

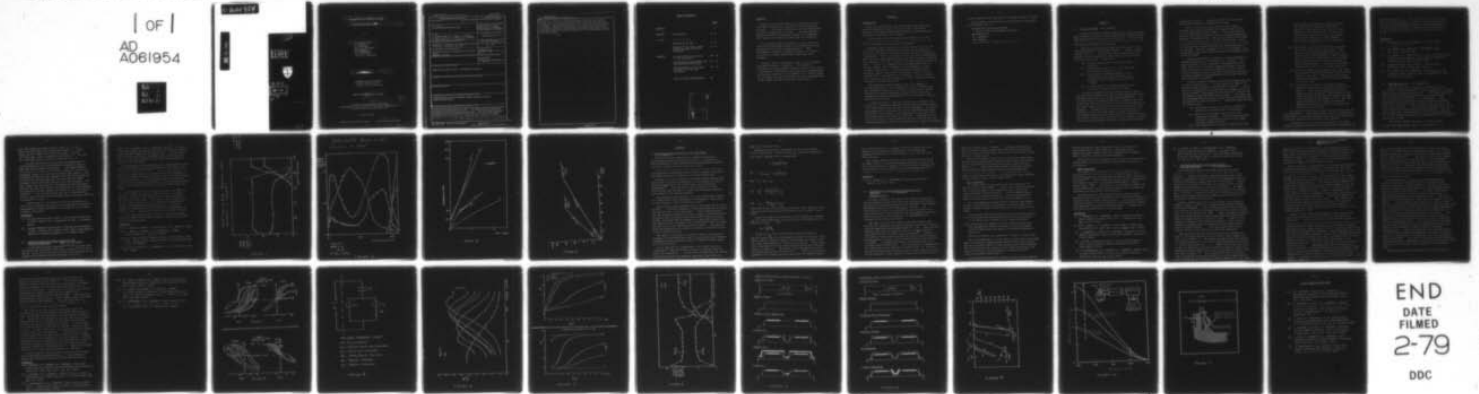
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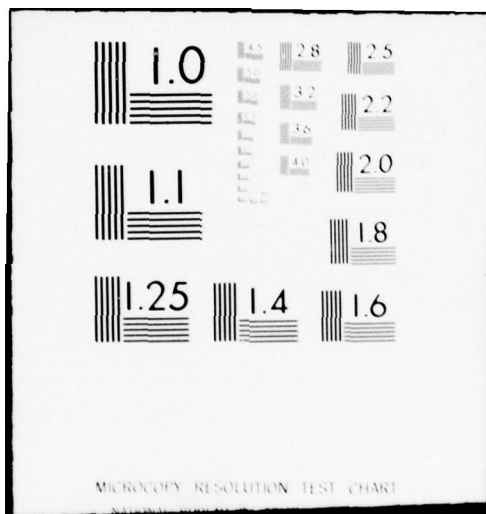
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6 NEW PASSIVATION METHODS OF GaAs

9 FINAL TECHNICAL REPORT 28 Jan-27 Sep

by

10 B./BAYRAKTAROGLU
P.A./BREEZE
A./COLQUHOUN
A.F.A.B./EL-SAFTI
S.J./HANNAH
*H.L. HARTNAGEL
B. LIVINGSTONE
MRS. A.J. AL-ADHAMY

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UNIVERSITY OF NEWCASTLE-UPON-TYNE, ^{univ. - (Engl. I)}
Department of Electrical and Electronic Engineering,
Newcastle upon Tyne, NE1 7RU, U.K.

* Grant Holder

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Further detailed ESCA studies are reported on native oxide MOS structures with and without the incorporation of Al_2O_3 . The anodic oxidation of InP was investigated from an electrolyte point of view. The oxide growth mechanism involving electrolytic oxidation of sandwiches of Al and Ge were studied by using GaAs as a substrate.

An attempt was made to formulate a simple electrical model of GaAs MOS diodes which fits reasonably well with experimental C-V and G-V characteristics. The development of further GaAs MOS transistors is reported, in particular charge storage MAOS transistors were measured.

Finally, the d.c. emission of white light from thin GaAs MOS structures is reported. The spectral distribution of this emission seems to indicate that either a substantial contribution is obtained by the recombination of electrons and holes in the amorphous oxide (whose energy gap is larger than that of GaAs) or by direct recombination of hot charge carriers in the space charge layer.

CHAPTER I

INTRODUCTION

This now represents the last report under this grant from Newcastle upon Tyne, since I have accepted a professorial position at the Hochfrequenztechnik Institut of the Technische Hochschule in Darmstadt, West Germany.

I am grateful to the European Research Office of the U.S. Army that I shall be able to continue my passivation work there with a continuation of this grant in Darmstadt.

Although several people have left this University since the news of my impending departure from Newcastle was announced, activities continued to be strong right to the end of my period in Newcastle. The first person to leave was Dr. B. Bayraktaroglu who joined the Wright-Patterson Air Force Base Avionics Laboratory for a period of two years in order to continue passivation work on GaAs. Then Dr. A. Colquhoun commenced employment with the Germany company, AEG-Telefunken, on the development of GaAs microwave transistors. Two Ph.D students completed their doctorate studies, namely Dr. T. Hutchinson and Mr. A.F.A.B. El-Safti whose oral examination is imminent. Dr. P.A. Breeze will leave us at the end of August this year in order to take up a position at Oxford University.

The remaining research students here, S.J. Hannah, B. Livingstone and Mrs. A.J. Al-Adhamy will in fact continue work here in Newcastle with Dr. C.A. Walley taking over as internal supervisor and myself remaining in close contact with these candidates by both correspondence and visits in order to help in supervising their work.

I should now like to take this opportunity of thanking the University of Newcastle upon Tyne for the generous help provided to me, particularly also with my research programme. The general environment among colleagues, both in this department and other departments, is certainly a very stimulating one and many first class experimental facilities are available. Gratitude is also expressed to those in this department who were involved with the research management, such as my secretary Mrs. B. Robson and members

of the technical staff such as Mr. T. Mitchell and Mr. E. Baker.

If any reader is interested in contacting me in the future my new address will be:-

Institut für Hochfrequenztechnik

an der Technische Hochschule Darmstadt,

Merckstrasse 25,

61 - Darmstadt,

West Germany.

Telephone number:- Darmstadt 06151 16 21 62.

CHAPTER 2

2.1 E.S.C.A. STUDIES (P.A. BREEZE)

In the last report (January 1978) E.S.C.A. depth profiling data, obtained from native oxide films on GaAs (anodised in the $(\text{NH}_4) \text{H}_2\text{PO}_4$ /glycol electrolyte) was discussed. It was shown that E.S.C.A. could provide valuable data relating to both bulk and interfacial regions of such films.

This work has now been extended to films grown in a 3% tartaric acid/glycol electrolyte (AGW)⁽¹⁾ Figure 1 shows a typical profile. The general features observed are very similar to those obtaining to films grown in the previously discussed electrolyte. These are:-

- (i) A surface composition differing from that in the bulk.
- (ii) A bulk As/Ga ratio of 0.5 - 0.6 (neglecting any selective etching effects).
- (iii) An oxide rich in Gallium adjacent to the semiconductor, the As_2O_3 interface region extending further into the oxide than that of Ga_2O_3 .
- (iv) A region close to the GaAs substrate where As starts to appear before Ga of GaAs.

A detailed comparison of the profile shown in Figure 1 with that for a similar film grown in $(\text{NH}_4) \text{H}_2\text{PO}_4$ /glycol shows the major difference to be in the surface composition. In the former case the surface As/Ga ratio is greater than that in the bulk indicating a relatively As rich surface whereas in the latter case the reverse is true, a Ga rich surface being obtained. Features (ii), (iii) and (iv) listed above are remarkably similar for films grown in either electrolyte. One conclusion which may be drawn from this comparison is that the surface composition of the oxide film is strongly influenced by the electrolyte whereas the bulk and oxide/GaAs interface

compositions are not. It seems probable that the growth mechanism is dominant in these last two cases.

Use of a non-linear, least squares curve fitting programme has made it possible to fit peaks to some of the E.S.C.A. spectra from these studies. These have indicated that the additional As observed close to the GaAs substrate is probably in elemental form. Due to the overlap of the peaks due to elemental As and As in Gallium Arsenide ($\Delta V = 0.5\text{eV}$) it is difficult to distinguish between the two types of As since we had previously thought that only the Gallium Arsenide As was present. However, assignment of this to elemental As is in agreement with recent results obtained from thermally grown oxides on GaAs.⁽²⁾

Using the data shown in Figure 1, we have attempted to estimate the amount of As present at the interface. The figure obtained, 2×10^{15} atoms/cm², (integrated over the whole width of the interface), probably represents an upper limit and could be an order of magnitude or more in error. It does indicate, however, that there may well be sufficient As atoms available to be responsible for the high interface state densities reported.⁽³⁾

We have recently examined an anodically grown oxide film on GaAs incorporating Al. A GaAs sample with 200Å of Al deposited onto the surface was anodised at 1mA cm⁻². After all the Al was oxidised, as indicated by a change in slope of the V/t curve,⁽⁴⁾ anodisation was continued for a further 30V to produce an expected 600Å of native oxide. The sample was then analysed using the E.S.C.A. depth profiling technique and the results are shown in Figure 2. There are several features of interest here.

- (i) The position of the Al 2p peak in the E.S.C.A. spectra was consistent with it being Al₂O₃.
- (ii) The native oxide grows both on top of and underneath the Al₂O₃. This was to be expected if growth is due to both metal ion and oxygen migration through the film.
- (iii) The composition of the native oxide film on top

of the Al_2O_3 was different to that underneath. The composition of the native oxide film adjacent to the substrate was similar to that observed in the absence of Al. However the native oxide film at the surface was significantly more As deficient. This suggests that surface interaction with the electrolyte may be very important.

- (iv) The amount of native oxide below the Al_2O_3 was much greater than that on the surface. This may mean that oxygen migration inwards is the faster process during growth.⁽⁵⁾ However account must also be taken of dissolution at the surface which could significantly affect the amount of native oxide observed. As noted in (iii) above surface effects seem to be of great importance.
- (v) The native oxide/ Al_2O_3 interfaces were very broad. In particular Al_2O_3 was observed all the way to the surface of the film. This may mean that during anodic growth migrating metal ions can interchange with lattice ions.
- (vi) The nature of the native oxide/GaAs interface appeared very similar to that observed with no Al present. This is quite consistent with the suggested growth mechanism⁽⁵⁾ where growth in this region is due to oxygen migration inwards and ought to be unaffected by the presence of Al. The same argument would apply to explain the observed composition of the native oxide film beneath the Al_2O_3 noted in (iii) above.

The effects of variation in growth current density have not been investigated. Such studies should indicate the importance or otherwise of surface dissolution effects since at lower growth rates these should become relatively more important.

In conclusion, we feel that our knowledge of the anodic oxidation process for GaAs has been significantly advanced by the

application of E.S.C.A. The use of this type of analytical tool is extremely important in order to gain a further understanding of microelectronic devices and fabrication processes. We hope that we have helped to demonstrate the applicability of one technique which has not yet been widely used in this field.

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2.2 ANODISATION OF InP (B. LIVINGSTONE)

Following work done previously by co-workers on the anodisation of indium phosphide,⁽¹⁾ indium phosphide was anodised using various electrolyte compositions. The basic electrolyte solution was 0.3M (NH₄) H₂PO₄ which was mixed with propan 1 : 2 diol (glycol) in the ratio of 1 : 2 respectively. This electrolyte was used previously in the anodisation of gallium arsenide⁽²⁾ and yielded a very low dissolution rate for a less concentrated solution. The electrolyte composition was varied by the addition of either glycol or basic solution. All the anodisations were performed using a constant current supply on 3 - 8 x 10¹⁷ n-type polycrystalline indium phosphide at a constant temperature of 12°C.

The results obtained are shown in Figure 3 where:

- (1) was 0.3M (NH₄) H₂PO₄ + glycol in the ratio of 1 : 2,

more $(\text{NH}_4) \text{H}_2\text{PO}_4$ was added to give a ratio of 1 : 1 (2), glycol was then added to bring the ratio back to 1 : 2 (3), glycol was then added until the ratio became 1 : 10 (4), (5) 50% H_3PO_4 + water and glycol in the ratio of 1 : 10, and (6) 0.02M $(\text{NH}_4) \text{H}_2\text{PO}_4$ + glycol in the ratio of 1 : 2.

As these and previous results show the anodisation of indium phosphide is not as well defined a process as that of GaAs when using aqueous electrolytes. The oxide tended to break down at low voltages with bubbles forming on the surface, presumably oxygen, for low ratios of glycol and solution. The overpotential-time plot also showed a very non-linear growth for low ratios but which became more linear and grew to larger voltages without break down when the ratio was increased. The most linear trace was obtained with phosphoric acid and glycol solution but this also departed from linearity for large voltages. The dissolution current density was found to be in the region of $100\mu\text{A}/\text{cm}^2$. This is much larger than that of GaAs but this would be expected with aqueous solutions due to the high solubility of phosphorus in water.

Further work is to be carried out on the anodisation of InP to try and improve the growth of the oxide and on electrical measurements to obtain information on the quality of the oxides formed by this method.

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2.3 STUDIES OF THE OXIDE GROWTH MECHANISM USING THIN FILM STRUCTURES (MRS. A.J. AL-ADHAMY)

It has been previously reported^{(1),(2),(3)} that both metal and oxygen ions are mobile during anodic oxidation of Al, Ta, W and Nb; if the metal ions alone are mobile the new oxide will be formed at the oxide/electrolyte interface on top of a

tagged layer (a marker layer of immobile ions)⁽²⁾; if oxygen is the only mobile species the new oxide will form at the metal/oxide interface underneath the tagged layer. It was found that the final location of the marker atoms within the oxide layer in some cases depended on the current density and the nature of the electrolyte.

The mechanism of oxide growth has been studied here by using thin film structures on GaAs consisting of a thin layer of Ge covered by a thin film of Al⁽⁴⁾. At high current densities the field is high enough to transport the Ge ions through the Al₂O₃ to the oxide electrolyte interface where it then dissolves in the electrolyte. At low current densities the field is insufficient to transport all the Ge to the electrolyte and hence some of the Ge may be oxidised by oxygen migrating across the Al₂O₃.

Figure 4 shows the V-t graph for anodisation of 100Å of Al^o on top of 100Å of Ga, both on a GaAs substrate, at current densities of 50μAcm⁻² and 100μAcm⁻² in the AGW electrolyte. The plots show two stages. The first is due to anodisation of Al, and the second is due to Ge anodisation which starts, at both current densities, after all the Al is anodised. Growth was stopped after the Ge anodisation had continued for about 10 volts. Beyond this Ge dissolution appears to start.

In the near future, Auger electron spectroscopy combined with ion etching, will be used to analyse samples of this type to obtain depth profiles.

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500 Å OXIDE ON GaAs. Anodisation in A.G.W.
 ELECTROLYTE AT 200 μ A cm^{-2} .

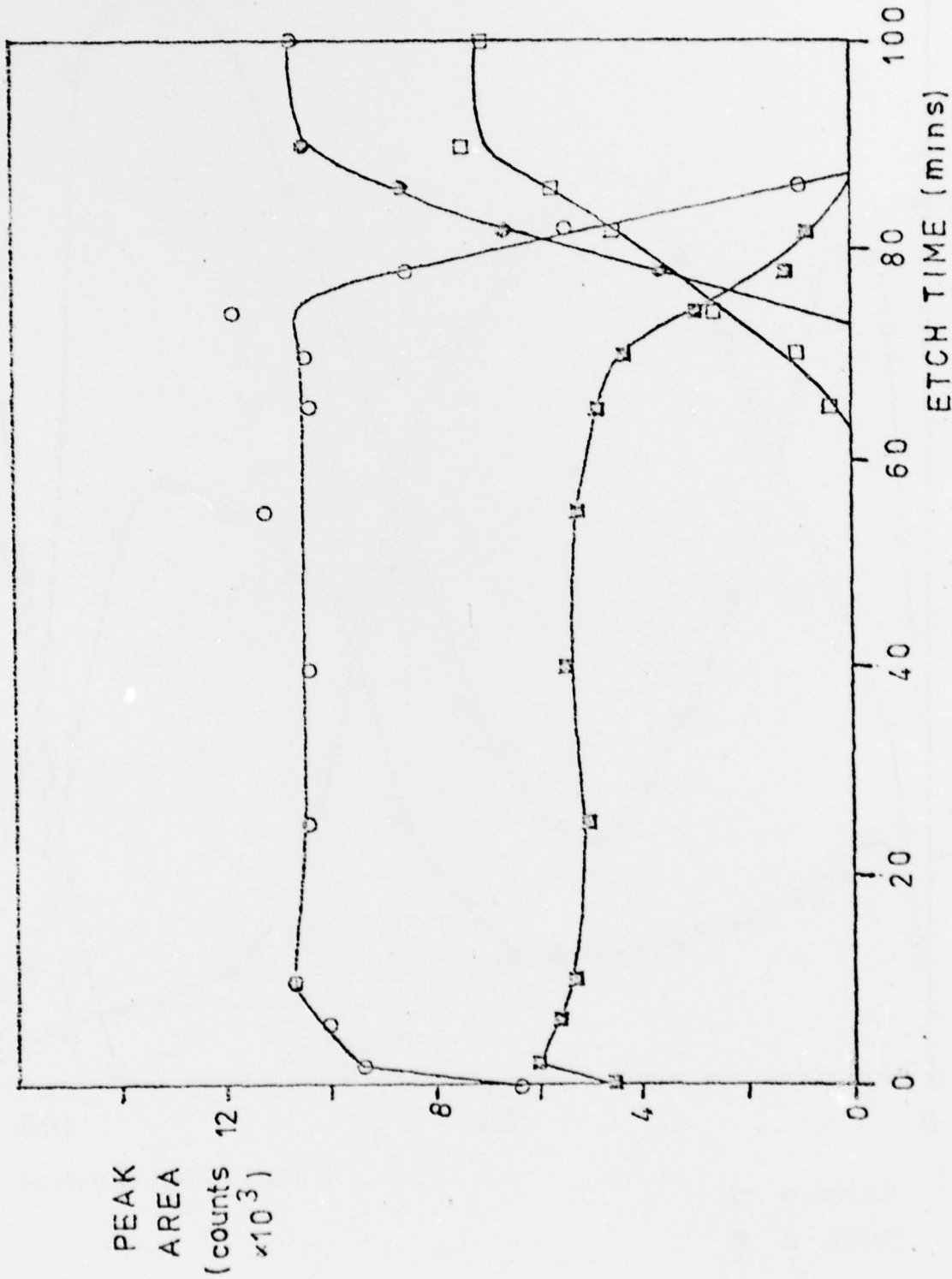


FIGURE 1

GaAs + 200 Å AL. ANODISED IN A.C.W.

ELECTROLYTE AT 1mA CM⁻².

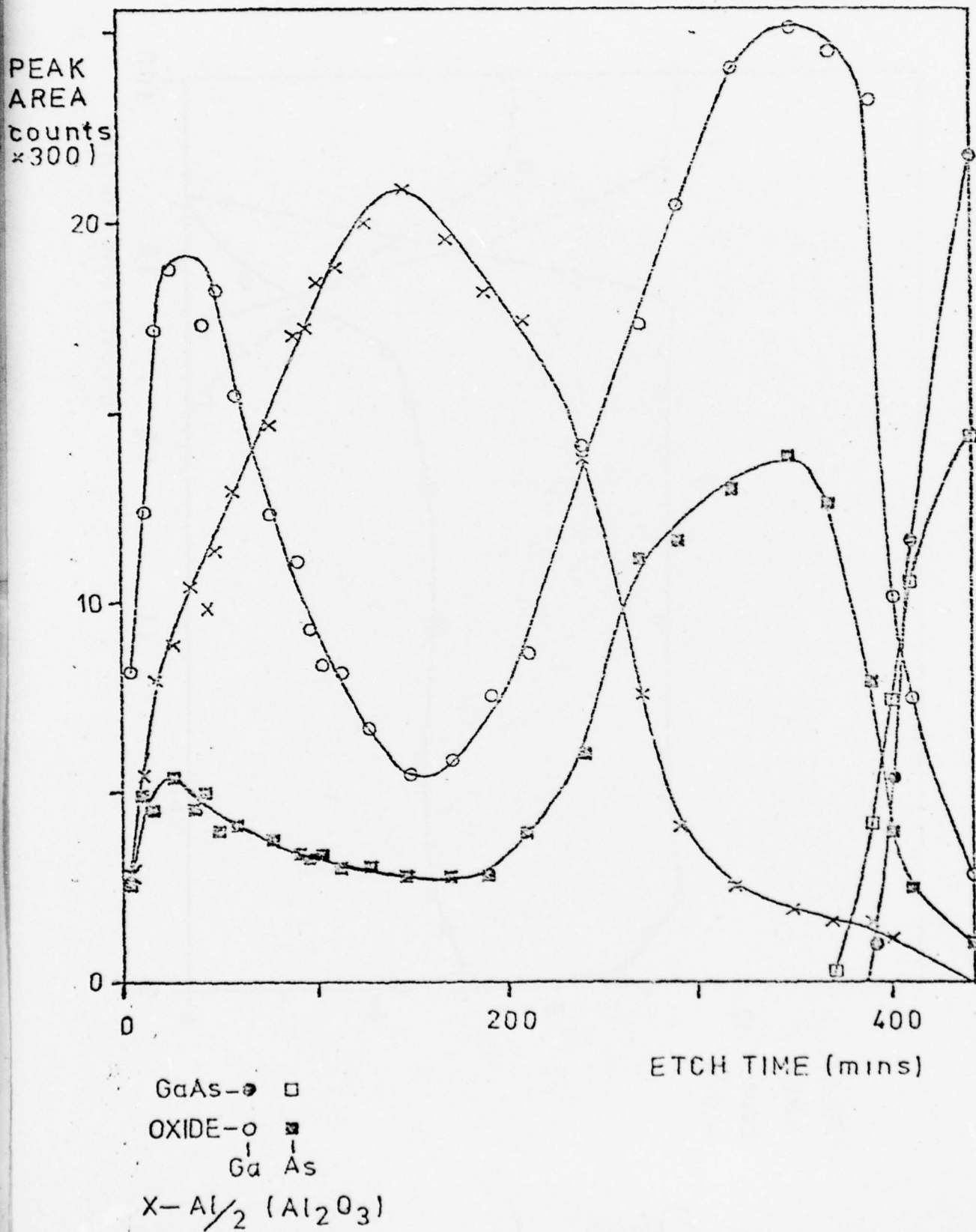


FIGURE 2

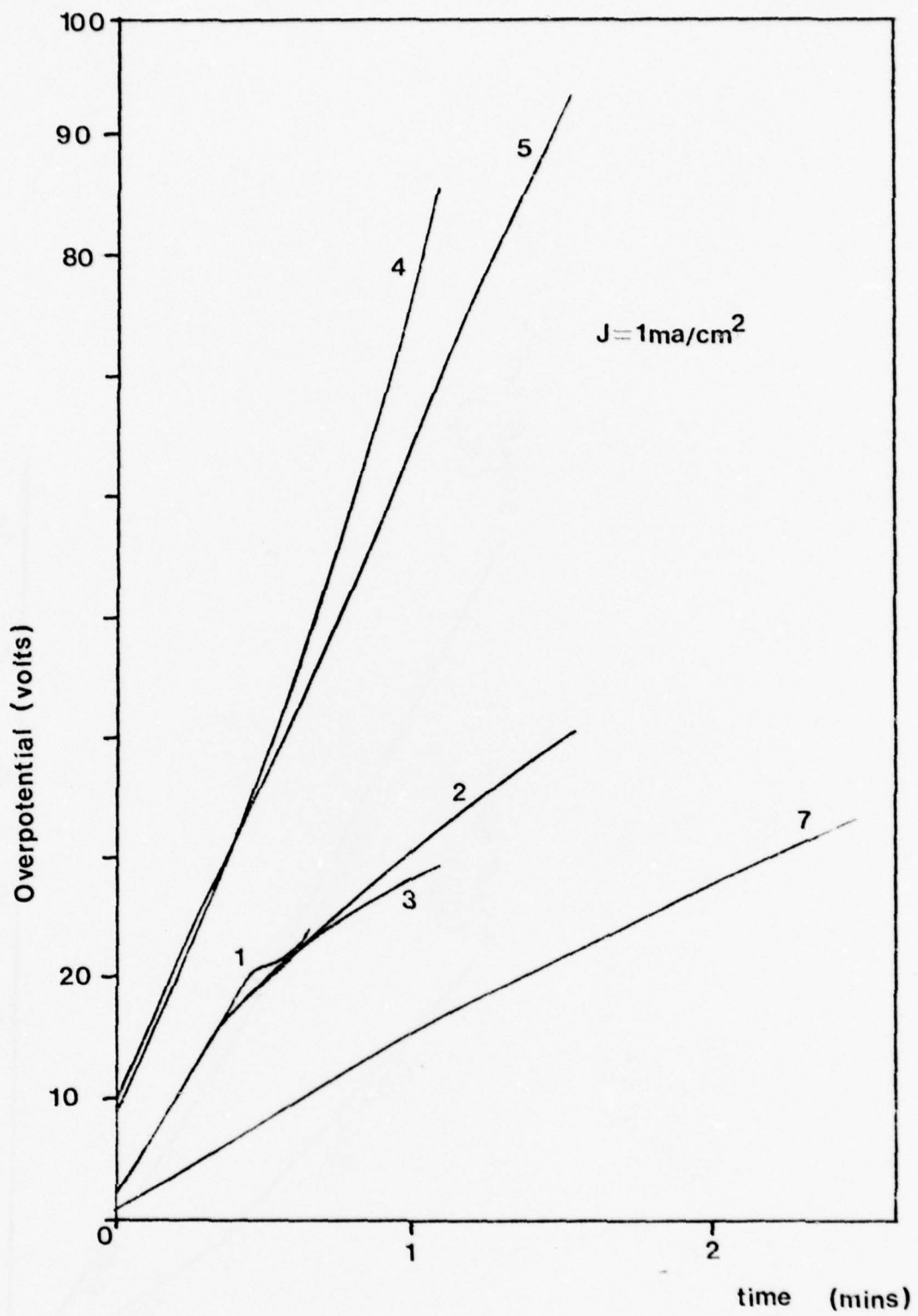


FIGURE 3

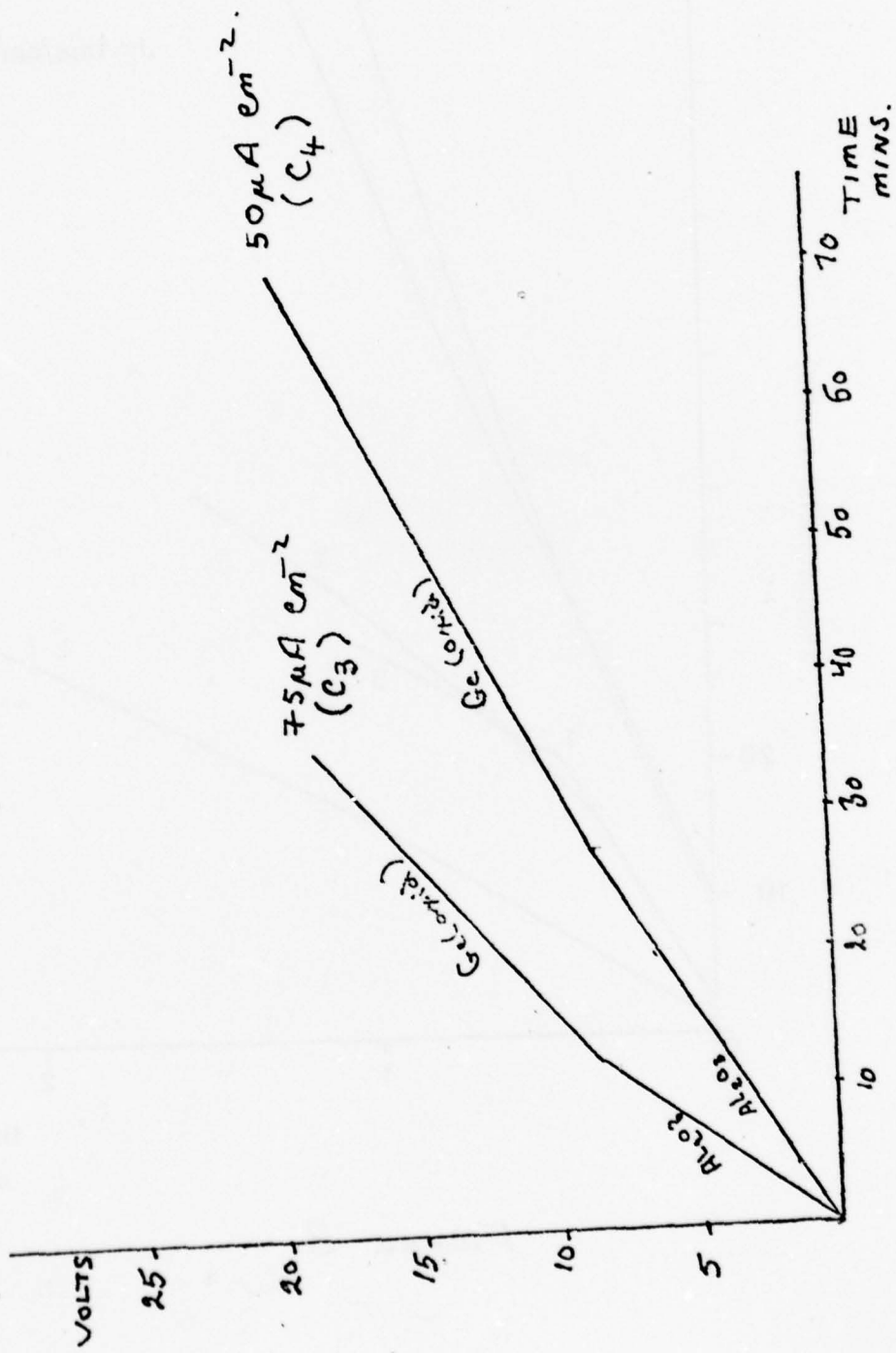


FIGURE 4

CHAPTER 3

3.1 C-V and G-V characteristics of GaAs MOS diodes (S. J. HANNAH)

In order to better appreciate the problems involved in producing good surface passivation of GaAs it is desirable to have a mathematical model which describes the presently observed electrical characteristics of GaAs MIS devices.

Silicon MIS technologists have successfully used capacitance and conductance measurements on MOS diodes to obtain information regarding surface state densities and surface potentials.

However, C-V curves obtained on GaAs MIS diodes incorporating native oxides on deposited insulators do not exhibit the type of behaviour predicted by the silicon model of Nicollian and Goetzberger.⁽¹⁾ This has led to some confusion as to the nature of the GaAs interface, in particular regarding interface state densities and surface potential.

An attempt has been made here to fit the experimental results to a simple model which incorporates surface state capacitance in a way similar to that of Nicollian et al.⁽¹⁾

The particular case considered here is for anodic native oxide on GaAs. C-V curves obtained for such MOS diodes are in Figures 1(b) and 2(b) for n- and p-type GaAs. These characteristics show the frequency dispersion reported by many research institutes. It appears that the p-type diodes behave better than the n-type. However, low temperature measurements for the diodes exhibit the same characteristic lowering of the "accumulation" level to almost the same level as the depletion/inversion level. Obviously at these low temperatures the surface potential is fixed for a large change in applied voltage.

The model of Nicollian et al predicts that if the measuring frequency is high enough the effect of the surface states becomes negligible and the equivalent circuit of Figure 3 becomes simply C_o in series with C_D .

For n-type GaAs the "accumulation" level falls flat at a frequency of ~ 1 MHz and further increase in frequency does not affect the capacitance. It can now be assumed that the surface states do not affect the measured capacitance (C_m) and C_m is

simply C_o in series with C_D .

Assuming now the simple equivalent circuit in an analysis of this circuit gives a condition that G_m/ω (measured conductance over angular frequency) has a maximum when

$$\omega = \frac{1}{R_T (C_T + C_o)}$$

and $G_m/\omega|_{\omega \text{ max}} = \frac{C_o^2}{2 (C_T + C_o)}$

when $C_T = C_D + C_{SS}'$

and $\frac{1}{R_T} = \frac{R_{SS} \omega^2 C_{SS}^2}{1 + R_{SS}^2 \omega^2 C_{SS}^2}$

and $C_{SS}' = \frac{C_{SS}}{1 + R_{SS}^2 \omega^2 C_{SS}^2}$

from which values of C_{SS} and R_{SS} the surface state capacitance and its associated capture/emission resistance may be found if C_D is known.

C_D may be found by measuring C_m at high frequencies, assuming C_{ox} , the oxide capacitance is given by C_m at low frequencies and high stress bias. Then:

$$C_D = \frac{C_o C_m}{C_o - C_m}$$

Values of G_m/ω were obtained using a Boonton Bridge 75C and the results for an n-type sample did indeed yield peaks, using these results as shown in Figure 4, and assuming C_D constant over the range investigated gave values for C_{SS} and R_{SS} . These values were then used to feed an Ecap program containing the equivalent circuit of Figure 3. The output of the Ecap program was used to predict the C-V curve for the diode. Figure 5(a) shows the experimental C-V curve and Figure 5(b) shows the computed result.

It is seen that there is a general agreement, inaccuracies being thought to be mainly due to calibration of the bridge and inadequate information at high frequencies, as it is this value which gives C_D .

The results indicate that the surface potential of the diode is almost fixed at a value near mid-gap and the C-V curve is not due to the variation of charges in the semiconductor surface, but is due mainly to the response of the surface states (which are causing the surface potential "pinning") with frequency.

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3.2 TECHNOLOGICAL DEVELOPMENTS FOR INTEGRATED GaAs MOS LOGIC (A. COLQUHOUN) Oxidation System

The anodic oxidation system has been studied in some detail. Various different electrolytes have been assessed, with the most promising two electrolytes being the AGW solution⁽¹⁾ and ammonium dihydrogen phosphate (0.02M $(NH_4) H_2PO_4$: Glycol 1 : 2). Electrically the oxides grown from these two electrolytes were rather similar but it was found that the $(NH_4) H_2PO_4$ electrolyte tends to form pores more easily and that phosphorus can be incorporated into the grown oxides.

AGW seems to be the most useful electrolyte found so far, with the best conditions for growth being at a low current density with the temperature maintained at about $0^\circ - 5^\circ C$. The electrical interface improves with annealing with the optimum conditions being about 10 mins. at $350^\circ C$ in N_2 ambient gas.⁽²⁾ Analysis of the oxide has been carried out using both ESCA and AUGER analysis and has shown a reasonably sharp interface between oxide and GaAs ($\sim 60-80\text{\AA}$). A typical ESCA profile is shown in Figure 6 for an oxide nominally 900\AA thick. From this profile it appears that the oxide close to the interface is arsenic rich, i.e. the oxide must be non-stoichiometric close to the interface and this may well be the reason for a large density of traps close to the interface

which our results seem to indicate. Variation of the growth current density and post growth treatments (such as annealing) have been used to try and adjust this profile, but so far we have had only limited success.

Growth at low current densities was found to be advantageous but presented a problem due to the dissolution rate of the oxide into the electrolyte which would then become faster than the growth rate. This problem was somewhat overcome by coating the surface with aluminium and then anodising through this aluminium to the GaAs surface. The dissolution rate of the Al_2O_3 was much lower than that of the GaAs native oxide and thus allowed growth at lower current densities.⁽³⁾

MOS transistors

The anodic oxide has been used as the gate insulator for various types of GaAs MOSFETs. Firstly in n channel inversion type transistors where it was found that only a very small transconductance could be achieved. This is probably due to trapping of any inversion charge at or close to the interface. Results from this type of device have been published in (4).

Due to the problems encountered with inversion type MOSFETs, efforts were concentrated on depletion type devices with the current through a channel being modulated by the depletion layer width.

Two different fabrication methods were used for these devices

- (a) a method using the anodisation process to etch the channel of the device and to produce a self-aligned structure,
- (b) a method where the channel is first chemically etched and then subsequently anodised to produce a non self-aligned structure which should be capable of increased enhancement operation.

Methods (a) and (b) are illustrated in Figure 7 respectively.

The transconductances of these depletion type devices have been quite high and have approached the theoretically expected values for the particular values of oxide thickness which were employed. This indicates that the depletion layer width and therefore the surface potential is being modulated as expected from the capacitance/voltage measurements.

Some microwave performance of these devices has been reported,

however the smallest gate structures that we have made were around 6 μ m which is rather large for a microwave device. Smaller structures would require more sophisticated equipment than is presently available in the laboratory.

Further work in this direction should be and is being directed towards double oxide gate structures and smaller geometries of the devices.

MAOS transistors

The search for a better anodic oxide using double oxide structures lead to the discovery of a charge storage effect at the interface between the Al₂O₃ and GaAs native oxide - both grown anodically.⁽³⁾ This effect has been used for the construction of a memory transistor - termed the MAOSFET (Metal - Al₂O₃ - native oxide - FET). This device is a non-volatile memory like silicon MNOS memory devices, but has proved to have considerable advantages over equivalent silicon devices.

The devices have a long charge retention time and can be operated without switch provided that the gate oxide field does not increase above a certain critical value. The system should be radiation hard due to the wide band gaps and final devices should have both fast read and fast switching times. Further work in this direction should prove to be rewarding.

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3.3 NON-DESTRUCTIVE WHITE LIGHT EMISSION FROM
GaAs MOS STRUCTURES (B. BAYRAKTAROGLU)

For field effect devices using MOS structures the oxide thickness is usually large enough to prevent excessive leakage currents. In GaAs technology, anodically grown native oxide thicker than 500Å were found to satisfy this requirement with its effective resistivity ranging from 10^{14} to $10^{16} \Omega \text{cm}$.⁽¹⁾ Leakage currents of the order of 10^{-10}A/cm^2 can be made to flow through 1000Å thick oxide layers at fields of $2 \times 10^6 \text{V/cm}$. At higher fields non-destructive "forming" occurs with the flow of currents up to 10^{-6}A/cm^2 . Fields in excess of $4 \times 10^6 \text{V/cm}$, on the other hand, cause permanent dielectric breakdown in the oxide layer changing its resistivity drastically.

However, anodic oxide films on GaAs thinner than typically 200Å were found to allow large leakage currents without causing destructive breakdown. The maximum permissible current density is a function of the oxide thickness and can be as high as several A/cm^2 for oxides thinner than 100Å. Light emission was observed from reverse biased MOS structures using such thin anodic oxides when the leakage current is typically 1A/cm^2 . Light is first produced in granular form around the periphery of the top metal contact but becomes almost continuous when the leakage current density is above 10A/cm^2 . The light appears white and is clearly visible to the naked eye in dim light conditions.

MOS structures used in this investigation were produced on both n- and p-type GaAs with carrier concentrations ranging from 10^{16} to $2 \times 10^{18} \text{cm}^{-3}$. The growth of anodic oxides was achieved in an AGW electrolyte⁽²⁾ or in 0.02M $(\text{NH}_4) \text{H}_2\text{PO}_4$ mixed with propan 1 : 2 diol in the ratio of 1 : 2. Growth of uniform oxide layers as thin as 11Å covering the whole surface of GaAs was shown to be possible in either electrolyte giving a high degree of control over the oxide thickness.⁽³⁾ The overpotential-time (V-t) curves were continuously traced during anodisation and the overpotential rise was used for the estimation of oxide thickness.⁽²⁾ The oxide

layers were dried in an N_2 ambient in the temperature range of $20^\circ C - 350^\circ C$ and finally metal contacts were established over the oxide layer by evaporating metal through a stainless steel mesh. Each metal contact was typically $10^{-3} cm^2$ in area and roughly circular in shape. The reverse side of the GaAs wafers were covered with appropriate ohmic contact metals.

The reverse-bias I-V characteristics of typical MOS structures are shown in Figure 9, produced on n-type GaAs with different carrier concentrations and different top metal contacts, as well as varying test temperatures. After an effective threshold voltage, the current is almost linearly dependent on the voltage and the resistance of the device in this region is only a function of the GaAs bulk resistance. The leakage current through the oxide layer (produced on n-type GaAs) becomes higher at the same bias when the top metal contact is In ($\phi_m \approx 3.8 eV$) than when it is Al ($\phi_m \approx 4.25 eV$) or Au ($\phi_m \approx 4.8 eV$). Consequently at a given bias more light is produced with structures having top metal contacts with lower work functions. The opposite argument is true for devices produced on p-type GaAs. The stability of the devices also depends strongly on the metal work function. Devices produced on n-type GaAs with Au contacts tend to breakdown much easier than those with Al or In, especially if the oxide thickness is $< 50 \overset{O}{\text{Å}}$.

Figure 10 shows 3 spectra (400-800nm) of the light emitted from a typical device with Al top contact at different bias conditions.* As the spectra do not shift with the increase in

*These spectra were taken by Professor H.L. Hartnagel at Tohoku University, Sendai, Japan, during a period of leave of absence from the University of Newcastle upon Tyne. The measurements were taken with small wavelength increments, step by step, (by determining the emission intensity for each step over an extended period of several minutes in order to eliminate the noise fluctuation of the experimental set-up. The noise amplitudes are given in Figure 10 by the error bars. A Shimadzu-Bausch and Lomb Monochromator, type No: 33-86-25, was employed together with a resistance of $10 M\Omega$ and a Vibrating Reed Electrometer by Takeda-Ricken (TR-84M). The photomultiplier used was HTV446 of the Hamamatsu Television Company. A schematic representation of the set-up is shown in the insert of Figure 10.

bias and also light emission is observed even when the oxide thickness is thicker than the tunnelling distance (e.g. $> 100\text{\AA}$) plasmon excitation is not likely to be the mechanism responsible for the light emission.⁽⁴⁾ The shape of the spectra suggests that the main part of the light emission is in the infrared region originating possibly from the band-to-band recombination radiation in GaAs. Radiation with wavelengths smaller than 700nm can not originate from the GaAs bulk since it corresponds to energies larger than the energy gap of GaAs and therefore can only be produced in the oxide layer or in the "metamorphic" layer between the oxide and GaAs.

I-V characteristics of these thin oxide MOS structures indicate that most of the applied bias is absorbed in the depletion layer near the GaAs surface with thermally generated minority carriers from the semiconductor inversion layer leaking through the oxide (see Figure 11). At higher bias values a large number of minority carriers becomes available due to avalanche multiplication in the depletion layer. Due to the edge effects avalanching first occurs around the edges of the top contact. If the leakage current through the oxide is lower than the creation rate of minority carriers an inversion layer near the surface forms which permits some further voltage drop across the oxide layer to be formed so that the metal Fermi level is moved with respect to the conduction band edge of GaAs. The probability of majority carrier flow from the metal to GaAs becomes higher as the metal Fermi level is closer or above the conduction band edge of GaAs. The light emission is believed to occur when the majority carriers originating from the metal combine with the minority carriers created by avalanche multiplication. Since the energy gap of the native oxide is $\sim 4.5\text{eV}$,⁽⁵⁾ electron-hole recombination occurring in the native oxide have the possibility of producing light with wavelengths smaller than 700nm. The negative resistance usually associated with electron-hole recombination is not observed with these devices, possibly due to the large series resistance of the depletion layer.

Light emission was also observed if the native oxide was

replaced by anodically grown Al_2O_3 or composite oxides i.e. Al_2O_3 and the native oxide of GaAs.⁽⁶⁾ The visible light intensity produced from such structures was found to be somewhat reduced as compared to structures using native oxide only at similar bias conditions. Furthermore, the stability of devices using Al_2O_3 or composite oxides decreases rapidly if the insulator layer becomes thicker than say 100\AA . MOS structures using thermally grown oxides or natural oxides, on the other hand, showed unstable light emission. The light could only be observed at localised spots and permanent breakdown was caused after an operation period of a few seconds.

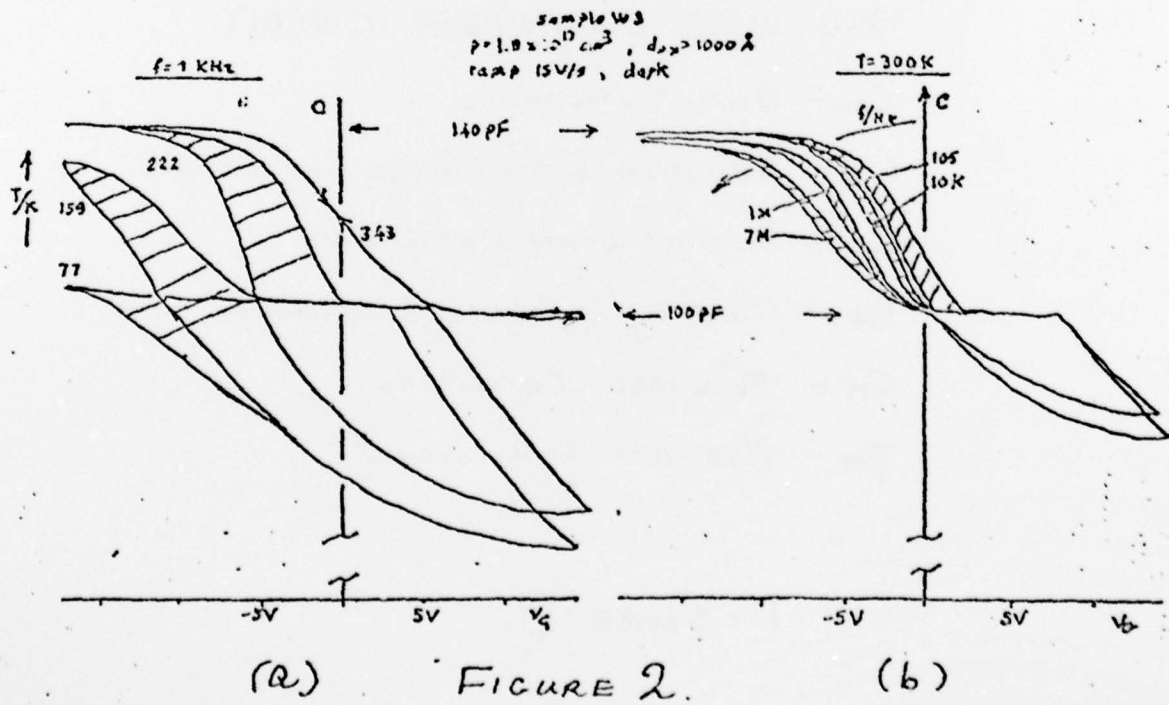
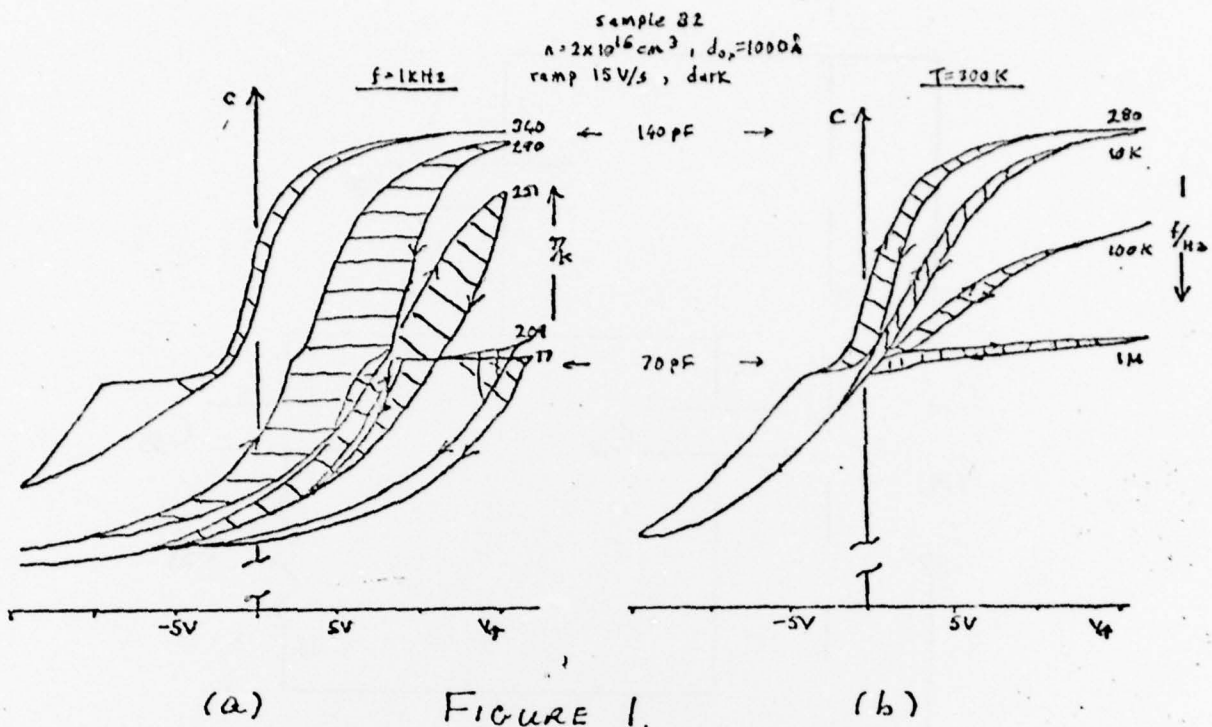
The stability of light emission from anodic oxide MOS structures was tested by operating two similar devices under d.c. bias, one embedded in high quality clear wax and one exposed to air. Each device was fabricated on n-type GaAs with a carrier concentration of $2.2 \times 10^{16} \text{cm}^{-3}$ and had 50\AA of native oxide. The metal contact was a 2000\AA thick In layer. A reverse bias of -17.0V was applied to the devices causing a current of 10mA to flow. In this condition the light emission could be observed in the darkness with the naked eye. Both devices have been operated so far for over 1500 hours without any visible change in the colour or the brightness of the light emission. Further stability tests were carried out on devices which are similar in structure to those described above, by operating them at 77K and also at 120°C for a 200 hour period. Again no noticeable change of their light emission properties was obtained.

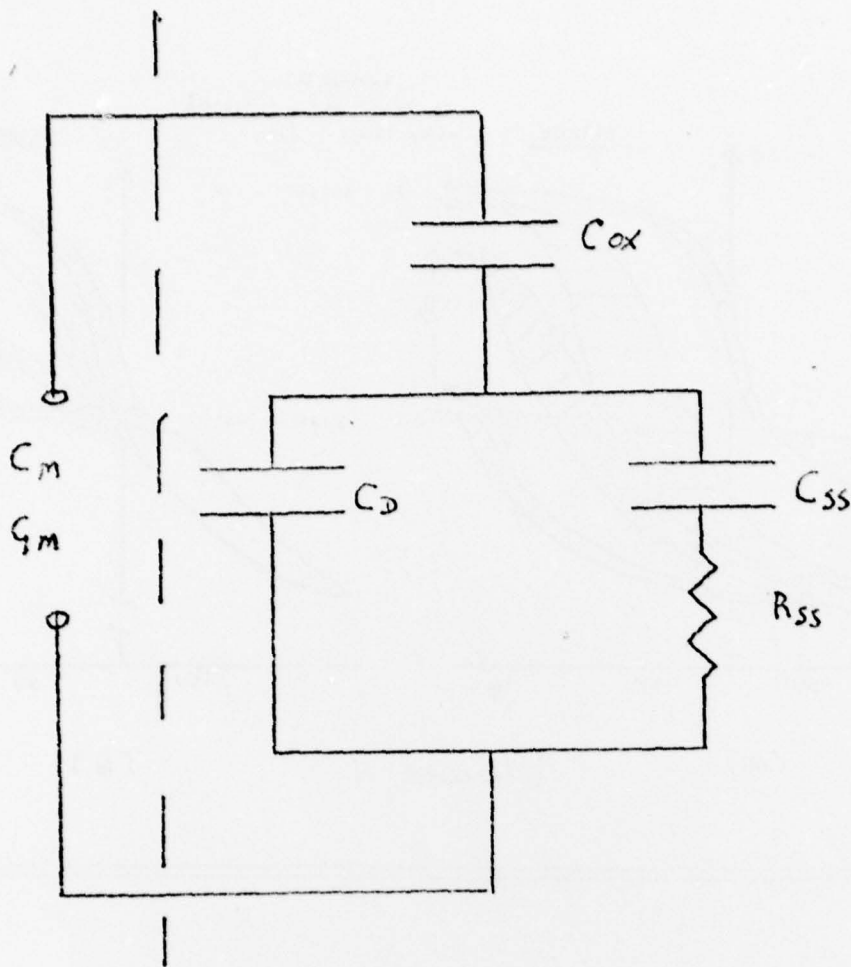
It can be expected that such devices can be developed into cheap light-emitting structures where it is not required to produce the technically more difficult p-n junction devices.

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MOS DIODE EQUIVALENT CIRCUIT

C_{ox} - OXIDE CAPACITANCE

C_D - DEPLETION SPACE CHARGE CAPACITANCE

C_{ss} - SURFACE STATE CAPACITANCE

R_{ss} - CAPTURE/EMISSION RESISTANCE

C_m - MEASURED CAPACITANCE

g_m - MEASURED CONDUCTANCE

FIGURE 3

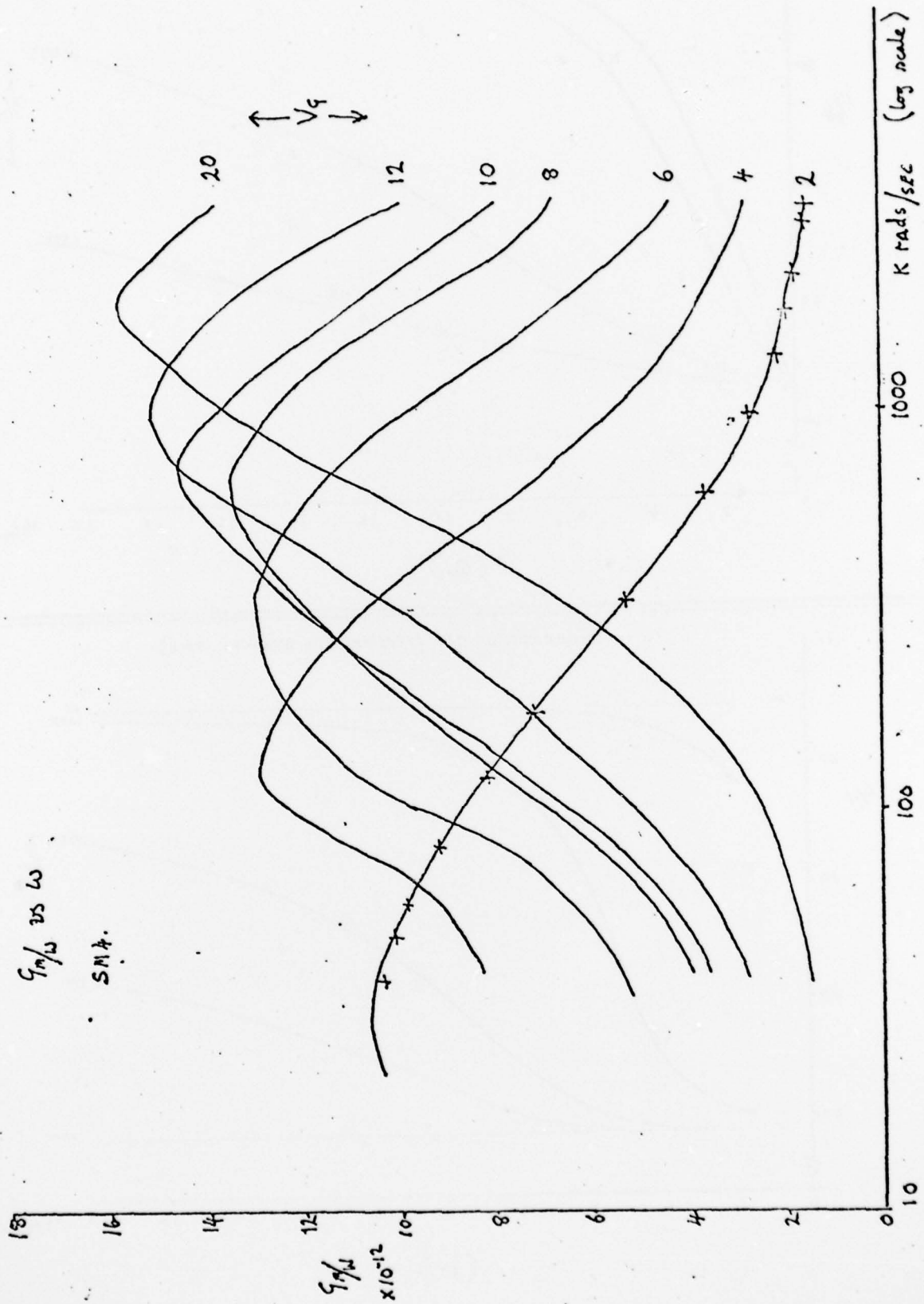
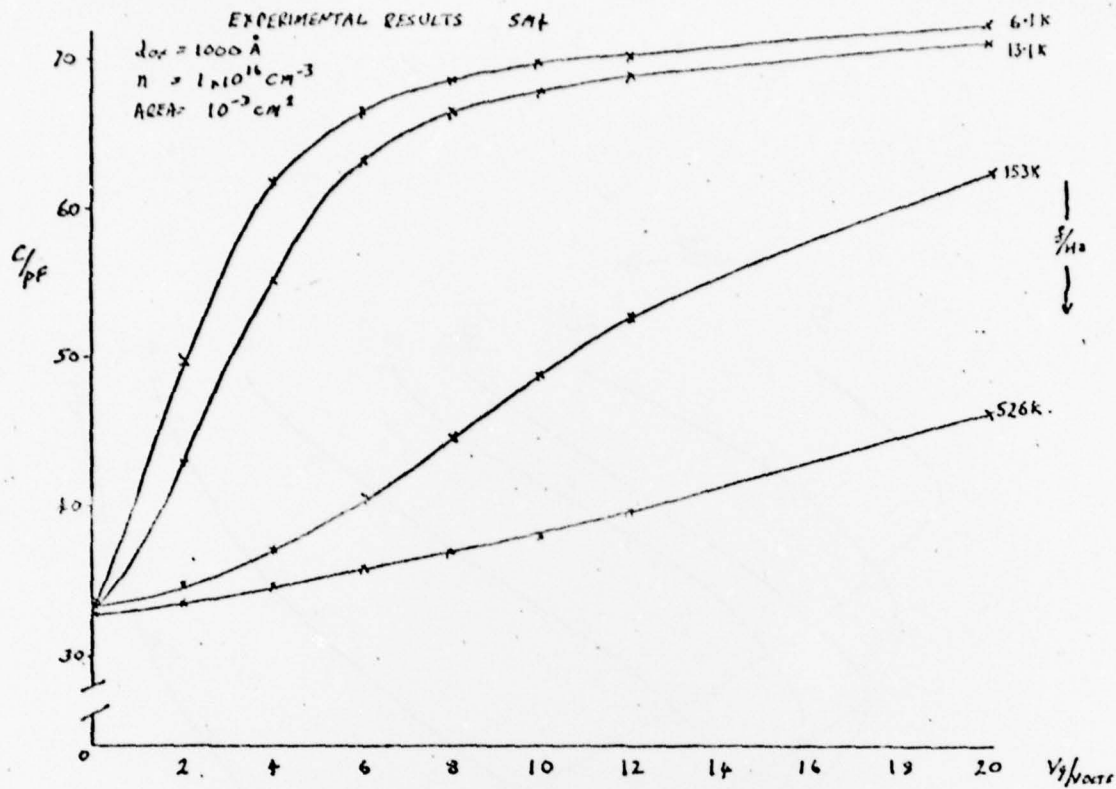
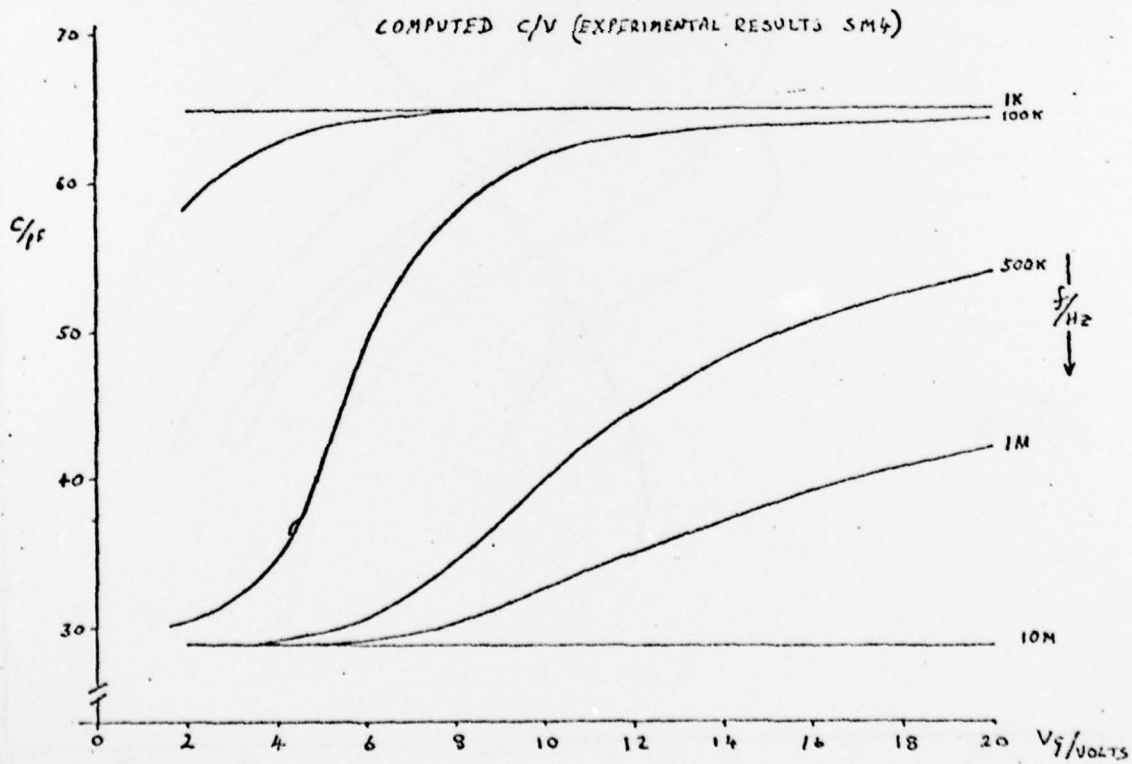


FIGURE 4



(a)



(b)

FIGURE 5.

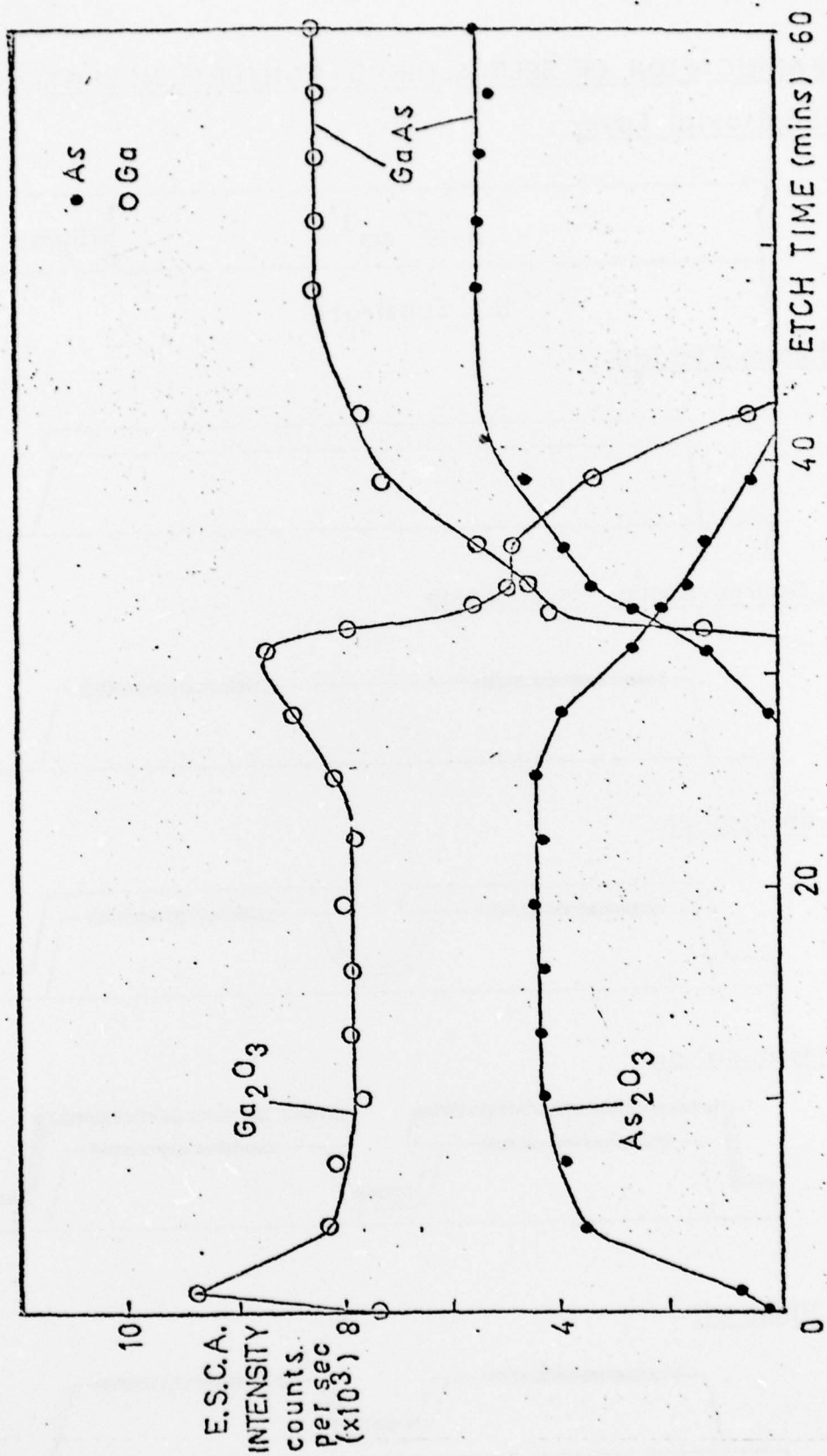
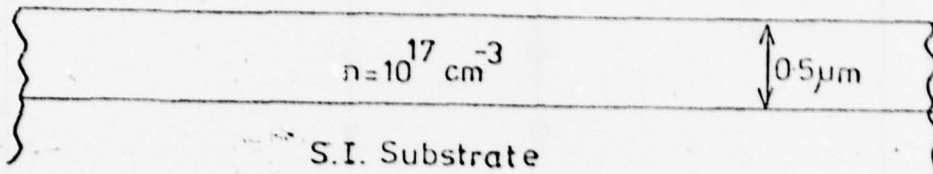


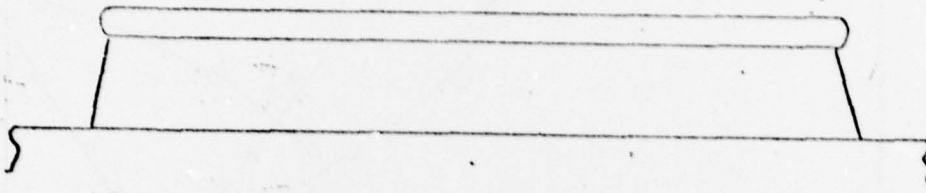
FIGURE 6.

FABRICATION OF SELF-ALIGNED ENH/DEP MOSFET

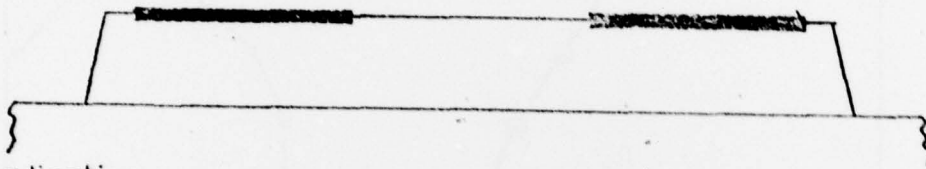
a) Epitaxial Layer



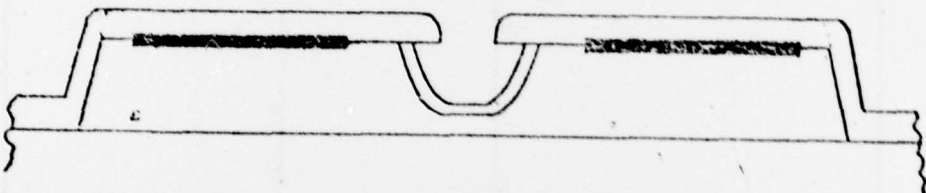
b) Mesa-Etching



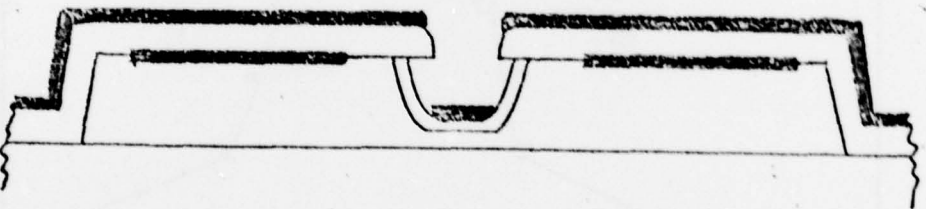
c) Source / Drain Metalisation



d) Anodisation



e) Metalisation



f) Float-off

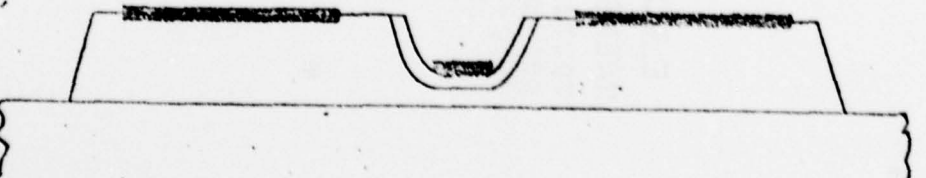
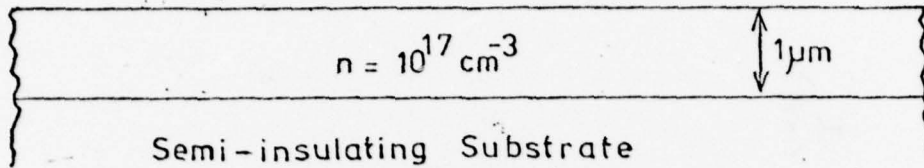


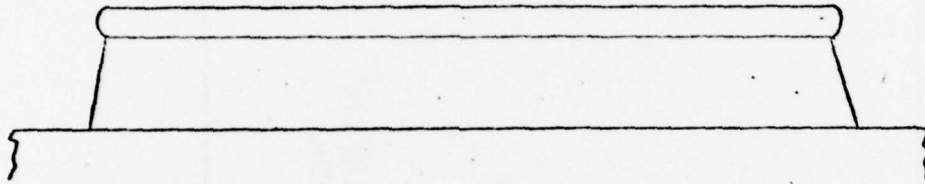
FIGURE 7

PROCESSING STEPS FOR ENHANCEMENT/DEPLETION MOSFET

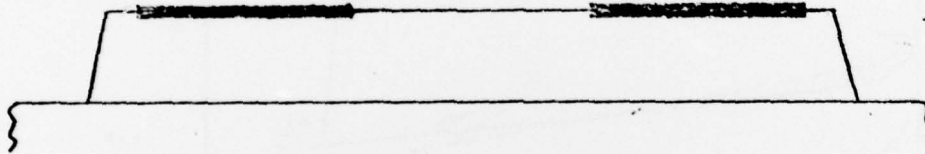
a) Epitaxial Layer



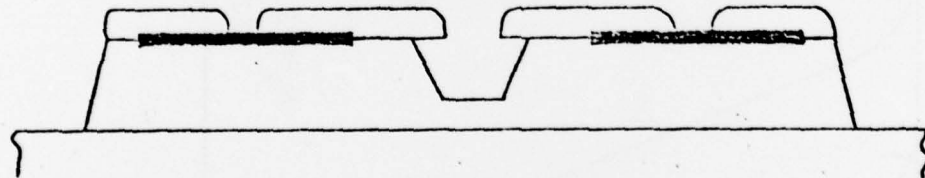
b) Mesa-Etching



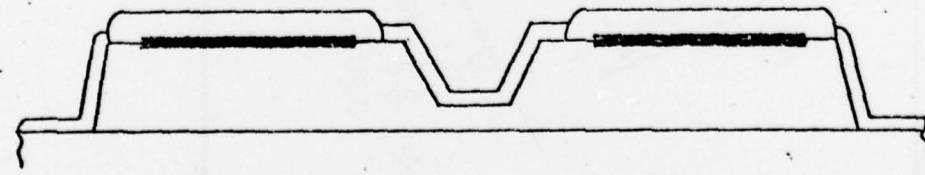
c) Source / Drain Metalisation



d) Channel Etching



e) Anodisation



f) Gate Metalisation

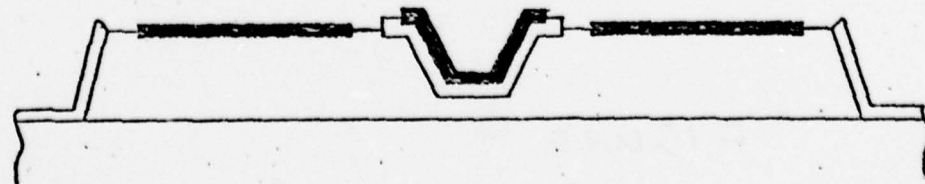


FIGURE 8

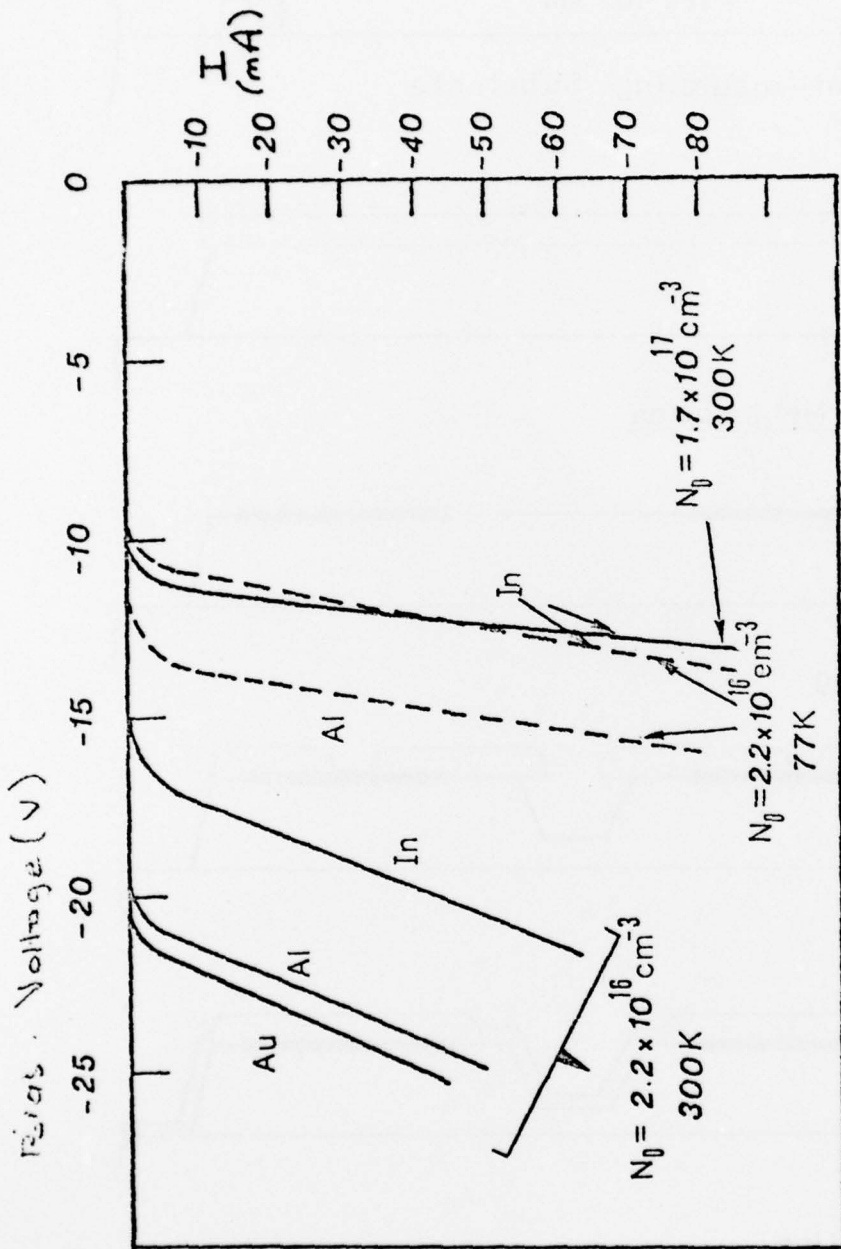


FIGURE 9

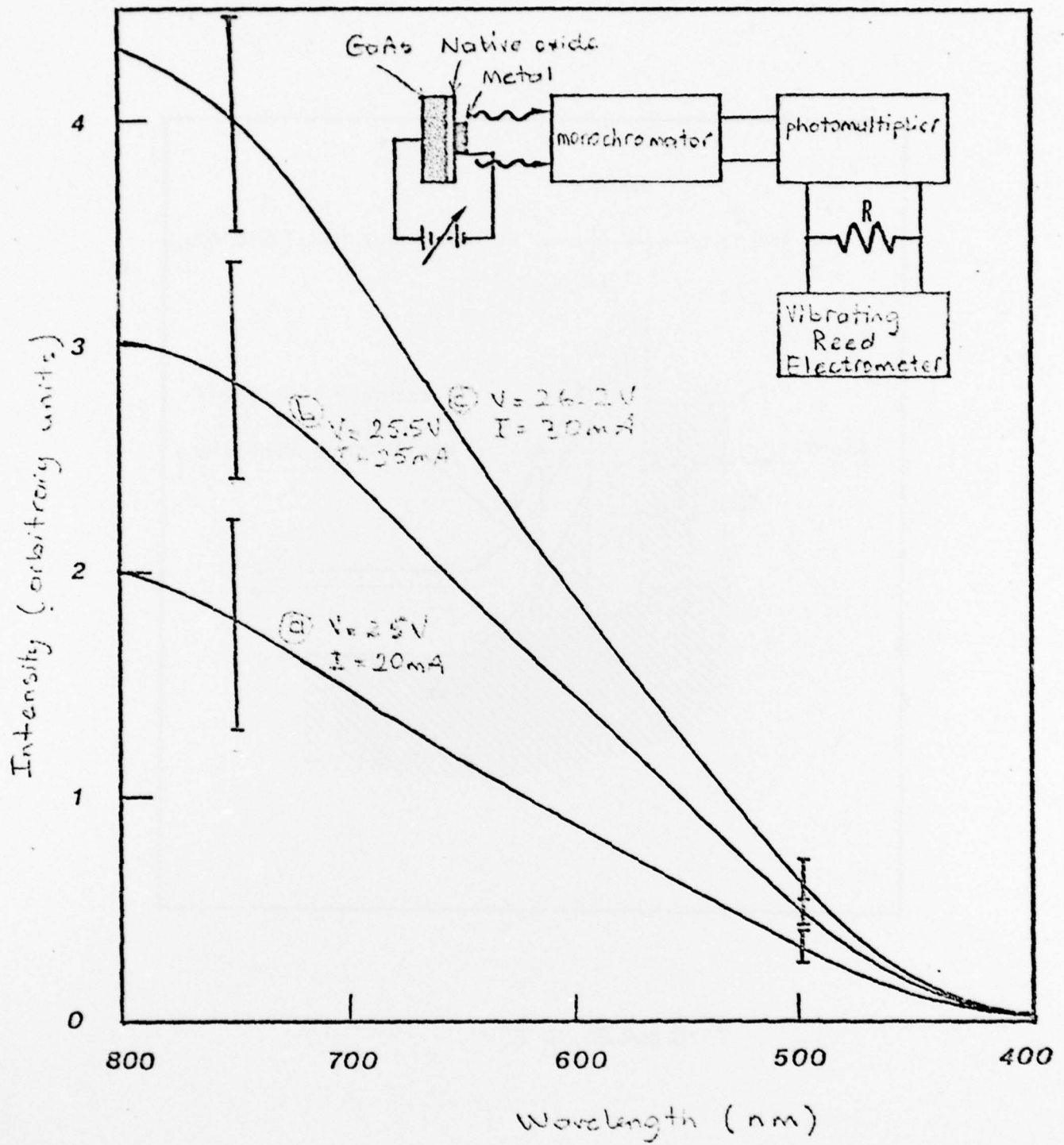


FIGURE 10.

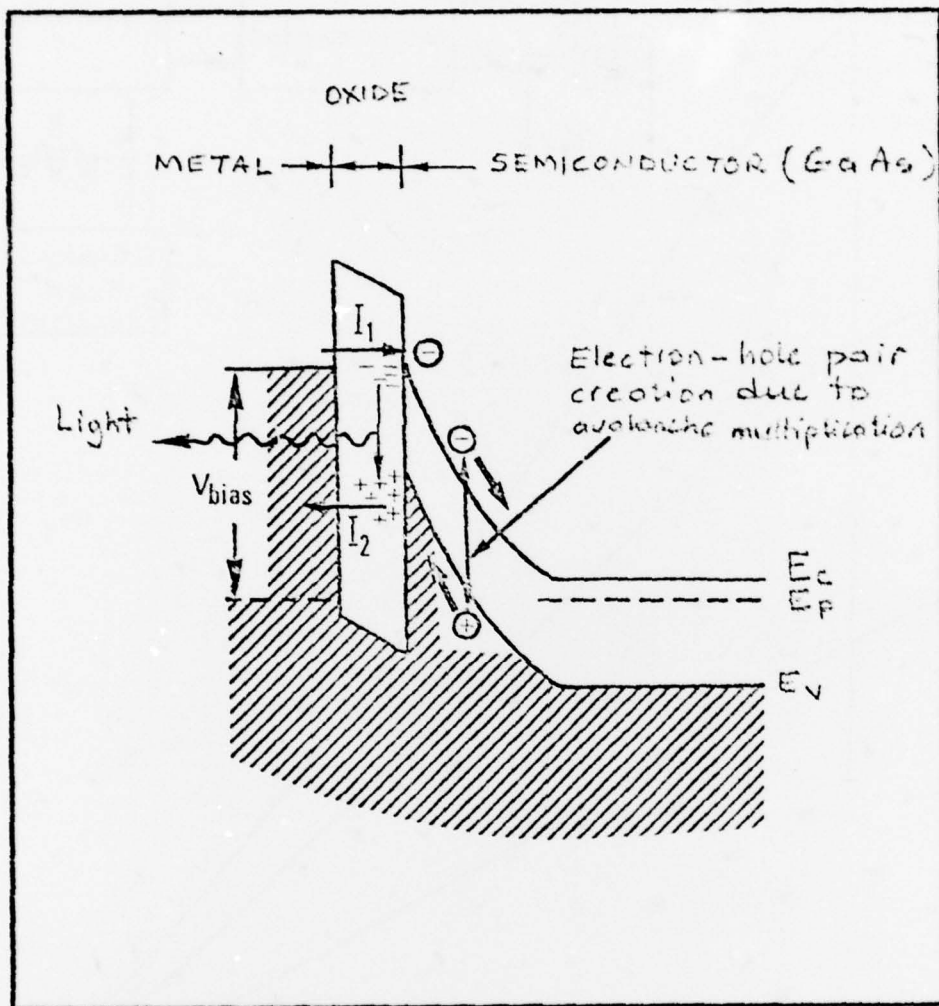


FIGURE 11.

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