CONCERNING FACTORS WHICH DETERMINE WHETHER FLUX-LATTICE SHEAR OR PIN BREAKING LIMITS THE CRITICAL CURRENT DENSITY OF SUPERCONDUCTORS

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Abstract --An elementary model is presented which illustrates the conditions under which flux-lattice shear, rather than pin breaking, limits the critical current density. An expression for the shear strength of the flux-lattice, based on the plasticity of metals and alloys, is used to derive the critical current density, including the effect of thermal activation in the flux creep regime.

I. INTRODUCTION

Details concerning the form and interpretation of scaling laws for the critical current density J_c of the type II superconductors have a long and contentious history, but the essential issues concern what physical barriers to flux motion limit the critical current density and consequently how much the introduction of pinning centers by various means can raise J_c . We focus here on two questions: (1) what characteristics of the pinning center distribution determine whether the breaking free of flux lines from the pinning centers or the plastic flow of the flux line lattice (FLL) limits J_c ? and (2) what is the effect of temperature on J_c if the plasticity of the FLL is the controlling factor? Some crucial issues in the questions include (1) the state of stress in the FLL as a function of the strength and distribution of pinning centers and (2) the density and arrangements of dislocations in the FLL, as well as their response to stress and temperature. In the first case we will utilize a highly simplified model to estimate the stress and in the second we will rely heavily on the theory and empirical knowledge of the plasticity of metals. The approach is a generalization and extension of the models used by Pruyboom et al.[1].

II. THE STRESS STATE AND FLUX-FLOW INSTABILITY OF THE FLUX LINE LATTICE

A state of stress in the FLL requires non-uniformity in the distribution of forces acting upon it: if every flux line is subject to the same force there will be no stress. Consequently, there must be a distribution in strength and/or density of pinning centers varying spatially if a stress is to arise. Hence, as a minimum, we need a characterization of the distribution of pinning strengths and densities and its spatial organization, i.e., its characteristic length scale. In general, given these specifications, it is still a difficult matter to derive the stress state. We use here for illustration a very simplified model: the spatial variation of pinning force density along a direction perpendicular to the direction of the Lorentz force is approximated as periodic with wavelength Λ , consisting of alternating regions with relatively weak and strong pinning, of width $q\Lambda$ and $(1-q)\Lambda$ respectively, i.e. q is the fraction of the FLL at the earliest

risk of flow. (We will denote the weak and strong regions by subscripts w and s , respectively.)

We assume for simplicity that the pinning force density, the effect of which we approximate by direct summation, is constant in each of the channels ("square-wave" approximation), which are assumed to form percolation paths through the material. Also, geometrical factors accounting for the orientation of the channels relative to the Lorentz force [1], and other factors describing the topology of the percolation network will be omitted here. Under these assumptions the net force N acting along the channel direction, per unit length of flux line and per unit length of channel, on the FLL in each of the regions is given at equilibrium by:

$$N_s = JBA(1-q) - (B/\phi_0)\Lambda(1-q)(gf_p)_s + 2\tau = 0 \quad (1)$$

$$N_w = JBAq - (B/\phi_0)\Lambda q(gf_p)_w - 2\tau = 0 \quad (2)$$

where J is the current density, B the magnetic induction, ϕ_0 the flux quantum, f_p the pinning force per unit length, g the probability that a given flux line is pinned, and τ is the shear stress acting at the interface between the weakly and strongly pinned regions. The pinning forces f_p and stress τ increase with increasing J , but can only sustain certain maximum values, $f_p^{\max}$ and τ_{flow} . Thus a depinning and/or plastic flow instability arises at the critical current density J_c . Information about the pinning force-displacement and FLL stress-strain relations are required, in addition to Eqs. (1) and (2), to fully solve the problem, but some general conclusions can be drawn. Before flow can occur $(f_p)_w$ must have reached its maximum value; the stress τ will then depend upon the difference between $(gf_p)_s$ and $(gf_p)_w^{\max}$ and will rise until either $(gf_p)_s$ reaches its maximum value or τ reaches the plastic flow stress τ_{flow} . In the first case, pin breaking is then the J_c -limiting process, while in the latter FLL plasticity is limiting. Thus, the criterion for FLL plasticity limiting the critical current is:

$$\tau_{\text{flow}} < (B/2\phi_0)\Lambda q(1-q)[(gf_p)_s^{\max} - (gf_p)_w^{\max}] \quad (3)$$

III. THE CRITICAL CURRENT DENSITY, NEGLECTING THERMAL ACTIVATION

To proceed further requires an explicit description of the flow stress in terms of the microstructure of the FLL. Pruyboom et al. [1] utilized the theoretical shear strength of a perfect FLL, $\sim 0.05C_{66}$ where C_{66} is the FLL shear

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modulus, recognizing however that the presence of FLL dislocations would reduce this value. Here we will quantify this dependence on dislocation density ρ by using the Bailey-Hirsch relation [2] between flow stress and dislocation density, which is obeyed for metals and alloys under a wide range of deformation conditions; in the context of the problem at hand, this becomes:

$$\tau_{\text{flow}} \equiv \tau_0 + \alpha C_{66} (\phi_0/B)^{1/2} \rho^{1/2} \quad (4)$$

where α is a constant of the order of unity (for atomic lattices but may differ from this for FLLs) and τ_0 is the dislocation friction stress; $(\phi_0/B)^{1/2}$ is the dislocation Burgers vector (the FLL spacing, in this case). Unlike the case of the atomic lattices in metals and alloys, where electron microscopy, x-ray methods, etch-pitting, etc. can be used to determine the dislocation density as function of plastic strain and other variables, we have no knowledge of this quantity for FLLs, and we must rely on plausibility arguments and analogies with atomic lattices. We expect that in this case ρ depends on both the pinning strength and density, since these parameters control the distortion of the FLL. We shall assume that there are nL^{-2} dislocation nucleation centers present in the pinning microstructure, where n is probably a small number and L is a length scale characteristic of the microstructure (e.g., the grain size or the interpin spacing in the case of point pins). It is known from experiment and theory [3] in the case of deformed metals and alloys, that there is a maximum value of the dislocation density attainable, of the order of $(.02 \text{ a}^{-1})^2$, where a is the interatomic spacing, and similar behavior should be expected for FLLs. Under these assumptions we find:

$$\rho \equiv \gamma^2 (B/\phi_0) [1 - \exp(-\gamma n L^{-2} \phi_0/B)] \quad (5)$$

where γ is the probability that a given dislocation nucleation center emits a dislocation into the FLL, and γ is a numerical parameter, probably about .02. We expect that γ depends on the pinning strength relative to the strength of the FLL: for example, a function of the ratio of the pinning energy per unit length, $\sim f_p^{\text{max}} r_p$ (where r_p is the pinning range), to the energy of the dislocation per unit length, $\sim C_{66} \phi_0/B$. A plausible function of this sort is:

$$\gamma \equiv [1 - \exp(-\alpha f_p^{\text{max}} r_p B/C_{66} \phi_0)] \quad (6)$$

where α is an unknown numerical parameter, probably of the order of unity.

Equations (4)-(6) must be combined with a description of the magnetic field and temperature dependence of C_{66} (e.g., that of Brandt [4] used by Pruyboom et al. [1]):

$$C_{66} = (B_c^2/4\mu_0)(\kappa/\kappa_2)^2 b(1-0.29b)(1-b)^2 \quad (7)$$

where B_c is the thermodynamic critical field, μ_0 the permeability of free space, κ the Ginzburg-Landau parameter, κ_2 a Maki parameter, and b is the reduced field B/B_{c2} , and, of course, the critical magnetic fields are themselves functions of the reduced temperature T/T_c . Combining Eqs.

(4)-(7) yields the desired description of the flow stress (neglecting thermal activation) of the FLL for a given state of pinning, magnetic field, and temperature. This resulting value of τ_{flow} is then inserted in Eq. (3) to determine whether plastic shear or depinning from the strongest pins limits J_c . Assuming that plastic flow is the limiting process, Eq. (2) yields the critical current density J_{c0} (neglecting thermal activation):

$$J_{c0} B \equiv \frac{2\tau_{\text{flow}}}{\Lambda q} + (B/\phi_0)(gf_p)w^{\text{max}} \quad (8)$$

The first term in the right-hand-side is essentially that of Puyboom et al [1], but in this case we describe the plastic flow stress using Eqs. (4)-(7). The second term is a frictional term resulting from the weak pins and is presumably smaller than the first term.

Wördenweber and Abd-El-Hamed [5] have used the first term above, together with a statistical distribution of critical magnetic fields in the channels, to derive simulated voltage-current curves which are very similar to those observed for $\text{YBa}_2\text{Cu}_3\text{O}_7$. This model, however, does not explicitly include thermally-activated processes, which many researchers feel to be very significant in cuprate superconductors at "high temperature." Thus, in the following, we will derive expressions which are explicitly based on thermally-activated dislocation motion.

IV. THE CRITICAL CURRENT DENSITY, INCLUDING THERMAL ACTIVATION

If, as in the case of high- T_c cuprates, flux creep can be a significant process, then thermally-activated processes must be specifically included. These processes can include thermally-activated FLL dislocation motion, as well as the more-commonly-considered thermal depinning of the flux lines themselves. This can be accounted for in a straightforward manner in same rate-theory approach that is often used for metals and alloys [6]. We assume that FLL plasticity is the factor limiting J_c and neglect the frictional term of Eq. (8) for simplicity. The plastic shear strain rate $\dot{\epsilon}$ due to thermally activated dislocation motion is calculated from dislocation rate theory [6] in the low-temperature creep regime (i.e., neglecting dislocation motion opposite to the driving force and neglecting the effect of dislocation climb). This yields the density of flux-lines which are moving and their velocity, thus yielding a flux-motion-induced electric field E :

$$E = B(q\Lambda)(\delta\Lambda/\Lambda)\dot{\epsilon} \quad (9)$$

where $q\Lambda$ is the width of each moving "block" of FLL and $\delta\Lambda$ is the width of the interface where the plastic shear is occurring between the strongly-pinned fraction $(1-q)\Lambda$ and the weakly-pinned fraction $q\Lambda$. The shear strain rate is

$$\dot{\epsilon} \equiv msp^{1/2} (\phi_0/B)^{1/2} v \exp[-U_+ (\tau/\tau_{\text{flow}})/kT] \quad (10)$$

where m is the fraction of FLL dislocations which are mobile, s is the number of inter-dislocation distances which a depinned dislocation moves before being repinned (thus ms is a number of the order of unity), v is the dislocation vibration frequency, and U_+ (τ/τ_{flow}) is the stress-dependent activation energy for dislocation motion in the direction of the driving force, given approximately by $U_0(1-J/J_{c0})^n$, with $n \approx 3/2$. The exact field and temperature dependence of U_0 depends on the particular mechanism of the dislocation motion; however, several possibilities [7] yield U_0 of the form $\beta C_{66}(\phi_0/B)\ell_c$ where ℓ_c is the length of dislocation line which moves in an activated depinning event and β is a numerical constant. Combining Eqs. (9) and (10) yields the electric-field vs. current-density curve:

$$E = [msq(\delta\Lambda)v\rho^{1/2}B^{1/2}\phi_0^{1/2}]\exp[-\beta C_{66}(\phi_0/B)\ell_c(1-J/J_{c0})^n/kT] \quad (11)$$

Choosing an electric field criterion E_c , which defines the critical current density, yields the desired expression for J_c when thermally-activated dislocation motion is appreciable:

$$J_c/J_{c0} = 1 - \{kT(B/\phi_0)\ell_n[(msq)(\delta\Lambda)v\rho^{1/2}B^{1/2}\phi_0^{1/2}/E_c]/\beta C_{66}\ell_c\}^{1/n} \quad (12)$$

where, as discussed above, β and (msq) are numerical constants, $\delta\Lambda$ is the interface width, and v is the dislocation vibrational frequency.

V. CONCLUSIONS

A simple model was used to describe the state of stress in a pinned flux line lattice and a criterion for the occurrence of plastic deformation was derived. Expressions were also derived for the flow stress and dislocation density of the FLL and were combined to yield J_{c0} , the magnetic field and temperature dependent critical current density in the absence of thermal activated flux creep. The rate theory of plastic deformation was used to derive the electric-field vs. current-density relation and J_c to be expected when thermally-activated processes make the dominant contribution to FLL shear. These results extend the earlier work of Pruymboom et al. [1] and Wördenweber and Abd-El-Hamed [5] to include a more realistic description of the effect of pinning strengths and densities on the plastic shear strength of the flux line lattice, as well as to include thermal activation explicitly. These new results thus permit an exploration of how adding or strengthening pinning centers is likely to alter the J_c when FLL shear is limiting.

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