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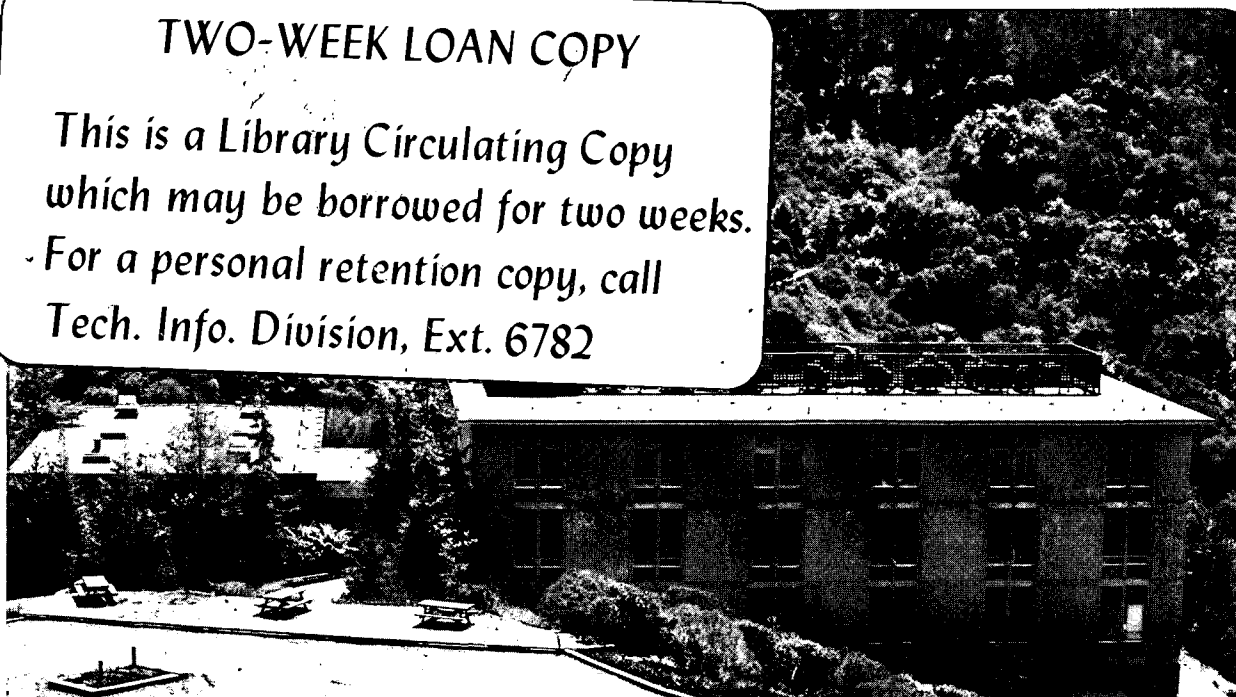
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ON THE LAW OF MIXTURES IN DUAL-PHASE STEELS

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Low carbon steels heat-treated to produce a mixed structure of ferrite and martensite are now known as dual-phase steels. The interest in these steels has grown enormously in the past five years and many of the underlying principles have already been documented, e.g., in two recent symposia.<sup>1,2</sup>

The strengthening principle of such composite structures involves the incorporation of inherently strong martensite as the load-carrying constituent in a ductile ferrite matrix. As such it has been experimentally shown that the tensile strength of dual-phase steels,  $\sigma_c$ , increases *linearly* with increasing volume fraction of materials,<sup>3-7</sup> and is given by the law of mixtures, viz:

$$\sigma_c = \sigma_\alpha(1 - V_m) + \sigma_m V_m \quad [1]$$

where  $\sigma_\alpha$  is the stress in the ferrite at the ultimate tensile strain of the martensite,  $\sigma_m$  is the tensile strength of martensite, and  $(1 - V_m)$  and  $V_m$  are their respective volume fractions. In a strict sense, Eq. [1] is valid 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.<sup>8</sup> This situation deviates significantly from the geometry of dual phase structures. Nevertheless, the mixture rule provides an important guideline which, in spite of the absence of an exact theoretical explanation, can predict with fairly good accuracy the strength level of dual-phase steels over a wide spectrum of martensite volume fractions, provided a minimum of two data points are known.

This paper discusses the applicability of Eq. [1] to dual-phase steels obtained by heat-treating 1010 and 1020 grades. The alloys used in this study were vacuum melted, high purity 1010 (0.1 w/o C, 0.5 w/o Mn, balance Fe) and 1020 (0.19 w/o C, 0.5 w/o Mn, balance Fe) steels. Oversize flat tensile specimens were austenitized in a vertical tube furnace under an argon atmosphere for 30 minutes at 1100°C, and quenched into agitated iced brine. The specimens were reheated to the desired temperature in the ( $\alpha+\gamma$ ) field, held for 20 minutes and then quenched directly into iced brine. For microhardness tests, the austenitized specimens were directly held in the ( $\alpha+\gamma$ ) region for 20 minutes and were subsequently quenched into iced brine. This transformation path was chosen to obtain a coarse dual phase structure<sup>9</sup> which enables valid microhardness measurements to be made. Martensite volume fractions were determined by quantitative optical metallography. Subsize tensile specimens with a gage length of 2.54 cm and a cross section of 0.254 x 0.254 cm were used. Tensile tests were performed at room temperature on an Instron machine at a cross head speed of 0.05 cm/min and full scale load of 1000 Kg.

Tensile test data of the dual-phase steels are summarized in Table 1, and are plotted in Fig. 1. Both yield and tensile strengths increase linearly with an increase in martensite volume fraction. In an attempt to determine the strengths of martensite at various volume fractions of martensite, diamond pyramid microhardness measurements with 25 g load were made for dual-phase 1010 steels having different volume fractions of martensite. The results are plotted in Fig. 2, where the relation between carbon content and hardness of the martensite in the dual-phase steels is

compared with that of as-quenched 100% martensite in iron-carbon alloys obtained by Jaffe and Gordon,<sup>10</sup> and Grange et al.<sup>11</sup> The lower hardness values of the martensite in the dual phase steels, compared to those of 100% martensite, may result from extensive auto-tempering<sup>12</sup> observed in the dual phase structures. The microhardness data were then converted to approximate tensile strengths, as seen in Fig. 3, using the ASM hardness-to-tensile strength conversion chart.<sup>13</sup>

Equation [1] can be rewritten in the convenient form

$$\sigma_c = \sigma_\alpha + (\sigma_m - \sigma_\alpha)V_m \quad [2]$$

Earlier experimental studies<sup>3,14-16</sup> on pearlitic and spheroidized carbon steel have shown that the flow stress of ferrite in these structures is controlled by the mean free path in the ferrite according to the Hall-Petch equation.<sup>17</sup> Ashby,<sup>18</sup> using his micromechanistic model, analyzed the flow stress of two-phase alloys and arrived at a similar equation. Therefore, in cases where the second phase particles are spaced at distances less than the mean dislocation slip length (in a metallic matrix, about 1-50  $\mu$  in most cases), enhanced strengthening of the matrix will result and its mechanical properties will be no longer those of the single phase matrix. Thus the relationship developed above may be extended to dual-phase steels as well, since the interparticle (martensite) spacing is typically of the order of 10  $\mu$  or less. As  $V_m$  increases the spacing ( $d$ ) decreases and it is necessary to account for dislocation strengthening of the ferrite;  $\sigma_y = \sigma_i + k_y d^{-1/2}$ , where  $\sigma_y$ ,  $\sigma_i$ , and  $k_y$  are yield stress, frictional stress, and a constant, respectively. Typically, the tensile strength of low carbon steels is approximately

10 ksi (69 MPa) greater than the yield strength.<sup>19</sup> Then we can write,  
 $\sigma_{\alpha}$  (ksi)  $\approx \sigma_i + k_y d^{-1/2} + 10$  (multiply 6.9 to obtain MPa unit). We assume that the variations in ferrite strength due to carbon are negligible, and also, that when necking occurs, the stress in the ferrite reaches its tensile strength.

The strength of martensite in dual phase steels also varies as  $V_m$  increases because the carbon content also then varies. The relationship can be derived from Fig. 3, where the strength of martensite is shown as a function of the first power of its carbon content,<sup>20</sup> viz.,  $\sigma_m = \sigma_o + k(\%C)$ , where  $\sigma_o$  and  $k$  are respectively the intercept at zero carbon and a constant. In this particular case,  $\sigma_m$  (ksi) = 110 + 470(%C). This result is similar to that obtained for 100% martensite in Fe-Ni-C steels.<sup>21</sup>

The actual carbon contents can be estimated from the tie line compositions corresponding to the annealing temperature in the ( $\alpha + \gamma$ ) phase field. Combining terms, and noting that the product (%C)  $\cdot V_m = a$ , the initial homogeneous carbon content, from a simple mass balance equation with the reasonable approximations that the density of the ferrite equals that of the martensite, and that the carbon content in the ferrite is relatively negligible, we can write

$$\sigma_c$$
 (ksi) =  $\sigma_i + k_y d^{-1/2} + 10 + 470a + [\sigma_o - \sigma_i - k_y d^{-1/2} - 10]V_m \dots$  [3]

(Multiply 6.9 to obtain the MPa unit.) All the terms on the right-hand side of Eq. [3] are constant except the terms  $d$  and  $V_m$ . For dual-phase 1010 steels,  $\sigma_i = 7$  ksi (49 MPa)<sup>22,23</sup> for extremely low carbon steels,  $k_y = 0.7$  ksi  $\cdot$  cm<sup>1/2</sup> (0.5(kg mm<sup>-2</sup>)  $\cdot$  cm<sup>1/2</sup>),<sup>22,23</sup>  $a = 0.1$ ,  $\sigma_o = 110$  ksi (770 MPa),  $k = 470$  ksi (3290 MPa), and  $d$  values were 12 - 20  $\mu$  for  $V_m$  30-45. The tensile



strengths of the dual phase 1010 steel calculated from Eq. [3] agree remarkably well, as seen in Fig. 4, with experimental data (Fig. 3). Equation [3] is based on the assumption that at the necking point of dual-phase steels the martensite constituent reaches its ultimate tensile strength. As a result the tensile strengths predicted from Eq. [3] for the dual-phase steels are slightly higher than those experimentally observed. On the other hand, using the yield strength of the martensite for  $\sigma_m$  instead of the ultimate tensile strengths, Eq. [3] predicts lower tensile strengths than those experimentally observed for the dual-phase steels. Yield strength is used for  $\sigma_m$ , in Eq. [3], under the assumption that martensite deforms only up to its yield strength when bulk necking occurs. From the preceding discussion it appears that when bulk necking occurs the stress state of martensite is somewhere between its yield stress and its ultimate tensile stress. This point will have to be clarified in future work, e.g., by relative strain measurements by X-rays in an instrument equipped with a position-sensitive detector.

The breakdown of the law of mixtures occurs in Fig. 2 where the increase in the tensile strength of dual-phase 1020 steels over that of dual phase 1010 steels at a given volume fraction of martensite is much smaller than that expected from Eq. [3]. The law of mixtures, therefore, is not quantitatively applicable to the case when the strength of martensite changes at a fixed volume fraction of martensite. This is presumably because the martensite constituent may not reach its ultimate tensile strain when bulk necking occurs, as discussed before. Furthermore, the general applications of the law of mixtures for many different dual-phase systems may require an introduction of a geometric factor multiplied

on the RHS of Eq. [1] in order to account for the difference in size, shape, and distribution of martensite constituent, as well as other considerations, e.g., variation in  $\sigma_{\alpha}$ .<sup>9</sup>

In summary, the law of mixtures can quantitatively describe the tensile strength of dual-phase 1010 steels over a wide range of volume fraction of martensite ( $V_m$ ), provided the dislocation hardening is included for the enhanced strength of ferrite as  $V_m$  increases. The application of the mixture rule extended to many other dual-phase systems may require modifications to account for differences in the important metallurgical parameters, e.g., geometry of two-phase structures, precipitation strengthening effects, possible premature failure of the martensite component having a high carbon content, etc.

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TABLE 1. Tensile test summary.

Steel	Designation	Two-phase Temp. (°C)	V <sub>m</sub>	$\sigma_y$ ksi(MPa)	$\sigma_{uts}$ ksi(MPa)
1010	1A	775	30	64(442)	99(683)
	1B	795	35	66(455)	104(718)
	1C	815	45	79(545)	113(780)
1020	2A	725	25	57(393)	96(662)
	2B	740	40	73(504)	122(842)
	2C	755	50	75(518)	127(876)

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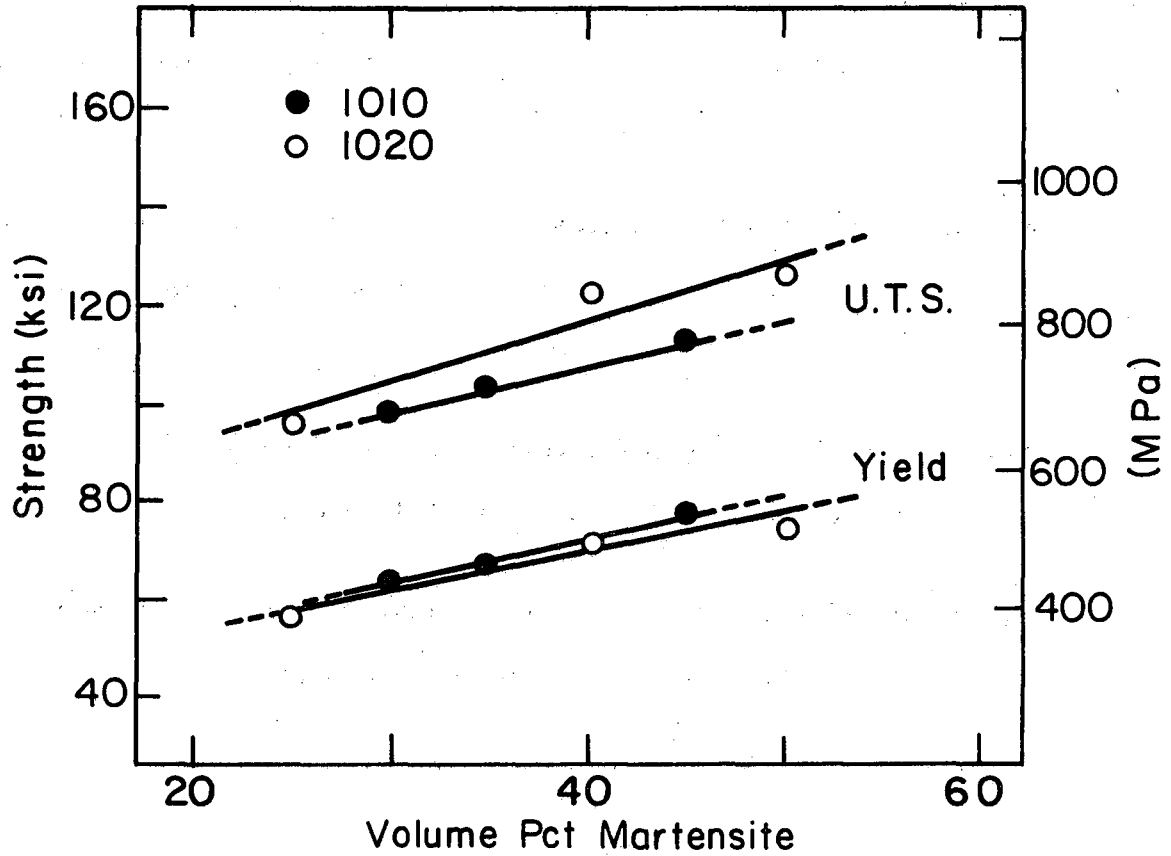
FIGURE CAPTIONS

Fig. 1. Variation of yield and tensile strengths as a function of volume fraction of martensite in dual-phase 1010 and 1020 steels.

Fig. 2. Hardness of the martensite in dual-phase 1010 steels, in comparison with that of as-quenched 100% martensite in Fe-C alloys.

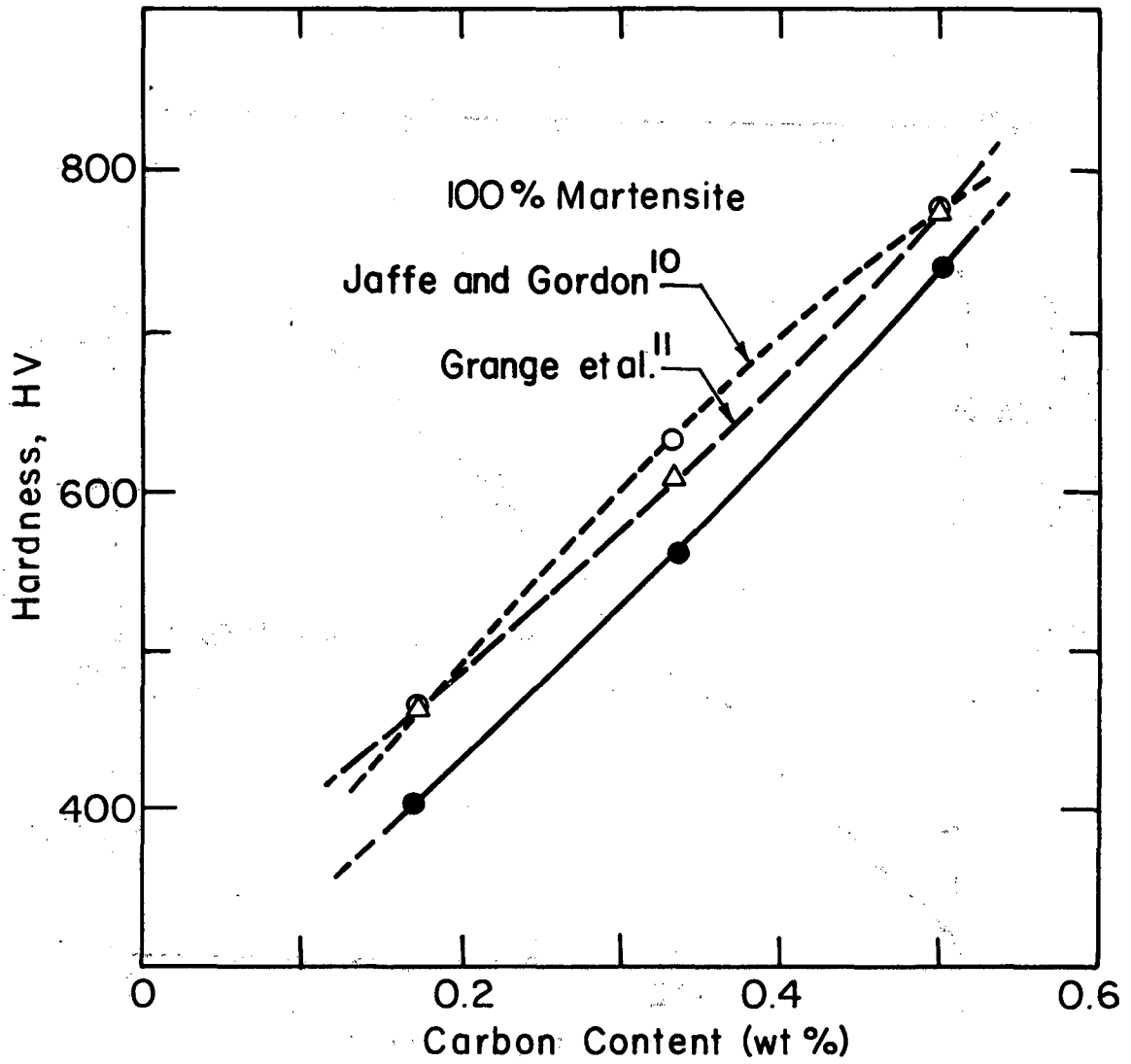
Fig. 3. Tensile strength (converted from Vickers microhardness) of the martensite in dual-phase 1010 steels.

Fig. 4. Tensile strengths of the experimental data are compared with those calculated from the law of mixtures.



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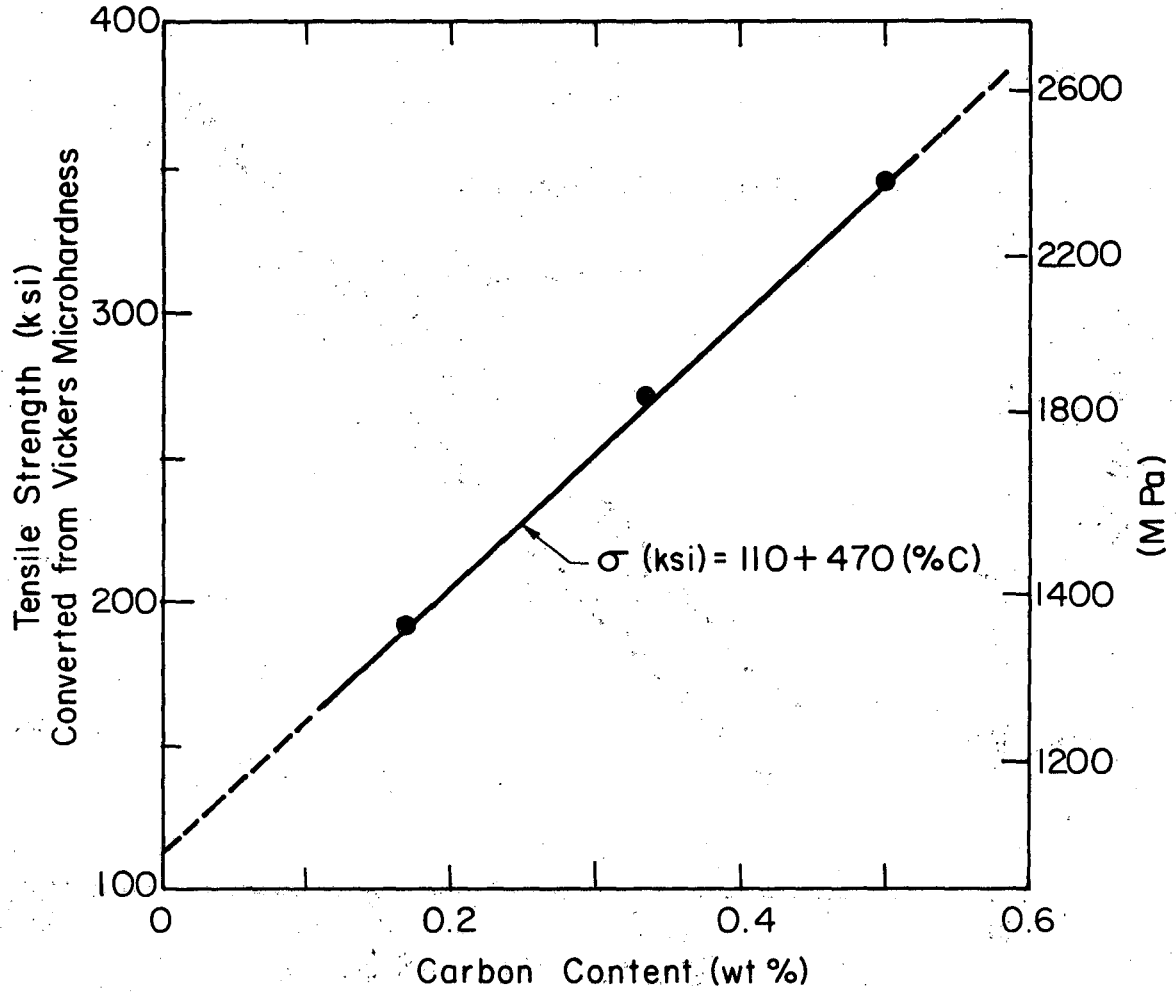
Fig. 1. Variation of yield and tensile strengths as a function of volume fraction of martensite in dual-phase 1010 and 1020 steels.



XBL797-6552

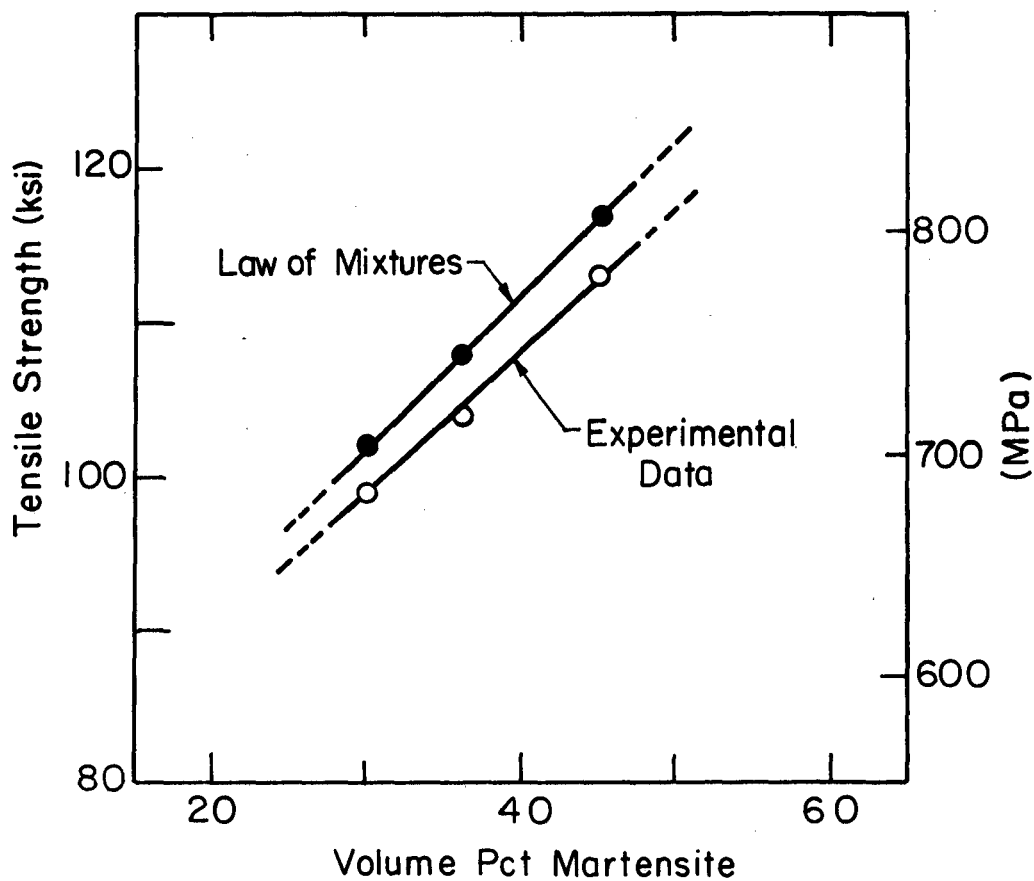
Fig. 2. Hardness of the martensite in dual-phase 1010 steels, in comparison with that of as-quenched 100% martensite in Fe-C alloys.





XBL797-655I

Fig. 3. Tensile strength (converted from Vickers microhardness) of the martensite in dual-phase 1010 steels.



XBL797-6553

Fig. 4. Tensile strengths of the experimental data are compared with those calculated from the law of mixtures.

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