the single phase niobium region, terminating the growth of the reinforcing carbide phase.

Banding effects as caused by inhomogeneities in the feed rod would be superimposed on banding effects caused by fluctuations in the growth rate, increasing the frequency and severity of banding. However, in our experiments it appears that inhomogeneities in the feed rod were the major factor leading to banding. Two zone levelling passes before composite growth were sufficient to control banding.

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Influence of Alloying Elements upon the Morphology of Austenite Formed from Martensite in Fe-C-X Alloys

M. R. PLICHTA AND H. I. AARONSON

As a result of the availability of a sensitive etching technique capable of revealing clearly the morphology of even small crystals of austenite in a martensitic matrix*,[1] some new observations have been made on the influence of alloying elements upon the morphology of austenite formed from martensite in the $\alpha+\gamma$ region.

Fig. 1—(*a*) Fe-0.13 pct C-0.39 pct Mo, reacted 20 s at 760°C, etched in BASP, Magnification 500 times, grain boundary allotriomorphs (1) and idiomorphs (2) at former austenite grain boundaries; (*b*) Fe-0.11 pct C-3.28 pct Ni, reacted 30 s, at 710°C, etched in BASP, Magnification 500 times, intragranular idiomorphs and martensite needle boundary allotriomorphs; (*c*) Fe-0.11 pct C-3.28 pct Ni, reacted 60 min at 710°C, etched in BASP and 2 pct nital, Magnification 2000 times, growth of MNB allotriomorph across a ferrite:ferrite boundary.

Table I. Compositions and Homogenization Anneals of Alloys

W/O C	W/O X	W/O Mn	W/O Si	W/O P	W/O S	Homogenization Time at 1300°C, Days
0.11	–	0.01	0.010	0.004	0.005	2.0
0.11	1.83 pct Si	<0.002	–	0.002	0.007	2.5
0.12	3.08 pct Mn	–	0.080	0.001	0.007	0.5
0.11	3.28 pct Ni	<0.002	0.001	0.001	0.004	2.5
0.12	0.99 pct Co	<0.002	0.001	0.001	0.006	1.5
0.13	0.39 pct Mo	<0.002	0.026	0.001	0.005	2.5
0.29	2.03 pct Al	<0.002	0.001	0.001	0.006	3.5
0.13	2.99 pct Cr	<0.002	0.001	0.001	0.006	2.5
0.12	2.98 pct Cu	<0.002	0.001	0.002	0.007	3.0

Previous investigators have shown that, in a martensitic matrix, formation of austenite begins at the former austenite grain boundaries.[2-5] Fig. 1(*a*), from the present study, shows that the morphology of the austenite precipitated at these boundaries varies from grain boundary allotriomorphs (1) to grain boundary idiomorphs (2). An acicular structure develops in the interiors of the austenite grains[3,5-8] (Fig. 1(*b*)). This structure is said to arise because austenite experiences difficulty in crossing boundaries between adjacent martensite needles or plates[6] and because the acicular morphology forms by a shear mechanism.[5,7] In addition to the acicular structure, intragranular idiomorphs have also been reported in the interiors of the former austenite grains[5,7,9,10] (Fig. 1(*b*)). At late reaction times, the idiomorphs appear to absorb the acicular structure.[7,9] Most of the steels on which these investigations have been conducted were complex; in one investigation, however, Cr was found to allow formation of the acicular structure at lower heating rates.[11]

The alloys used in the present investigation are listed in Table I. These alloys were vacuum-melted and cast, and then homogenized as indicated in Table I. Specimens 0.8 mm. × 6 mm. × 6 mm. were machined

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*The austenite formed during reheating above the eutectoid range is, of course, transformed to martensite again upon quenching. Similarly, the martensitic matrix into which the austenite grows is rapidly converted into a mixture of ferrite and dissolving carbides. For clarity of presentation, however, the transformation product will be described as "austenite" and its matrix as "martensite."

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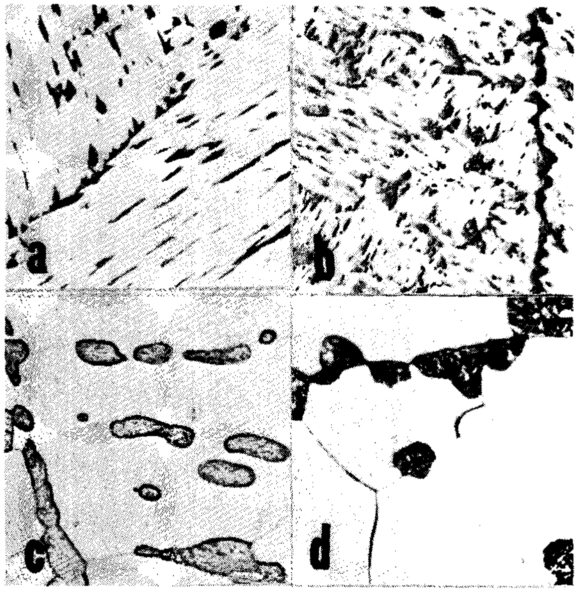

Fig. 2—(*a*) Fe-0.11 pct C, reacted 15 s at 740°C, etched in BASP, Magnification 500 times, typical early reaction time microstructure in a Group I alloy; (*b*) Fe-0.11 pct C-3.28 pct Ni, reacted 30 s at 710°C, etched in BASP, Magnification 500 times, illustrating typical early reaction time microstructure of a Group II alloy; (*c*) Fe-0.11 pct C, reacted 30 min at 740°C, etched in BASP, Magnification 500 times, typical late reaction time structure of an Fe-C Group I alloy; (*d*) Fe-0.13 pct C-0.39 pct Mo, reacted 10 min at 740°C, etched in BASP and 2 pct nital, Magnification 500 times, late reaction time microstructure of an Fe-C-X Group I alloy.

Fig. 3—(*a*) Fe-0.11 pct C-3.28 pct Ni, reacted 60 min at 710°C, etched in BASP, Magnification 500 times, typical late reaction time Group II microstructure; (*b*)Fe-0.12 pct C-3.08 pct Mn, reacted 60 min at 735°C, etched in BASP, Magnification 500 times, typical Group II microstructure at a late reaction time in another alloy; (*c*) Fe-0.29 pct C-2.03 pct Al, reacted 60 min at 750°C, etched in BASP, Magnification 500 times, late reaction time structure of a Group III alloy.

from the homogenized bars, austenitized for 10 min at 1300°C, quenched directly into iced 10 pct brine to form martensite, reacted in lead baths (usually at temperatures not far above that of their eutectoid range in order to reduce the rates of austenite formation), and finally quenched again in iced brine. After mounting in diallyl phthalate,[1] grinding and chemical polishing,[12-14] the specimens were etched in boiling alkaline sodium picrate (BASP).[1] This etch blackens the large number of fine carbides precipitated from the freshly formed martensite (austenite at elevated temperatures) during mounting of the specimens, leaving the essentially ferritic matrix substantially unaffected. Nital etching was used to reveal boundaries within this matrix.

Before considering the alloying element effects, it should be noted that the so-called acicular morphology within the austenite grains is not the product of Widmanstatten-type growth, as this name implies, but rather of allotriomorphic growth along the long faces of parallel martensite needles[15] or laths.[16] The arrowhead in Fig. 1(*c*) shows such a martensite needle boundary (MNB) allotriomorph growing along both sides of a martensite: martensite boundary, and hence simultaneously in two martensite needles. This morphology, and the observation that MNB allotriomorphs formed in the $\alpha + \gamma$ region evolve gradually with time, militate against the view[5,7] that these crystals form by a shear mechanism.

Considering now the observations made upon the influence of alloying elements upon austenite morphology, on the basis of *late* reaction time microstructures the alloys studied have been classified into three groups: I—Fe-C, Fe-C-Mo, Fe-C-Cr; II—Fe-C-Mn, Fe-C-Ni, Fe-C-Cu; III—Fe-C-Si, Fe-C-Al, Fe-C-Co. During the early stages of reaction (times usually less than one min at temperatures just above the eutectoid range), the microstructures of all alloys are similar. Because of the coarse former austenite grain size (ASTM nos. 2 to >1), MNB allotriomorphs were the principal morphology. Those of Group I, however, had a lower density of MNB allotriomorphs than the other alloys (cf. Figs. 2(*a*) and (*b*)). After longer reaction times (10-60 min), marked differences developed among the three Groups. In Group I, the MNB allotriomorphs virtually disappeared; grain boundary allotriomorphs and idiomorphs at the former austenite grain boundaries and intragranular idiomorphs predominated (Figs. 2(*c*) and (*d*)). The microstructure of the Group II alloys, on the other hand, was largely unchanged (cf. Fig. 2(*b*) with Figs. 3(*a*) and (*b*)). The behavior of the Class III alloys was intermediate. MNB allotriomorphs remained prominent primarily in complex patterns developed in limited areas of the specimens (Fig. 3(*c*)); elsewhere the microstructure tended toward that of Group I. The complex structures clearly resulted from preferential precipitation at the irregular boundaries formed where packets of martensite needles oriented parallel to different habit directions came into contact.

Nital etching provided a valuable clue to the differences which developed amongst the three Groups. In Group I, the ferrite grain structure is coarsely equiaxed and bears no resemblance to the pattern of martensite:martensite boundaries from which it evolved (Fig. 4(*a*)). Grain growth especially, but likely also secondary recrystallization appear responsible for this marked change. As a result of this change, many of the austenite crystals are no longer in contact with ferrite grain boundaries. The Group II structures, on the other hand, display a fine, complex, generally acicular network of boundaries in which the influence of the mar-

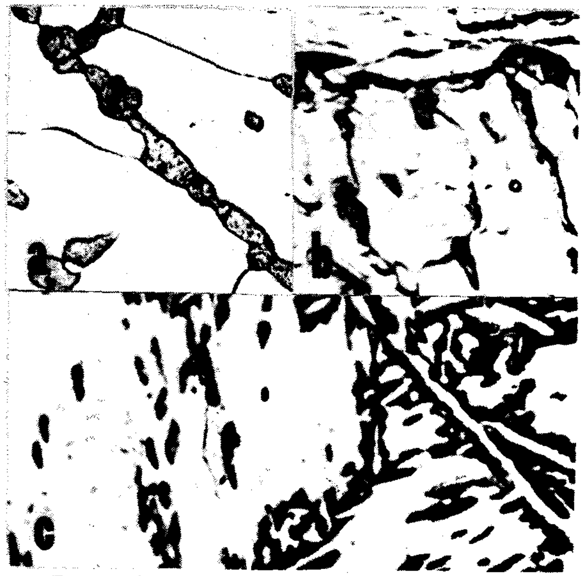

Fig. 4—Ferrite:ferrite boundaries in late reaction time specimens after etching in BASP and 2 pct nital. (*a*) Fe-0.11 pct C, reacted 30 min at 740°C, Magnification 500 times, Group I; (*b*) Fe-0.11 pct C-3.28 pct Ni, reacted 60 min at 710°C, Magnification 2000 times, Group II; (*c*) Fe-0.29 pct C-2.03 pct Al, reacted 60 min at 750°C, Magnification 1000 times, Group III.

tensitic initial structure is still clearly reflected (Fig. 4(*b*)). Note the numerous austenite crystals, often quite small, along these boundaries. In the Group III alloys, only in the areas of complex structure are $\alpha:\alpha$ boundaries, and individual austenite crystals along them, still present in large numbers (Fig. 4(*c*)).

Explanations for the observed alloying element effects were considered in terms of the influence of X upon the martensitic initial structure, the recrystallization behavior of α Fe-X alloys[17] and the effects of X upon carbide precipitation from martensite[18] (since carbides at $\alpha:\alpha$ boundaries provide particularly favorable sites for austenite nucleation[3]); all proved to be unsatisfactory. A more useful approach appears to be the following. The nucleation and growth of austenite at boundaries between martensite needles and migration of these boundaries into lower energy configurations are competitive processes. The more rapid the rate of nucleation of austenite at martensite:martensite boundaries, the more effectively will the motion of these boundaries be inhibited. It appears that in the Group I alloys, where the density of MNB allotriomorphs is initially lower than in the others, boundary migration occurred with sufficient rapidity relative to the kinetics of austenite formation not only to eliminate many boundaries but also to allow the surviving boundaries to be sufficiently free of austenite so that they were able to break away from many of the MNB allotriomorphs which did form. When no longer in contact with a ferrite:ferrite boundary, austenite crystals rapidly spheroidize because of the larger driving force for this process which is then available, and are thus much more susceptible to dissolution by coarsening. Hence only a relatively small number of austenite crystals remain in the interiors of the former austenite grains, and the survivors are largely idiomorphs. In the Group II alloys, on the other hand, rates of austenite nucleation appear to have been sufficiently high to restrict significantly the amount of martensite:martensite boundary migration which could occur. The resulting inhibition of breakaway from MNB allotriomorphs further ensured retention of a martensite-like arrangement of the ferrite boundaries. The MNB allotriomorph morphology effectively resists coarsening because of the relatively large radii of curvature of its $\alpha:\gamma$ boundaries. Evidently the nucleation rates of austenite were intermediate in the Group III alloys, being high enough to produce Group II-type pinning only amongst the higher energy boundaries produced by the transformation into contact of martensite needles parallel to different habit directions. The nucleation and growth kinetics of austenite in Group II alloys relative to those in Fe-C can be explained by the higher driving forces for both of these processes which follow directly from the lower temperature range of the no-partition Ae3 in these alloys—just the reverse of the situation in the proeutectoid ferrite reaction.[19] In Group I, Mo slightly decreases the driving force; Cr may increase it a little, but the "solute drag-effect" upon growth, which occurs when X is a strong carbide-former,[20] can cause both to retard reaction kinetics. On the other hand, the Group III elements should decrease the driving forces for austenite formation, yet appear to increase somewhat the kinetics of the reaction. These elements may, however, decrease the interfacial free energy required for nucleation; even a small change of this type would more than offset the lower driving force.

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