

**Short
Communication**

**Growth rate data
on bainite in alloy
steels**

On the basis of a theoretical analysis of published experimental data on the lengthening rate of sheaves of bainite in alloy steels, it is concluded that they lengthen at rates much faster than expected from paraequilibrium carbon diffusion controlled growth. The results are interpreted in terms of the mechanism of the bainite reaction and in the context of any solute atom-transformation interface interactions.

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Introduction

In a critical assessment of the diffusion controlled growth of ferrite plates in plain carbon steels it was inferred that in such steels, plates of Widmanstätten ferrite lengthen at a rate which is approximately consistent with growth controlled by the diffusion of carbon in the austenite ahead of the transformation interface.¹ On the other hand, the growth of bainite sheaves was found to occur at a rate much faster than expected from carbon diffusion controlled transformation. These conclusions are consistent with current models for transformations in steels, in which it is assumed that Widmanstätten ferrite grows by a displacive paraequilibrium transformation mechanism, whereas bainite grows by displacive transformation in the form of sheaves of platelets called subunits. Each subunit inherits the composition of the parent phase and any excess carbon is subsequently partitioned into the residual austenite or precipitated in the form of carbides within the bainitic ferrite.²

There is some early work on Fe-Ni-C alloys containing relatively high concentrations of nickel, in which the experimentally measured lengthening rates of bainite were found to very much lower than expected from carbon diffusion controlled growth.^{3,4} It has been suggested that relatively large concentrations of nickel may, via some kind of an interface-solute interaction, retard growth kinetics.^{4,5} However, since the original work^{3,4} was published, there have been many refinements to the theory and other aspects concerning the diffusion controlled growth of plates.¹ Consequently, the purpose of the present work is to reassess the experimental data on bainite growth kinetics in nickel rich steels in particular and any other relevant data on alloy steels in general. The early experiments are all based on measurements carried out using light microscopy, a technique which has limited resolution as far as the structure within sheaves of bainite is concerned. Consequently, it is assumed that they refer to the growth rate of sheaves of bainite, rather than to the subunits within the sheaves.

Method

Trivedi⁶ has obtained a solution for the diffusion controlled growth of parabolic cylinders (an approximation to a plate morphology). The steady state growth velocity v_{dc} , controlled by the diffusion of carbon in the austenite (γ) ahead of the interface, is obtained by solving the equation

$$f_1 = (\pi p)^{1/2} \exp(p) \operatorname{erfc}(p^{1/2}) \left[1 + \frac{r_c f_1 S_2(p)}{r} \right] \quad (1)$$

where p is the Péclet number, r_c is the critical plate tip radius at which the growth rate becomes zero, r is the radius of curvature at the tip of the plate, and

$$f_1 = \frac{\bar{x} - x^{\gamma\alpha}}{x^{\alpha\gamma} - x^{\gamma\alpha}}$$

In this equation, \bar{x} is the undisturbed carbon concentration in the austenite, well ahead of the plate tip, and can be taken to be equal to the average carbon concentration in the alloy if soft impingement effects are absent; and $x^{\gamma\alpha}$ and $x^{\alpha\gamma}$ are the paraequilibrium carbon concentrations in the austenite and ferrite, respectively.

The function S_2 in equation (1) has been evaluated numerically by Trivedi⁶ and the Péclet number p is a dimensionless velocity given by

$$p = \frac{v_{dc} r}{2\bar{D}} \quad (2)$$

where v_{dc} is the diffusion controlled growth velocity, r is the radius of curvature at the tip of the plate (the value of r used in the present work is chosen to give the maximum growth rate), and \bar{D} is the average diffusion coefficient. Since the diffusion coefficient $D(x, T)$ depends on the carbon concentration x , which in turn is a function of the distance ahead of the plate tip, the diffusivity used in equation (2) is a weighted average⁷

$$\bar{D} = \int_{x_r}^{\bar{x}} \frac{D(x, T)}{\bar{x} - x_r} dx \quad (3)$$

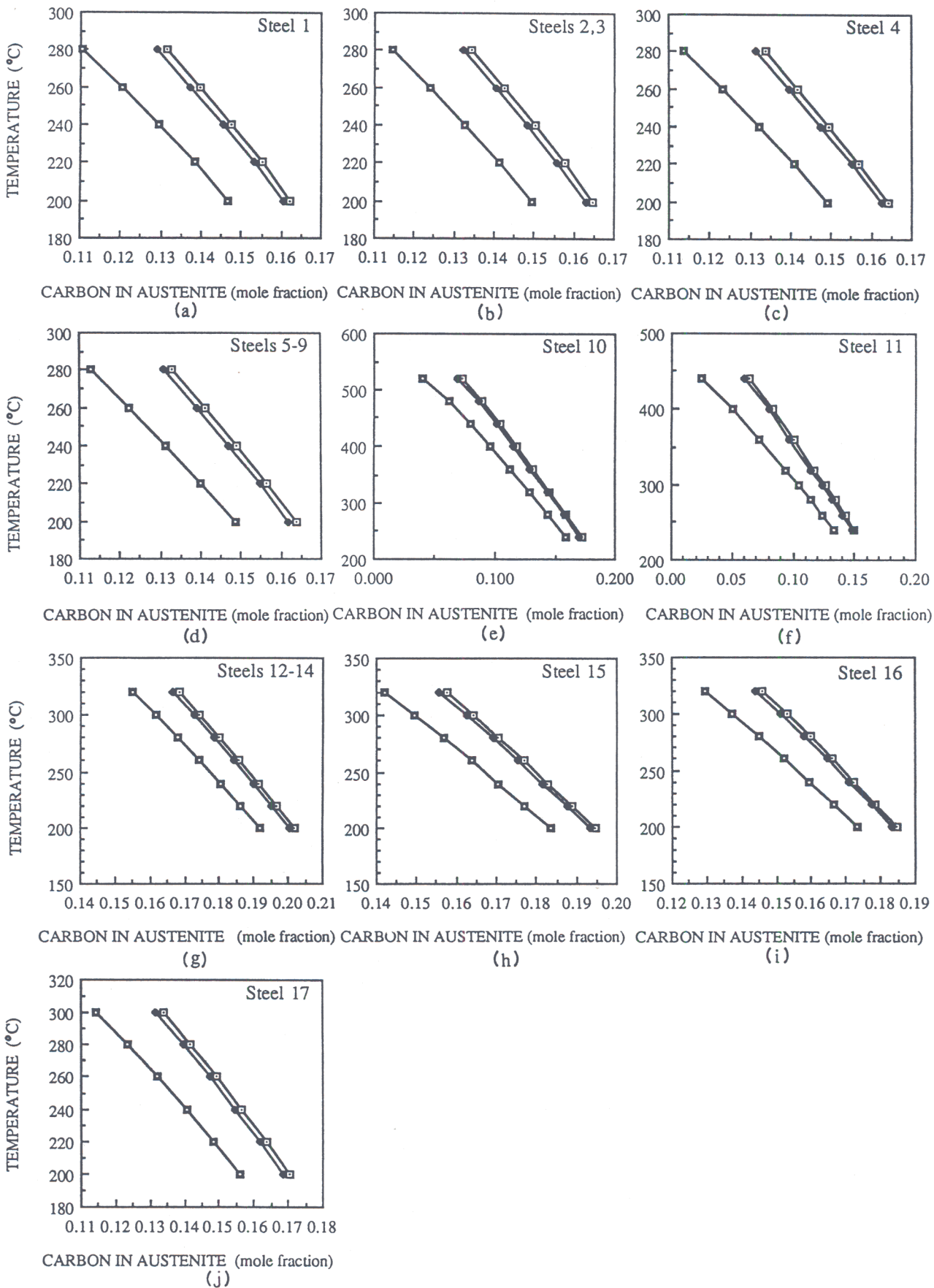
where D is evaluated using the Siller and McLellan model⁸⁻¹⁰ for the diffusion of carbon in austenite and x_r is the carbon concentration in the austenite at the plate tip and differs from the paraequilibrium concentration because of the curvature of the interface at the plate tip. It can be calculated using the following equation¹¹

$$x_r = x^{\gamma\alpha} \left(1 + \frac{\Gamma}{r} \right) \quad (4)$$

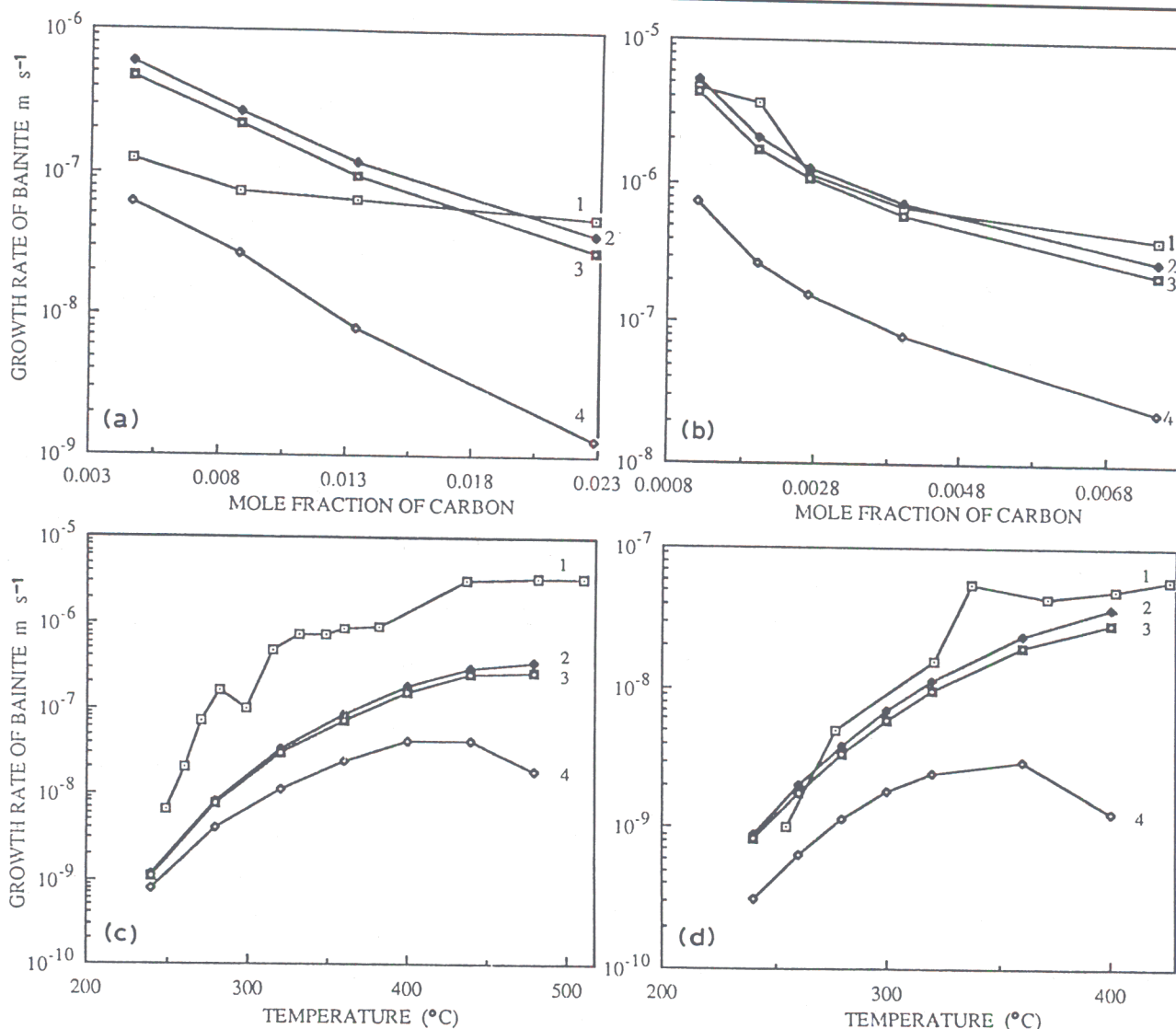
where

$$\Gamma = \frac{\sigma V_m (1 - x^{\gamma\alpha}) / (x^{\alpha\gamma} - x^{\gamma\alpha})}{RT \left\{ 1 + \frac{d[\ln(f)]}{d[\ln(x^{\gamma\alpha})]} \right\}}$$

σ is the interface free energy per unit area, taken to be 0.2 J m^{-2} (Ref. 1), f is the activity coefficient for carbon in austenite, R is the universal gas constant, T is the absolute temperature of transformation, and V_m is the molar volume of ferrite.¹ The paraequilibrium carbon concentration in the austenite (i.e. $x^{\gamma\alpha}$) and the activity coefficient f were calculated using the Lacher-Fowler-Guggenheim quasichemical thermodynamic model¹² and the thermodynamic data detailed in Ref. 1. The free energy data for pure iron were obtained from Kaufman *et al.*¹³ This is believed to make



1 Effect of stored energy on carbon content of austenite which is in paraequilibrium with ferrite



experimental data due to *a* Goodenow *et al.*³ (steels 1–4); *b* Rao and Winchell⁴ (steels 5–9); *c* Oblak and Hehemann¹⁵ (steel 10), *d* Oblak and Hehemann¹⁵ (steel 11)

2 Comparison of experimental (curve 1) and calculated bainite growth rates: curves 2–4 are calculated using stored energy values of 0, 50, and 400 J mol⁻¹, respectively

the analysis more reliable for extrapolation to low temperatures.¹ The paraequilibrium carbon concentration x^{27} is always so small that it was approximated by the values for plain carbon steels as deduced by Bhadeshia.¹⁴ Finally, because both Widmanstätten ferrite and bainite grow by a displacive transformation mechanism,² the strain energy

resulting from the invariant-plane strain shape change accompanying their formation must be taken into account in the calculations. The stored energy terms for Widmanstätten ferrite and bainite are about 50 and 400 J mol⁻¹, respectively,² and are taken into account as in Ref. 1. For illustration purposes, the calculations have been carried out using three different stored energy values of 0, 50, and 400 J mol⁻¹, although only the last set of calculations is strictly relevant for the present work, which deals with bainite.

The chemical compositions of the steels^{3, 4, 15–17} studied are given in Table 1. In the thermodynamic calculations all the elements listed are taken into account and an effect on the diffusion coefficient of carbon in austenite is included through their influence on the activity of carbon in austenite.⁸ Note that some of the steels are not substitutionally alloyed, but were missed in the original assessment¹ and have been included in the analysis for completeness.

Table 1 Chemical composition of steels used, wt-%

Steel no.	C	Si	Mn	Cr	Mo	Ni	Ref. no.
1	0.10	0.18	0.79	0.21	0.07	8.70	3
2	0.50	0.36	0.68	8.70	3
3	0.19	0.29	0.60	8.90	3
4	0.29	0.31	0.69	8.80	3
5	0.026	10.6	4
6	0.043	10.6	4
7	0.058	10.6	4
8	0.086	10.6	4
9	0.162	10.6	4
10	0.69	0.70	0.85	0.85	0.81	1.80	15
11	0.50	0.36	0.68	8.70	15
12	0.96	16
13	1.16	16
14	1.43	16
15	1.14	2.70	16
16	1.12	5.28	16
17	0.51	9.10	17

Results and discussion

The calculated $\alpha + \gamma/\gamma$ phase boundaries, representing the variation in x^{28} as a function of temperature and stored

energy, are shown in Fig. 1. The corresponding growth rate calculations are presented in Fig. 2, for most of the steels listed in Table 1, with the exception of steels 12–17. For steels 12–16, the transformation to bainite was carried out under conditions where the diffusivity of carbon in austenite had to be calculated for exceptionally high values of $x^{\gamma\alpha}$. The Siller and McLellan theory for the diffusion coefficient of carbon,^{8–9} although known to be highly reliable in both the thermodynamic and kinetic sense, is not able to cope with carbon concentrations in excess of about $\frac{1}{2}$ mole fraction. Consequently, growth calculations could not be carried out for steels 12–16. For steel 17, only one experimental growth rate of $5 \times 10^{-8} \text{ m s}^{-1}$ was reported, for a transformation temperature of 380°C; the calculated growth rates for stored energy values of 0, 50, and 400 J mol⁻¹ are found to be 5.178×10^{-8} , 4.255×10^{-8} , and $6.874 \times 10^{-9} \text{ m s}^{-1}$, respectively.

Measurements of bainite growth are not easy to make and it is likely that the experimental data summarised in Fig. 2 show a fair degree of scatter. Also, the procedures available for the calculation of bainite growth rates are by no means totally rigorous. For example, in the theory a shape of a parabolic cylinder is assumed which is not likely to be a perfect representation of the morphology of bainite. Therefore, the calculated data must also show a degree of uncertainty and these factors must be borne in mind in assessing the data.

In Fig. 2, for the sake of completeness, the calculations have been carried out for stored energy values of 0, 50, and 400 J mol⁻¹. Widmanstätten ferrite formation involves the simultaneous, back-to-back growth of two plates with mutually accommodating shape deformations, so that a stored energy term of only 50 J mol⁻¹ is associated with Widmanstätten ferrite.^{18, 19} However, for bainite, the shape deformation is identical for adjacent subunits, so the stored energy term is greater at $\sim 400 \text{ J mol}^{-1}$ (Ref. 20). Hence, it is appropriate for the experimental data for bainite to be assessed against the curves labelled '4' in Fig. 2. It will be noted that in some cases (e.g. Fig. 2b) the agreement between experimental data and curves calculated using lower values of stored energy appears to be rather good. However, that agreement seems fortuitous, since the stored energy values used are not physically significant for bainite.

It is evident from a comparison of the experimental data with curve 4 on each figure, that the growth rate of bainite sheaves always exceeds that expected on the basis of carbon diffusion controlled paraequilibrium transformation. This might intuitively be expected to be the case since there are now considerable data to indicate that bainite forms by diffusionless transformation, the excess carbon in the bainitic ferrite being partitioned or precipitated subsequent to transformation.^{2, 21} However, the lengthening rate of a sheaf of bainite does not depend only on the rate of subunit lengthening, but also on the rate at which new subunits are nucleated. The overall lengthening rate of a sheaf may in fact be very low if the nucleation rate of subunits is sufficiently small, even though individual subunits may grow rapidly.

If it is assumed that a subunit reaches its limiting size in a time interval t and that the time Δt elapsed before another subunit nucleates at the tip of the original subunit, then the lengthening rate v_s of a sheaf is given by:

$$v_s = \frac{v}{1 + (\Delta t/t)} \quad \dots \dots \dots (5)$$

where v is the average lengthening rate of a subunit. Because the growth of a subunit is diffusionless, v is expected to exceed the rate v_{dc} calculated on the basis of carbon diffusion controlled growth, although v_s will only be greater than v_{dc} if the factor $1 + (\Delta t/t)$ is sufficiently small (i.e. if the time interval between the nucleation of subunits is small in comparison with the time required for the subunits

to reach their limiting size). The results in general indicate that $v_s \gg v_{dc}$ so it can be safely concluded that the growth rate of the subunits is much faster than expected from carbon diffusion controlled transformation. This is consistent with direct high resolution observations on the growth of bainite subunits in an Fe–Mn–Si–C alloy.²²

The results also indicate that solute–interface interactions are not strong enough to slow down interface motion to velocities less than v_{dc} and that there is no special nickel atom–interface interaction in Fe–Ni–C alloys, as has been suggested previously.^{4, 5} This conclusion also applies to chromium which is a significant alloying element in some of the steels analysed in the present work.

Conclusions

It is found that the lengthening rates of sheaves of bainite in alloy steels are far greater than expected from calculations based on carbon diffusion controlled paraequilibrium transformation. Furthermore, it is found that the lengthening rates of individual bainite subunits within the sheaves must also be faster than carbon diffusion controlled growth. The results are consistent with the proposed mechanism of bainite transformation in which the growth of the bainite subunits is diffusionless, with the partitioning of carbon into the residual austenite occurring immediately after transformation.

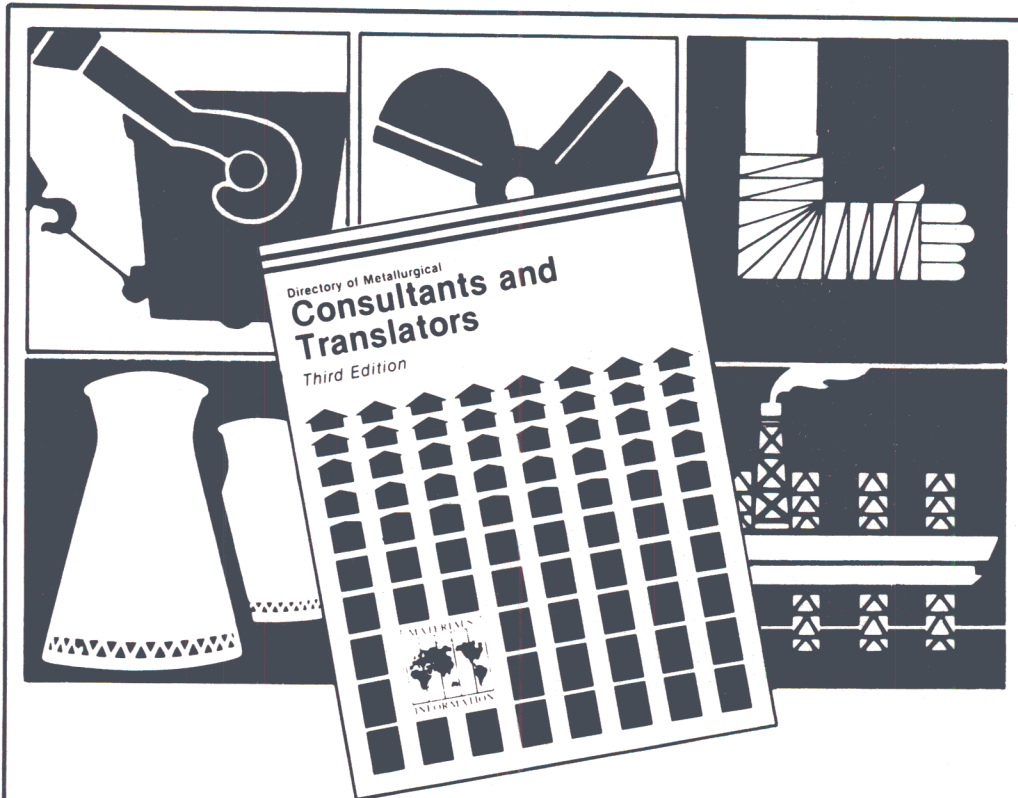
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