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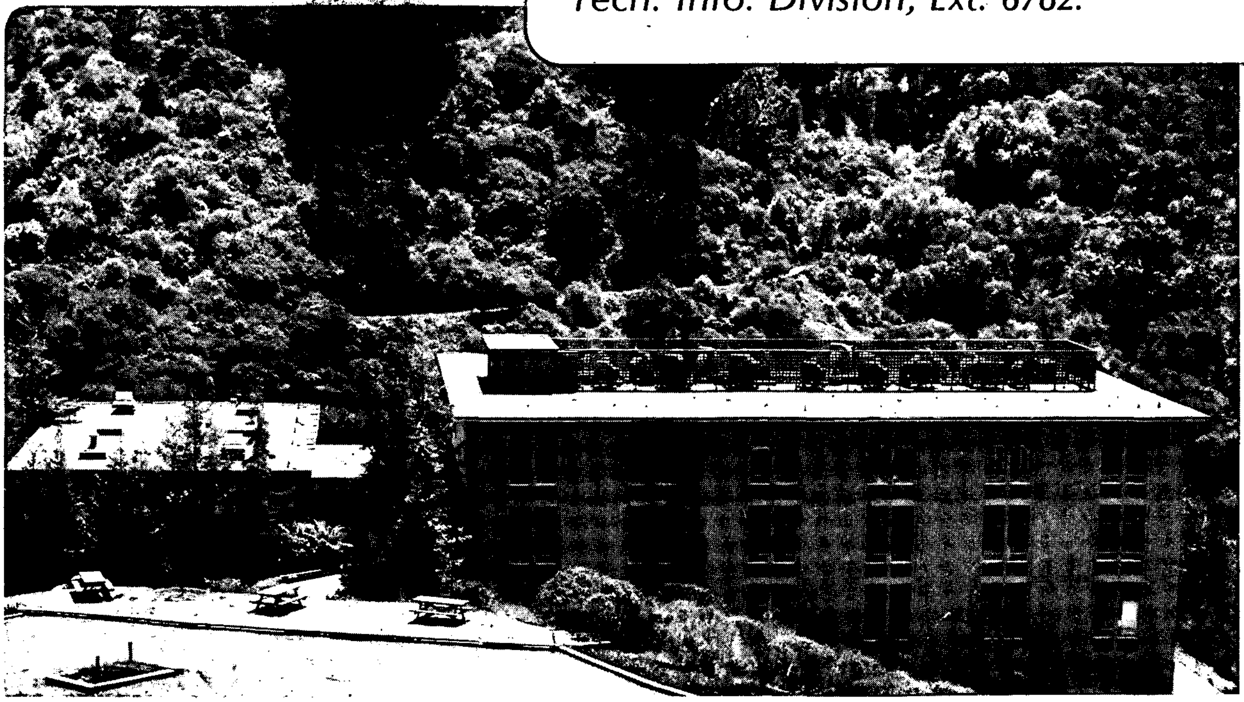

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The Physical Metallurgy and Alloy Design of Dual Phase Steels

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INTRODUCTION

The terminology "dual phase" has become accepted to define a class of low-carbon, low-alloy steels [1] that generally contain two phases, ferrite and martensite or ferrite and bainite, which can be obtained by relatively simple processing involving quenching from the two phase ferrite-austenite field (Fig 1). Strictly speaking, of course, these steels may actually contain more than two phases. If the martensite is low carbon lath martensite, it will contain interlath austenite (see Fig 4(b)), and if high temperature decomposition of austenite has occurred during quenching, i.e., if bainite is present, there will be bainitic carbides. In microalloyed steels the ferrite phase may also contain the corresponding alloy carbides or nitrides. Thus, whilst the term "dual phase" represents the simple principle of obtaining two phases, the actual microstructures may be multiphase and quite complex. It is the object of this paper to discuss the general physical metallurgy of "dual phase" steels and to indicate some potential applications to developments besides flat rolled products. Since in principle these steels can have simple compositions and are economically attractive, and can be designed to provide excellent combinations of cold formability, strength and ductility, they represent an exciting development in steel metallurgy over the past decade [2-5].

A number of factors have triggered intensive development programs resulting in the introduction of new steel compositions and processing techniques. The need for economical higher strength steels with good formability in transportation industries to achieve weight reductions and fuel savings as well as today's energy and resource conservation requirements are among the factors generating strong incentives to produce better steels than those in general use at minimum cost penalty.

Pursuant to these increased requirements, the alloy design trends for structural applications have focussed attention on the detailed physical metallurgy of low-carbon, low-alloy steels, thus enabling greater exploitation of their potential for higher strength along with sufficient ductility for a wide range of applications. As an outgrowth of this movement, for example, conventional hot rolled low-carbon steels which were adequate to fill the material requirements for the majority of structural applications for many years are progressively being replaced by high-strength, low-alloy (HSLA) steels. The principal strengthening of these HSLA steels is derived from precipitation of finely dispersed alloy carbides and grain refinement [6]. However, their overall mechanical properties are not always satisfactory for many applications (e.g., formability).

The search for an alternative has spurred the recent development of duplex ferritic-martensitic (DFM) steels. These are a new class of HSLA steels whose approach to strengthening contrasts markedly with microalloyed HSLA steels in chemistry as well as processing technique. Interest in the DFM steels has arisen since:

(1) The required composite microstructures can be produced solely by simple heat treatment. The alloy composition and heat treatment may be varied to give different transformation paths and hence different morphologies and mechanical properties [Fig 1].

(2) Dual phase steels have mechanical properties which are characterized by continuous yielding with high initial work hardening rates, large uniform strains and high tensile to yield ratios. These factors account for their better formability compared to that of ferrite, pearlite or HSLA steels of similar strengths (see Fig 2) and allow attractive combinations of strength and ductility to be obtained.

(3) Carbide forming elements as in commercial HSLA steels are not always essential (but may be utilized, as will be discussed later), so simple alloy chemistries are involved.

The major source of strengthening in the DFM structure arises from the presence of inherently strong martensite as a load carrying constituent in a soft ferrite matrix which supplies the system with the essential element of ductility. The resulting mixture is analagous to that of a composite but which can be obtained solely by heat treatment. The concept of fiber-composite strengthening is thus useful in qualitatively understanding dual phase steels as will be discussed later.

Undoubtedly, the occurrence of DFM aggregates is one of the oldest phenomena in the history of martensitic transformations in steel since these mixed microstructures are produced as an unavoidable consequence of incomplete austenitization and/or lack of sufficient hardenability. In this regard, earlier investigators, e.g., Herres and Lorig [7] considered two phase aggregates to be undesirable microstructural features rather than potential strengthening devices in low carbon steels. The idea of the DFM structures as possible materials of technological interest was recognized by Cairns and Charles [8] in 1967 who produced controlled microstructures of elongated regions of martensite in a ferrite matrix either synthetically or by a combination of cold deformation and rapid reheating. However, this technique suffers from various disadvantages and the resultant mechanical properties were no better than could be obtained from suitably heat-treated alloy steels. In contrast, Grange [9] utilized the phase transformation occurring in the two phase ($\alpha+\gamma$) range in Fe-C phase diagram to obtain fibrous mixtures of martensite and ferrite in various carbon steels by thermomechanical treatments. Again, the processing technique is complex and only limited success in improving mechanical properties was achieved.

There is no doubt that the fuel crisis and pressure on the automotive industry to improve fuel economy spurred the initial development of dual phase steels for flat rolled sheet products. Owen [10] has given a review of the dual phase steel potential for automotive industry and no attempt will be made to review this again here. Very recently, the potential of dual phase steels for applications other than flat rolled products, e.g., line pipe, rod, wire, etc., has been realized because these steels can be designed to optimize the ever conflicting property requirements of strength and ductility. Some discussion of this potential will be given later in this paper.

II. FUNDAMENTAL CONSIDERATIONS OF THE SELECTION OF DUAL PHASE STEELS

A. Nature of Two Phase Materials

Most of the modern engineering materials of technological importance are aggregates of two or more phases, or composites of great variety. For metallurgical applications, these composite materials can be divided into two main categories in terms of the strengthening mechanisms involved and associated manufacturing techniques. These are fiber or particle reinforced composites, and dispersion strengthened materials, both of which contain a second phase(s) which is harder than the matrix.

In the former case, two physically separable materials, i.e., particles or fibers and metal matrix are compounded into an aggregate. Its strengthening mechanism is principally governed by the well established fiber-loading concept. On the other hand, the latter alloy is a type of composite in which the second phase particles are usually non-deformable, and the Orowan mechanism for yielding is operative when the particle size is less than several microns (see e.g. Ref 11). However, when the size of the second phase is in the range of ~1μ50 , which corresponds to the transition between fiber reinforced strengthening and dispersion hardening, there exists no simple model for a strengthening mechanism that governs the flow property of the alloy. This situation involves complications in theoretical analysis such as the particle/particle interactions and stress concentration effects caused by discontinuities at the particle. Therefore, the specific strengthening mechanism governing a two-phase material is determined by the size and morphology of the second phase. However, one thing that is common in the nature of two phase materials is the way in which the second phase particles enhance the strength of the composite system, i.e., they exert constraints upon the matrix. Accordingly, upon deformation, the material in their neighborhood is restricted from elongating freely, resulting in an increase in shear stresses required to produce further deformation. Thus the characteristic mechanical behavior of the composite includes early and fairly extreme work hardening rates which suppress mechanical instability as well as a considerably greater flow stress than that of the pure matrix. While the better engineering properties of the strong particles are utilized in this way, the effect of some of their less desirable features, such as brittleness, is simultaneously mitigated by the presence of a ductile matrix that binds the particles together. Thus attractive tensile properties can be achieved as illustrated in Figs 1,2.

B. Two Phase Mixture Rule

Most experiments and theoretical analyses have been performed in composites containing either unidirectionally aligned fibers or very finely dispersed particles. Thus the understanding of the principles and practical utilizations of these two well defined composites have been fairly well established, and comprehensive reviews on the subject appear in the literature. In the case of randomly oriented coarse two phase structures, however, current understanding of their mechanical behavior is far from complete. Duplex structures of this kind occur in many technologically important alloys, and thus have received much attention in the past. As a result, much data concerning their mechanical properties has been reported in the literature (e.g., refs. 1-5). Much of the empirical evidence accumulated to date suggests that, in spite of the absence of an exact theoretical explanation, the strengthening of the duplex alloys can be predicted with reasonable accuracy from the volume fraction and strength of the components by the mixture rule of the form:

$$\sigma_c = \sigma_f V_f + \sigma_m V_m \qquad [1]$$

where the subscripts c, f, m refer to composite, ferrite and martensite, V_m and V_f are the volume fractions of martensite and ferrite respectively. σ_f is the stress carried by the ferrite matrix when the composite is strained to its ultimate tensile stress [1-5, 12-16]. In a strict sense, such an equation as [1] holds only for describing the mechanical behavior of unidirectionally aligned, continuous fiber composites assuming that the fibers and matrix are well bonded and the fiber strain, matrix strain and composite strain are all equal. The law of mixtures predicts a dependence on V_m which can be positive, constant, or negative depending on the magnitude of the other parameters. The attractive feature of dual phase steel is that all of these parameters can be controlled to a considerable extent by simple heat treatments and/or thermal-mechanical processing. However, V_m and σ_m are interrelated as discussed below.

C. Selection Principles

By adding second phase particles to ductile matrices, ductility is usually sacrificed for strength. The implication of the mixture rule is that the loss of ductility is proportional to the volume fraction of martensite (V_m). However, one can imagine many duplex distributions of martensite particles that, when properly spaced in dual phase alloys, will result in not only strengthening but also minimizing loss in ductility. This goal can be achieved by proper control of parameters such as the size, shape, and distribution of the martensite particles, and the properties of the individual phases [14]. The following guidelines for developing desirable dual phase structures are thus suggested:

1. *Optimum Volume Fraction*

The volume fraction and morphology of the martensite particles plays a major role in the control of mechanical properties of DFM alloys [Figs 1,6]. V_m must exceed the lower limit below which substantial contribution to strengthening does not occur. At the same time, V_m must be less than the upper limit beyond which the failure of the particles immediately leads to the failure of the duplex structure. In other words, if the fracture of one particle occurs at some weak point, it will cause stress to be transferred back to the matrix near the particle break. Ideally, however, these cracks will be subcritical in size and their local stress intensification will not cause other particles to fracture immediately in the same cross section.

Crack propagation may be delayed considerably if V_m is in the optimum range range so that the matrix can work harden locally and carry the transferred load. In addition, the alloy system must be chosen such that a slight temperature variation during heat treatment will not significantly alter V_m so as to ensure reproducibility of the materials.

2. *Size and Shape Factors: Finer Scale of the Substructure and Fibrous Particles are Desirable*

If martensite particles are sufficiently small (on the order of a μ or less), and if the spacing of the particles is such that the yield of the matrix is controlled by the micro-mechanistic model instead of a fiber loading concept, then an increased yield stress of the matrix coupled with an increased rate of work hardening is expected to be observed [16]. With regard to shape factors, load transfer is most efficient when particles are present in the form of fibers rather than spheres. This is primarily because the transfer of load occurs by shear action along the paticle/matrix interfaces and, for a given volume fraction, and the same number of particles, more interfacial area is available in the case of a fibrous morphology.

3. *Distribution: Discontinuous Particles in a Continuous Matrix*

Practical considerations argue that unidirectionally aligned martensite particles throughout a ferrite matrix cannot be produced by simple heat treatment alone. It is more likely that a mixture of macroscopically randomly oriented DFM structure will be obtained, thus resulting in isotropic mechanical properties. This is desirable in a sense that for most engineering materials at least partial isotropy will be required. DFM structures where the harder particles are connected throughout the matrix should be avoided since they will be directly loaded at their ends. This will cause the particles to break in the early stage of plastic deformation and prevent the ferrite from attaining its full ductility (Fig 1 - Alloys A,B).

4. *Carbon Content*

Martensite is inherently strong, because of its defect and alloy content. It is a diffusionless transformation product of carbon-enriched austenite. Hence, the average carbon content in the initial alloy should be kept sufficiently low to maintain a carbon level of about 0.3 pct. in austenite, resulting in dislocated martensite of the optimum V_m. Higher carbon contents can drastically degrade the toughness of martensite by forming brittle twinned plates (Fig 3). This is immediately apparent from the partitioning of carbon (Fig 4a) during intercritical annealing. Figure 4b shows the desired lath martensite, with interlath retained austenite films.

5. *Microstructural Features*

The stress in the continuous matrix is transferred to the strong particles by shear at the particle/matrix interface. As a consequence, the nature of the interface and nearby microstructural features are of critical importance. Since the composition of a dual phase alloy should essentially be low in carbon and other alloying elements, even thoughsince partitioning occurs during intercritical annealing, the resultant lack of sufficient hardenability may cause a pearlitic or bainitic reaction to occur with (Fe_3C or M_3C formed) during quenching from austenite to martensite. These carbides, if any, will be located near the interface where stress concentrations can originate formation of cracks within the carbides, eventually leading to complete failure of the specimen.

6. *Nature of Interfaces*

A high energy interface is more likely to fail under stress than a low energy one, since a high interfacial energy substantially reduces the work required to separate a particle from the matrix, and thus creates a void. Ideally, therefore, the martensite/ferrite interface should have good atomic fit. This can be ensured with dislocated packet martensite in ferrite because of the Kurdjumov-Sachs orientation relation which exists between the close packed planes of the prior austenite and its subsequent transformation product, martensite. This is especially true for fibrous microstructures as shown in Fig 5.

7. *Processing and Choice of Alloying Elements.*

The considerations above suggest desirable microstructural characteristics of a dual phase steel that is to be both strong and ductile. The most difficult challenge will be to produce DFM steels with these considerations for a sufficiently small cost premium to make them economically competitive. As is common practice in alloy design, simple ternary Fe/X/C alloys should be considered prior to advancing to a more complex system. The success achieved will then depend largely on a favorable choice of the substitutional alloying element, X, and processing.

III. HEAT TREATMENTS, ALLOYING, MICROSTRUCTURES

A. Transformation Path

Controlled amounts of martensite can be produced in a ferrite matrix using many different heat treatments (e.g., Figs 6,7), all of which involve phase transformation following holding in the ($\alpha+\gamma$) region. Choice of the specific heat treatment will be dependent on the alloy composition, property requirements, and production capabilities.

Of particular interest is an intermediate quenching treatment (Fig 7I). This treatment was developed so as to fully exploit the characteristic nature of the initial martensite structure prior to subsequent annealing in the ($\alpha+\gamma$) range [5]: i.e., (i) it provides sufficient heterogenous nucleation sites for austenite (martensite at room temperature) during two phase annealing, and (ii) parallel laths of extremely small width within a prior austenite grain can be utilized to produce a fine, fibrous distribution of martensite in a ferrite matrix (Figs 4b, 7). Moreover, these advantages allow ease of control of the size, distribution, and shape of the martensite particles, depending on the amount of type of the alloying element X.

Silicon and aluminum have been found to be favorable for obtaining fibrous structures. Silicon also has the following advantages:

1. Silicon broadens the ($\alpha+\gamma$) temperate range when added to the Fe-C system, and increases the slope of the $\gamma/\alpha+\gamma$ solvus, thereby facilitating practical control of the martensite volume fraction and the martensite carbon content. This can be understood by referring to Fig 8.

2. Promotion of a fine, fibrous distribution of martensite in a continuous ferrite matrix (see Fig 1, Alloy C).

3. Maintains coherency of the ferrite/martensite interface by inhibiting the formation of coarse carbides during the final quench.

4. Provision of very effective solid solution strengthening in the ferrite.

5. Low cost and availability.

B. Phase Diagrams

Recognizing that the main objective in dual phase steels is the interfacing of the soft ductile phase with the hard (tough) martensite phase, it is necessary to first consider the appropriate phase diagrams. The basic principles for heat treatment can be easily seen from Fig 8. Referring to the upper phase diagram and following the lever rule, if a steel with, say, 0.075%C is heated to about 840°C and equilibrium is attained, the microstructure will consist of 40% γ of composition 0.16wt% carbon, as given by the tie-line at this temperature. If the same steel is equilibrated at 805° C, the microstructure will consist of 20% γ of composition 0.31 wt%C. On rapid quenching, the austenite will be transformed to martensite. Hence, the volume fraction can be controlled merely by controlling the temperature of holding in the $(\alpha+\gamma)$ range, i.e., the so-called "intercritical annealing" condition.

The following important points can be made from this consideration:

1. In plain carbon, lean alloys such as 1010 steel, small temperature changes during intercritical annealing could have large effects on the volume fraction and carbon content of the martensite. However, to a first approximation these two changes have opposite and largely compensating effects on properties since for dislocated martensite its strength σ_m is linearly dependent on % carbon viz $\sigma_m = G\varepsilon^2 C$. Where ε is the strain, C is the carbon content and G shear modulus (ref 11).

Comparing the case consdered above, i.e., 40% and 20% martensite, the ratio of strengths of the martensite is:

$$\frac{\sigma_m(40) = G\varepsilon^2(0.16)}{\sigma_m(20) = G\varepsilon^2(0.31)} \text{ is} \cong \frac{1}{2}$$

and the ratio of volume fractions is

$$\frac{V_m\ (40)}{V_m\ (20)} = \frac{2}{1}$$

so the dominent martensite term $\sigma_m V_m$ in equation [1] is essentially a constant, over this range of martensite volume fraction.

2. Alloying additions which affect the slopes of the $\gamma/(\alpha+\gamma)$ and $\alpha/(\alpha+\gamma)$ solvus lines will thus change the temperature ranges required for appropriate intercritical annealing to achieve the volume fraction, carbon content, and strength of martensite desired.

Alloys can thus be designed to take advantage of these changes, e.g., referring to the lower part of Fig 8, ferrite stabilizers increase the slopes of both solvus lines under consideration. What this means is that changes in intercritical annealing temperature have less effect on martensite volume fraction and martensite strength through its carbon content, e.g., to obtain the same 2:1 change in martensite volume fraction for 0.075% carbon steel the annealing temperature must be raised from 810°C to 890° C rather than 840°C for the case of plain carbon steel (upper figure). From a practical viewpoint, therefore, the temperature control to attain a given volume fraction is less critical for steeply sloping $\gamma/(\alpha+\gamma)$ solvus lines and this is one of the values of silicon additions.

The change in slope of the $\gamma/(\alpha+\gamma)$ solvus means that in alloys quenched from lower temperatures, e.g., compare again the 0.075% carbon steels (Fig 8), the carbon content in ferrite on quenching from 820° C is twice that in ferrite quenched from 890° C, these temperatures giving twice the martensite volume fraction. Hence, the lower volume fraction martensite steels contain supersaturated ferrite, so that subsequent carbide (or nitride) precipitation could occur in the ferrite and compensate for possible lower strengths associated with lower volume fractions of martensite. Obviously, some sacrifice in ductility of ferrite will also occur. Figure 9 illustrates such carbides in a Fe/0.15C/1.55Si/0.03Nb Steel [17].

Returning then to the law of mixtures:

$$\sigma_c = \sigma_f V_f + \sigma_m V_m = \sigma_f(1-V_m) + \sigma_m V_m$$

it can be seen in principle that three main variables σ_m, V_m, σ_f, can be manipulated to achieve desired properties. In fact, one can obtain a decrease in strength with an increase in V_m (Fig 10), ref [17].

C. Retained Austenite: "Multiple Phase" Steels

Depending on composition, heat treatment temperature (Fig 8) and cooling rate from this temperature, it is possible to obtain more than the "normal" two phase microstructures of martensite (vol. fraction ~10-30%) and ferrite because retained austenite and bainite can be produced [18-20] in addition to these two phases, particularly in commercial "dual phase" steels derived from microalloyed or HSLA compositions especially with vanadium. The retained austenite is in the form of blocky grains (Fig 11) and is not to be confused with the interlath retained austenite which is a normal component of the dislocated packet martensites. The volume fraction and composition of the microconstituents and hence mechanical properties can be altered by varying processing parameters and composition. Steels with about 0.1%V and 0.1%C can be air cooled or oil quenched to yield 3-10% blocky austenite [20], but if more rapidly quenched (~ 80° C per sec.) [21-22], the normal ferrite-martensite mixtures are obtained.

The possible ranges of microstructures attainable with a microalloyed duplex steel (e.g., Van 80 type or experimental Fe/Mn/Si steels with V, Nb, Mo etc.) can be understood qualitatively from the schematic transformation diagram [22] shown in Fig 12.

In addition to retained austenite and "retained ferrite" (i.e., ferrite formed during intercritical annealing), some austenite may transform to bainite or the interphase type reaction to produce arrays of alloy carbides or carbonitides [21] dispersed in the "new ferrite". This latter reaction is shown schematically in Fig 12. Such reactions can of course raise the yield strength of these steels compared to those with precipitate free ferrite.

Since the cooling rate has important effects on the isothermal or non-martensitic decomposition products as well as on solute partitioning, it is clear that a wide range of microstructures and properties can be obtained by varying this parameter. Likewise, in a given steel product non-uniformity in structure and properties is likely in parts of the product that may be cooled at different rates because of different section thicknesses. For this reason, it is unlikely that low carbon dual phase steels would be used for sections much thicker than one-half inch. Recent in-situ straining studies [23] have proved that the TRIP phenomenon [24] occurs readily for blocky austenite particles, $\gtrsim$ 1-6 μ in size whereas smaller, submicron, austenite particles resist transformation. The interesting observation is that the TRIP martensite is of the twinned plate variety indicating solute enrichment of the prior austenite occurs by alloy partitioning especially manganese and silicon, during intercritical annealing. Of course, the carbon content of austenite is also enriched as is clearly evident from the phase diagram (Fig 4a). Even if Ms is below room temperature, M_d may be above room temperature so that the TRIP phenomenon can occur if the martensite particles are not stabilized (the well known size effects on stabilization [e.g., see ref 25]).

In this context, it is very significant that due to heterogeneous chemistry and size distribution, individual particles of retained austenite in air cooled steels will have different stabilities and therefore undergo the martensitic transformation to provide plasticity over a range of initial strains rather than at one time. The low yield strength and high initial strain hardening obtainable in such steels can be explained in part by the retained austenite transformation to martensite and the resultant strengthening of the adjacent ferrite. The stress-induced transformation of the retained austenite could trigger flow in the surrounding ferrite at a lower stress than would otherwise be possible. This is also supported by the observation that a strictly ferrite-15 wt pct martensite steel has a higher yield strength than that of the air cooled steel (also with 15 vol pct martensite). [20]

The transformed ferrite could be beneficial because it has a higher strength than the retained ferrite and could provide an intermediate load transfer step before deformation is induced in the martensite. The strength of this ferrite is attributed to the resistance to dislocation motion provided by aligned, banded carbonitrides which are observed [21] and are explained by Fig 12.

The presence of a lower amount of martensite with air cooling provides an advantage, since martensite formation is associated with a substantial volume change. A large volume change requires accomodation by the ferrite which causes pre-yielding of ferrite and a reduction in formability.

D. The Martensite Phase

Since the mechanical properties of martensite depend upon its substructure and composition, control of these factors is necessary if one is to optimize the benefits of the martensite phase (see e.g. ref 26 for summary). To do this, one calls upon the vast amount of research done on martensitic steels. The important features that need to be remembered are: 1) strong, tough martensites are associated with packets of dislocated laths surrounded by stable interlath films of austenite; 2) these structures are obtained where the carbon level is below about 0.4%, depending on the other solutes; 3) twinned plate martensites obtained when carbon is 0.4% tend to be brittle. These effects are summarized in Figs 3,4.

In the alloy design program at Berkeley, care has been taken to obtain dual phase steels with tough martensites so that alloy compositions and heat treatments are utilized to keep the carbon content of martensites well below 0.4%. Obviously, this sets the lowest intercritical annealing temperature for the alloys considered and hence the lowest volume fraction of martensite. For plain carbon steels this temperature is 760° C (Fig 8, upper diagram).

Another important aspect of utilizing fiber composite principles for dual phase steels is that of applications involving large strain deformations. The coherency of the interface between ferrite and martensite as well as the incompatibility in properties will depend on morphology and the parameters discussed above. Fibrous morphologies can be obtained with silicon additions and in these cases excellent coherency is achieved at the two phase interfaces. This is readily revealed by high resolution electron microscopy (Fig 5) where the slip planes (110) are parallel in both phases. This crystallographic compatibility will minimize void formation during plastic flow and becomes an important consideration in rod and wire drawing where true strains of up to 5 or more can be achieved.

E. Morphology

The morphology of the dual phase microstructure of a particular steel is controllable by heat treatment and thermal-mechanical treatment, in addition to the effects of alloying elements. From the point of view of heat treatment, the transformation path by which ferrite and martensite is achieved is important, e.g., as shown in Figs 6,7, viz.,

Continuous (or Step) Cooling: austenite $\rightarrow$ $(\alpha+\gamma)$ $\rightarrow$ $(\alpha+Ms)$
Heating: martensite or pearlite $\rightarrow$ $(\alpha+\gamma)$ plus
quenching $\rightarrow$ $(\alpha+ Ms)$

Fig 7, obtained for Fe/2%Si/0.1%C steel shows the large variation in microstructures that can be produced by these different heat treatments. The properties resulting from these microstructures at the same volume fractions can be quite different, especially in fatigue (see below Sec. IV (d)).

IV. STRUCTURE-PROPERTY RELATIONS

A. General Considerations: Morphology

The attractive tensile mechanical properties of dual-phase steels are illustrated in Figs 1,2,13,14. These properties include continuous yielding, a high work hardening rate, low yield strength, high tensile strength, good ductility and good formability [27,28]. The continuous yielding results from residual stresses and the high density of mobile dislocations formed during the austenite to martensite transformation (Figs 5,11). These dislocations and the soft ferrite cause the low yield strength. The incompatibility between the hard and soft phases causes a rapid rate of work hardening leading to a high tensile strength and suppresses necking, resulting in good ductility and formability [28].

Good ductility appears to be associated with the connectivity of the phases present, especially interconnected martensite regions along the prior austenite grain boundaries. Thus the transformation path by which the final ferrite (α) + martensite (Ms) is obtained is very important. Also, the influence of alloying elements on the kinetics and morphology of the "step quenching" [$\gamma \rightarrow (\alpha+\gamma) \rightarrow (\alpha+ Ms)$] or "intermediate quenching" [(α+ pearlite) or Ms $\rightarrow (\alpha+\gamma) \rightarrow (\alpha+ Ms)$] transformation paths must be considered (Fig 6). The detailed microstructural characteristics will depend on the specific alloying elements present. Examples are illustrated in the optical micrographs shown in Figs 1,7. The microstructures differ in their respective macroscopic morphologies such as in fineness of islands in the ferritic matrix. This geometric difference will influence the homogeneity of plastic flow in the ferrite region upon deformation. As a result, significant changes in the mechanical properties of the duplex structures are observed.

It is important to emphasize here that the other microstructural features, e.g., precipitates and substructure, must be carefully identified and characterized if valid correlations between the microstructures and properties are to be established. In addition, composition changes due to solute partitioning should be identified if possible [29-31]. The presence of any fine precipitates in the ferrite, for instance, can affect the mechanical properties of duplex steels, as discussed above. In most alloys studied, both uniform and total elongations at the same volume fraction of martensite of the intermediately quenched structure are superior to those of the step quenched structure (Fig 7) over a wide range of martensite volume fractions examined. From a continuum mechanics viewpoint, this is to be expected, since the coarse and the more connected martensite geometry in the ferrite causes more severe inhomogeneous deformation and restricts initial plastic flow to a smaller fraction of the total volume of the ferrite matrix [32]. Also, void growth will occur at a faster rate when martensite is more interconnected, but with less plastic strain. As a result, the overall ductility measured by tensile elongation and reduction in area will be lower [32,33]. A similar trend, but a more drastic decrease in ductility has been found for the duplex Fe/2Si/0.1C steel subjected to the $\gamma \rightarrow (\alpha+\gamma)$ transformation path compared to that of the (α+ pearlite) → $(\alpha+\gamma)$ treatment.

B. Strength and the Law of Mixtures

Strength seems to be less sensitive to morphology than ductility, and depends to a good approximation on the "law" of two phase mixtures, given in Equation 1. Figures 1, 13, 14 shows the linear increase in strength as a function of volume fraction martensite. It in fact suggests that the mixture rule, which is based on continuum mechanics, should be applied to the duplex steels in conjunction with dislocation theory (the micromechanical viewpoint) since the latter emphasizes the characteristic slip distance available in the ferrite. The slip distance will in turn be determined by the substructure and macroscopic geometry of the constituent phases. An analysis of this has been given by Koo et.al.[16].

Nonetheless, the mixture law provides important guidelines which can predict with fairly good accuracy the strength level of duplex steels at the various volume fractions of martensite, provided a minimum of two data points are known. However, deviations can occur in the above relationship of the law of mixtures when third phase precipitates, especially in the ferrite, are present (Fig 10).

C. Fracture

1. *Tensile Fracture*

Coarse dual phase structures obtained by continuously annealing in the two phase region directly from austenite results in poor elongation ductility with relatively high strength. However, by obtaining a fine fibrous or fine globular dual phase structure by following different transformation paths with alloying control, significant improvements can be obtained in elongation ductility without much sacrifice in strength. The poor elongation ductility of the coarse dual phase structure is due to the initiation of cleavage cracks in regions in the ferrite where intense localized stress concentration developed during straining. But in steels with both fine fibrous or globular morphologies, fracture occurs by void nucleation and coalescence only after large amounts of plastic deformation: these effects are illustrated by the composite of Fig 15.

2. *Impact Toughness*

Control of morphology, combined with grain size can be very effective in providing dual phase steels with excellent low temperature properties. This has been achieved by controlled rolling as schematically shown in Fig 16, such that either or both prior austenite and austenite-ferrite grain sizes are refined, depending on the alloy being treated. In the case of Fe/1.5Mn/0.06C steel, the ductile-brittle transition temperatures have been lowered to below -100°C, with little loss in strength (Figs. 17,18) [34].The possible benefits of such treatments for line pipe applications have been reviewed by Kim [35].

D. Fatigue

Fatigue resistance of dual phase steels is very important with regard to many applications, particularly automotive [36]. In general, there are two types of fatigue testing: (i) the classical S/N at low cycle fatigue approach which in essence gives information on crack initiation and (ii) fatigue crack propagation and threshold stress intensity measurements on pre-notched or pre-cracked samples [37], which in many cases is probably more representative of real engineering conditions. As yet, the amount of literature on microstructural effects on fatigue of dual phase steels is relatively small [38]. Current research on Fe/2Si/0.1C steels [39,40] heat treated as shown in Fig 7 indicate extremely attractive fatigue crack growth properties as shown in Figs 19,20. Figure 19 shows the variation in crack propagation rate (da/dn) as a function of the stress intensity range ΔK (ΔK= K maximum/K minimum) for load ratios R= K minimum - K maximum of 0.05 and 0.75. It can be seen that the coarse microstructures show exceptional high threshold ΔKo values as compared for small load ratios, without compromising strength. Thus, when compared to existing steels (Fig 20) the fatigue crack growth resistance of the dual phase Fe/Si/C steel looks very promising. The current interpretation of these results is based upon crack deflection and toughness-induced crack closure mechanisms [41] so that this phenomenon only applies to crack propagation rather than crack initiation criteria. In the latter case, S/N behavior of dual phase steels is similar to conventional steels in that fatigue resistance is related to tensile strength.

V. POTENTIAL FOR COLD FORMING APPLICATIONS

The formability of dual phase steels make them ideal candidates for almost all cold forming operations involving section sizes which can easily be heat-treated to obtain the necessary starting two phase microstructures. Commercial applications to automobile parts, especially sheet, involving also the development in Japan of new continuous annealing-quenching-cooling lines, have already occurred, but the potential of dual phase steels for general applications does not yet seem to be widely explored. Wire drawing is one example of current research at Berkeley [43].

The capability of dual phase steels to be heat-treated as rods and then cold drawn to wires is illustrated by the micrographs shown in Figs 21,22. Not only can dual phase steels be drawn to strength levels and ductilities found in conventional pearlitic drawn steels (~0.7%C) but no intermediate heat treatments such as the so-called patenting treatments (Fig 23) seem to be necessary. Consequently, dual phase steels show very good promise as a new economical approach to bars, rods, and high tensile strength wires. One problem, as exists in all steel for wires, is the incompatability to drawing encountered at non-metallic inclusions or large second phase particles. Figure 24 is a scanning electron micrograph showing fracture at one side of a large martensite particle which did not deform in contrast to the nearby finer acicular martensite particles The electron micrograph of Fig 25 shows the development of dense elongated dislocation cells after drawing and it is difficult to distinguish between ferrite and martensite. These examples illustrate again the need for control of microstructure (morphology, volume fraction, etc.) if the advantages of dual phase steels are to be fully exploited. Current work indicates wires of tensile strengths up to 400,000 psi can be obtained; this is remarkable for such low carbon steels (~0.08%C). Another potential application is in line-pipe steels. Such steels must have high strength as well as high toughness (low ductile-to-brittle transition temperature). The alloy design principle for pipeline application [26] is that high toughness can be obtained primarily by grain size control, while maintaining a dual-phase structure to simultaneously obtain high strength as shown in Fig 18. Controlled rolling of an Fe/1.5Mn/0.06C alloy, followed by direct quenching produces a duplex microstructure in which upper bainite particles are uniformly distributed within the fine, ferrite matrix. The mechanical properties of this ferrite-bainite steel, in the as-hot-rolled condition, far exceed the property specifications for arctic pipeline, as shown in the table. A further advantage of this ferrite-bainite steel is that there is an abrupt increase in strength during pipe-forming due to its high work-hardening rate. A review of the physical metallurgy of HSLA steels for line-pipe has recently been given by Kim (35], who points out that it is possible that the development of improved strength in ferrite-pearlite steels has reached a limit. While several bainite/acicular ferrite steels are available with promising properties, multiphase steels appear to show more scope for development because they possess the advantages of both ferrite-pearlite steels and bainite/acicular ferrite steels. They have high toughness due to their fine polygonal ferrite matrix while the strong second phase gives high strength. However, they may have some limitations in applications, especially for sour gas and oil services, because of the well-known susceptibility of martensitic and bainitic steels to hydrogen induced cracking and stress corrosion cracking [44]. While there is limited evidence that strong second phases, when embedded in a ferrite matrix, are not preferential sites for crack nucleation [45], the effect of second phases on environmental degradation should be well understood before application of multiphase steels to potentially aggressive environments can be undertaken.

VI. SUMMARY

A. Effects of Alloying Elements

The overall effects of alloying elements on the microstructure and properties of dual phase steels can be interpreted in terms of the phase transformations that occur on subsequent cooling from the intercritical annealing temperature. These transformations depend in turn upon solute partitioning. At equilibrium (Fig 8) the partitioning is given by the respective tie-line. On cooling fast enough to suppress isothermal decomposition products of austenite, the martensite can be low carbon lath type or high carbon plate type (Fig 4). The substitutional elements can also partition, although so far there is limited data on direct spectroscopic measurements of such partitioning [17,19,20,23,29]. More experimental work is needed to utilize modern high resolution facilities such as analytical microscopy (X-ray and electron energy loss), microdiffraction and atom probe spectroscopy. (Figure 26 shows an example of the use of X-ray microanalysis.) The difficulty of analyzing for interstitial elements (C, N, H, O) are formidable, however [30].

The main alloying elements for dual phase steels include Si, Al, Mn, and the carbide/nitride formers V, Nb, Mo. Considerations of these elements for alloy design have been described by Thomas and Koo [15], but it should be noted that both Si and Al increase the slope of the A_3 line (Fig 8) thus allowing for flexibility in heat treatment (temperature control to obtain a given vol-fraction of martensite is less critical than with e.g., Fe/Mn/C alloys). The effects of carbide/nitride forming elements appear to have limited value except as noted above with regard to retained austenite [20,31] precipitation in ferrite [20-22] and in the tempering response of dual phase steels.

B. Mechanical Properties

The absence of discontinuous yielding, a low yield stress and the high initial work hardening rate of dual phase steels arise from the fact that transformation of austenite on cooling from the intercritical temperature generates a high density of fresh mobile dislocations in the surrounding ferrite. The presence of transformation products then provide the necessary barriers to dislocation motion for strengthening as evidenced by the linear relationships observed between strength and % martensite, at least in the simple ferrite-martensite composites. Precipitation in ferrite will lead to higher yield strengths but lower ductility and retained austenite grains could contribute to increased ductility through the TRIP mechanism. It is clear that a range of properties is achievable in dual phase steels by appropriate choice of alloying elements, heat treatments and thermo-mechanical processing. Attractive combinations of strength ductility and formability can be easily achieved economically with simple compositions such as Fe-Si-Mn-C, and much promise for the future for such alloys can be envisioned.

VII. CONCLUSIONS

From the standpoint of their superior properties, and simplicity in composition and processing, dual-phase steels show great promise, not only for the transportation industry, but for general structural applications, e.g., pipelines, wires, rods and bars. Their advantages include economy of material and enormous flexibility wherein a single alloy can be manipulated to provide a wide range of products and properties. The main limitations of low carbon dual phase steels for general applications are material dimensions, i.e., sufficient hardenability must be achieved (~0.5in thickness/diameter) to obtain martensite/bainite on quenching from the $(\alpha+\gamma)$ field, and problems of microstructure control if welding must be done after the dual phase treatment.

ACKNOWLEDGEMENTS

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TABLE 1

PROPERTY SPECIFICATIONS FOR ARCTIC PIPELINE AND TYPICAL MECHANICAL PROPERTIES OF THE AS-HOT ROLLED Fe/1.5Mn/.06C STEEL WITH AND WITHOUT COLD WORKING

	UTS $MN.m^{-2}$	*Charpy Shelf Energy* Transverse	*Charpy Shelf Energy* Longitudinal
Arctice Pipeline	585.7	92 J. at -60° C	136 - 170 J. at -60° C
Fe/1.5Mn/.06C (as hot rolled)	642.2	149 J. up to -100° C	183 J. up to -100° C
Fe/1.5Mn/.06C (with 2% cold reduction)	649.0	129 J. up to -100° C	170 J. up to -150° C

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FIGURE 1. Principles of Dual Phase Steels: composite shows a typical heat treatment schedule, microstructure and tensile properties.

FIGURE 2. Stress-strain curves for different classes of low/medium carbon steels: notice the low yield stress, absence of discontinuous yielding, excellent ductility and high initial work hardening rates of dual phase steels

FIGURE 3. The microstructure of medium carbon steels ~0.3%C) has a major influence on fracture toughness at the same yield strength as illustrated. Dislocated lath martensites (upper micrograph)(%C~0.3% Ms≈250° C) are tougher than twinned plate martensites (lower micrograph) so control of composition, and Ms is essential (ref. courtesy of Metall. Soc. of AIME)

FIGURE 4. As can be seen from the Fe-C diagram (above), during intercritical annealing an alloy of composition C_A will phase separate into 1 low carbon ferrite at the intercritical annealing temperature B and higher carbon austenite. If the carbon content in austenite C_B exceeds about 0.4wt%, brittle twinned plates are formed as the Ms temperature is lowered. If the initial carbon content is much above 0.1wt%, it will be difficult to avoid twinned martensite in dual phase structures, if the annealing temperature B is chosen to obtain 20-30% martensite.

FIGURE 4b. Bright and dark field electron micrographs of lath martensite in 1010 steel - notice retained films between the dislocated laths. Composition corresponding to B in Fig 4a.

FIGURE 5. Transmission electron micrographs of ferrite-martensite in a dual phase Fe/2%Si/0.1%C (ref 15). In the lattice image (right side) the resolved (110) planes in ferrite and martensite are parallel showing excellent coherency parallel to the long axis of the martensite. Arrows indicate interface dislocations. From the spacing of the (110) fringes, the carbon content in martensite is estimated to be about 0.4wt%, assuming a linear relationship between lattice fringe spacing, lattice parameter and carbon content with % carbon in ferrite assumed to be essentially zero.

FIGURE 6. *Schematic diagram illustrating the various kinds of two phase morphologies that can be produced by varying the transformation path (vol. fraction approx. constant)*

A continuous cooling from γ
B step cooling $\gamma \rightarrow (\alpha + \gamma)$ *+ quench*
C starting from normalized ferrite-pearlite state
D starting from 100% *martensite (intermediate quench)*

FIGURE 7. *Actual examples of ferrite-martensite dual phase microstructure produced in Fe/*2*Si/*0.1*C steel by different processing routes shown at the top of the figure (Ref.* 40).

FIGURE 8. *Illustrating the (partial) influence of alloying elements on the Fe-C phase diagram. Notice that the slope of the* $\gamma/\alpha+\gamma$ *solvus and* $\alpha/\alpha+\gamma$ *solvus is important with regard to carbon partitioning, choice of tie line and hence composition and volume fractions of the two phases ferrite and martensite after quenching.*

FIGURE 9. *Fine scale precipitation in ferrite in dual phase Fe/*1.5*Si/*0.15*C/*0.03*Nb intermediate quenched. The decrease in strength of the steel with increasing % martensite is accounted for by this precipitation (ref* 17): *Courtesy of Scripta Met.*

FIGURE 10. *Variation in yield and ultimate tensile strength with % martensite in the dual phase Fe/*1.5*Si/*0.15*C/*0.03*Nb steel. Compare to Figs* 1, 2 *(ref.* 17): *Courtesy of Scripta Met.*

FIGURE 11. *Retained austenite in air cooled specimen characterized by martensite (bainite)(M)/austenite(A) transformation front. (a) Bright field and (b) the correspondig austenite dark field micrographs revealing the large volume fraction of austenite present. Both micrographs show more than* 1 *variant of the accomodation stacking faults (ref.* 20): *Courtesy of Met. Trans. AIME.*

FIGURE 12. *Schematic transformation diagram and microstructures possible in microalloyed steels (example is for Nb with carbon nitrogen) showing how interphase alloy carbide structures can form (ref* 22): *Courtesy Met. Trans. AIME*

FIGURE 13. Summary of strength-ductility properties of a range of experimental dual phase steels containing up to ~ 30% martensite from research at U.C. Berkeley.

FIGURE 14. Summary of strength-ductility-volume fraction - intercritical annealing temperatures for the intermediate quench treatment of Fe/2Si/0.1C steel.

FIGURE 15. Composite illustrating the effects of transformation path and morphology on fracture resistance of Fe/2Si/0.1C dual phase steel (Courtesy: N.J. Kim).

FIGURE 16. Schematic representation of structural changes which occur on rolling steel in different temperature regimes. Crystallization of austenite leads to fine packets of martensite upon quenching. (Courtesy: C. Shiga et.al., Kawasaki Steel Tech Report 4, 97 (1981)

FIGURE 17. Ductile-brittle impact transition curves of as hot-rolled and quenched dual phase Fe/1.5Mn/0.06 steel (Courtesy: N.J. Kim, Ph.D. Thesis, University of California, LBL 12661 (1981)

FIGURE 18. Mechanical properties of the same steel as Fig 17 as a function of grain size following controlled rolling in the ($\alpha+\gamma$) range and quenching.

FIGURE 19. Fatigue crack propagation results for Fe/2Si/0.1C steel treated in the three ways shown in Fig 7 (Courtesy: V. Dutta et.al), Ref. 40

FIGURE 20. Summary of fatigue data on a range of steels (Ref 40). Notice the exceptional properties of dual phase steels tested as shown in Fig 19. Previous results marked X from Refs 39, 42.

FIGURE 21. Light optical micrographs of dual phase Fe/1.5Mn/0.1C before (left) and after (right) cold drawing 99% is a strain of 4.7. Notice the alignment of martensite particles after drawing.(Courtesy: A. Nakagawa)

FIGURE 22. Scanning electron micrograph corresponding to Fig. 21 (right side) showing some tendancy for non-uniformity in plastic flow nearer non-fibrous shaped martensite (see Fig 24) (Courtesy: A. Kakagawa)

FIGURE 23. Results of continuous cold drawing tests on dual phase wire with 30% martensite. Data for 0.7%C patent wire shown for comparison. Starting material is 0.25in. dual phase rod (Courtesy: A. Nakagawa, R.M. Fisher)

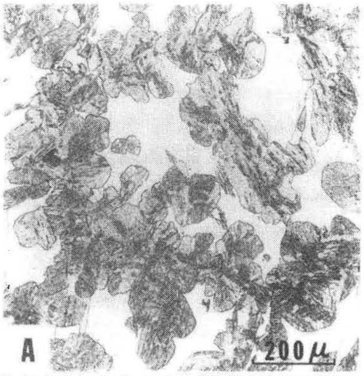
FIGURE 24. Scanning electron micrograph showing cracks developed at a large martensite particle. Similar effects occur at inclusions (Courtesy: A. Nakagawa)

FIGURE 25. Transmission electron micrograph of dual phase wire after a total drawing strain of 3.7 (97.4% reduction). Notice the dense dislocation networks and alignment into the drawing direction (Courtesy: A. Nakagawa).

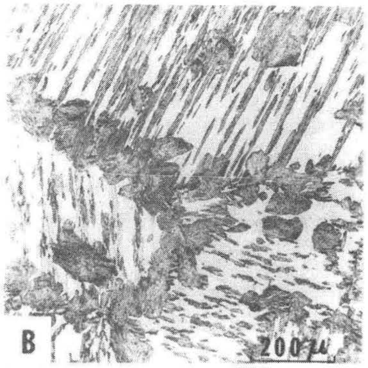
FIGURE 26. X-ray microanalytical results from small vanadium containing particles extracted by the carbon replication technique and examined in an analytical electron microscope. Such fine particles are difficult to extract but this method avoids difficulties such as flourescence as in thin foil studies. Thin foil microstructure is similar to that shown in Fig 9.

DUPLEX FERRITE/MARTENSITE STRUCTURES DEVELOPED IN Fe/0.1C/X ALLOYS

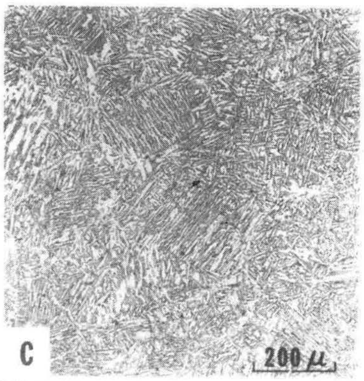
OPTICAL MICROGRAPH



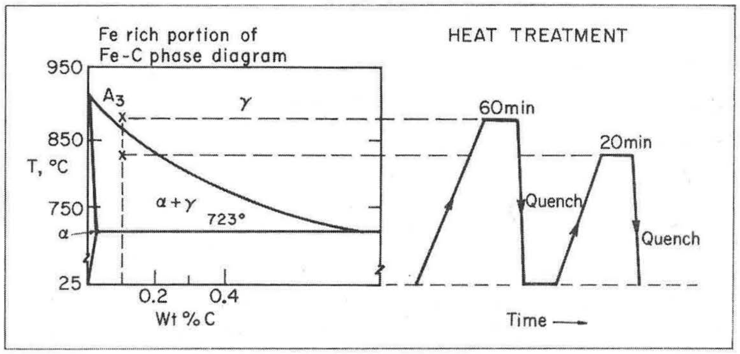
Duplex ferrite/martensite structures developed in Fe/0.1C4Cr steel. Globular shape of martensite contrasts with light background ferrite.



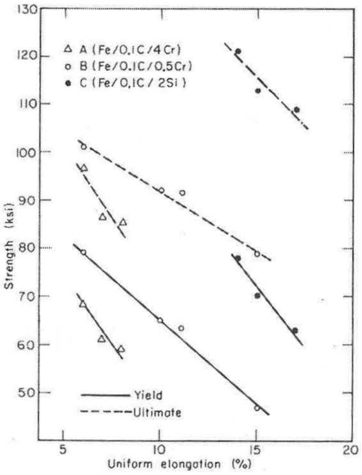
Duplex structures developed in Fe/0.1C/0.5Cr steel. Predominantly needle-like martensite is shown within each prior austenite grain. Prior austenite grain boundaries are decorated with continuous martensite phase.



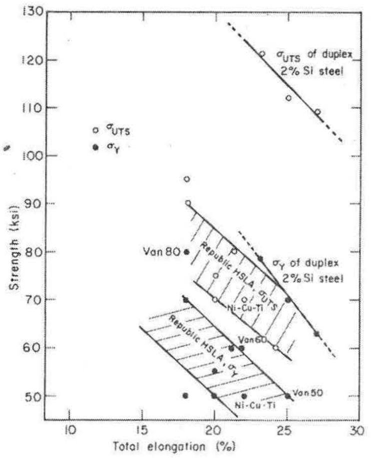
Duplex structures developed in Fe/0.1C/2Si steel show a fine, aligned, and acicular shape of martensite within each prior austenite grain.



MECHANICAL PROPERTIES



Depending on the annealing temperature in the $(\alpha + \gamma)$ range, various volume fraction of martensite and a wide range of strength and elongation combinations are obtained.



Tensile properties of the duplex 2% Si steel are compared with those of commercial HSLA steels.

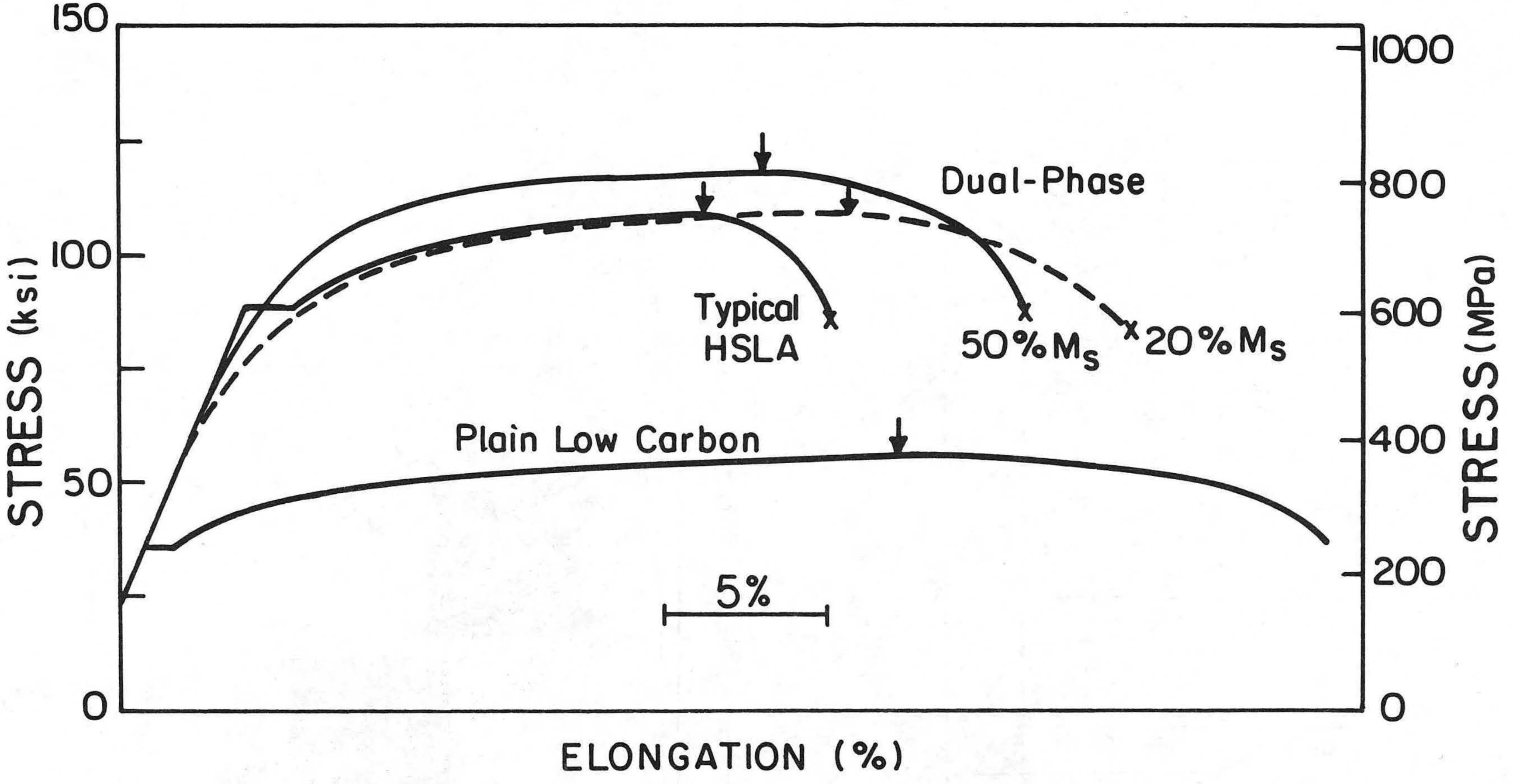
The incorporation of the strong phase martensite as a load carrying constituent in a ductile ferrite matrix can strengthen the composite alloy. The production of such duplex ferrite/martensite structures in Fe/0.1C/X alloys (X being a small amount of a substitutional alloying element) is carried out by an initial austenitizing treatment followed by annealing in the $(\alpha + \gamma)$ with intermediate quenching. Depending on X and the exact heat treatment, various morphologies and shapes of martensite in the ferrite matrix are obtained (Figs. A, B and C) and these seriously affect the corresponding mechanical properties.

A globular shape of the second phase martensite, combined with continuous martensite along the prior austenite grain boundaries (Fig. A), results in relatively poor tensile properties. With microstructures containing a combination of needle-like martensite within each prior austenite grain (Fig. B), while still showing continuous globular martensite along the prior austenite grain boundaries, somewhat improved tensile properties are exhibited. However, upon obtaining a complete needle-like martensite (Fig. C) a superior strength and elongation-ductility combination is found (these properties are compared with these of commercial HSLA steels).

Other factors, such as carbide morphology, solid solution strengthening, etc. were not found to be as significant in determining mechanical properties as the shape and distribution of the martensite phase. Photos were made with polaroid film at 130 ×; 2% nital etch.

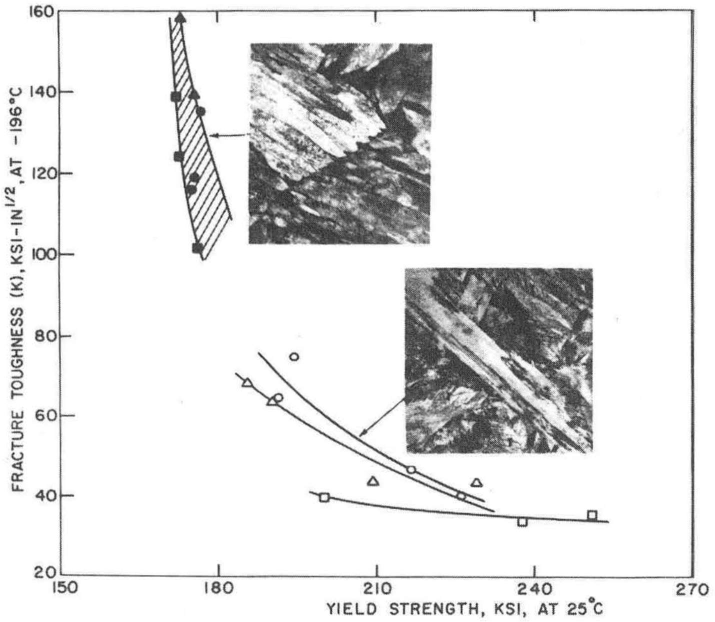
XBB 766-5888

Fig. 1



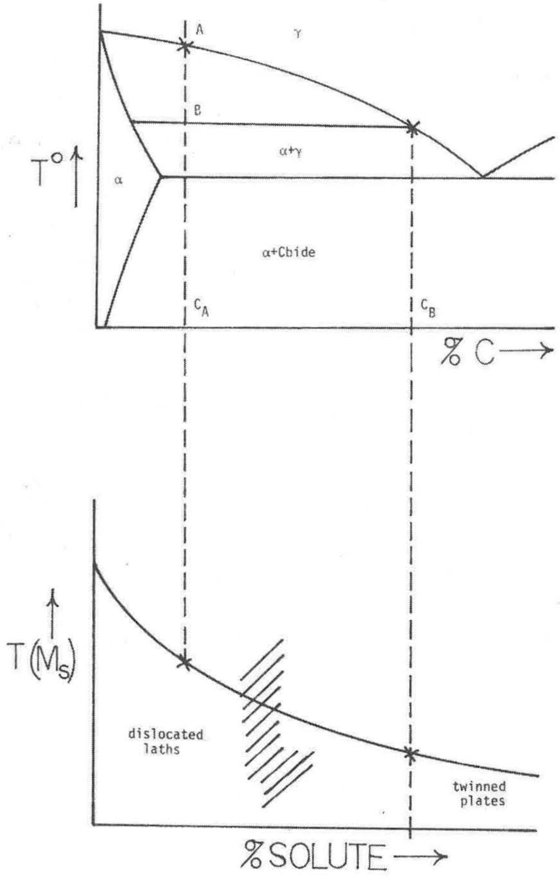
XBL 838-6214

Fig. 2



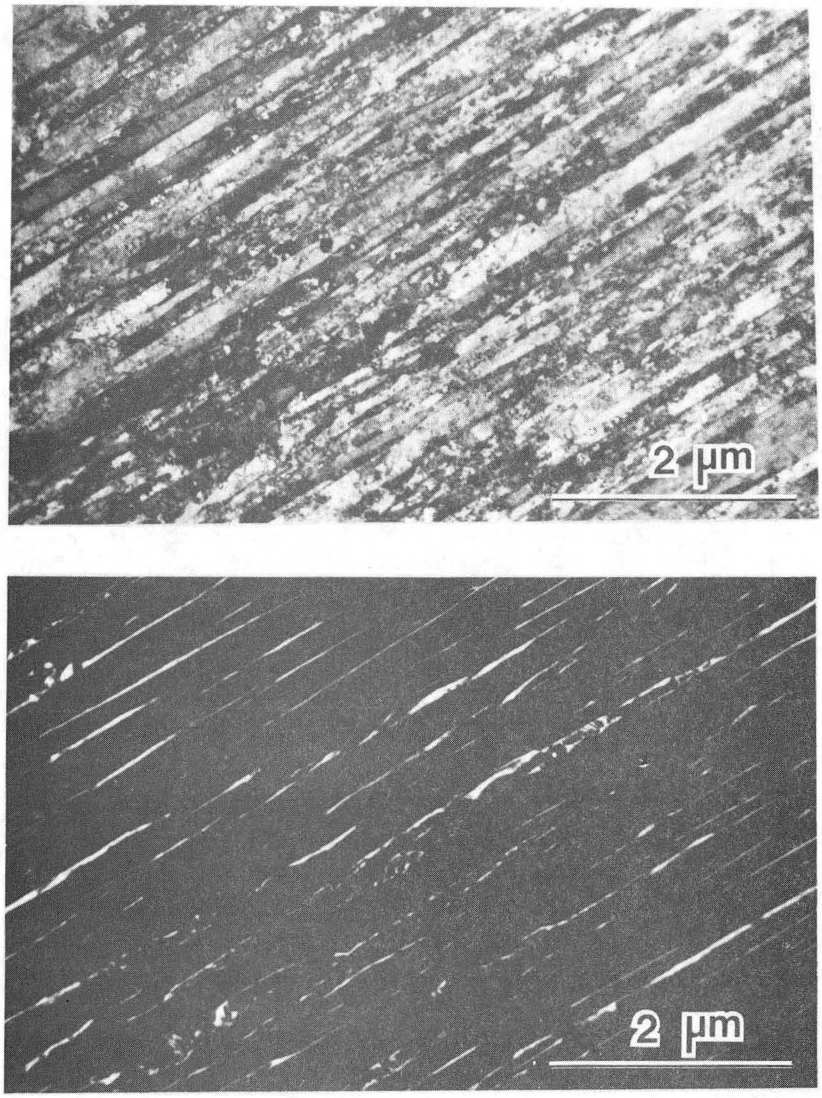
XBB 701-475

Fig. 3



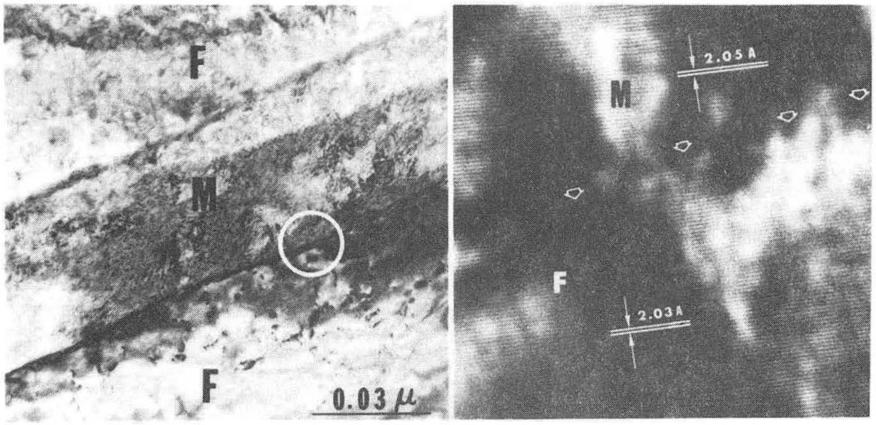
XBL 838-11227

Fig. 4a



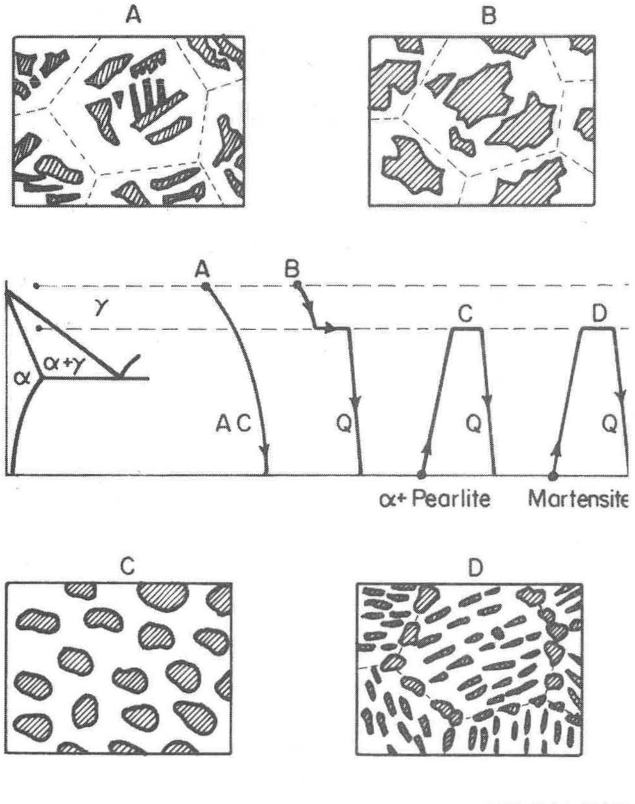
XBB 800-14347

Fig. 4b



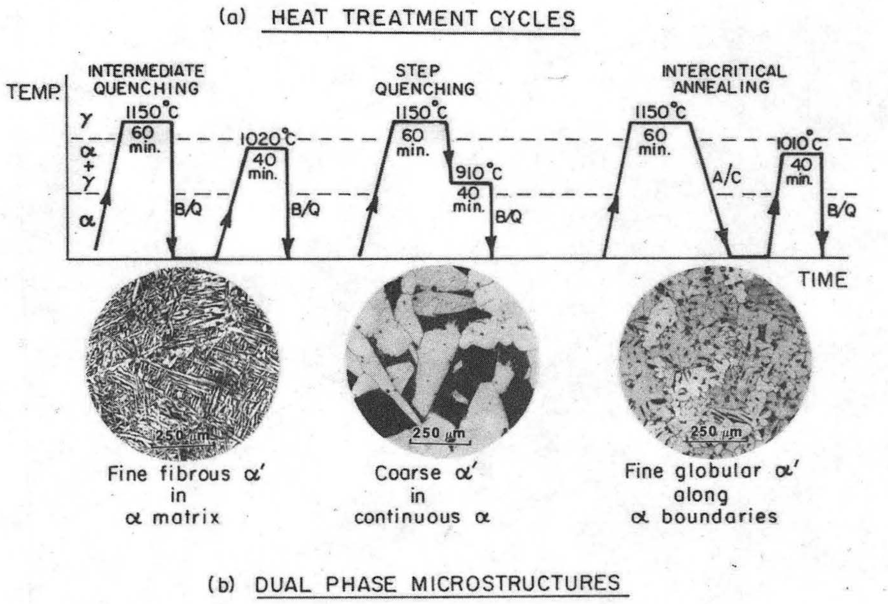
XBB 773-2297

Fig. 5



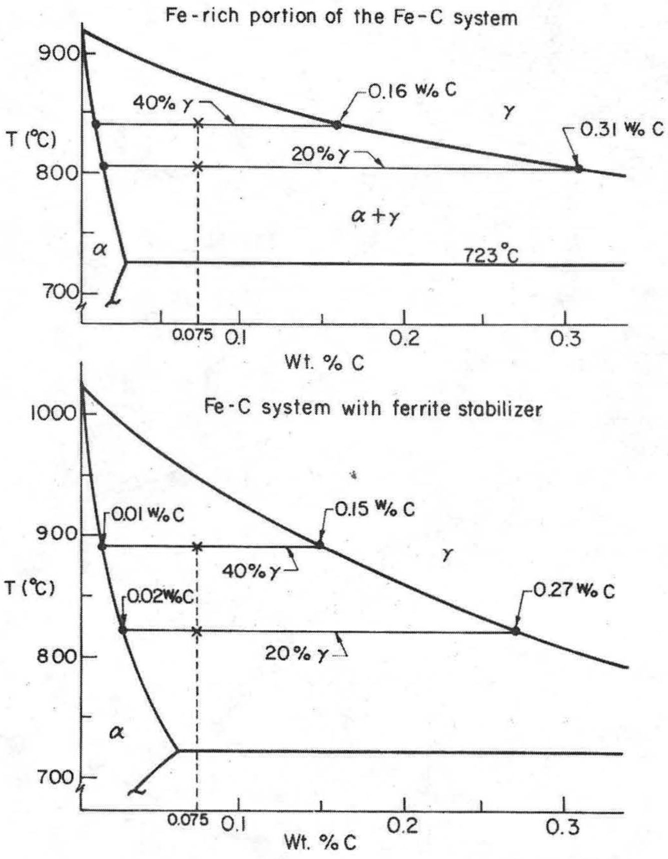
XBL 792-568

Fig. 6



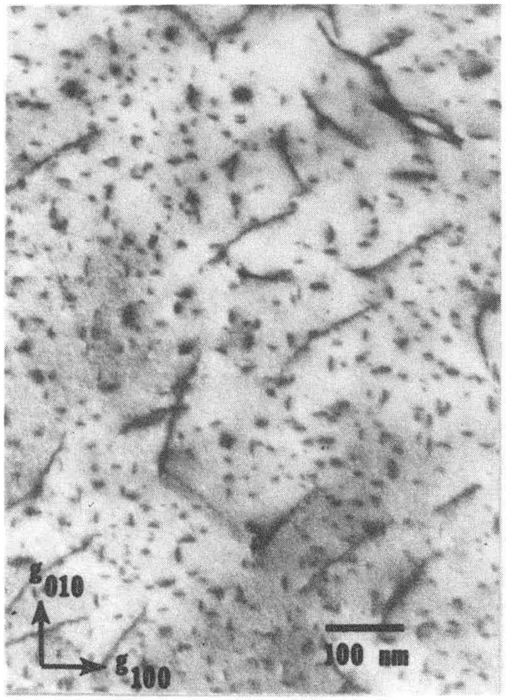
XBB 833-2300

Fig. 7



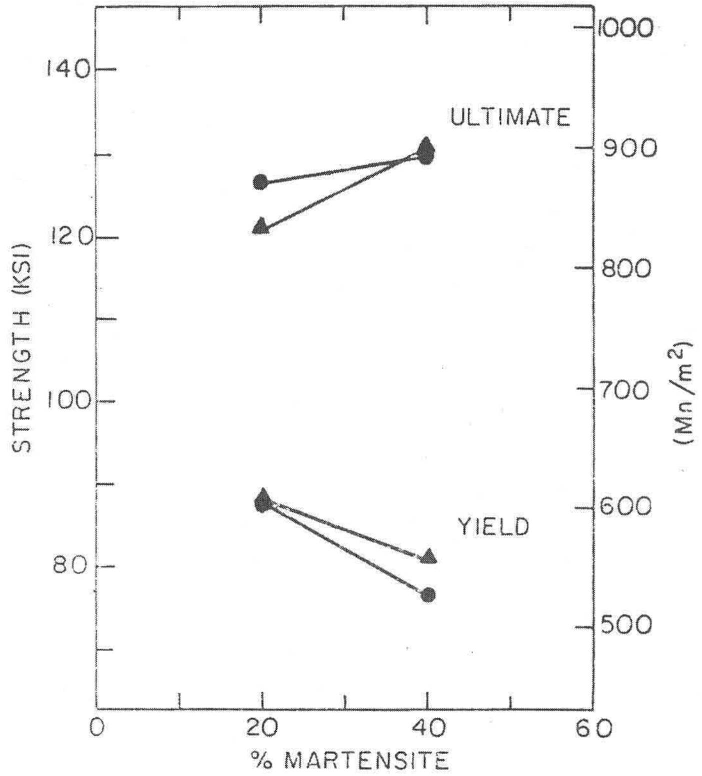
XBL792-5746

Fig. 8



XBB 806-7299

Fig. 9

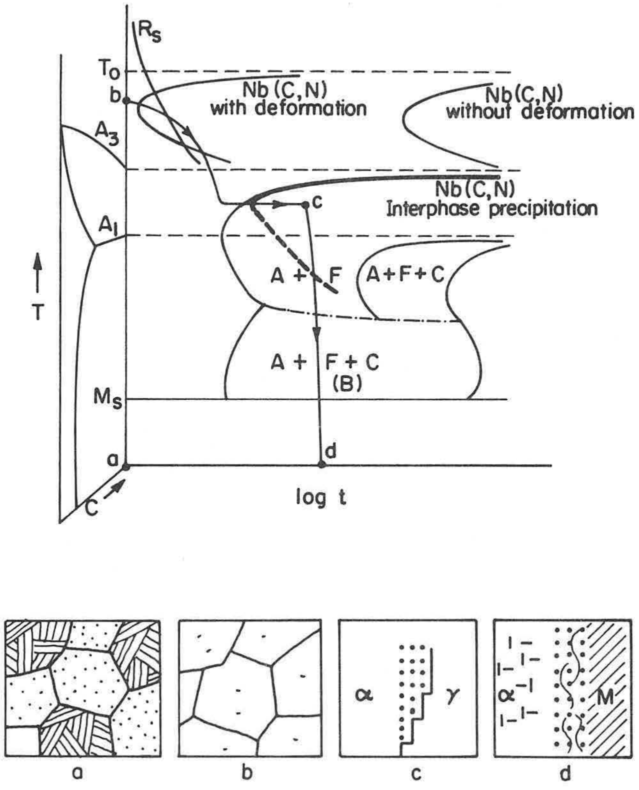


XBL 806-10321

Fig. 10

XBB 831-9444

Fig. 11



T_o = Precipitate Solubility Temperature

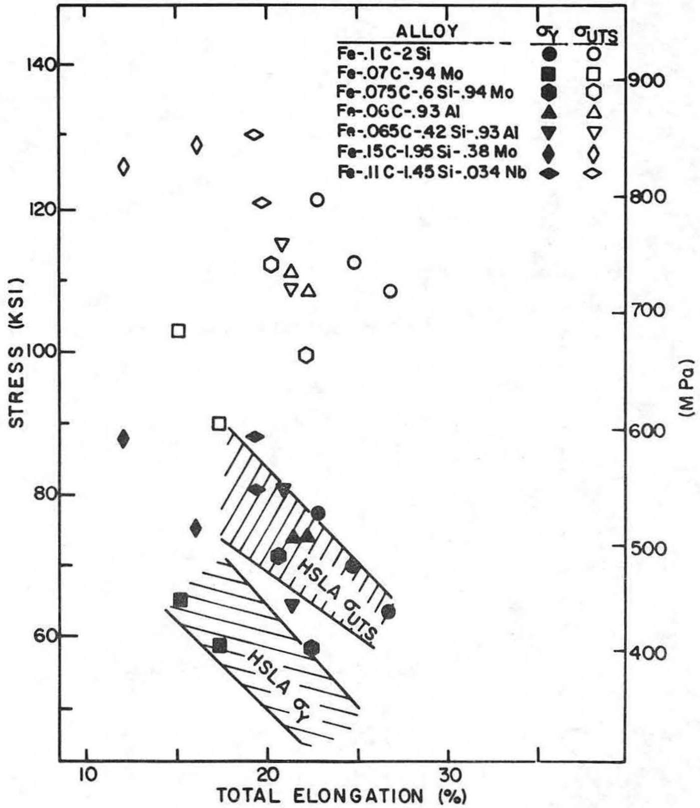
M_s = Martensite Start Temperature

A = Austenite, F = Ferrite, C = Carbide

R_s = Recrystallization Start Curve

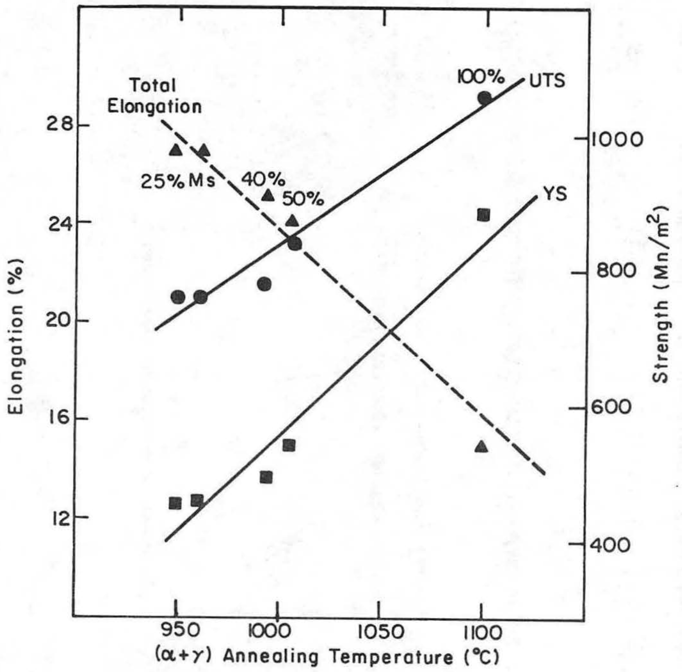
XBL 819-6637

Fig. 12



XBL 7812-6281

Fig. 13



XBL 8210-6617

Fig. 14

FRACTURE BEHAVIOR OF DUAL PHASE Fe/Si/C STEELS

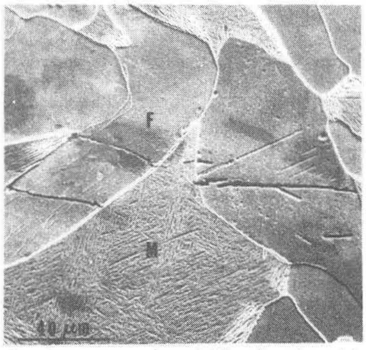
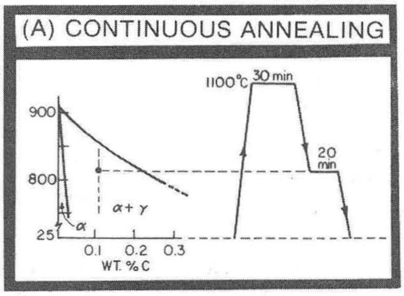


Fig. 1. Scanning electron micrograph of the continuously annealed Fe/2 Si/0.1C dual phase structure showing the mophology of ferrite and martensite and the cracks developed in the ferrite. 5% nital etch. M - martensite region, F - ferrite region.


(A) CONTINUOUS ANNEALING

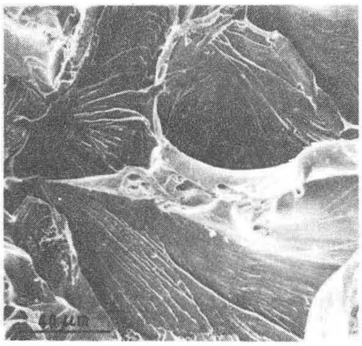


Fig. 2. Cleavage fracture observed in the broken tensile specimen of the continuously annealed Fe/2Si/0.1C dual phase steels.

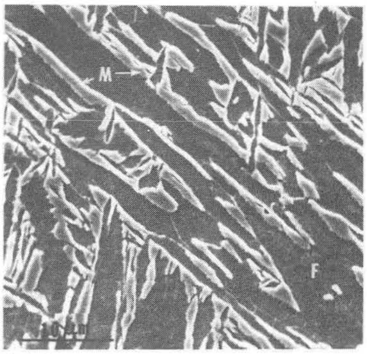
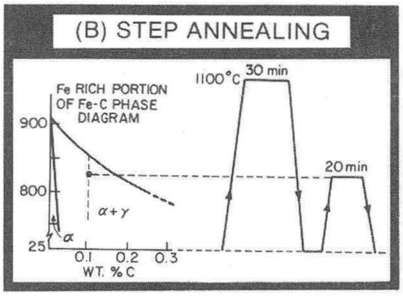


Fig. 3. Scanning electron micrograph of the step annealed Fe/2Si/0.1C dual phase structure showing a fine dispersion of martensite in the ferrite. 5% nital etch.


(B) STEP ANNEALING

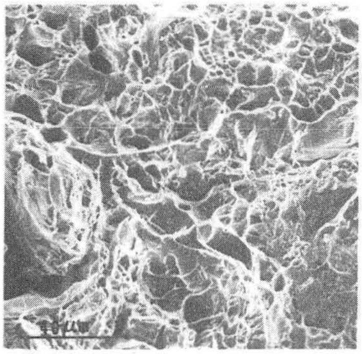


Fig. 4. Ductile fracture (dimpled rupture) observed in the broken tensile specimen of the step annealed Fe/2Si/0.1C dual phase steels.

Dual phase ferrite-martensite steel is a "composite" which has emerged as a new generation of HSLA steels. The fracture characteristics of these dual phase steels are very sensitive to the transformation path by which the composite structure is obtained. A Fe/2% Si/0.1% C steel when subjected to the continuous annealing" treatment (Fig. A), which is a commercially attractive process, resulted in a very coarse dual phase structure as shown in Fig. 1. A careful scanning electron microscopy study indicated that in this case a premature failure occurs by the initiation of cracks in the ferrite region where maximum localized stress concentration takes place. This is illustrated in the scanning electron micrograph (Fig. 1) of the cross sectional surface of a tensile test specimen just before final fracture. The fracture surface of a broken tensile specimen revealed predominantly cleavage fracture (Fig. 2). As expected this dual phase structure resulted in very poor ductility (Fig. 5) which places a limiting factor on the potential structural applications.

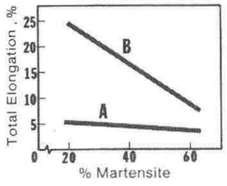


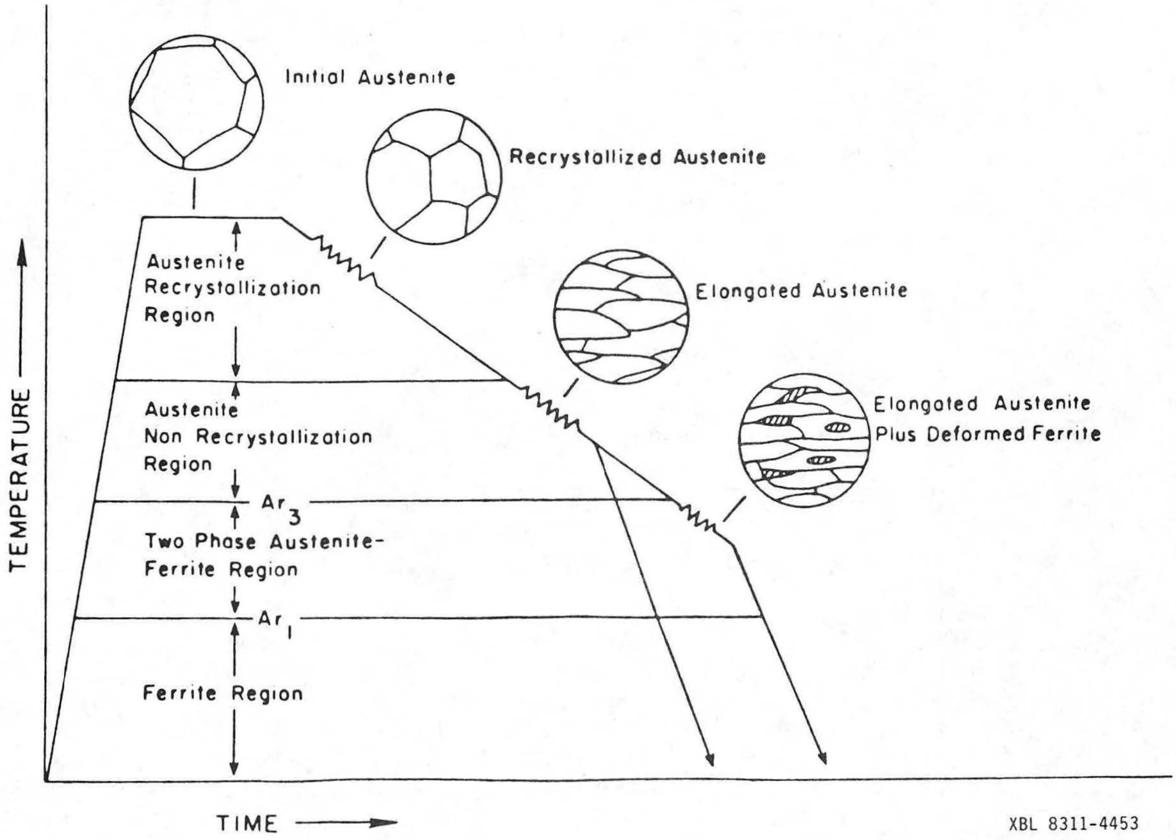
Fig. 5 Plot of elongation ductility vs volume fraction of martensite for heat treatments A and B

A modified heat treatment, "step annealing", was designed to improve the ductility. Following the step annealing process (Fig. B) a remarkable refinement in dual phase structure was obtained and no evidence of crack formation was found in the tensile specimen just prior to the final fracture. This is shown in Fig. 3. The fracture appearance of the broken tensile specimen exhibited a primarily dimpled rupture (Fig. 4). Thus the step annealing process allows a significant improvement in ductility with no sacrifice in strength, as illustrated in Fig. 5.

The drastic change in the fracture mode from brittle (heat treatment A) to ductile (heat treatment B) is attributed to the refinement of dual phase structure by the step annealing process which appears to be highly promising for the maximum utilization of the composite steels.

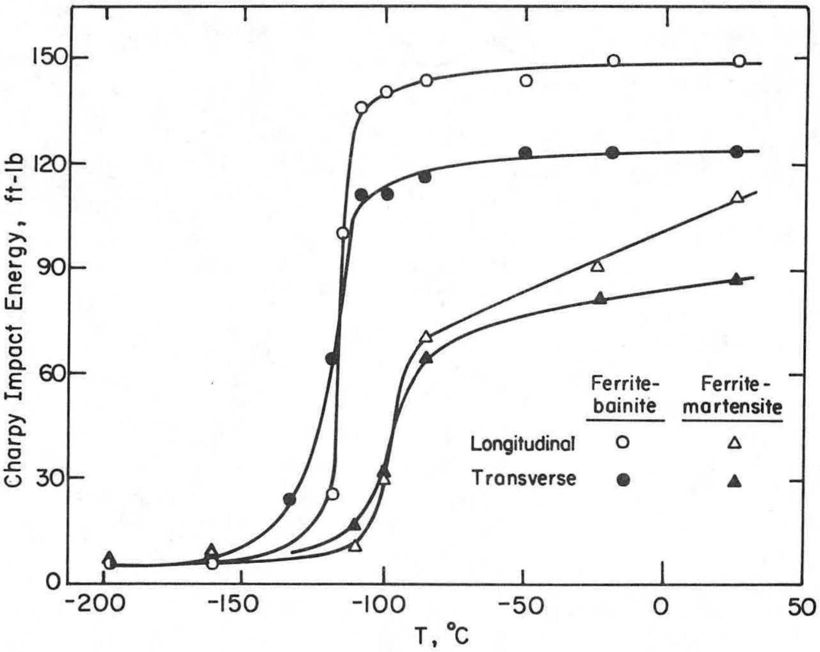
XBB 797-9313

Fig. 15



XBL 8311-4453

Fig. 16



XBL 8012-13468

Fig. 17

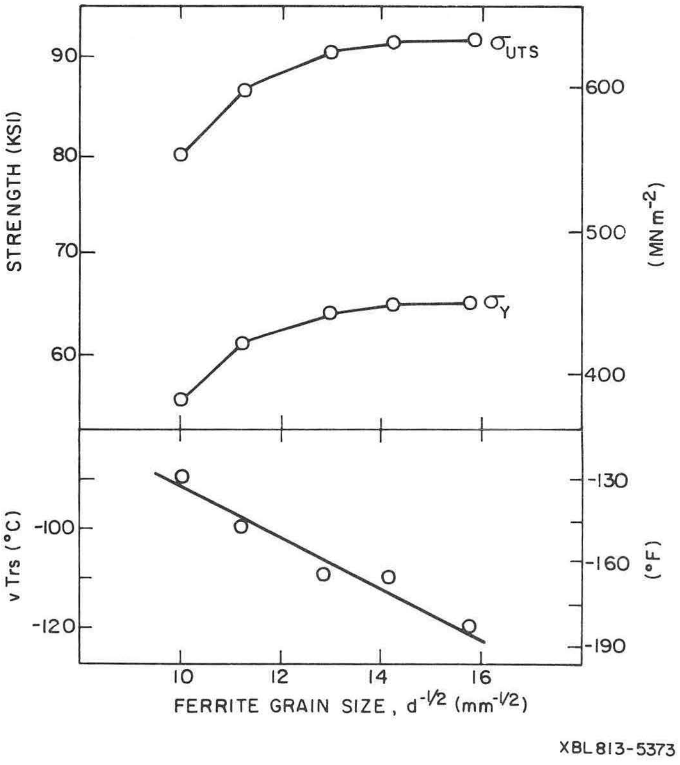
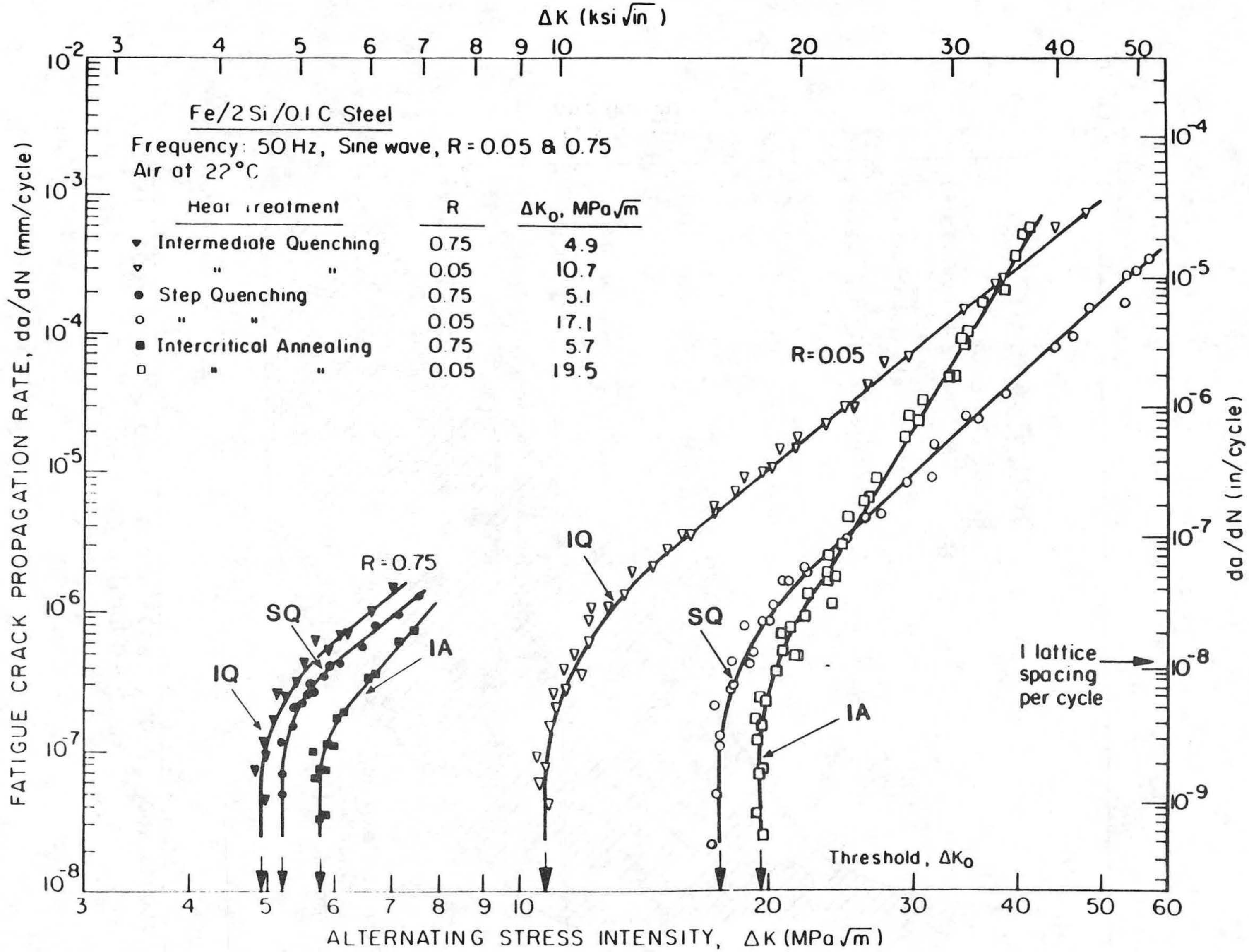
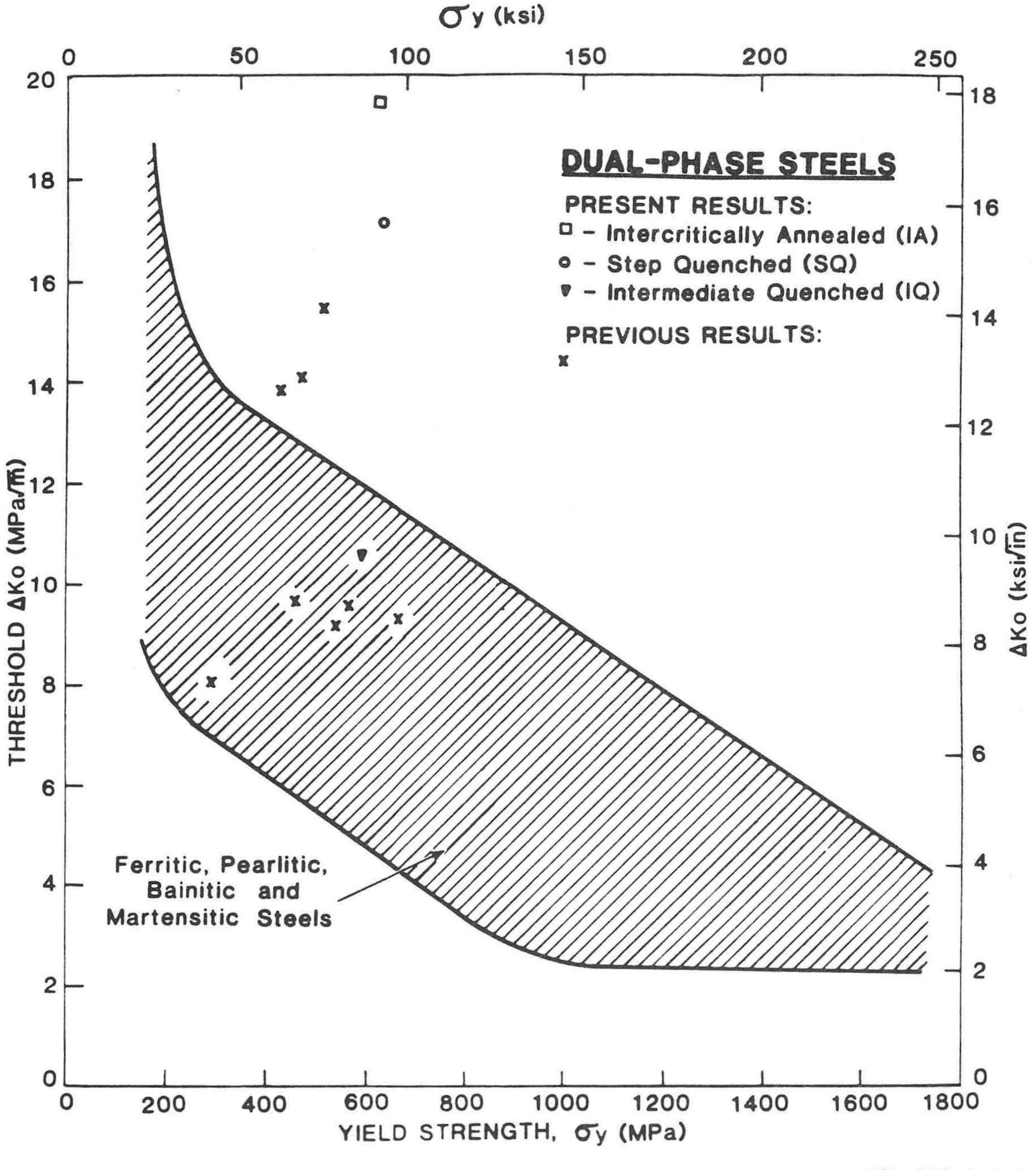

STRENGTH (KSI)
90
80
70
60
σ_{UTS}
σ_Y
600
500
400
(MN m^{-2})
vTrs (°C)
-100
-120
-130
-160
-190
(°F)
10
12
14
16
FERRITE GRAIN SIZE, d$^{-1/2}$ (mm$^{-1/2}$)
XBL 813-5373

Fig. 18



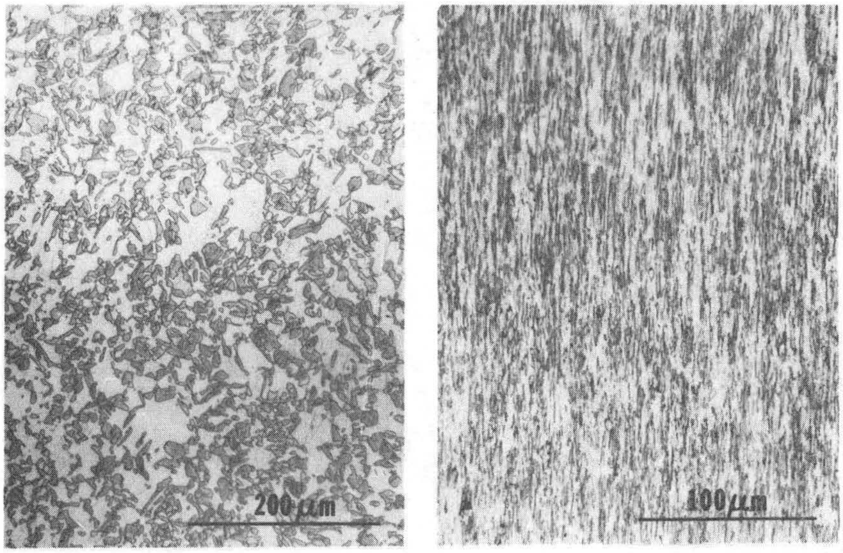
XBL 838-10897

Fig. 19



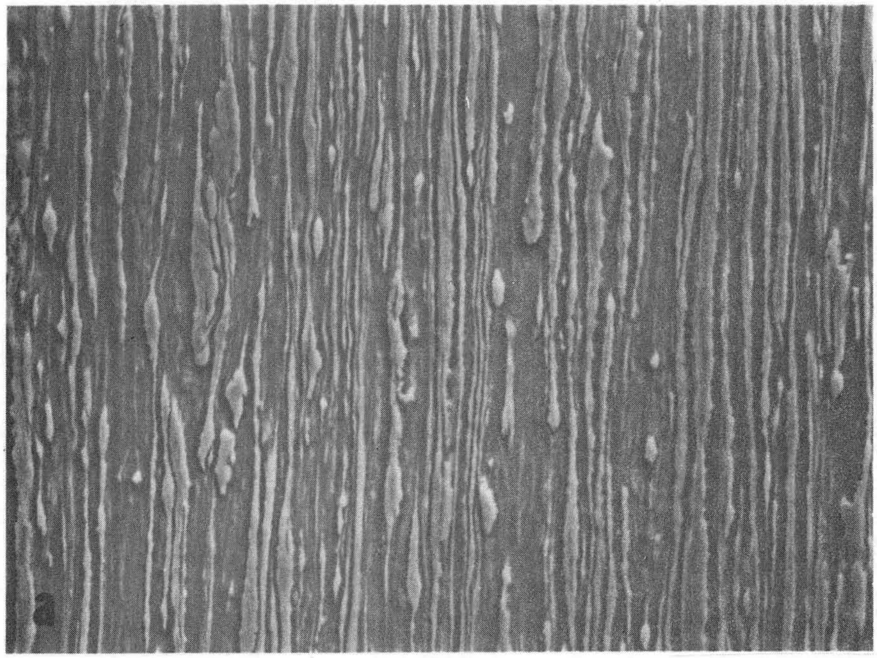
XBL 838-10895

Fig. 20



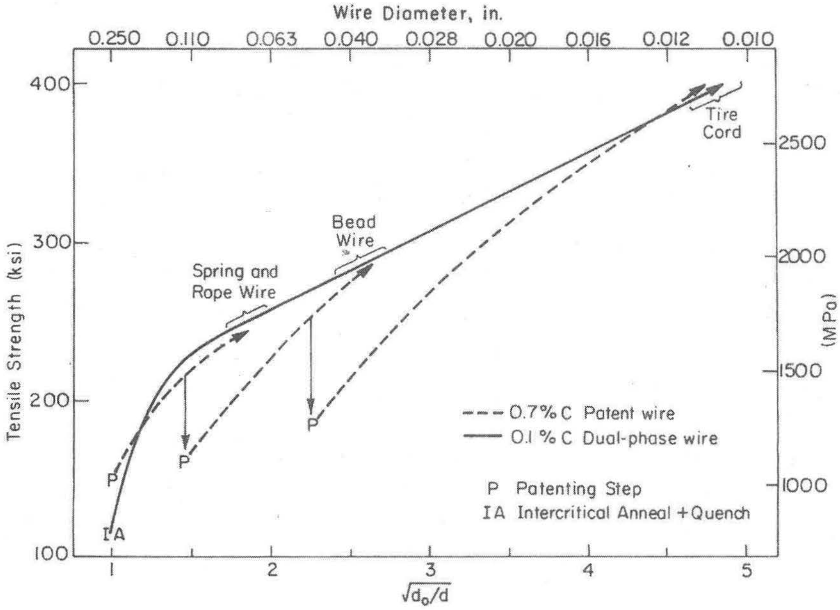
XBB 810-10279

Fig. 21



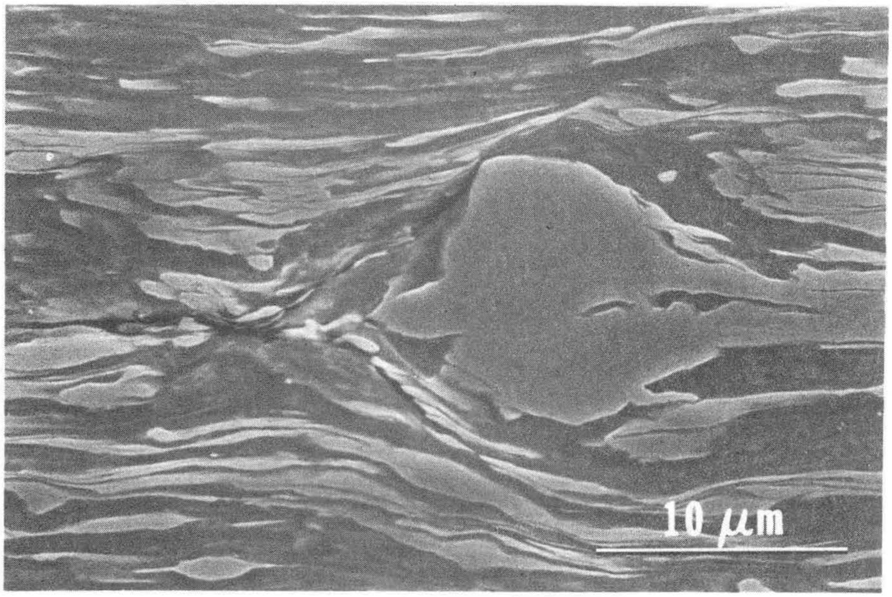
XBB 820-9787

Fig. 22



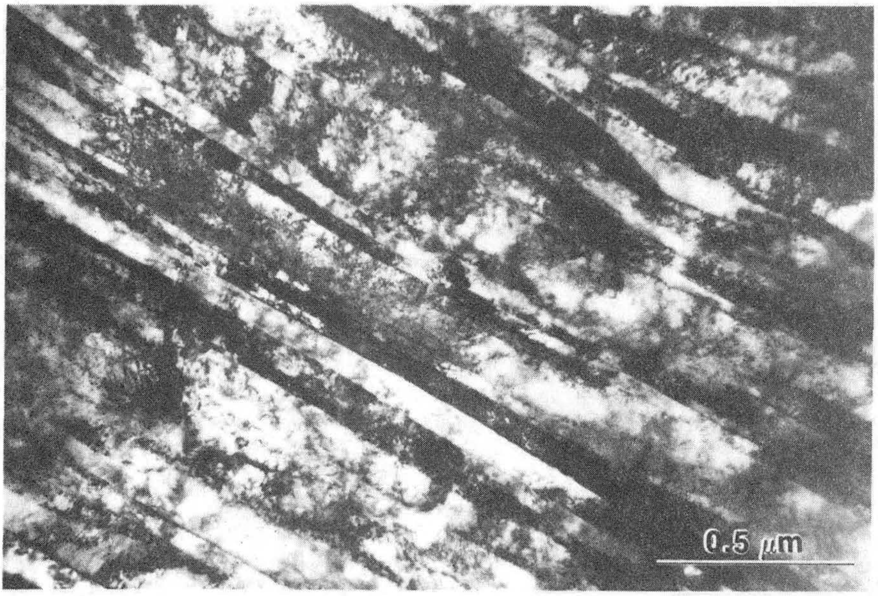
XBL 8210-6760

Fig. 23



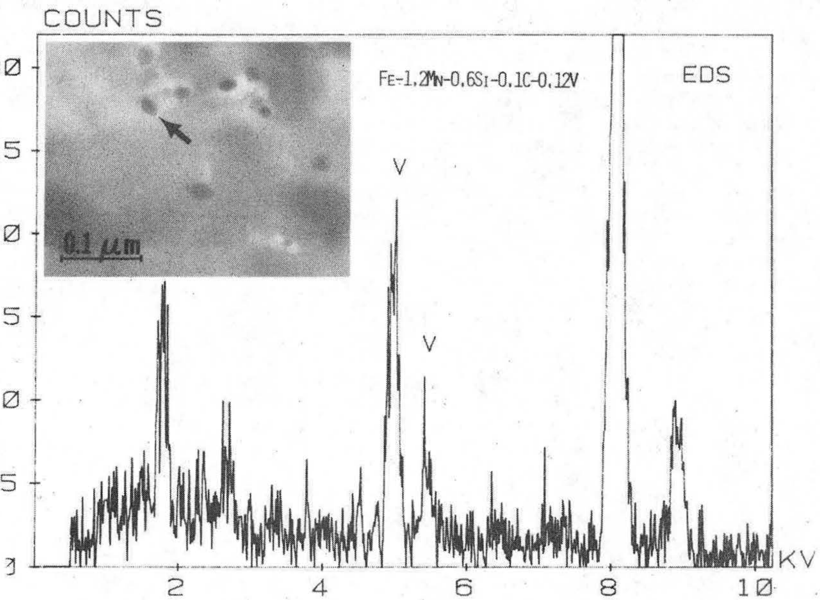
XBB 832-1670

Fig. 24



XBB 832-2272

Fig. 25



XBB 823-2182

Fig. 26

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