ni content. A practical means of estimating the temperature of the cessation of partition in such alloys is considered. The growth kinetics of alloy element partitioned and no-partitioned ferrite allotriomorphs are briefly reviewed. In alloys with X which has a strong affinity with carbon in austenite, such as Cr, Mn, Mo, etc., the growth kinetics differ remarkably from both the LE and the PE theories, the reasons for which are yet to be explained adequately.

Key words: phase transformation; growth kinetics; partition; local equilibrium; paraequilibrium; proeutectoid ferrite; Fe-C-X alloy.

I. Introduction

To understand how each of alloy elements of different diffusivities are partitioned during the precipitate growth in multicomponent systems and thus how the growth kinetics are affected by the mechanism of such partition is one of the most important problems in materials science. Kirkaldy¹⁻³⁾ and Coates⁴⁻⁷⁾ have presented a lucid theory to tackle this problem. Under the assumption that each component species is in local equilibrium at phase interfaces, the theory has proved to be quite successful at least qualitatively in predicting the microstructural changes during various metallurgical reactions, as reviewed by Hillert.⁸⁾

More than one decade ago Coates^{4,6,7)} considered the effects of various alloy elements upon the hardenability of steel using his theory and a limited amount of thermodynamic data available at that time. In the course of time the solution models and the thermodynamic data which permit direct evaluation of phase equilibria in many Fe-C-X systems to be made have become available.⁹⁻¹⁴⁾ Analyses in 6 Fe-C-X systems ($X = \text{Mn, Ni, Cr, Si, Mo}$ and V) which utilized them¹⁵⁾ have demonstrated that the Coates model predicts various types of partition behavior depending upon the species of alloy elements; in some alloys partition occurs in a wide temperature range and in others the growth without alloy partition predomi-

nates. Also, in this model detailed analyses of the parabolic rate constant of the growth under local equilibrium (LE) can be made.

The purpose of this paper is to make comparison of experimental data on alloy element partition and growth kinetics of proeutectoid ferrite allotriomorphs with the theory in as many ternary Fe-C-X systems as possible. Except the one at the very initial period¹⁶⁾ the partition data have been taken mostly with EPMA,¹⁷⁻²¹⁾ and also with SEM or STEM-EDX.²²⁻²⁴⁾ The concentration of alloy elements has been reported to be fairly uniform in the ferrite particles even on a STEM scale.²³⁾ Further, the measured variation of the alloy element concentration with the reaction time is very sluggish.^{18,23)} It can be regarded as constant in the time range in which the growth kinetics are usually measured.

Three methods have been utilized to study the growth kinetics of ferrite: austenite interphase boundaries. One is the diffusion couple technique. The experimental condition which occurs in this technique is most favorable for comparison with theoretical analyses, and thereby, it has been used extensively by some authors.^{25,26)} However, this technique is applicable primarily to the reaction which occurs above the eutectoid (A_1) temperature. The second is the *in-situ* observation of migration kinetics of ferrite: austenite boundaries under the thermionic emission microscope (THEEM),^{27,28)} or the photoemission electron microscope.²⁹⁾ The third method is the isothermal reaction and quench method in which the thickness of the thickest and the length of the longest ferrite particles are measured as a function of isothermal reaction time with an optical microscope.^{23,24,30-35)} As a matter of fact, analyses and comparisons of growth kinetics with experiment have already been reported in many Fe-C-X systems. Here, the temperatures of measurement are collated to the calculated thermodynamic (denoted as T_{PE} , see Chap. II) and the kinetic (T_{LE} , also Chap. II) boundary temperatures between the growth modes with and without partition. Hence, comparisons may be made in a somewhat better perspective than those previously made.

Recently computer control of many industrial processes to achieve cost and energy saving operation have attracted considerable attention in steel industry. Under this circumstance it may be useful to evaluate the capability of predicting the transformation kinetics from these models in view of the potential application to the computer simulation of microstructural

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evolution during various manufacturing processes.

II. Summary of Theory of Diffusional Growth in Ternary Alloys

In the Coates model^{4,6)} alloy partition and growth kinetics of ferrite allotriomorphs in Fe-C-X alloys are described by a set of mass balance equations for X and C wherein only the flux of solute in austenite is assumed to contribute to the mass balance at the interface. The equations are written as,

$$\Omega_2 = f(\lambda_2), \dots\dots\dots(1-A)$$

and

$$\Omega_3 = f(\lambda_3) - \frac{mD_{32}}{D_3(1-D_2/D_3)}[f(\lambda_2) - f(\lambda_3)], \dots\dots\dots(1-B)$$

- where, $\lambda_i = \alpha/2\sqrt{D_i}$ ($i=1,2$ and 3 for Fe, X and C, respectively)
- α : the parabolic rate constant of thickening of allotriomorphs
- $\Omega_i = (x_i^f - x_i^\infty)/(x_i^f - x_i^a)$: the supersaturation of the i -th solute
- $m = (x_2^f - x_2^a)/(x_3^f - x_3^a)$: the slope of the tieline which connects the compositions in the two phases at the interface
- x_i^ν ($\nu = \alpha$ for ferrite and γ for austenite): the concentration of the i -th solute at the interface
- x_i^∞ : the concentration in the matrix
- D_i : the volume diffusion coefficient of the i -th solute in austenite
- D_{32} : the cross term diffusivity of carbon due to the X concentration gradient.

When both C and X concentrations are small, the following relationship has been proposed.³⁶⁾

$$\frac{D_{32}}{D_3} \sim \frac{\epsilon_3^{(2)}x_3^f}{1 + \epsilon_3^{(3)}x_3^f}, \dots\dots\dots(2)$$

where, $\epsilon_i^{(j)}$: the Wagner interaction parameter. The function $f(\lambda)$ is derived from the solution to the diffusion equation and its form depends upon the particle shape. Here, ferrite allotriomorphs are assumed to be an ellipsoid of aspect ratio 1/3. Under such an assumption $f(\lambda_i)$ becomes,³⁷⁾

$$f(\lambda_i) = 2\lambda_i^2[1 - \sqrt{\pi}\lambda_i e^{\lambda_i^2} \text{erfc}(\lambda_i)]. \dots\dots\dots(3)$$

Equations (1-A) and (1-B) predict that in case that the diffusivity of two species of solute differs by several orders of magnitude as in many Fe-C-X alloys, there are two modes of ferrite growth depending upon the amount of undercooling.^{4,6)} One is the partition local equilibrium (P-LE) mode in which bulk partition of X occurs between the ferrite particles and the austenite matrix and the growth kinetics are controlled by diffusion of X. The other is the negligible-partition local equilibrium (NP-LE) mode in which parti-

tion occurs only to an extremely small (practically null) amount and the growth kinetics are controlled by C diffusion. The former mode is operative at smaller undercoolings and the latter, at larger ones. The temperature at which a transition from the P-LE to the NP-LE modes occurs is hereafter denoted as T_{LE} . It is the upper limiting temperature at which no-partitioned ferrite (also called para ferrite) can be formed under the kinetic constraint imposed by the large difference in magnitude of diffusivity of C and X.

As described in the previous reports,¹⁵⁾ Eqs. (1-A) and (1-B) are solved by approximating the $\alpha/(\alpha+\gamma)$ and $\gamma/(\alpha+\gamma)$ phase boundaries by polynomials of one of the four terminus compositions of the interfacial tie-line. The coefficients of each term in the polynomials are determined by regression from the ortho-equilibrium* phase boundaries which are calculated with the use of an appropriate thermodynamic model. For the 6 Fe-C-X systems previously analyzed, the Hillert-Staffanson (HS) regular solution model^{9,10)} is utilized again. Of the 3 systems added in this report, calculations in an Fe-C-Cu alloy are also conducted from the HS model. The central atoms (CA) model^{13,14)} is used for Fe-C-Co and Fe-C-Al alloys. In the Fe-C-Al system the values of the interaction parameter are determined from the compilation by Kaufman³⁸⁾ and from the C activity data in austenite at 1 000°C.³⁹⁾ They are $\epsilon_2^{(2)} = 23.95 - 2 496/T$ and $23.95 - 14 090/T$ for the Al-Al interaction in austenite and ferrite, respectively, and $\epsilon_6^{(2)} = -2.1$ for the Al-C interaction in austenite ($\epsilon_6^{(2)}$ in ferrite is assumed to be zero). These are all assumed to be temperature independent in the calculation. The thermodynamic properties *etc.* of alloy elements are listed in Table 1.

When the X diffusion zone ahead of the interface is smaller than the atom distance, an alternative mode called paraequilibrium (PE) in which only equilibrium with respect to C is maintained at the interface has been proposed to operate.^{23,32)} The thermodynamic conditions for paraequilibrium in ternary Fe-C-X systems are well established.⁴²⁾ For ternary Fe-C-X systems, they are written as

$$\mu_3^a = \mu_3^f, \dots\dots\dots(4-A)$$

$$(\mu_1^a - \mu_1^f) + \frac{x_2^\infty}{x_1^\infty}(\mu_2^a - \mu_2^f) = 0, \dots\dots\dots(4-B)$$

and

$$\frac{x_2^a}{x_1^a} = \frac{x_2^f}{x_1^f} = \frac{x_2^\infty}{x_1^\infty}, \dots\dots\dots(4-C)$$

- where, μ_i^* : the chemical potential of the i -th component species in ferrite or in austenite
- x_i^* : the composition at the interface.

The temperature at which a certain alloy composition crosses the paraequilibrium $\gamma/(\alpha+\gamma)$ boundary is designated as T_{PE} . It is the maximum temperature at which the formation of no-partitioned ferrite is thermodynamically possible. The parabolic rate constant

* This is the ordinary full equilibrium in which the chemical potential of all component species is equal between the two phases. The prefix "ortho" is used to give a clear contrast to the paraequilibrium.

Table 1. Thermodynamic properties of alloy elements and experimental data sources.*

Alloy element	γ or α stabilizer	Interaction with carbon in austenite*2	Thermodynamic model used in the calculation	References*3	
				Partition	Growth kinetics
Al	α	—*4	CA	18	28
Si	α	+	HS	18	26, 28, 32
V	α	—	HS	—	33
Cr	γ	—	HS	17, 18	26, 29, 31, 32
Mn	γ	—	HS	18, 20, 21, 23, 24	23, 24, 25, 26, 27, 32
Co	α	+	CA	18	26, 28, 30
Ni	γ	+	HS	18, 19, 24	24, 26, 30, 32
Cu	γ	+	HS	18	—
Mo	α	—	HS	16, 18	26, 27, 30, 34, 35

* References in which only *TTT* curves are shown are not included in this table.
*2 The symbols, + and —, indicate repulsive and attractive interactions, respectively.
*3 For underlined references, data are compared with calculations in this report.
*4 In liquid iron the interaction is reported to be repulsive.¹³⁾

for ellipsoids which grow under the paraequilibrium mode can be calculated solely from the equation,

$$\Omega_3 = 2\lambda_3^2[1 - \sqrt{\pi} \lambda_3 e^{\lambda_3^2} \text{erfc}(\lambda_3)] . \quad \text{.....(5)}$$

The concentration dependence of carbon diffusivity can be incorporated, though not exactly, by taking the average of D_3 from x_3^∞ to x_3^i .^{15,43)} A more sophisticated approach has been proposed which takes full account of the concentration dependence of D_3 .^{44–46)}

Of the 4 solute concentrations which appear in Eqs. (1-A) and (1-B), the X concentration in ferrite (x_2^a) can be readily compared with measurement. Under the assumption that the flux of solute in ferrite does not contribute to the mass balance at the interface, x_2^a is equivalent to the bulk composition in the ferrite particles. At present the composition in a few atom planes immediately ahead of the phase interface is very difficult to be analyzed, and thereby, the other concentrations are not compared in this report. In case of paraequilibrium by definition x_2^a is the same as in the matrix.

III. Comparison of Observed Partition Behavior with Calculation

1. α -former

Figures 1 to 5 show the measured¹⁸⁾ and calculated concentrations of X =Al, Si, V, Co and Mo in ferrite allotriomorphs isothermally formed. The standard deviation of these concentrations are reported to be a few hundredths of percent. Experimental data sources are shown in Table 1. Since these elements are all α -formers, they are predicted to be concentrated in austenite at higher temperatures. Actually, however, the enrichment is observed only at the highest temperature in the Al and Si alloys. At all other temperatures and in the other alloys X is not partitioned. The temperature $Ae_3(\text{exp})$ in the figure was determined by extrapolating the lower temperature portion of measures *TTT* curves to the upper temperature side in each alloy.¹⁸⁾ By comparison, the calculated boundary temperatures (Ae_3 , T_{PE} , etc.) appear to con-

Fig. 1. Concentration of Al in ferrite allotriomorphs as a function of reaction temperature.

Fig. 2. Concentration of Si in ferrite allotriomorphs as a function of reaction temperature.

tain an error of less than 15°C. Because differences in T_{LE} and T_{PE} of these alloys are usually small, it is not possible to decide which one agrees better with the observed cessation of X partition.

Fig. 3. Concentration of V in ferrite allotriomorphs as a function of reaction temperature.

Fig. 6. Concentration of Cr in ferrite allotriomorphs as a function of reaction temperature.

Fig. 4. Concentration of Co in ferrite allotriomorphs as a function of reaction temperature.

Fig. 7. Concentration of Mn in ferrite allotriomorphs as a function of reaction temperature.

Fig. 5. Concentration of Mo in ferrite allotriomorphs as a function of reaction temperature.

Fig. 8. Concentration of Ni in ferrite allotriomorphs as a function of reaction temperature.

2. γ -former

Figures 6 to 9 compare the measured¹⁸⁾ and calculated x_2^a in alloys which contain a γ -former. In the Cr alloy (Fig. 6) partitioned ferrite is formed only at the highest temperature of measurement, consistent with the results of calculation that both T_{PE} and T_{LE} are very close to A_{e3} . Thus, Cr shows partition behavior very similar to Al or Si except for the phase to which enrichment occurs. The fact that the predicted P-LE range is somewhat higher than the temperature at which partition does take place may be attributed to the failure of coincidence between the experimental and calculated A_{e3} .

A large difference from the LE prediction is ob-

served in the Mn alloy (Fig. 7). According to the results of calculation P-LE is operative at all temperatures in this alloy. Though the measured x_{Mn}^a lies close to the LE curve at higher temperatures, it begins to deviate from this curve, and the Mn partition ceases to occur at 600°C, as much as about 100°C below T_{PE} . Similar behavior has been observed in an Fe-0.6C-4.5Mn alloy.²⁴⁾ On the other hand, it has been reported that the Mn partition can be well described by the LE theory at the late stages of reac-

tion in an alloy with lower Mn content (Fe-0.45C-1.5Mn alloy).²⁰⁾

Nickel is second to Mn in respect to the width of the temperature range in which partition is observed to occur.¹⁸⁾ In this system the temperature of the cessation of partition is usually slightly lower than T_{LE} and as much as 50°C below T_{PE} as seen in Fig. 8. It is noted that in this system also a similar situation to the 3.15 Mn alloy (Fig. 7) occurs as the Ni content is increased: whereas x_{Ni}^{α} is close to LE at highest temperatures, it becomes equal to the matrix concentration at a temperature about 50°C below T_{PE} in an Fe-0.6C-6.5Ni alloy.²⁴⁾ It has also been reported that the LE theory accounts well for the Ni partition at very long reaction times in Fe-1.02C-0.54Ni and Fe-0.92C-2.08Ni alloys.¹⁹⁾ The partition of Cu did not occur at any temperature of measurement in

Fig. 9. Concentration of Cu in ferrite allotriomorphs as a function of reaction temperature.

Fig. 10. Parabolic growth rate constant of ferrite allotriomorphs in an alloy with Al.

○, ●: Respectively, as-measured and corrected (see text) values³²⁾
Fig. 11. Parabolic growth rate constant of ferrite allotriomorphs in alloys with Si. Experimental data were obtained from (a) thermionic emission microscopy (THEEM)²⁸⁾, and (b) isothermal reaction and quench method.

contrast to all other γ -forming elements.

These results indicate that T_{LE} gives a somewhat better accounting for the experimental temperature at which the formation of no-partitioned ferrite occurs. However, large discrepancies between the theory and the observations appear as the bulk content of some elements, *e.g.*, Mn and Ni, is increased.

IV. Growth Kinetics

1. Comparison in the No-partition Regime

1. α -former

As seen above, in alloys containing an α -former the formation of proeutectoid ferrite allotriomorphs and presumably Widmanstätten ferrite sideplates occurs overwhelmingly without alloy element partition. Figures 10 to 14 show comparisons of parabolic rate constant of thickening of allotriomorphs, α , predicted from the LE and PE models with measurements.^{28,32,33} Symbols designated as THEEM are the data obtained by the thermionic emission microscopy.²⁹ The scatter of the data, indicated by error bars, was attributed to the variation of the aspect ratio of allotriomorphs: ones of larger aspect ratio tend to exhibit a larger α value. For triangles the data are reported for only one aspect ratio at each temperature. Open circles are the data obtained by the isothermal reaction and quench method. In this set of data correction has been made for the inhibition of growth caused by the development of partially coherent facets during the growth of allotriomorphs (faceting correction).³² The corrected α is shown by solid circles. Though not shown in the figures, the error limit of each experimental point is typically within a factor of two to three in this method.

Fig. 12. Parabolic growth rate constant of ferrite allotriomorphs in an alloy with V.

It is seen that the measured α agrees reasonably well not only with the PE, but also with the LE curves in all these alloys. Because α calculated under the PE and NP-LE modes is very similar, it is not possible to determine the operating growth mode solely from the comparison of growth kinetics. More detailed comparisons have been made by Aaronson and his co-workers.^{28,32,33}

In the Mo alloy the data points are in excellent agreement with the calculation at the two highest temperatures of measurement (Fig. 14). However, they

Fig. 13. Parabolic growth rate constant of ferrite allotriomorphs in an alloy with Co.

Fig. 14. Parabolic growth rate constant of ferrite allotriomorphs in an alloy with Mo.

show a sharp minimum around 650°C. It has been proposed that this anomaly in α is the principal reason for the bay formation in the TTT -curve for the initiation of proeutectoid ferrite reaction in this alloy system.³⁴⁾ It has been reported that higher V content alloys also show a bay in the TTT -curve for the initiation of the ferrite formation.⁴⁷⁾ Because Mo and V are similar in terms of the propensity of carbide formation, the kinetic anomaly and the bay formation are

considered to be the results of the segregation of these elements to the advancing ferrite: austenite interfaces (the solute drag-like effect, SDLE).^{34,48)}

2. γ -former

In Figs. 15 to 18 are shown the measured³²⁾ and calculated α in alloys containing Cr, Mn, Ni and Cu, respectively. It is noted that the X partition did not take place at any these data points except those at the highest temperature in the Mn and Ni alloys.

In the Cr alloy the measured α is considerably

Fig. 15. Parabolic growth rate constant of ferrite allotriomorphs in an alloy with Cr.

Fig. 17. Parabolic growth rate constant of ferrite allotriomorphs in an alloy with Ni.

Fig. 16. Parabolic growth rate constant of ferrite allotriomorphs in an alloy with Mn.

Fig. 18. Parabolic growth rate constant of ferrite allotriomorphs in an alloy with Cu.

smaller than the PE predicted values (Fig. 15). The same is true with the Mn alloy (Fig. 16).^{*} Both Cr and Mn are known to have an attractive interaction with C atoms in austenite like Mo. The data on C diffusion in alloyed austenite indicate that alloy elements of these levels of content have usually little influence upon the diffusion of C atoms.⁴⁹⁾ Thus, the same mechanism as the one which caused the kinetic anomaly in the Mo and V alloys (SDLE) has been considered to be responsible for the deviation from PE.³²⁾ One point to be added is that in this Mn alloy P-LE is also supposed to operate at all temperatures if LE is achieved. Thus, a difference by 3~4 orders of magnitude between the measured data and the LE curve resulted. On the other hand, the calculated α is in much better agreement with the measured one in the Ni alloy (Fig. 17).

2. Comparison in the Partition Regime

Because only a few alloy elements exhibit a significantly wide range of partitioned growth, studies on the growth kinetics in this regime are rather scarce. The experimental data on the parabolic rate constant in the Mn and Ni alloys are all higher by about 2 orders of magnitude than the values obtained from the P-LE mode in which volume diffusion of X in austenite is assumed to be rate controlling.^{23,24)} To account for this difference it was necessary to incorporate the flow of X into the austenite matrix *via* diffusion along the austenite grain boundaries (the rejector/collector plate mechanism).²³⁾

V. Discussion

1. Temperature Range of Alloy Element Partitioned Growth

We consider here how the range of ferrite formation with and without partition is affected by the thermodynamic properties of alloy elements. The region in which the P-LE and NP-LE modes operate is schematically illustrated for various alloy elements in Figs. 19 and 20. The boundary which divides the $(\alpha+\gamma)$ two phase region into the two modes starts at the $\gamma/(\alpha+\gamma)$

Fig. 19. Division of $(\alpha+\gamma)$ two phase region into the P-LE and NP-LE regions for a γ -former which has (a) attractive and (b) repulsive interaction with carbon.

composition on the Fe-C side and ends up at the $\alpha/(\alpha+\gamma)$ composition on the Fe-X side. As a result, for a γ -former the temperature range of the partitioned growth is considered to be generally wide. When the C-X interaction is attractive, this boundary becomes convex downward so that the P-LE region is somewhat widened (Fig. 19(a)). This is the case with Mn. On the other hand, when the interaction is repulsive, the boundary becomes convex upward (Fig. 19(b)). In such a system, T_{LE} is reached at smaller undercoolings from A_{e3} at a fixed alloy composition than in the systems which belong to the former case. A typical example for this is Ni. Particularly, it is readily expected that the P-LE region is decreased at an increasing rate as the bulk C content is increased in this system.

Chromium is the same as Mn in terms of the interaction with C and the effect upon the phase stability of ferrite and austenite. However, the $(\alpha+\gamma)$ region in the Fe-Cr system is narrow compared to that in the Fe-Mn system. As a result, the temperature range of Cr partition is remarkably narrower than that of Mn. It should be noted that at lower temperatures the two phase region in the Fe-C-Cr system once shrinks and expands again as the Cr content is increased. Thus, the P-LE to NP-LE boundary is expected to have a peculiar curvature which follows the path of the $\gamma/(\alpha+\gamma)$ boundary. Copper has also a small composition range of the $(\alpha+\gamma)$ region and the interaction with C atoms is known to be repulsive. Accordingly, the range in which partition occurs may be small like Cr.

When X is an α -former, the two phase region in Fe-C-X systems is extended as shown in Fig. 20. It is immediately seen that the P-LE region is generally narrow in such alloys. The sign of the C-X interaction is seen to influence the width of the two regions in a manner opposite to the case of a γ -former; whereas repulsive interaction with C (*e.g.*, Si and Co, Fig. 20 (a)) widens the region of partitioned growth, attractive interaction with C (*e.g.*, Mo and V, Fig. 20 (b)) tends to reduce it. Though Al has weak attractive interaction with C in austenite,³⁹⁾ the interaction in liquid iron is reported to be repulsive.¹³⁾ The negative sign of the Al-C interaction in austenite is not well established.

Fig. 20. Same figures as Fig. 19 for an α -former of (a) repulsive and (b) attractive interaction with carbon.

* It is noted that in this figure the bulk content of C is different from that in Fig. 7.

Fig. 21. Diffusion zone thickness at reaction times at which ferrite allotriomorphs grow to a fixed thickness, S , in an Fe-0.6C-4.5Mn²⁴⁾ alloy plotted against the reaction temperature.

2. Temperature of Cessation of Partition

For such alloys as shown in Figs. 7 and 16 it is important to know how much undercooling is necessary for the onset of the formation of no-partitioned ferrite. The criterion which has been frequently used is the thickness of the diffusion zone in front of the ferrite: austenite interface, $W_{\alpha:\gamma}$. It is usually several orders of magnitude smaller than the atom distance in austenite.³²⁾ Though, as seen in Chap. III, the calculate T_{LE} does not much differ from the observed temperature of the transition of partition in many alloys, the present author has proposed²⁴⁾ that because the X partition is likely to occur *via* the rejector/collector plate mechanism, in the alloys in which T_{LE} is non-existent (P-LE operates at all temperatures) the thickness of the diffusion zone on both sides of grain boundary, W_{gb} , is more appropriate for discussing the boundary temperature between the partition and no-partitioned growth.

In Fig. 21 are plotted $W_{gb} (\sim S\sqrt{D_{Mn}/2\alpha})$ and $W_{\alpha:\gamma} (\sim SD_{Mn}/2\alpha^2)$ at the times when the particle grows to a fixed thickness, S , against the reaction temperature. They are calculated from the previously reported data on α in an Fe-0.6C-4.5Mn alloy.²⁴⁾ The error limits of the thickness value at each temperature are due to the error in the measurement of α . In this alloy the boundary temperature between partitioned and no-partitioned growth lies somewhere between 550 and 570°C. However, x_{Mn}^{α} begins to deviate from the LE prediction around 600°C. It is observed that W_{gb} for $S \sim 1 \mu m$ becomes an order of the atom distance in the transition region between them. Similar results are obtained in the 6.5 Ni alloy. Thus, it seems appropriate to use this criterion for the onset of no-partitioned growth in these alloys.

VI. Concluding Remarks

The above comparisons have shown that in most alloys the temperature dependence of the X concentration in ferrite allotriomorphs is explained well in qualitative manner from the Coates local equilibrium (LE) model, and is consistent with the results of considerations in terms of the nature of C- X interactions and the influence of X upon the stability of ferrite and austenite.

The growth kinetics of no-partitioned ferrite allotriomorphs are also in fair agreement with the theory in Al, Si, V, Co and Ni alloys. In the Cr, Mn and Mo alloys the measured growth kinetics are significantly slower than the paraequilibrium (PE) and/or the NP-LE predictions. In the partition regime, the growth kinetics are faster than those calculated from the LE model.

In alloys with relatively high Mn or Ni contents a major deviation from LE occurs in the partition behavior. In such an alloy a practical means for estimating the temperature of the cessation of alloy partitioning was considered in terms of the diffusion zone thickness on both sides of the austenite grain boundary near the allotriomorphs.

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