

MASTER**EVIDENCE FOR MICROSTRUCTURAL EFFECTS UNDER STRAIN IN BRONZE****PROCESS Nb₃Sn**

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The possible effects of externally applied strain on the flux pinning process can be divided into two groups according to whether the significant property changes produced by strain are associated with the crystalline defects which act as flux pinners, or with the bulk superconducting material which carries the supercurrent. Changes in the latter are reflected in the equilibrium properties such as the upper critical field, B_{c2} , and the Ginzburg-Landau parameter, κ . To first approximation, changes in the former are not seen in the equilibrium properties, but may affect the bulk pinning force, F_p , by changing the number of pinning centers or their strength. For want of a better term, we have called changes associated with the flux pins microstructural effects, even though they may or may not involve gross changes in structure such as the martensitic transformation which is known to occur in Nb₃Sn at low temperature.

A number of investigators¹⁻³ have concluded, primarily on the basis of the similarity of the dependence on strain of the critical current density, J_c , to the strain dependences of B_{c2} , and the transition temperature, T_c , that, for strains small enough that cracks and filament breakage do not occur^{4,5} equilibrium property changes are chiefly responsible for the strain dependence of J_c . However,

* Research sponsored by the Division of Material Sciences, U.S. Department of Energy under contract W-7405-eng-26 with the Union Carbide Corporation.

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since this conclusion is based on comparisons of J_c and T_c or B_{c2} measurements made on different samples in different apparatuses, or on B_{c2} values obtained by extrapolation of J_c data, these results do not exclude the occurrence of microstructural change as an additional effect altering the strength or density of flux pins.

Whether even gross microstructural changes occur in the Nb_3Sn when uniaxial strain is applied to a bronze diffusion process conductor is not easily determined, because any change which may occur takes place at low temperatures, under cover of the bronze matrix, making the Nb_3Sn all but inaccessible to analytical tools. In the absence of the bronze matrix, little stress can be applied to the brittle Nb_3Sn , and removal of the matrix after straining would change the stress state of the Nb_3Sn .

Since direct evidence is not easily obtained, we chose to investigate the question of whether there is a microstructural component to the strain dependence of F_p by carefully determining whether the bulk property changes which occur can adequately account for all of the variation of F_p with strain. In addition to comparisons of the dependences on strain of J_c , T_c , and B_{c2} , determinations were made of the strain dependences of the parameters which enter the semi-empirical expression for F_p . Examination of these results in light of current theories of flux pinning and type-II superconductivity indicates that bulk property changes do not account for all of the change in F_p , suggesting that microstructural effects are also important.

COMPARISON OF STRAIN DEPENDENCES OF J_c , T_c , B_{c2}

Our first effort was to compare the dependences of J_c and B_{c2} on strain. Figure 1 shows such a comparison for a conductor from Harwell Laboratory, in England, with 1024 filaments, reacted for only 1 h at 700°C, to produce a 0.4 μm thick layer of Nb_3Sn . Conductors with such thin Nb_3Sn layers have been found to exhibit strong I_c variation with strain⁶. The peak in I_c/I_{c0} occurs at a somewhat higher strain than does the peak in B_{c2}/B_{c20} , suggesting that another factor in addition to the variation of B_{c2} is affecting I_c . However, this conclusion requires additional confirmation because, for experimental convenience, I_c and B_{c2} were determined on similar but separate samples, in different apparatuses. The values of B_{c2} at 4.2 K were estimated from measurements of $B_{c2}(T)$ near T_c . These measurements were made resistively, in vacuum to facilitate temperature control. However, the relatively high critical current of this conductor (~70 A at 4.2 K and 7 T) could not be introduced to the specimen in this apparatus without heating in the joints and possible interference with the measurement, so the I_c measurements were made on a different specimen in liquid He at 4.2 K while strain was applied by means of an Instron.

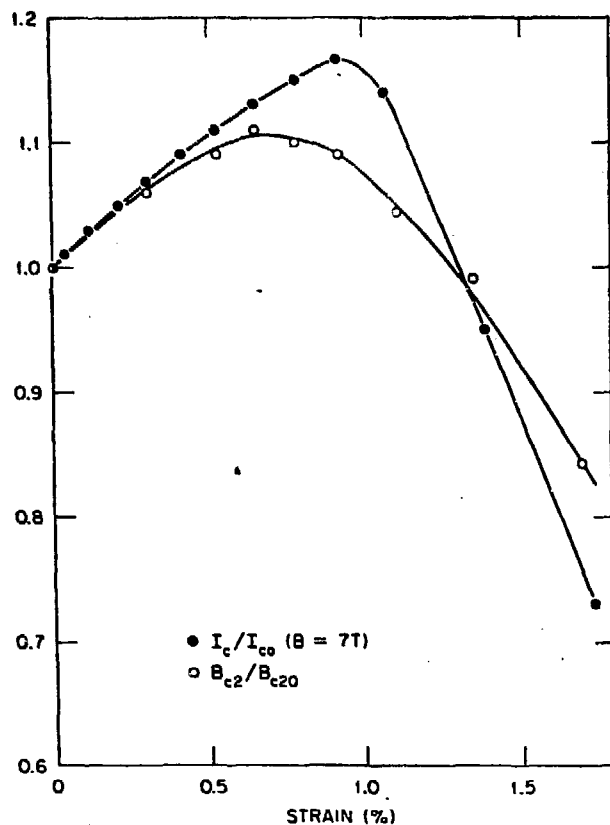


Fig. 1. Comparison of dependences of I_c and B_{c2} on strain in a 1024 filament conductor. Note that I_c peaks at a higher strain than B_{c2} .

tensile testing machine (see Ref. 7,8 for description of the procedure). Thus, uncertainty exists about the comparison of strain measurements. To remove this difficulty, a monofilament conductor with much smaller Nb_3Sn cross-section than the Harwell material, and therefore smaller I_c , was made. The bronze-to-niobium ratio was 34/1, the specimen diameter was 0.08 cm, and the reaction time at 700°C was 8 h, producing a Nb_3Sn layer approximately 1 μm thick. Using the apparatus described in Ref. 9, resistive measurements of $I_c(T)$, $B_{c2}(T)$ and T_c as functions of strain could all be made on the same sample, in the same apparatus, and without intervening stress or temperature cycles. For reasons which will be made clear in the discussion below of scaling behavior of F_p , all T_c and B_{c2} values reported here are what we have called "finish" values, i.e., the values of T and B at which the sample exhibits zero resistance (within limits of detection) to the flow of a small test current. From Fig. 2 we see that T_{cf} and B_{c2f} peak at the same strain, but, as with the multifilament

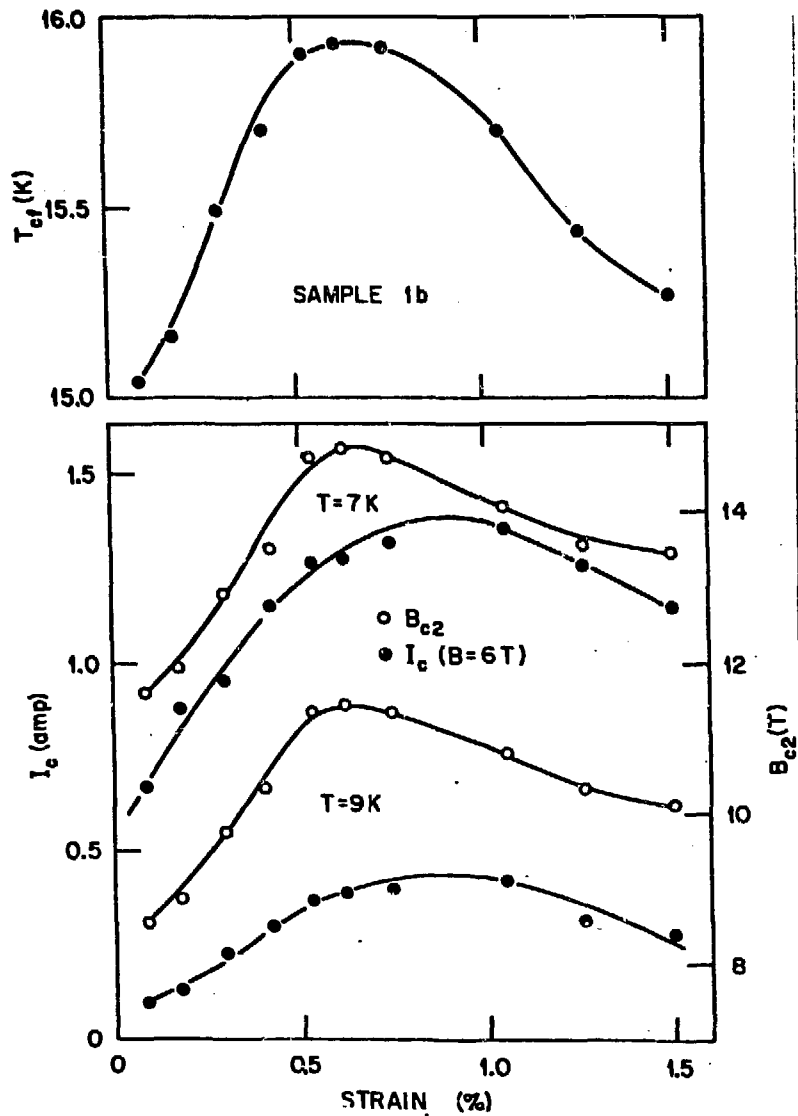


Fig. 2. Comparison of strain dependences of I_c , T_c and B_{c2} for monofilament conductor. The T_c and B_{c2} peaks occur together, but, as for the commercial conductor of Fig. 1, I_c peaks at a higher strain.

conductor, I_c peaks at a somewhat higher strain. These results indicate that B_{c2} variation is not solely responsible for the change in I_c . As will be discussed below, flux pinning theory indicates that F_p also depends on the Ginzburg-Landau parameter, κ , so these results suggest, but do not prove definitely, that additional effects beyond equilibrium property changes must be invoked.

COMPARISON OF STRAIN DEPENDENCES OF T_c AND $(dB_{c2}/dT)_{T_c}$

Further suggestion of microstructural change is obtained by comparing plots of T_{cf}^* and $(dB_{c2}/dT)_{T_c}$ versus strain, as in Fig. 3. Here, T_{cf}^* is obtained by extrapolating the linear portion of the B_{c2} vs T curve to $B_{c2} = 0$, thereby ignoring the curvature which occurs very near T_c . T_{cf}^* and $(dB_{c2}/dT)_{T_c}$ behave rather similarly at low strains ($\epsilon < \epsilon_{maxT_c}$), but at high strains there is a region where $(dB_{c2}/dT)_{T_c}$ is increasing while T_c is decreasing. Since, in type-II superconductors $(dB_{c2}/dT)_{T_c} \propto \gamma \rho_n$ (γ = electronic specific heat coefficient and ρ_n = normal state resistivity), and T_c is determined by the density of states at the Fermi level and is thus a function of γ , one would expect a unique relationship between T_c and $(dB_{c2}/dT)_{T_c}$. Figure 3 indicates that strain changes this relationship.

STRAIN DEPENDENCE OF PARAMETERS IN $F_p(B, T)$

Background and Procedure

Flux pinning studies¹⁰⁻¹⁴ have demonstrated that, for many type-II superconductors, the bulk pinning force density, $F_p = |\mathbf{J}_c \times \mathbf{H}|$, varies with field and temperature according to the equation

$$F_p = A B_{c2}^n(T) f(b) \quad (1)$$

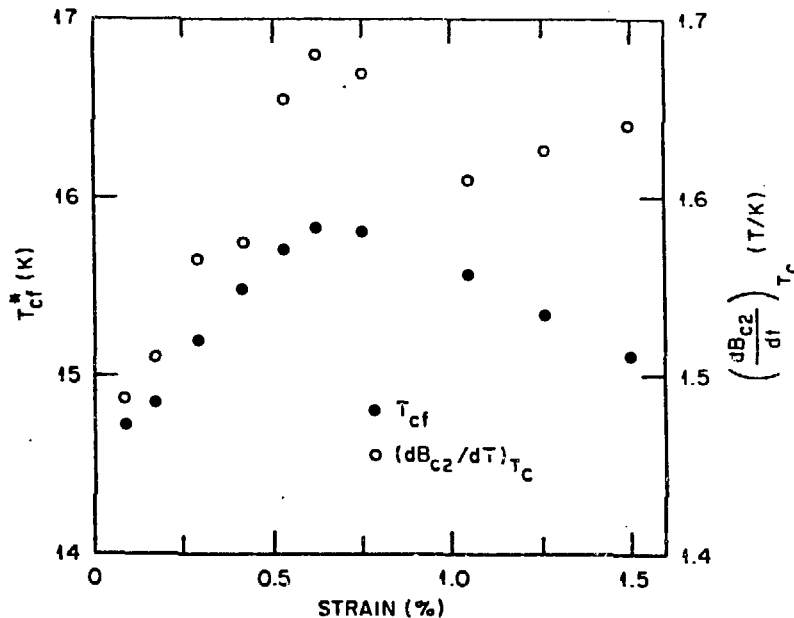


Fig. 3. Comparison of the strain dependences of $(dB_{c2}/dT)_{T_c}$ and T_c .

where A is temperature and field independent, and $b = B/B_{c2}$. A , n , and $f(b)$ are all sensitive to the microstructure of the specimen. Furthermore, expressions for F_p based on fluxoid interactions with the various microstructural features thought to be responsible for flux pinning indicate $f(b)$ should have the form $f(b) = b^{\ell} (1 - b)^m$, where $\ell = 1/2 - 3/2$ and $m = 1 - 2$. Measurements often fit functions of this form reasonably well, although the values of the exponents, ℓ , m , and n , are sometimes not easily rationalized by theory and models. When F_p has the form of Eq. (1), the constant A and the exponents, ℓ , m , and n contain all of the influence of microstructure on pinning force, such as the volume density of pinning centers, pinning center morphology and distribution, and stress fields about microstructural features. Thus, if F_p in bronze-process Nb_3Sn conductors should have the form of Eq. (1), then any influence of stress upon F_p through microstructure should be reflected in these quantities.

On a given specimen, sufficient data were obtained, at each of several applied loads, to test for conformity of F_p to Eq. (1), and to determine values for the constants, A , n , ℓ , and m . For this purpose, I_c and B_{c2} were measured as functions of temperature. Magnetic field limitations confined our measurements to temperatures above 60-70% of T_c . All measurements at a given load were made without cycling the temperature beyond the neighborhood of T_c and without change of load. $B_{c2}(T)$ was determined resistively by fixing B and slowly sweeping T while passing a 3mA test current through the specimen. A voltage criterion of 0.2 $\mu V/cm$ was used. The transition widths observed were about 1 K, presumably reflecting inhomogeneity within the Nb_3Sn layer. Since the transition width varied with applied strain, and tended to be smallest when T_c was a maximum, stress gradients in the Nb_3Sn may also contribute to the width. In general, F_p was found to conform well to Eq. (1) provided B_{c2} was determined from the point in the transition at which the sample voltage went to zero, what we have called the finish point, rather than the (higher) onset point, at which full normal resistance is restored. The B_{c2} 's corresponding to finish and onset points we call B_{c2f} and B_{c2o} .

In Eq. (1), B_{c2} refers to the upper critical field of the bulk of the superconductor. Regions of limited extent in which the superconducting properties differ from those of the bulk act as pinning centers. If the superconducting layer were perfectly homogeneous, and if the voltage criterion were arbitrarily small, B_{c2f} would be the field at which I_c is reduced to the constant 3mA test current, and B_{c2o} would be the field at which $I_c = 0$. For an inhomogeneous specimen, B_{c2f} is the field at which I_c for the continuous path having the highest $I_c(B)$ curve has been reduced to 3mA, and B_{c2o} is the field at which no bulk superconductivity remains in the specimen. The 1 K width of the transition shows that the Nb_3Sn layer is inhomogeneous, but the fact that F_p scales

with $B_{c2f}(T)$ indicates the existence of a connected phase (or composition) which can be characterized by the transition curve $B_{c2f}(T)$ and which carries the transport current.

Results

Figures 4 and 5 show plots of F_p/F_{pmax} vs b for the same specimen at two different strains, $\epsilon = 0$ and $\epsilon = 0.64\%$. The latter is approximately the strain at which T_c and B_{c2} were maximized. In both cases, the data tend to lie on a single curve, i.e., F_p scales with B_{c2f} . In Fig. 4, normalized curves of $f(b) = b^\ell(1-b)^m$ for various values of ℓ and m are drawn. In this way, very approximate determinations of the values of ℓ and m can be made. In Fig. 4, the data fall between the curves for $\ell = 1/2, m = 2.5$, and $\ell = 1/2, m = 3$. In Fig. 5, all except the 14.25 K data are closest to the $\ell = 1/2, m = 3$ curve. The maxima of $f(b) = b^\ell(1-b)^m$ occur at $b_{max} = \ell/(\ell + m)$, so that increasing ℓ moves the maximum to larger b and increasing m moves it to smaller b . In all cases measured,

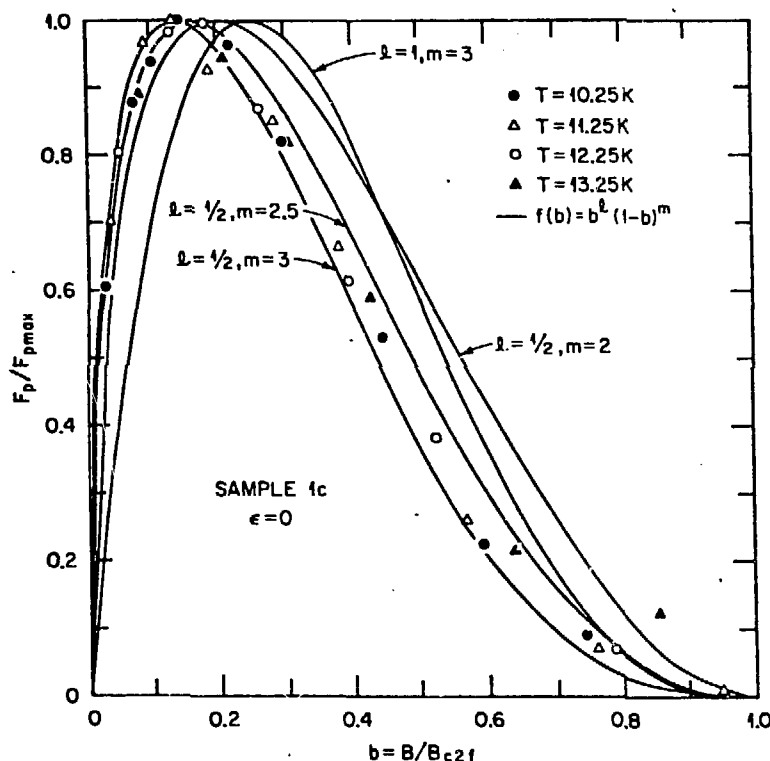


Fig. 4. Plots of F_p/F_{pmax} vs B/B_{c2} for various temperatures tend to fall on the same curve. Plots of $f(b) = b^\ell(1-b)^m$ are shown for various values of ℓ and m . These data, for $\epsilon = 0$, fall between the $\ell = 1/2, m = 2.5$ and $\ell = 1/2, m = 3$ curves.

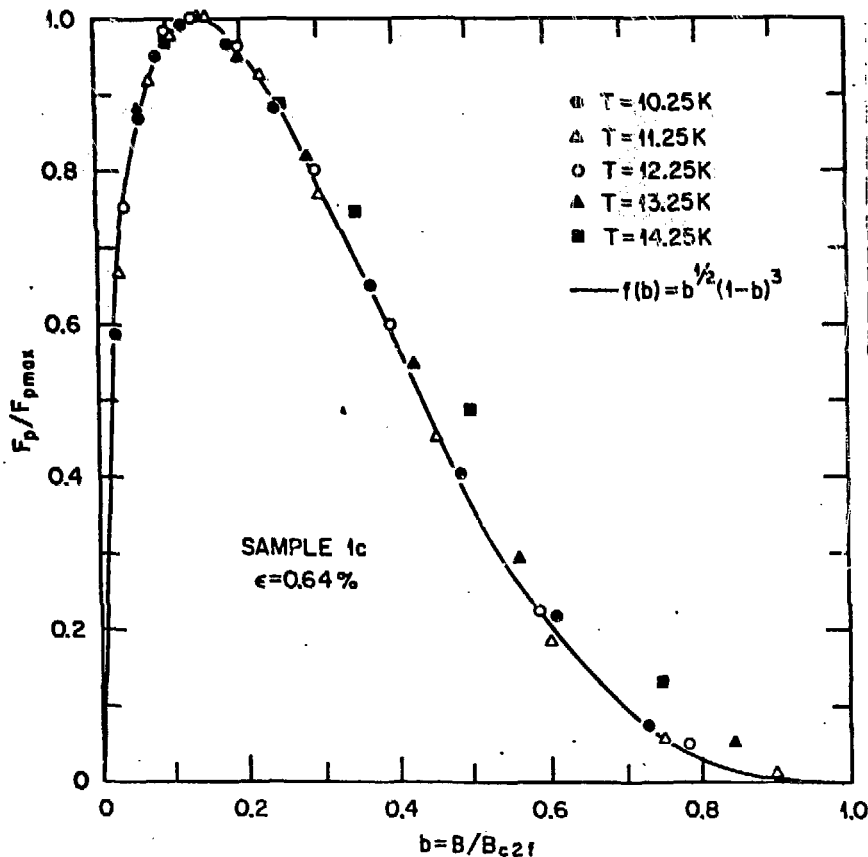


Fig. 5. At 0.64% strain, F_p scales well with B_{c2f} , and the data fall approximately on the $f(b)$ curve for $\ell = 1/2$, $m = 3$.

the position of the maximum indicates that the appropriate value of ℓ is $1/2$ if very large values of m (≈ 5) are to be avoided. Values of m all lie between 2.5 and 3, with m tending to increase slightly as the applied strain nears $\epsilon_{\max T_c}$; that value at which T_c and B_{c2} are maximized. This trend can be seen in Figs. 4 and 5.

Scaling was not observed in all cases. Fig. 6 shows $F_p/F_{p\max}$ vs b for the same sample as in Figs. 4 and 5, but at a strain greater than $\epsilon_{\max T_c}$. The absence of scaling is apparently not due to the development of major cracks or break up of the Nb_3Sn layer, since the I_c values were higher at this strain than at the lower strains of Figs. 4 and 5. This loss of scaling above $\epsilon_{\max T_c}$ was not observed in two other specimens, results for one of which are shown in Fig. 7 at $\epsilon = 1.1\%$, which is well beyond $\epsilon_{\max T_c}$. The scaling is excellent. Also note that the best value of m is 2.5 rather than 3 for this specimen.

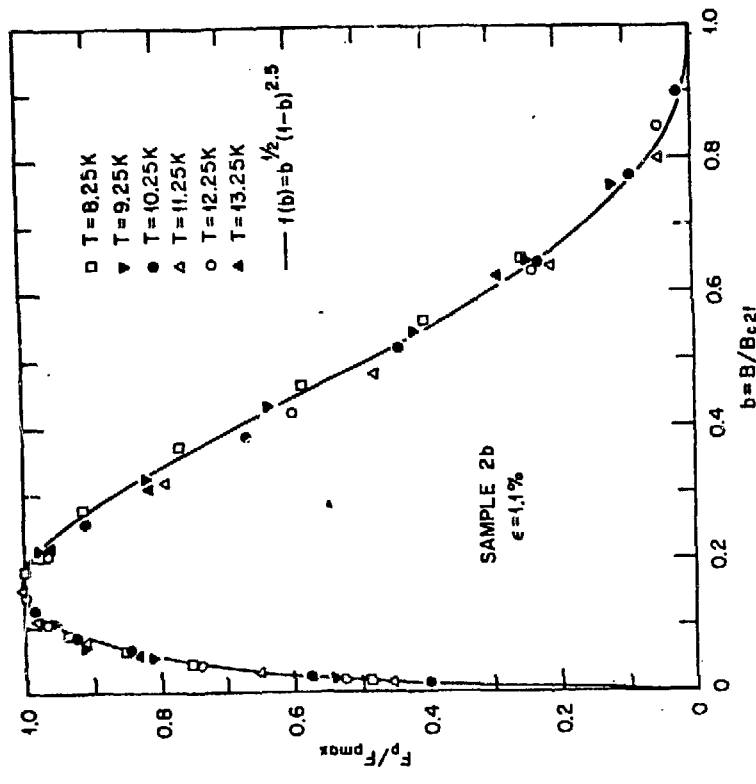


Fig. 7. For sample 2b, F_p scaled well with B_{c2f} even at $\epsilon = 1.1\%$, which is well beyond the strain at which B_{c2} and T_c reached maxima.

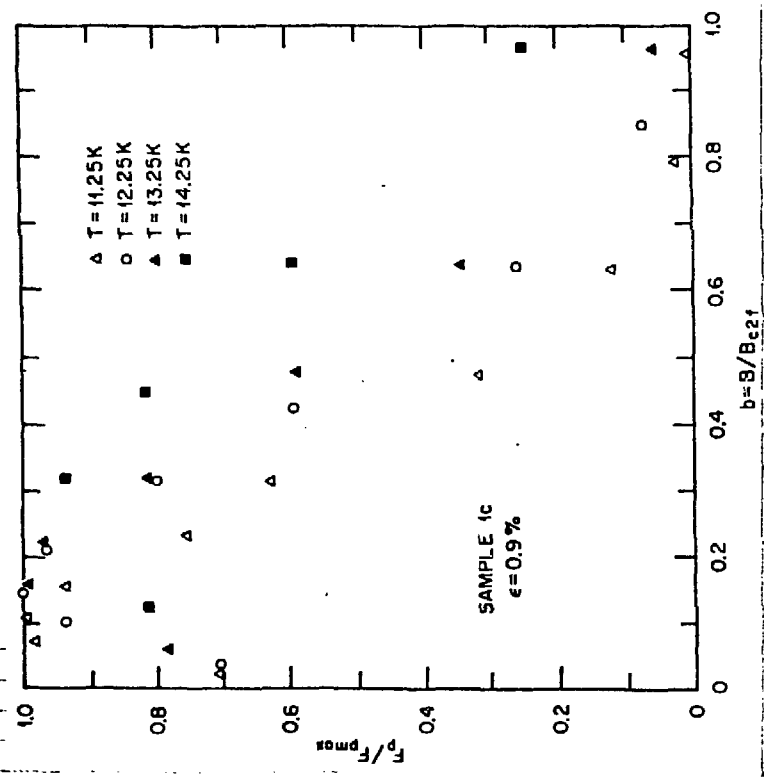


Fig. 6. At the higher strain of 0.9%, F_p no longer scales with B_{c2f} for sample 1c.

The exponent n of B_{c2} in Eq. (1) can be determined by plotting $\log F_p$ at a particular b vs $\log B_{c2}$ for the various temperatures at which data were taken. It is convenient and customary to use F_{pmax} . Figure 8 shows such plots for the data of Figs. 4 and 5. The plots are linear, their slopes being equal to n . For sample 1c, sufficient data were obtained to determine n at five loads, and all of the values were between 2.9 and 3.2. For sample 2b (Fig. 7), n was determined at two loads, giving values of 2.5 and 2.6. Sample 2a, on which data were taken only for zero load, also gave $n = 2.5$. The values of n were determined by least squares fit to the data.

These results, as well as the loss of scaling for $\epsilon > \epsilon_{maxT_c}$ in sample 1c and the variation of m from sample to sample, illustrate the variability of behavior under stress exhibited by this material. This variability may itself be regarded as suggestive that a stress-mediated phase transformation, or other microstructural change, is occurring under stress, since electronic properties would not be expected to exhibit such irregularity.

How rapidly the sample is cooled, and whether it is under stress during cooling may also be involved in sample-to-sample

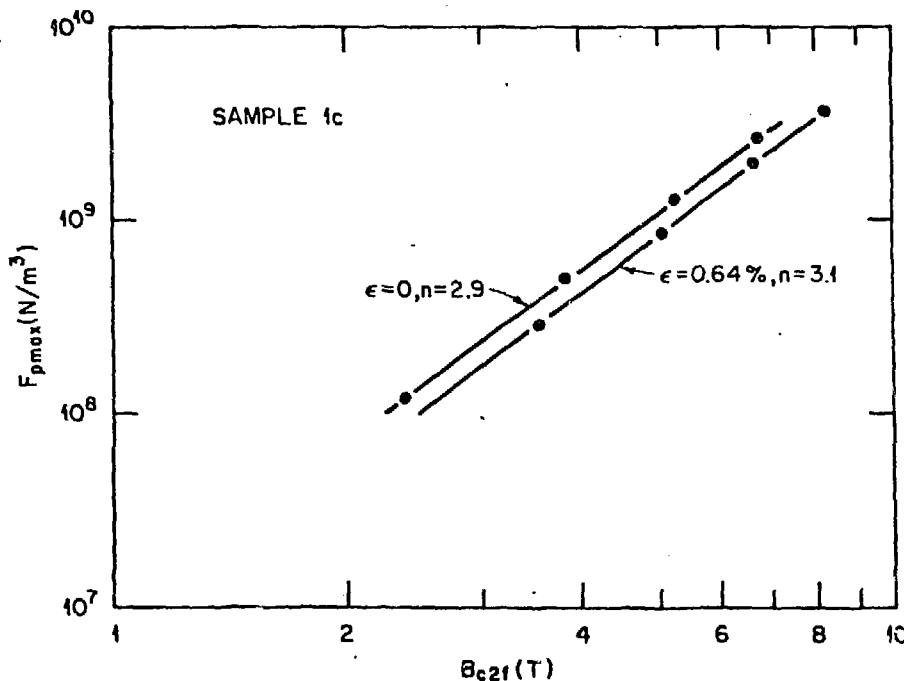


Fig. 8. $\log F_{pmax}$ vs $\log B_{c2f}$ was linear for all strains at which F_p scaled with B_{c2f} , but the lines for different strains are displaced, indicating that A depends on strain.

variation, because of the following unusual effect. On occasion specimens were left overnight in the cryostat, under load. During this time, the temperature of the specimen rose to 30-35 K. The first measurement of T_c made the next morning was invariably different from the value obtained the previous afternoon, sometimes by as much as 0.5 K. Generally, if the load was less than that required to produce maximum T_c , then T_c increased overnight, and if the maximum T_c had already been passed, T_c decreased. Also, some drop in load occurred overnight. Thus, the direction of the change in T_c was such as to suggest that additional strain occurred overnight under constant or slightly decreasing load. A controlled experiment, in which the sample was held under constant load at temperatures up to about 35 K, with a strain gage attached to the specimen, confirmed that the changes in T_c were due to changes in length of the specimen which occurred under constant load and temperature as a function of time. This effect was studied only briefly, but one rough determination of rate was made; at 35 K, with a load of 200 Mpa, approximately 0.04% strain occurred in 1 h, with a consequent increase in T_c of 0.07 K. At temperatures under about 25 K the rate of elongation was quite small. Measurements were not made at temperatures above 35 K. Since the components of the conductor are always under load during cooling from room temperature, different cooling schedules may result in different stress states for the Nb₃Sn.

Clearly, much more work is needed to understand these observations, but we may note that in these specimens the 1 μ m thick layer of Nb₃Sn represents a very small percentage of the total cross-section, and the larger Nb core is still only about 3% of the total. Therefore, it is reasonable to assume that this effect is due to a process occurring in the bronze matrix. The amount of tin removed from the bronze during reaction to form the Nb₃Sn layer is nearly negligible in this conductor, contrary to the case of commercial conductors, for which the matrix may have only 2-3% tin remaining after reaction. But if this effect should be found in commercial materials, it may have significance for operating procedures for Nb₃Sn magnets.

The fact that the lines in Fig. 8 are not coincident indicates that the factor A is different for the two data sets. Values of A can be determined for each load from the constant term in a linear fit of $\log F_{pmax}$ vs $\log B_{c2}$. The filled circles in Fig. 9 show how A varies with strain in Sample 1c. Values for strains larger than approximately ϵ_{maxT_c} were not obtained because, as previously stated, the F_p data did not scale.

Data are also shown in Fig. 9 for sample 1b. Measurements sufficient for determination of all the parameters in the expression for F_p were not obtained for this sample. Instead, at each of several loads, $B_{c2}(T)$ was measured, along with I_c at several

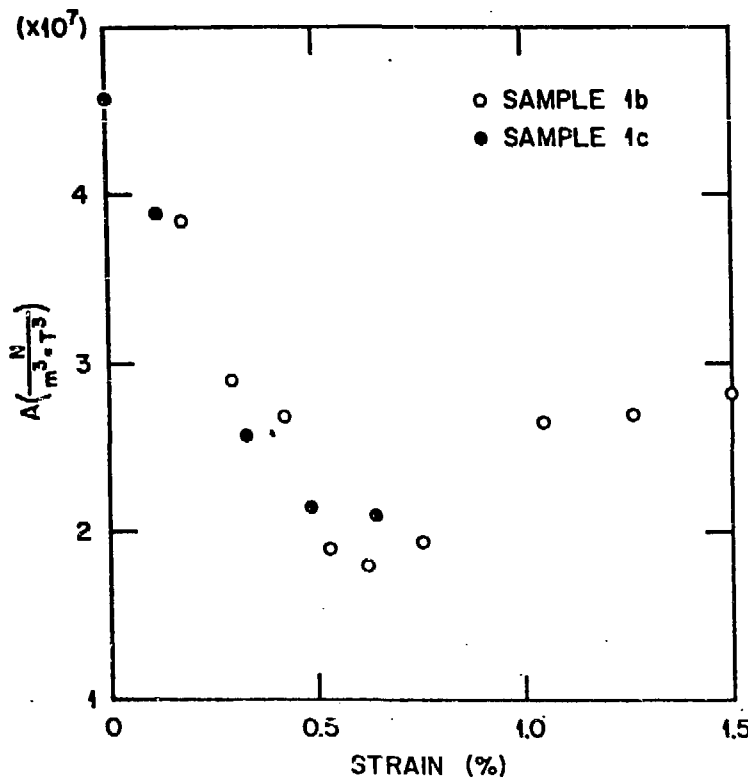


Fig. 9. A decreased by more than a factor of two between $\epsilon = 0$ and $\epsilon \approx 0.6\%$, the latter being near that value of strain at which B_{c2} and T_c were maximized.

selected fields and temperatures. Plots of F_p vs $B_{c2}^3 b^{1/2} (1-b)^3$ made from these data were found to be linear, indicating that the behavior of this sample is similar to that of sample 1c. However, there was no apparent change in behavior around ϵ_{maxT_c} comparable to the loss of scaling which occurred in sample 1c. The slopes of the F_p vs $B_{c2}^3 b^{1/2} (1-b)^3$ plots are of course equal to the fore-factor A, and the values of A so determined are plotted in Fig. 9. It should be pointed out that if n and m are not precisely 3, as has been assumed, plots of F_p vs $B_{c2}^3 b^{1/2} (1-b)^3$ might still be approximately linear, but with slopes which differ from the values of A which would be obtained from analysis of a full set of $J_c(B,T)$ data.

It should be stressed that the variation of A with strain represents an effect on J_c which is independent of the change in B_{c2} with strain. A has a minimum which approximately coincides with the maxima in T_c and B_{c2} so that the effects of A and B_{c2} on J_c are in opposite directions. The B_{c2} effect dominates, but for

Sample 1c, F_p calculated at $\epsilon = 0.64\%$ for $B = 3$ T, assuming that A and n have the values determined at $\epsilon = 0$, was $\sim 40\%$ higher than the measured value. Thus, if the observed B_{c2} change were the sole effect of straining the sample, J_c at $\epsilon = 0.64\%$ would be $\sim 40\%$ greater at 3 T than was measured.

Effect of Strain Dependence of κ on A

Regardless of the pinning species, or mechanism, or the manner of summation of the elementary pinning force, the forefactor A should be proportional to the number, N , of pinning centers per unit volume. It also depends on field and temperature independent parameters which determine the strength of a pin, such as its size, or in the case of pinning through the elastic energy, the stress field of a defect. Therefore, the observed variation of A may imply changes in either or both of those microstructurally determined quantities. However, A also depends inversely on the Ginzburg-Landau parameter, κ , raised to a power p , which, according to most theories, is in the range 2-4. Therefore, to discuss the question of microstructural change we must evaluate the effect of κ on A .

If we assume that A depends linearly on a parameter, a , which is related to the pin strength, then $A \propto Na/\kappa^p$. Furthermore, $\kappa \propto \gamma^{-1/2} (dB_{c2}/dT)_{T_c}$ where γ is the electronic specific heat coefficient. Therefore, the fractional changes in A , N , a , γ and $B_{c2} = (B_{c2}/dT)_{T_c}$ associated with a small change in strain are related by

$$\delta A/A = \frac{\delta N}{N} + \frac{\delta a}{a} + \frac{p}{2} \frac{\delta \gamma}{\gamma} - p \frac{\delta B'_{c2}}{B'_{c2}} \quad (2)$$

The first two terms are the microstructural component of $\delta A/A$ and the latter two are the κ or electronic component. Note that as the sample is strained from 0 to $\sim 0.6\%$, A decreases by more than a factor of two, and T_c increases by ~ 1 K. If this change in T_c is associated with a change in γ , then $\delta \gamma$ is small but positive, i.e., its sign is opposite to that of δA . Thus, the effect of κ on A will be overestimated if we ignore the $\delta \gamma$ term. The change in B'_{c2} with strain was determined for both samples of Fig. 9, using the $B_{c2}(T)$ data and ignoring the slight upward curvature which appears very near T_c . For sample 1b, when A changes from 3.8×10^7 N/m³T³ to 1.8×10^7 N/m³T³, B'_{c2} changes from ~ 1.5 T/K to ≈ 1.7 T/K, giving $\delta A/\langle A \rangle = -0.7$ and $\delta B'_{c2}/\langle B'_{c2} \rangle = 0.12$. For sample 1c, the corresponding numbers are $\delta A/\langle A \rangle = -0.74$ and $\delta B'_{c2}/\langle B'_{c2} \rangle = 0.06$. Depending on the value of p , the fourth term in Eq. (2) is in the range of 15-50% of $\delta A/\langle A \rangle$, and, as was pointed out above, the third term is likely to be opposite in sign to $\delta A/A$. Therefore, it does not appear that

the κ dependence of A can adequately account for its change with strain, and that the microstructural component must be responsible for a substantial portion of $\delta A/A$.

However, firmness of this conclusion must be tempered by the realization that our understanding of flux pinning may yet be so imperfect that data of this kind cannot be adequately interpreted. It may be important in this connection to note that while F_p follows the form of Eq. (1), the values of the exponents n and m are, as in some other flux pinning studies, larger than can be accounted for by theory. Also, while numerous studies have been made of the dependence of F_p on B , B_{c2} , and microstructure, direct tests of the dependence of A on κ have not, to our knowledge, been reported. Nevertheless, interpretation of our results according to current understanding of flux pinning leads to the conclusion that microstructural factors affecting F_p change when stress is applied.

MICROSTRUCTURAL EFFECTS WHICH COULD ACCOUNT FOR RESULTS

In Nb_3Sn , grain or interphase boundaries, or second phase particles are considered to be the microstructural features responsible for flux pinning. Consideration of the ways in which applied stress might alter either N or a has led to a limited number of possibilities. A brief discussion of each follows.

A change in the number of microstructural features available to act as pinning centers could be caused by strain through a stress-induced transformation, reversible twinning, and the formation of microcracks, which, in early stages of development, may be of an appropriate size to pin flux. As strain is increased, and the microcracks grow, their effect on J_c would become deleterious, in accord with the rapid decline of J_c observed at large strains. If a stress induced transformation is involved, it would not necessarily be the tetragonal transformation known to occur in unstressed Nb_3Sn at about 42 K.

External application of stress may alter the stress distribution about a pinning site, if the crystalline defect which acts to pin flux also acts as a point of stress concentration, or if the applied stress results in other microstructural change such as a phase transformation or generation of defects such as twins or dislocations. Therefore, if the crystalline defect-flux line interaction through the elastic energy is an important component of the pinning, then the strength of a pin could be altered by external stress.

If two superconducting phases are present, and pinning is due to differences in superconducting properties (κ , H_{c2}) between the

two phases, then the pin strength would be altered if, as one may expect, the superconducting properties of the two phases are affected differently by strain. Similarly, if pinning is due to anisotropy of H_{c2} , stress may alter pin strength by altering the degree of anisotropy.

It is possible that more than one type of pin is operative, or that the same defect pins by more than one mechanism. If so, applied stress could alter the relative strengths of pins or pin mechanisms, thereby changing not only the pin strength but the dependences of F_p on B_{c2} and B . Bartlett, et al.¹⁷ put forth this interpretation of changes with strain which they observed in F_p vs b curves. Additionally, if a pinning threshold exists, so that pins weaker than some threshold value are ineffective, then the population of operative pins could be changed by strain if the strength of the pins were affected.

SUMMARY

In both mono- and multifilament conductors, we have found that the maxima in J_c and B_{c2} as functions of strain do not coincide. This indicates that variation of the bulk property, B_{c2} , does not account for all of the variation of J_c with strain, and suggests that microstructural effects, associated with the pinning centers, are involved.

Comparison of the strain dependences of T_c and $(dB_{c2}/dT)_{T_c}$ indicates that the relationship between them is not unique, i.e., that the relationship changes when strain is increased beyond that required to produce maxima in T_c and $(dB_{c2}/dT)_{T_c}$. Since $(dB_{c2}/dT)_{T_c} \propto \gamma \rho_n$ and T_c is determined by the density of states, or γ , the relationship between them should be determined uniquely by crystal structure and resistivity.

The samples exhibited variability of behavior under stress, such as the loss of scaling of F_p with B_{c2f} at large strains in sample 1c (but not in other specimens), and variation from sample to sample, as well as with strain, of the exponents n and m . The loss of scaling might result from an abrupt change in bulk properties, and the sample to sample variability may be related to the time dependent strain which was observed to develop at constant load and temperature.

From measurements of J_c and B_{c2} as functions of strain and temperature, the forefactor A in the expression for F_p was found to depend strongly on strain. A is determined by pin strength, pin density, and κ . Measurements of $(dB_{c2}/dT)_{T_c}$ and T_c as functions of

strain indicate that variation of κ with strain does not account for the change in A. Therefore, a microstructural effect, associated with the pins, is implied.

Microstructural effects which could account for these the results include a stress-induced martensitic transformation, twinning, microcracks, and changes in the stress distribution about pinning sites. If pinning involves a second superconducting phase, change in the superconducting properties of the second phase with strain could alter the pin strength.

ACKNOWLEDGMENTS

The authors would like to thank J. P. Charlesworth of A.E.R.E. Laboratory, Harwell, England for providing multifilamentary specimens, J. O. Scarborough for performing many of the measurements, and Connie Harrison for efficient preparation of the manuscript.

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