

Fig. 3—Cavitation in specimen B near the fracture tip, after testing at an initial strain rate of $6.6 \times 10^{-5} \text{ s}^{-1}$ to fracture at ~ 650 pct. The stress axis is horizontal. Magnification 65 times.

to observations on various Cu alloys,^{7,8,10,13} where cavitation is more extensive at the high strain rates, but it is similar to the conclusions reached through density measurements on two microduplex steels.^{15,17} The difference in behavior probably arises because of the considerable grain growth occurring in some of the Cu alloys, since the presence of substantial grain boundary migration would tend to relieve the localized stress concentrations which are a prerequisite for cavity formation. For example, specimens of an α - β brass (Cu-40 pct Zn) contained very few cavities but exhibited grain growth of up to an order of magnitude after testing at low strain rates,⁸ whereas in the present work the increase in grain size, even at the lowest testing strain rate, was always less than a factor of 2.

Although no detailed measurements have been taken to determine the average aspect ratio of the cavities at failure, A_v (defined as average length parallel to stress axis/average length perpendicular to stress axis), it is clear from Fig. 3 that $A_v > 1$. This suggests that fracture occurred at low strain rates by a combination of external necking and the internal linkage of voids by an intergranular void sheet (IVS) process.^{11,21} The present results therefore provide the first evidence both of cavitation in Zn-22 pct Al and of a contribution from the IVS fracture mechanism in a superplastic alloy deformed at low strain rates in Region I.

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Design of Duplex Fe/X/0.1C Steels For Improved Mechanical Properties

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The incorporation of the strong phase martensite as a load carrying constituent in a ductile ferrite matrix can strengthen composite steel alloys.¹⁻⁴ Such duplex ferritic-martensitic (DFM) structures offer good combinations of strength and ductility as a useful method of strengthening plain, low-carbon steels.⁴ In view of the attractive mechanical properties and the simplicity in composition and heat treatment already achieved,²⁻⁴ the idea of duplex simple low carbon steels is now receiving much attention. This approach contrasts with high strength, low-alloy (HSLA) steels microalloyed with columbium (niobium), vanadium or titanium in which the principal strengthening is derived from precipitation of finely dispersed carbides and whose manufacture is associated with some economical problems discussed elsewhere (*e.g.* Ref. 5).

Earlier experimental work on DFM steels^{1,3,4,6,7} has primarily been concerned with the influence of the volume fraction or strength of the second phase (martensite) on mechanical properties. From the viewpoint that the mechanical properties of two phase materials in general are determined by the shape, distribution and properties of the individual phases, the present paper reports on the design considerations for desirable DFM structures which in turn result in desirable mechanical properties. This work is based on an earlier report⁴ in which improvements in 1010 steels by duplex heat treatments alone were achieved.

Experimental. The chemical compositions of the ternary alloys used in this investigation are given in Table I. X represents the substitutional solute Cr or

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Si. The design basis was to increase the extent of the $(\alpha + \gamma)$ field shown in Fig. 1, increase hardenability, and vary the DFM morphology by comparison to 1010 steels.⁴ The alloys were vacuum melted into 33 lb heats, hot forged to 0.6 in. diam rods, homogenized at 1200°C in vacuum for 24 h and subsequently furnace-cooled.

The heat treatment to produce controlled DFM structures consists of austenitizing and quenching to 100 pct martensite, followed by annealing in the $(\alpha + \gamma)$ range as shown in Fig. 1. By holding in the two phase range, martensite transforms partially to austenite and the residual martensite becomes ferrite as the two phases attain the composition specified by the tie line corresponding to the holding temperature. The alloy will then consist of low carbon ferrite (α) and higher carbon austenite (γ). Upon quenching, the austenite transforms to martensite (strong phase), and the ferrite becomes heavily dislocated due to plastic deformation as a result of the austenite \rightarrow martensite transformation strain. This structure has been confirmed by transmission electron microscopy (see Fig. 3).

Design of heat treatment is done so as to control the martensite properties through its substructure by control of prior γ composition (e.g. Ref. 6). Thus temperatures in the $(\alpha + \gamma)$ field are chosen to maintain carbon levels of about 0.3 pct in austenite, which allows dislocated martensite to form. The α, γ volume fractions then depend upon the total alloy composition for this temperature. The morphology of the duplex microstructural constituents in the alloy is strongly influenced by the substitutional solute (X) (Ref. 8) and is a very important factor, as will be shown later.

Table I. Chemical Compositions of Fe-C-X Alloys (Wt Pct)

Alloy	C	Cr	Si	Fe
A	0.073	4	—	Balance
B	0.06	0.5	—	Balance
C	0.065	—	2	Balance
D	0.075	—	0.5	Balance

Tensile properties were measured using both 1.25 in. and 2 in. gage round specimens, the latter for comparison with the tensile properties of commercial steels. The testing procedures and the experimental procedures for optical and transmission electron microscopic examinations have been described previously.⁴

Results. Marked differences develop in the morphology of the DFM structures, depending on the amount and type of alloying element X in Fe/X/0.1C alloys (hereafter, "0.1C" represents for convenience the carbon content in each alloy). These are illustrated in the optical micrographs, Figs. 2(a) to (c). The acicular shape of martensite developed in the DFM Fe/0.5Si/0.1C steel is virtually identical to that exhibited in the Fe/2Si/0.1C alloy.

The ferrite regions in the DFM 0.5Si and 0.5Cr steels revealed coarse carbides in the immediate vicinity of the α /martensite boundaries. These carbides formed during final quenching from the $(\alpha + \gamma)$ range results from the $\gamma \rightarrow \alpha +$ carbide reaction products due to the lack of sufficient hardenability in the above two steels.

The corresponding tensile properties reflecting the changes in the geometries of martensite phase are plotted in Fig. 4. A globular shape of the second phase martensite, combined with continuous martensite along the prior austenite grain boundaries (Fig. 2(a)), results in relatively poor tensile properties.

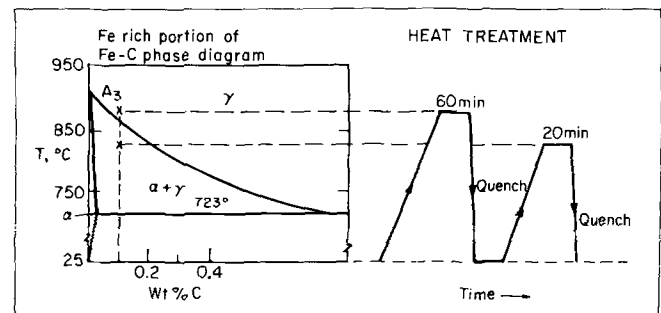


Fig. 1—Schematic representation of heat treatment to produce controlled DFM structures.

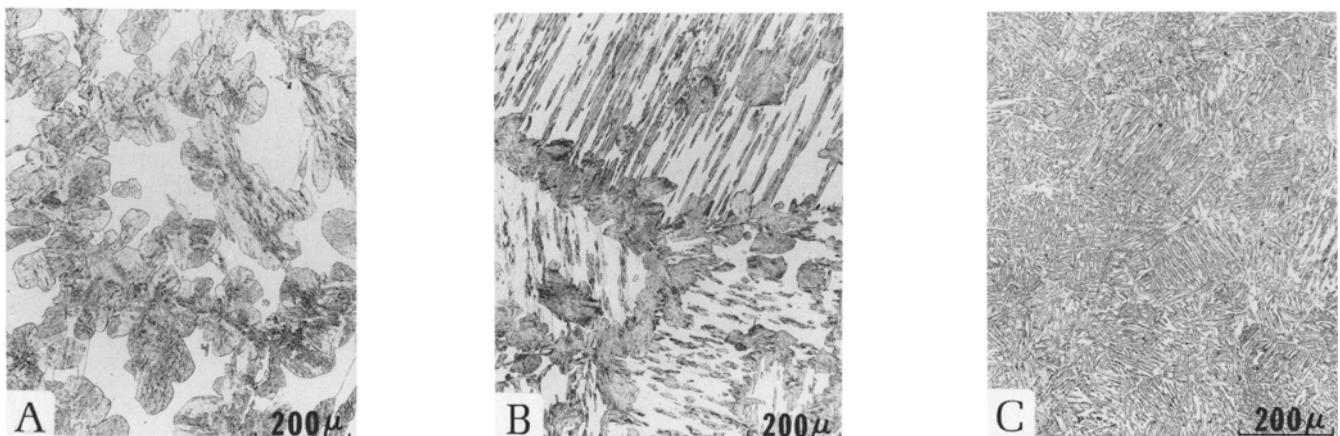


Fig. 2—Optical Micrographs. (a) DFM structures developed in Fe/4Cr/0.1C steel. Globular shape of martensite contrasts with light background ferrite. (b) DFM structures developed in Fe/0.5Cr/0.1C steel. Predominantly needle-like martensite is shown within each prior austenite grain. Prior austenite grain boundaries are decorated with continuous martensite phase. (c) Duplex structures developed in Fe/2Si/0.1C steel show a fine, acicular shape of martensite within each prior austenite grain. A magnified view of the martensite morphology is shown in Fig. 3.

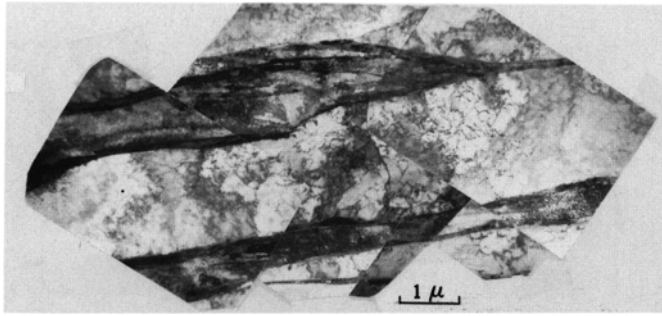


Fig. 3—Composite transmission electron micrograph of the DFM structure developed in 2 pct Si steel. Two parallel needles are martensite phase surrounded by ferrite with high density of dislocations.

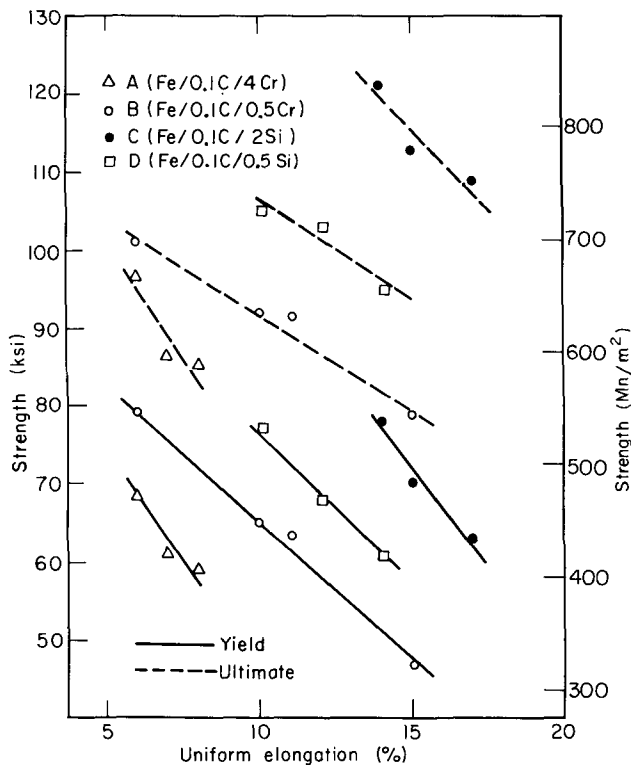


Fig. 4—Strength vs uniform elongation is plotted for the duplex Fe/X/0.1C alloys. Depending on the annealing temperature in the ($\alpha + \gamma$) range, various volume fractions of martensite and a wide range of strength and elongation combinations are obtained.

With microstructures containing a combination of needle-like martensite within each prior austenite grain (Fig. 2(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. 2(c)) superior strength and elongation ductility combinations are found. These properties are compared, in Fig. 5, with those of some commercial HSLA steels. One notes that the duplex 2 pct Si steel shows notably high uniform elongation and superior tensile strength which presumably are the consequences of its morphological features.

Discussion. Many factors affect the mechanical properties of a simple ternary Fe/X/0.1C alloy with

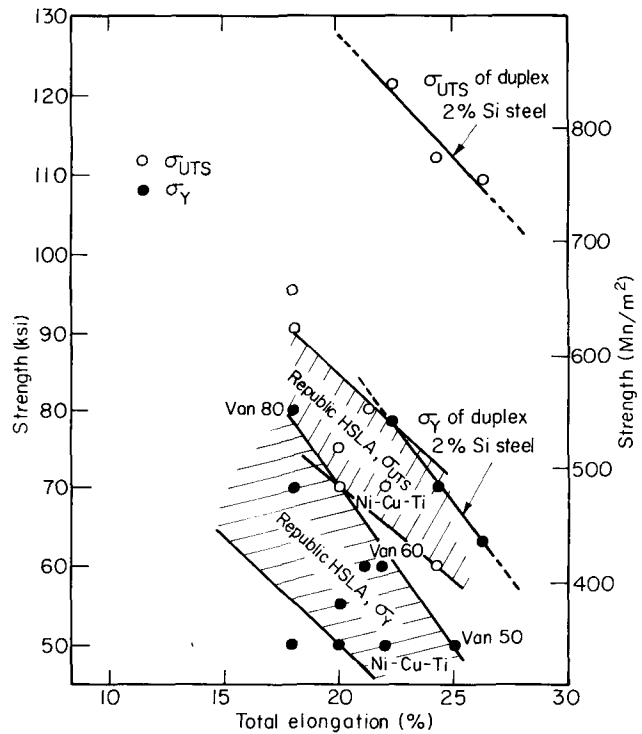


Fig. 5—Tensile properties of the duplex 2 pct Si steel are compared with those of commercial HSLA steels.

DFM structures, amongst which are, 1) shape and distribution of martensite in the ferrite matrix, 2) properties of the martensite, 3) properties of the ferrite, 4) volume fraction of martensite, and 5) microstructural features, such as the presence of coarse carbides. The alloying addition, X, and the exact heat treating conditions control these main structural features.

The superior tensile properties found in the duplex 2 pct Si steel are attributed to silicon being one of the alloying elements which most favorably controls these mutually interacting variables: promoting dislocated martensite in a fine, fibrous distribution, providing solid solution strengthening in ferrite, and improving the ferrite/martensite interface by inhibiting the formation of coarse carbides during the final quench.

The strengthening component obtained from a fine, fibrous martensite morphology appears to be two-fold: it offers more effective barriers to the motion of dislocations, and simultaneously provides more efficient composite strengthening for a fixed volume fraction of martensite as given by the theory of discontinuous fiber composites.^{9,10}

The data in Figs. 6 and 7 show that the range of martensite percentage for improved combinations of tensile properties varies from ~20 pct to ~60 pct, below or above which the morphological features of the DFM structures become undesirable. These values are thus optimum for these Fe/Si/C steels. For the range of martensite percentage shown in Fig. 7, the values of elongation in silicon-bearing DFM steels are higher than those in chromium-containing alloys.

Current research is concerned with optimizing the microstructure-property relations with especial

emphasis being placed on the morphologies of the DFM alloys and with understanding the principal strengthening mechanisms involved.

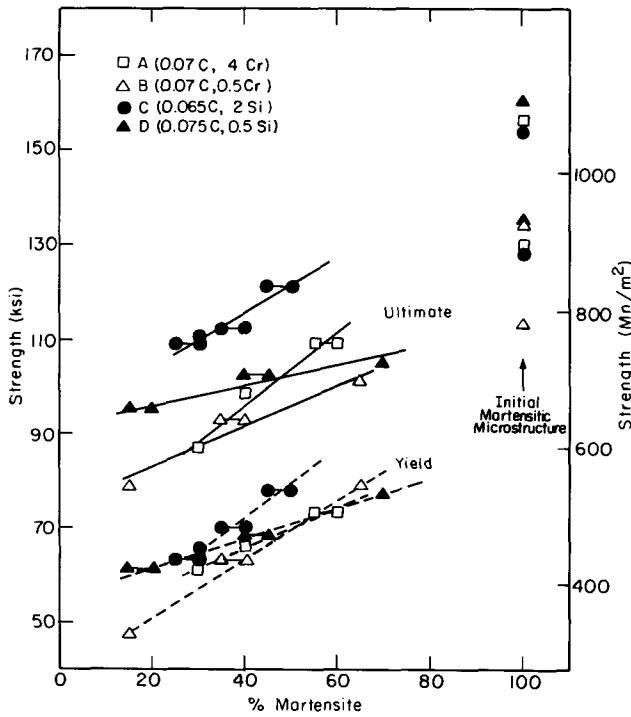


Fig. 6—Yield strength and ultimate tensile strength are plotted against volume percent martensite, for the duplex Fe/X/0.1C alloys.

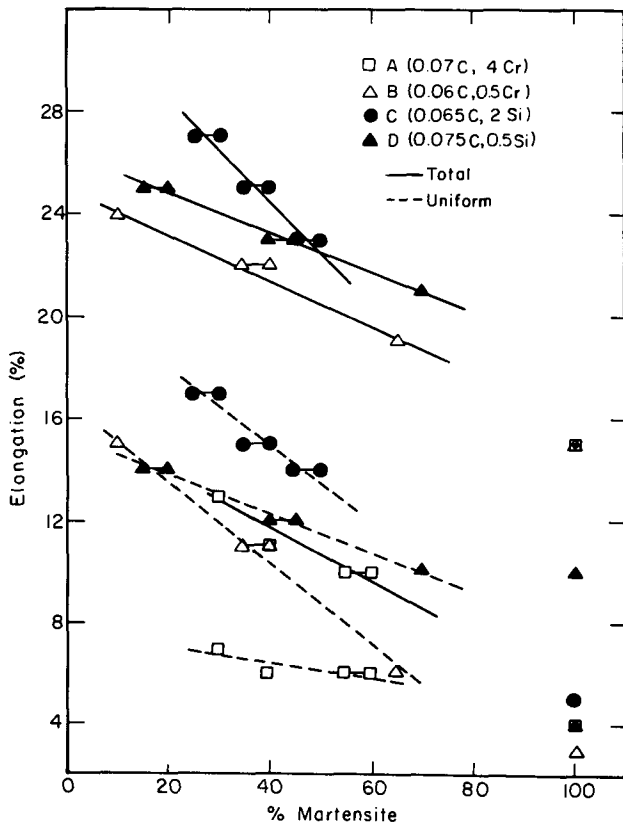
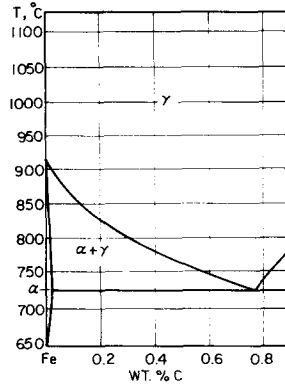


Fig. 7—The values of uniform elongation and total elongation are plotted against volume percent martensite for the duplex Fe/X/0.1C alloys.

Fe-RICH PORTION OF THE Fe-C PHASE DIAGRAM



Fe-RICH PORTION OF THE 2.4 WT. % Si SECTION OF THE Fe-Si-C PHASE DIAGRAM

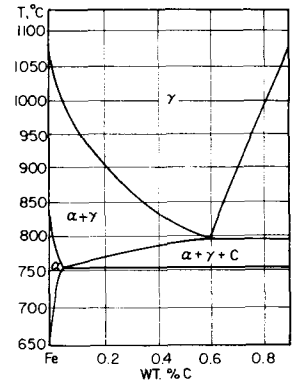


Fig. 8—Phase diagrams showing the expansion of the $(\alpha + \gamma)$ range when silicon is added to the Fe-C system.

Conclusion. From a practical point of view, silicon is a valuable alloying element since it opens up the $(\alpha + \gamma)$ range¹¹ when added to Fe-C system,¹² as in Fig. 8. As a result small variations in annealing temperature in the $(\alpha + \gamma)$ range do not change composition significantly, thereby assuring reproducibility of the material. Silicon is cheap and is an effective solid solution strengthening element. Thus Fe/Si/0.1C steels appear to be very promising for duplex treatments as noted also in Japan.³ However, the published work on these³ does not appear to take into account the desirable macro- and microstructural morphologies which are obtainable under the limitation of practical feasibility. The desirable features have been illustrated in the present work in the duplex Fe/2 pct Si/0.1C pct steel which exhibits better tensile properties than those of Van 80 which is considered to be one of the best commercial HSLA steels. The possibility of using these DFM steels for automotive applications *e.g.* to replace conventional 1010 steels is apparent.

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