NMR data show no evidence of a transition down to 1.25 K. It is clearly of interest to extend the experiment to lower temperatures in order to determine whether ordering actually takes place or whether fluctuations in this 2D system reduce the transition temperature to zero.

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Threshold for Optically Induced Dislocation Glide in GaAs-AlGaAs Double Heterostructures: Degradation via a New Cooperative Phenomenon?

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We have observed a sharp threshold for the process of optically induced glide at which the velocity changes by more than a factor of 10^3 when the excitation intensity changes only 20%. This threshold is insensitive to doping and to the presence of a p-n junction. The effect is shown not to be related to recombination-enhanced motion or to local heating. An explanation in terms of the reduction of frictional forces by interaction with unrecombined carriers is offered.

It has been reported previously[1,2] that optical excitations of intensity in excess of 10^5 W cm^{-2} can lead to rapid (50 μm s^{-1}) dislocation glide in AlGaAs double-heterostructure lasers. The observations were made using photoluminescence topography. A laser wafer, without the normal capping layer of GaAs, is illuminated by the focused 647.1- and 616.4-nm radiation of a Kr^+ laser, to which the upper waveguide layer is transparent but the active layer is not. The active layer may be observed by the photoluminescence produced in it. Defects are apparent as dark regions because of the higher probability of nonradiative recombination events in their vicinity. Thus the motion of a gliding dislocation may be observed directly as a moving dark spot in the photoluminescence field.

The glide process does not occur at dislocations which were present during growth, but only at fresh dislocations. These are introduced by gently scratching the surface of the upper waveguide layer. On applying an optical excitation of sufficient duration, a dense dislocation network develops within a few microns of the scratch. Similar networks have been observed in studies of moderate-intensity 10.6-μm-laser damage in GaAs.[3]

A secondary stage of damage may now be produced if the optical excitation is sufficiently intense. Then we find that dislocation loops will glide away from the damaged area. This results in a portion of the dislocation threading from the bottom waveguide layer through the active layer and the upper waveguide layer to the surface.

This new threading dislocation may be made to glide in a ⟨110⟩ direction. As it does so, it lays down an increasing length of misfit dislocation. If one assumes the scratch punches out interstitial dislocations, then the extra half-planes may relieve the misfit between the active layer, which has a smaller lattice constant, and the underlying waveguide. Therefore, misfit dislocations are expected to occupy this interface. Transmission electron microscopy has been used to confirm this location.[4]

Dislocation glide in double-heterostructure lasers has also been reported under conditions of electrical excitation and simultaneous stress.[5,6] The same phenomenon has been observed in GaP light-emitting diodes.[7] In order to explain this remarkably enhanced dislocation velocity both local heating effects and recombination-enhanced motion have been suggested tentatively.[1,2,6] In this Letter we report additional observations of these effects. We find a sharp threshold of excitation intensity, above which the dislocation moves extremely rapidly. The nature of the change about the threshold is so great as to imply a critical phenomenon and to discount the previously suggested mechanisms. We provide both a description of the observations made on dislocation movement and a new tentative theory which may explain the sharp threshold we have observed. We conclude it is likely that some cooperative phenomenon is active in producing the observed critical phenomenon.

The materials used for the present study consisted of three layers grown on GaAs(100) substrates by liquid-phase epitaxy. These layers were the lower waveguide, the active layer, and the upper waveguide. The normal GaAs capping layer was omitted. Details of both the Al contents of the three layers and the doping of the layers are given in Table I.

The exciting radiation was focused to a spot whose diameter was 30 μm. The intensities used in this investigation are within the range of those at which semiconductor lasers are operated. (The absolute limit of optical flux which may be obtained from such a laser is that which causes catastrophic facet damage. This occurs when the emergent flux is greater than[8,9] 3×10^6 W/cm^2. The internal flux is greater than this.) It is also worth noting that the samples were not lasing. Although the local flux was above threshold, only a small fraction the total cavity was being pumped

As dislocations were seen to move in the irradiated region, the sample was tracked to keep the moving point central in the field.

The nature of the movement was complicated. At very high excitation levels, $\sim3\times10^5$ W cm^{-2}, the dislocation glides with a fairly uniform rate across the excited area. As the intensity is reduced a threshold region is reached where the dislocation velocity diminishes by at least three orders of magnitude for a 20% change in the exciting intensity. The motion of the dislocations also becomes irregular in this region. When the intensity is below threshold, the speed is 60 nm/s or less and is in a ⟨100⟩ rather than a ⟨110⟩ direction. This behavior of velocity as a function of intensity, as intensity is reduced, is shown in Fig. 1.

The growth in the ⟨100⟩ direction at low intensity has been investigated by transmission electron microscopy.[10] A complex structure is produced as a result of dislocation climb. Such structures are evidenced as dark line defects. Once the dislocation has grown a few microns by the climb mechanism, it will not glide again.

It has been proposed that the glide mechanism is thermally activated.[6] If this were the case, the dislocation velocity, v, would be given by

$$v = \nu b B \tau^m e^{-U/kT}, \quad (1)$$

where ν is an attempt frequency, b the Burgers vector of a dislocation, U the activation energy, τ the local stress, m an exponent in the range[11] from 1 to 2, and B is a constant. It has been estimated[12,13] that the temperature rise under

TABLE I. Sample parameters (layers numbered from substrate) and measured threshold intensities, I_t.

Wafer	I_t (10^5 W/cm^2)	Percentage Al (layers 1/2/3)	Carrier type, number/cm^3 (dopant)	Width (μm)
A	0.8	32/1/32	[*n*, 10^{15}(Sn)]/[*n*, 10^{18}(Te)]/[*n*, 10^{15}(Sn)]	3.1/0.7/0.9
B	1.4	41/1/41	[*n*, 10^{17}(Te)]/[*n*, 10^{17}(Sn)]/[*n*, 10^{16}(Sn)]	5.0/0.7/1.5
C	1.6	40/1/40	[*n*, 10^{17}(Te)]/[*n*, 10^{15}(—)]/[*p*, 10^{17}(Ge)]	4.5/0.5/0.8
D	2.0	40/1/40	[*n*, 10^{17}(Te)]/[*p*, 10^{18}(Ge)]/[*p*, 10^{17}(Ge)]	3.6/0.6/0.6

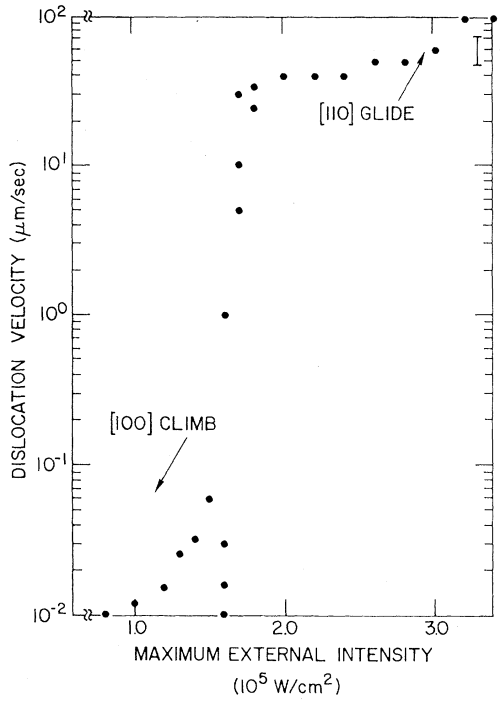

FIG. 1. Semilog plot of velocity of photoluminesence spot due to threading dislocation versus external intensity of 647.1- and 676.4-nm laser light. External intensity is controlled to about 10%; internal intensity may vary more due to variations in sample surface. At the highest intensities, velocity uncertainty becomes as large as indicated. Just below threshold the scatter in data is large, but it is difficult to separate any glide component of the predominant climb motion. Only the highest velocity values observed for several dislocations are shown below threshold whereas several measurements are shown for each intensity in the threshold region.

these experimental conditions may be as great as 200°C. Three scratched samples were heated to temperatures of 250, 400, and 500°C, respectively, for periods of 1 min. No dislocation glide was observed to have taken place. Upon optical excitation, the samples behaved exactly like unannealed samples. In Eq. (1) τ would not be expected to be more than linear in the intensity I, so that one would not expect the preexponential to be more than quadratic in I. Note that the effect of the nonlocal heating due to the absorption of the light only in the GaAs is to *reduce* the misfit stress that drives the glide. This is because the GaAs has the smaller lattice constant and expands as it is heated. A quantitative analysis of the nonlocal heating shows that it reduces the misfit stress less than 10% in our experiment.

A further characteristic of thermally activated glide is that it is strongly dependent upon the doping of the crystal.[14] Therefore, the different compositions of the structures listed in Table I ought to be expected to show large differences in threshold levels or velocities. Neither is observed.

The other previously proposed mechanism for glide is that energy from recombining electron-hole pairs[15] enhances the motion of the dislocation.[1,2] In this case, one may write

$$v = \nu b(e^{-U/kT} + Ale^{-(U-R)/kT}), \quad (2)$$

where A is a constant and R is the energy donated by a recombination event. According to this hypothesis v would simply be linear in I.

We will now present a tentative explanation for optically induced glide which we find consistent with our observations. It is proposed that the threading dislocation glides locally when the instantaneous concentration, N, of unrecombined electron-hole pairs in its vicinity fluctuates from its mean value, M, to such a high value, N_c, that the frictional forces of the lattice are weakened below the level required to prevent the dislocation from gliding under its resolved stress. We recall[16-21] that the excitation of an electron from the bonding states near the top of the valence band to the antibonding states of the conduction band severely reduces the frequency of transverse acoustic lattice modes and the shear strength. When $N_c \gg M$, the probability that an instantaneous fluctuation in N will drive it above N_c is a very rapidly increasing function of M, and thus of I, for virtually all plausible models of the statistical distribution.

We illustrate our proposal with an admittedly oversimplified model in which the number, n, of electron-hole pairs in a critical volume, V, about the dislocation obeys the Poisson distribution,

$$P(n) = m^n e^{-m}/n!, \quad (3)$$

where m is the mean number of pairs, and assume

$$v = \nu b \textstyle\sum_n P(n) e^{-U(n,T,\tau)/kT}, \quad (4)$$

where the activation energy, U, is now taken to be a function of n, of temperature, and of strain. Note that even at the highest velocities the dislocation jumps only very infrequently (its velocity is much less than the speed of sound). Taking $\nu = 5\times10^{12}$/sec and $b = 0.4$ nm, one would have $\nu b = 2.0\times10^3$ m/sec, a factor of 10^7 faster than the

highest velocity observed above threshold and at least 10^{10} faster than below. With a 1-ns lifetime, M is of order 10^{19} cm^{-3} in the bulk of the active layer at threshold. Because of recombination at the dislocation, let us suppose it is 5×10^{18} in the region V about the dislocation. We may also suppose V has a radius of 1.0 nm and a length of 1.0 μm so that its volume is 3×10^{-18} cm^3 and therefore $m=15$. If we further suppose that the critical number $n=45$, the intrinsic value at the maximal melting point, we find $P(45)=2.1\times10^{-10}$. Now if the intensity and thus m is increased by 20% to 18, we find that $P(45)=7.9\times10^{-7}$, an increase of more than three orders of magnitude. While this choice of suppositions serves to produce from Eq. (4) a predicted onset of the glide consistent with the lower bound for the sharpness of the observed threshold, it is obvious that a number of other choices of approximations and parameters would also serve the purpose. We suspect that the threshold is actually much sharper than this lower bound. (Observation of lower rates of glide is obscured by the dislocation climb.) The threshold would be sharper if a cooperative phenomenon among the unrecombined carriers operates to cause their concentration to increase more than linearly with I near the critical intensity. Possible cooperative phenomena might include a saturation of the recombination channels about the dislocation or a polarization of the dislocation which might attract carriers from the bulk or strain or field gradients produced by the carriers that might separate electrons and holes spatially to prevent their recombination.

The difficulty in causing the dislocation to glide once it has climbed may be explained either by the reduction in carrier lifetime, and thus in M and m, in the vicinity of a climb network due to recombination at it, or by a pinning effect of the climb network upon the threading dislocation. The inability to glide grown-in dislocations may similarly be explained by the recombination due to defects which congregate about dislocations at growth temperatures, or by a difference in the geometries of the two dislocations. We note that the maximum n- or p-type doping in our samples was about 1×10^{18} cm^{-3}, which is a factor of 15 less than our assumed value for N_c, so that the observed insensitivity of threshold to doping seems also to be consistent. Moreover, the intrinsic carrier concentration reaches N_c only at the congruent melting point of the material; so the lack of thermally activated migration at 500°C is also consistent.

There remain three questions: (a) What limits the velocity of the dislocation above the threshold; (b) why does the portion threading through the top layer, where it does not relieve any misfit stress, glide along with the portion at the lower GaAs-alloy interface; and (c) is there any relation between this phenomenon and the increased plasticity found when metals enter the superconducting state[22]?

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Transition from Pauli Paramagnetism to Band Ferromagnetism in Very Thin Ni Films

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Ni, Co, and Fe films of a few atomic layers are condensed in UHV at 10 K on metallic substrates. Anomalous-Hall-effect measurements are used to determine the magnetization and the susceptibility of the films. Ni films with a thickness smaller than two atomic layers possess no magnetic moment but show an enhanced susceptibility. Between two and three atomic layers of Ni the susceptibility diverges and thicker films possess a magnetic moment. For Co and Fe, the first monolayers already shows a magnetic moment.

Thin magnetic films or surfaces of magnetic materials are modified in their magnetic properties. Of particular interest is the existence of so called "dead layers" in which the ferromagnetism and even the magnetic moments are suppressed. A discussion of "dead layers" has to distinguish two completely different situations and problems: (i) dead layers at the surface of a bulk magnetic material; (ii) dead layers in a thin film of a magnetic metal on top of a nonmagnetic one.

The existence of dead layers in the surface of Fe, Co, and Ni has been experimentally disproved. Mössbauer measurements of ferromagnetic Co[1] demonstrated that the surface atoms possess an unchanged magnetic moment. Spin-polarized electrons[2,3] did not find dead surface layers for Fe, Co, and Ni.

The history of magnetism of thin films is rather eventful[4-12] (see Gradmann[13]) and demonstrates in particular the important role played by sample preparation conditions. For the most promising candidate, Ni, Liebermann *et al.*[9] extrapolated from their data the existence of two magnetically passive monolayers at liquid He temperature. These films were formed by electroplating from an aqueous solution of Ni salts. Their results have been contradicted by Pierce and Siegmann[10] who condensed thin Ni films onto Cu and measured the spin polarization of photoemitted electrons. They concluded from their results that ferromagnetism already occurs in Ni films with one to two atomic layers.

In the present work we examine experimentally the basic question: At which film thickness do Ni, Co, and Fe *atoms* begin to possess a magnetic moment? How does the occurrence of the magnetic moment happen, i.e., with a divergence of the susceptibility as the theory of band magnetism suggests or by formation of magnetic clusters?

In the investigation of the magnetic properties of thin films of a few atomic layers one has to overcome a number of technical difficulties: (i) avoiding oxidation and impurities, (ii) diffusion into the metallic substrate, (iii) the formation of islands, (iv) the exact determination of the thickness of the films, and (v) the sensitivity of the equipment.

I detect the magnetic moments of the Ni (Co, Fe) and their ordering by a rather unconventional but extremely sensitive method, the anomalous Hall effect. Atoms with a magnetic moment scatter the conduction electrons asymmetrically. This asymmetric scattering yields a voltage perpendicular to the current, like the Hall effect, and is called the "anomalous Hall effect." The anomalous Hall resistivity is proportional to the magnetization of the sample[14] so that

$$\rho_{xy}^{\mathrm{an}} = R_S J_z,$$

where R_S is the anomalous Hall constant and J_z is the z component of the magnetization; the magnetic field is in the z direction. It is additive to the normal Hall resistivity $R_0 B_z$.

In recent investigations[15] I demonstrated that in amorphous ferromagnets this anomalous Hall effect is a factor of 100–1000 larger than the

