This article is an accepted manuscript

It has been peer reviewed and accepted for publication but has not yet been copyedited, house styled, proofread or typeset. The final published version may contain differences as a result of the above procedures.

It will be published in the July 2018 issue of the Johnson Matthey Technology Review.

Please visit the website http://www.technology.matthey.com/ for Open Access to the article and the full issue once published.

Note: the doi will not resolve until the final version is published.

Editorial team

Manager Dan Carter
Editor Sara Coles
Editorial Assistant Ming Chung
Senior Information Officer Elisabeth Riley

Johnson Matthey Technology Review
Johnson Matthey Plc
Orchard Road
Royston
SG8 5HE
UK
Tel +44 (0)1763 253 000
Email tech.review@matthey.com
Inter-diffusion of iridium, platinum, palladium and rhodium with germanium

A. Habanyama\textsuperscript{(a,c)} and C. M. Comrie\textsuperscript{(b,c)}

\textsuperscript{a}Department of Physics, Copperbelt University, P. O. Box 21692, Jambo Drive, Riverside, Kitwe 10101, Zambia

\textsuperscript{b}iThemba LABS, National Research Foundation, P.O. Box 722, Somerset West 7129 South Africa

\textsuperscript{c}Department of Physics, University of Cape Town, Rondebosch 7700, South Africa

Corresponding author: Adrian Habanyama
Email: adrian.habanyama@cbu.ac.zm
Tel: +260 963313923
Synopsis

The down scaling of nano-electronic devices to ever smaller dimensions and greater performance has pushed the Si-based devices to their physical limits. A lot of effort is currently being invested in research to examine the feasibility of replacing Si by a higher mobility semiconductor, such as Ge, in niche high performance metal-oxide semiconductor (MOS) devices. Before germanium can be adopted in industry, a suitable contact material to the active areas of a transistor must be identified. It is proposed that platinum-group metal germanides be used for this purpose, in a similar manner as metal silicides are used in silicon technology.

Implementation of germanium based technology requires a thorough understanding of the solid state interactions in metal/germanium systems in order to foresee and avoid problems that may be encountered during integration. We present a systematic study of the solid state interactions in germanide systems of the four Platinum-Group Metals: Iridium, Platinum, Palladium and Rhodium.

Our approach was essentially twofold. Firstly, conventional thin film couples were used to study the sequence of phase formation in the germanide systems. Conventional thin film couples were also used to identify and monitor the dominant diffusing species during the formation of some of the germanides as it can influence the thermal stability of a device. Secondly, we observed and analyzed several aspects of the lateral diffusion reactions in these four systems, including activation energies and diffusion mechanisms. Lateral diffusion couples were prepared by the deposition of thick rectangular islands of one material on to thin films of another material.

Rutherford backscattering spectrometry (RBS) and microprobe-Rutherford backscattering spectrometry (µRBS) was used to analyze several aspects of the thin film and lateral diffusion interactions respectively. X-ray diffraction (XRD) and scanning electron microscopy (SEM) were also employed.

Keywords: Thin film; diffusion; ion beam analysis.
1. Introduction

Germanium has several attractive properties such as high mobility of charge carriers and very low carrier freeze-out temperatures [1-4]. There is currently a lot of research on high mobility semiconductors, such as Ge, with the view of using them to replace silicon in niche high-performance metal-oxide-semiconductor (MOS) devices [5-7]. Before germanium can be adopted by industry a suitable contact material to the active areas (source, drain and gate) of a transistor must be identified. It is proposed that platinum-group metal germanides be used for this purpose, similar to the manner in which metal silicides are used in present silicon technology [6, 8]. Our work investigates the solid state interactions in germanide systems of the four Platinum-Group Metals: Iridium, Platinum, Palladium and Rhodium, in thin film and lateral diffusion couples.

In the design of transistors the contact material should be stable over a wide temperature range. Conventional thin film couples are well suited for investigating the phase formation sequence and temperature stability of the phases of a system. We have also used thin film couples to identify some of the dominant diffusing species during the phase formation. For device integrity it is important to identify the dominant diffusing species during the formation of the respective germanides as this can influence their thermal stability.

The samples used for studying lateral diffusion reactions were composed of a thick island of one material on top of a much thinner film of another material. Upon annealing the island material would react with the underlying film through vertical diffusion, going through a sequence of phases until the most island-material rich phase is formed. Since no further vertical reaction with the underlying film is possible, the most island-material rich phase may then grow laterally until it attains a critical width, after which other phases appear and grow simultaneously. This is a case of multiple phase formation as would be found in bulk diffusion couples. Lateral diffusion couples thus provide the transition between thin film and bulk behavior.

Since the island material is abundant for diffusion in the lateral diffusion couples, phase formation and reaction kinetics can be studied to a greater extent than in thin film planar
structures. Lateral diffusion structures can be used to simulate bulk diffusion couples because phase formation could extend to lengths of around 100 µm [9]. In kinetic studies of thin film planar structures the diffusion lengths are typically less than 0.5 µm. One can therefore study the transition from thin film to bulk diffusion couple behavior. The study of lateral diffusion couples is particularly well suited for dealing with the challenges of achieving the required lateral abruptness of semiconductor junctions. Excessive diffusion of the substrate element, in this case Ge, during germanide formation could result in overgrowth and bridging in devices [10].

Various early techniques were developed to study lateral diffusion couples [11-20]. In later studies, microprobe-Rutherford backscattering spectrometry (µRBS) was used to study them [9, 21, 22]. The major advantage of this technique is its ability to give depth information.

Some previous work has been carried out in the research field of platinum-group metal/germanium junctions. Saedi et al. [23] reported a scanning tunnelling microscopy and spectroscopy study of the formation of platinum/germanide phases on Ge (111). This study gave a demonstration of the structural dependence of electronic properties in the Pt-Ge system. Schottky barrier diodes have been used in many applications such as gates for metal-semiconductor field-effect transistors, solar cells and detectors [24-27]. A reduction of the PtGe/Ge electron Schottky barrier height by rapid thermal diffusion of phosphorous was reported by Henkel et al. [28]. The results showed that rapid thermal diffusion from a solid diffusion doping source was effective in reducing Schottky barrier heights of platinum germanide Schottky barrier diodes on Ge. Chawanda et al. [29] investigated the change in the (I-V) electrical properties of Pt Schottky contacts on Ge (100) at different annealing temperatures. Their results showed that the as-deposited barrier heights had values that were near the band gap of Ge for Pt/Ge (100) Schottky diodes resulting in good Schottky source/drain contact materials in p-channel Ge-MOSFETS, for the hole injection from source into inverted p-channel [30]. Chawanda et al. [31] also studied the electrical properties of palladium Schottky contacts with Ge (100). Current-voltage (I-V) and capacitance-voltage (C-V) measurements was performed under various annealing conditions. Only one palladium germanide phase, PdGe was formed. A hole trap at 0.33 eV above the valence band was observed after
annealing at 300 ºC. In another study Chawanda et al. [32] investigated the change in the electrical properties of Ir Schottky barrier diodes on Ge (100). Electrical characterization of these contacts using current-voltage (I-V) and capacitance-voltage (C-V) measurements was performed under various annealing conditions. Thermal stability of the Ir/Ge (100) sample was observed up to an annealing temperature of 500 ºC. The results also showed that the onset temperature for agglomeration (binding of primary particles leading to phase formation) in 20 nm Ir/Ge (100) samples occurs between 600-700 ºC.

Gaudet et al. [7] carried out a systematic study of thermally induced reactions of 20 transition metals with Ge substrates. They monitored metal/germanium reactions in situ during ramp anneals at 3 ºCs⁻¹ using time-resolved X-ray diffraction and diffusion light scattering, they also carried out resistance measurements. Their results showed that the platinum group metals Pd and Pt were among the six most promising candidates for microelectronic applications, the other candidates being Ni, Co, Cu and Fe. Nickel is the most used metal for reducing contact resistance, an example of previous research in the area of Ni/Ge junctions is the work reported by Peng et al. [33] on the I-V characteristics of Ni/Ge (100) Schottky diodes and the nickel germanide induced strain after subjecting the Schottky contacts to rapid thermal annealing in the temperature range of 300-600 ºC. Their results showed that the orthorhombic structure of NiGe induces epitaxial tensile strain on Ge substrates due to the difference in lattice constants. They also suggested that the increase in barrier height with increasing annealing temperature may have been due to the conduction band edge shift by the strain after the germanidation process. Hallstedt et al. [34] studied the phase transformation and sheet resistance of Ni on single crystalline SiGe(C) layers after annealing treatments at 360-900 ºC. The role of strain relaxation or compensation in the reaction of Ni on Si₁₋ₓₓGeₓCy layers due to Ge or C out-diffusion to the underlying layer during the phase transformation was investigated. Formation of crystalline Ni(SiGe) was complete at 400-450 ºC but the thermal stability decreased rapidly with increased Ge amount due to agglomeration. This thermal behaviour was shifted to higher annealing temperatures when carbon was incorporated, Ni(SiGeC) layers formed at 500-550 ºC after which there was Ge segregation to the underlying layer.
and carbon accumulation at the interface. Thanailakis et al. [35] established a relationship between as-deposited Pd/Ge (111) and Ni/Ge (111) Schottky barrier height values, the metal work functions and the density of surface states of the germanium substrate.

2. Experimental summary

A study of the phase formation sequence, using conventional thin film couples, was carried out prior to the lateral diffusion study. This study was carried out using RBS and X-ray diffraction (XRD) for phase identification. Thermally oxidized single crystal silicon wafers with a (100) crystal orientation, were used as substrates in all studies. In the studies of the Ir-Ge system, a thin layer of titanium (2 nm) had to be deposited onto the SiO₂ prior to the deposition of the coupling materials; this Ti layer reacted with SiO₂ forming a 'glue' without which the structure could not adhere. Electron beam vacuum deposition of coupling layers of metal and Ge was carried out at pressures in the low 10⁻⁵ Pa range.

In a further preliminary investigation a marker technique was used to monitor atomic mobility during phase formation. The term marker refers to a material deposited in the sample as a reference plane in monitoring the direction of flow of atoms. A thin layer of Ti acted as an inert marker interposed between coupling layers of metal and Ge; the Ge layer being at the surface of the sample. Upon annealing of this structure both Ge and metal atoms could have diffused across the marker at different rates causing it to shift in position towards the dominant diffusing species. The marker Ti signal was monitored by RBS for different annealing times. The dominant diffusing species (DDS) during phase formation was determined by observing the relative shift of the marker.

The lateral diffusion couples were also prepared by electron beam evaporation at a base pressure in the low 10⁻⁵ Pa range. A thin film of one material was deposited first. An ordinary silicon wafer with an array of 390×780 μm² rectangular windows (referred to as a silicon mask), made by photolithographic techniques and selective etching, was then brought into contact with the substrate without breaking vacuum. Island material was deposited through the mask to form structures consisting of metal islands on germanium films and germanium islands on metal films.
Fig. 1 shows a schematic illustration of the sample preparation setup. The source metal or germanium sample pieces were each placed in one of the three crucibles on a water-cooled copper hearth. The positions of the crucibles could be adjusted from the outside of the electron-beam evaporation chamber so that each material to be evaporated could be placed in the path of the electron beam in turn. Each crucible was shielded from the adjacent one by a 2 cm high partition to prevent cross-contamination during deposition. Above the crucibles was a shutter that could be opened and closed from outside the chamber using a bar magnet. The sample changer, which accommodated a maximum of...
three sample holders, hovered above the shutter. The sample changer could be rotated using an external handle in such a way as to place one sample holder at a time in the line of sight of the target vapor. In this way, it was possible to prepare up to three sets of different samples (each on one sample holder) in one experimental run. Between the shutter and the sample changer hung the silicon mask holder which had a provision where a silicon wafer with several rectangular 390×780 μm² openings could be placed. The mask holder could be swung from side to side without breaking the vacuum. Likewise, its height could be adjusted externally. The main features of the bottom compartment of the chamber were a system of vacuum pumps and a cryopanel. A baffle valve is closed between the upper and lower compartments to ensure that the lower section of the system is maintained under vacuum when the system is not in use or during sample changing.

After removing the samples with islands from the evaporation chamber they were cleaved into twelve identical samples, each with two or three islands. Furnace-annealing was used to activate solid state interaction after which the samples were analyzed. The samples were examined using scanning electron microscopy (SEM) to distinguish the various reaction regions and measure their diffusion lengths. Representative samples were selected for further analysis by µRBS. The distribution of elements as a function of lateral position was obtained using µRBS. This technique also provided information regarding the elemental distribution as a function of depth and the thickness of the films. A 2 MeV α-beam focused down to a pixel size of about 1×1 μm was scanned across a well-defined area of the samples. This area, typically of 400×400 μm², was chosen to include all reaction regions observed in the SEM micrographs. Sample orientation was adjusted in such a way that the interfaces of the regions of interest lay horizontally in line with the original island edge so that the microprobe beam scanned parallel to the original island interface. RBS spectra were recorded along with position information. Typically, 128×128 spectra were generated in each run and once it had been established that no variation in composition was observed parallel to the interface, spectra were summed along this direction. This reduced the analysis to a one-dimensional traverse of 128 spectra perpendicular to the interface, thereby optimally exploiting beam position while improving on statistics. The RBS data was analyzed using RUMP [36] simulations.
3. Results

(a) Ir-Ge System

(i) Thin Film Couples

Ir-Ge conventional thin film couple samples for the study of the phase formation sequence had the structure: SiO$_2$/Ge (550 nm)/Ir (90 nm). The samples were annealed in vacuum for various periods of time at temperatures ranging from 350 ºC to 800 ºC. The results showed the appearance of the compounds IrGe and Ir$_4$Ge$_5$ at annealing temperatures of around 350 ºC. Observation of the separate formation of either compound as a first phase could not be achieved. An XRD spectrum obtained following annealing at 400 ºC for 80 minutes is shown in Fig. 2.

![X-ray diffraction spectrum](image)

Fig.2. X-ray diffraction spectrum, of a sample of composition SiO$_2$/Ge (550 nm)/Ir (90 nm) after annealing at 400 ºC for 80 minutes.

The figure shows the presence of IrGe and Ir$_4$Ge$_5$ together with unreacted Ir indicating co-existence. After these two phases, Ir$_3$Ge$_7$ formed; from our data it was not possible to tell whether all the Ir was consumed before Ir$_3$Ge$_7$ appeared. The phase IrGe$_4$ is the most Ge...
rich in the Ir-Ge system, it was the final phase observed and only formed above 800 °C. There was no evidence of the presence of Ir₃Ge₄ during the whole reaction.

A 1.2 nm layer of Ti acted as an inert marker interposed between coupling layers of Ir and Ge to monitor atomic mobility during phase formation. This structure was annealed at 400 °C for different times. Fig. 3 is a diagram showing the results obtained from a RUMP simulation [36] on an RBS spectrum of a sample annealed at 400 °C for 20 minutes. The results show the Ti marker as lying between the phase Ir₄Ge₅ and the unreacted Ge at the surface.

![Fig. 3. A diagram showing the results obtained from a RUMP simulation on an RBS spectrum of a sample annealed at 400 °C for 20 minutes. The insert shows the result of the simulation with thickness in units of 10¹⁵ at/cm².](image)

Fig. 4 shows the Ti ‘glue’ and Ti marker signals before and after annealing. The Ti marker signal shifted by 0.02 MeV to higher energies, i.e. towards the surface. From these results a conclusion can be drawn if certain assumptions are made. Firstly, if IrGe was the first phase to form, then Ge would be the sole moving species for both IrGe and Ir₄Ge₅ formation. This is so because had Ir been a moving species during either IrGe or Ir₄Ge₅ formation, it would have been observed to diffuse across the marker, which was not the case.
case. Secondly, if Ir₄Ge₅ was the first phase to form, then all we can be sure of is that Ge was the sole moving species during Ir₄Ge₅ formation. During IrGe formation there would be no discernable indicator as to which species was moving, the marker being at the interface between Ir₄Ge₅ and Ge. The only firm conclusion we can draw from the marker results is that Ge was the sole diffusing species during Ir₄Ge₅ formation.

(ii) Lateral Diffusion Couples

The lateral diffusion couple results with iridium islands on germanium films showed little germanide growth upon annealing with a reaction region that was too narrow to properly resolve and monitor. The reverse configuration with germanium islands on iridium films showed substantial lateral diffusion. Fig. 5 shows an SEM micrograph of part of a lateral diffusion couple, with a Ge island (250 nm) on an Ir film (25 nm), which had been annealed at 800 °C for 30 minutes.

Four different reaction regions were observed and are labeled A to D in Fig. 5. The original edge of the island is clearly visible as the right edge of the white vertical strip in region C.
This sample then underwent μRBS in order to determine the composition and thickness of the various regions. An area to be scanned was chosen to include all regions observed.

Fig. 5. SEM micrograph of part of a Ge island (250 nm) on an Ir film (25 nm) annealed at 800 °C for 30 minutes showing all the different reaction regions observed.
Fig. 6. Superposition of selected RBS spectra from each of the reaction regions and from the unreacted Ir film. Ir peak heights of the various phases and surface positions of Ge and Ir are indicated.

A representative RBS spectrum picked from each of the four reaction regions labeled A to D is shown in Fig. 6. A spectrum picked from the unreacted Ir film is also included in Fig. 6. Peak heights of the spectra taken from regions D, C and B show the phases Ir₄Ge₅, Ir₃Ge₇ and IrGe₄ respectively. The germanides in these three regions are seen at the surface position. On the other hand, the Ir peak from region A lies below the surface position, indicating that there was no Ir at the surface. Region A therefore consisted of a layer of unreacted Ge on top of the IrGe₄ phase.

The RBS data from the scanned area were analyzed to get stoichiometric information from integrated counts of the Ir and Ge peaks. Fig. 7 shows the results as a function of lateral position.
Fig. 7. Stoichiometric information of a Ge island (250 nm) on an Ir film (25 nm) annealed at 800 ºC for 30 minutes, derived from integrated counts of the Ir and Ge peaks as a function of position.

Fig. 7 shows a region of the phase Ir₃Ge₇ inside the original island interface. The width of this Ir₃Ge₇ region was observed to increase with annealing time. This suggests a gradual decomposition of the phase IrGe₄ into Ir₃Ge₇ by the reaction 3IrGe₄ → Ir₃Ge₇ + 5Ge.

As in the thin film work, the major difference between the lateral diffusion couples prepared by annealing at temperatures below 800 ºC and those at and above 800 ºC was the absence of the IrGe₄ phase below 800 ºC; hence no region of IrGe₄ decomposition was observed in the lateral diffusion couples below 800 ºC.

(b) Pt-Ge System

(i) Thin Film Couples
Pt-Ge conventional thin film couple samples for the study of the phase formation sequence had the structure: SiO$_2$/Ge (500 nm)/Pt (120 nm) or equivalently, SiO$_2$/Ge (2270×10$^{15}$ at/cm$^2$)/Pt (780×10$^{15}$ at/cm$^2$). These were annealed in vacuum for various periods of time at temperatures ranging from 150 °C to 300 °C.

Fig. 8. RBS spectrum of a sample of composition SiO$_2$/Ge (2270×10$^{15}$ at/cm$^2$)/Pt (780×10$^{15}$ at/cm$^2$) after annealing at 190 °C for 2 hours, together with its RUMP simulation. The insert shows the result of the simulation with thickness in units of 10$^{15}$ at/cm$^2$. The phase Pt$_2$Ge is observed to form first.

The results showed the appearance of Pt$_2$Ge as the first phase formed at an annealing temperature of around 190 °C. Fig. 8 shows an RBS spectrum obtained following annealing at 190 °C for 2 hours, together with its RUMP simulation.

The insert in Fig. 8 shows that at this stage of the reaction the sample consisted of Si covered with SiO$_2$ as the substrate, followed by Ge then Pt$_2$Ge and some unreacted Pt at the surface. The results shown in Fig. 8 are consistent with results obtained from the XRD analysis performed on this same sample.

The phase Pt$_3$Ge$_2$ was observed as the second phase to form at 220 °C. Fig. 9 shows an XRD spectrum of a sample annealed at 220 °C for 80 minutes indicating the presence of
Pt$_3$Ge$_2$. The result was consistent with the result of a RUMP simulation on RBS data obtained from the same sample.

Fig. 9. X-ray diffraction spectrum of a sample of composition, SiO$_2$/Ge (500 nm)/Pt (120 nm) after annealing at 220 ºC for 80 minutes. The non-congruent phase Pt$_3$Ge$_2$ is observed to form second.

From our data it was not possible to tell whether all the Pt was consumed before Pt$_3$Ge$_2$ appeared. The next detected phase was PtGe at 250 ºC. The last phase observed was PtGe$_2$ which formed from the interaction of PtGe with unreacted Ge.

A 1.2 nm layer of Ti acted as an inert marker interposed between coupling layers of Pt (300×10$^{15}$ at/cm$^2$) and Ge (430×10$^{15}$ at/cm$^2$) to monitor atomic mobility during phase formation. Fig. 10 shows the results obtained from a RUMP simulation on an RBS spectrum of a sample annealed at 250 ºC for 20 minutes.
Fig. 10. RBS spectrum and RUMP simulation of a marker structure of composition SiO$_2$/Pt (300×10$^{15}$ at/cm$^2$)/Ti (1.2 nm)/Ge (430×10$^{15}$ at/cm$^2$) after annealing at 250 ºC for 20 minutes. The insert shows the result of the RUMP simulation with thickness in units of 10$^{15}$ at/cm$^2$.

The simulated result illustrated by the insert in Fig. 10, shows the Ti marker as lying between two layers of Pt$_2$Ge. We observe some unreacted Ge at the surface and some unreacted Pt below the marker. The results show a significant amount (235×10$^{15}$ at/cm$^2$) of PtGe above the marker, between Pt$_2$Ge and unreacted germanium, there is no PtGe below the marker. Fig. 11 shows the Ti marker signals before and after annealing at 250 ºC for 20 minutes.

The Ti marker signal shifted from A to B by 0.025 MeV to lower energies, i.e. towards the substrate. It is clear that both Pt and Ge atoms migrated across the marker to interact, forming Pt$_2$Ge on either side of the marker.
Fig. 11. Upon annealing at 250 °C for 20 minutes, the marker Ti signal shifts by about 0.025 MeV from channel 342 to channel 335 (A to B).

The amount of platinum that crossed the marker and was observed above it, in units of $10^{15}$ at/cm$^2$, is:

$$\# \text{Pt} = \frac{2}{3} \times 55 \text{ (from Pt$_2$Ge)} + \frac{1}{2} \times 235 \text{ (from PtGe)} = 154. \quad (1)$$

The amount of germanium that crossed the marker and was observed below it is:

$$\# \text{Ge} = \frac{1}{3} \times 110 \text{ (from Pt$_2$Ge)} = 37. \quad (2)$$

The atomic diffusion ratio of Pt to Ge is therefore about 4 to 1.

There are two possible mechanisms by which the PtGe above the marker could have been formed. These are: Pt$_2$Ge $\rightarrow$ PtGe + Pt and Pt$_2$Ge + Ge $\rightarrow$ 2PtGe. No PtGe was seen below the marker; the first of these two processes did not take place in that region as it would have left some PtGe there. It is therefore unlikely for the first process to have taken place above the marker. The second reaction is the one most likely to have taken place above the marker. The PtGe was therefore formed from Pt$_2$Ge and Ge above the marker where there was no discernable indicator as to which species was moving during its growth. The phase Pt$_3$Ge$_2$ was not observed to form in the presence of the Ti marker; PtGe was formed after Pt$_2$Ge skipping Pt$_3$Ge$_2$, which was observed in our earlier thin film work on the system.


https://doi.org/10.1595/205651318X696639
(ii) Lateral Diffusion Couples

The lateral diffusion couple results with platinum islands on germanium films showed little lateral diffusion upon annealing with a reaction region that was too narrow to resolve and monitor properly. The reverse configuration with germanium islands on platinum films showed substantial lateral germanide growth. Fig. 12 shows an SEM micrograph of part of a lateral diffusion couple of a Ge island (145 nm) on a Pt film (35 nm) annealed at 450 ºC for 24 hours. The figure shows a part of the island that is close to one of its corners.

![SEM micrograph](image)

Fig. 12. SEM micrograph of a Ge island (145 nm) on a Pt film (35 nm) annealed at 450 ºC for 24 hours, showing representative regions in a part of the island which is close to one of the corners.

Five different regions, labeled A to E are observed where A starts in the middle of the island and E is the outer most region. The original edge of the island is clearly visible between regions C and D.

The sample in Fig. 12 underwent μRBS analysis to obtain stoichiometric information by RUMP simulation of the Pt and Ge peaks. The scanned area was chosen to include all
regions observed in the SEM micrograph. Fig. 13 shows the results as a function of lateral position.

![Graph showing stoichiometric information of a Ge island on a Pt film annealed at 450 °C for 24 hours derived from RUMP simulation of the Pt and Ge peaks, as a function of lateral position.]

The regions B, C, D and E are found to consist of PtGe₂, Pt₂Ge₃, PtGe and unreacted Pt respectively. Region A appeared to be composed of a mixture of PtGe₂ and unreacted Ge. To show this more clearly, a representative spectrum picked from each of the five regions labeled A to E is shown in Fig. 14.

Downward pointing arrows are used to indicate the surface positions of Ge and Pt. The spectrum from region E shows a peak of unreacted Pt and no Ge. Peak heights of the spectra taken from regions D, C and B show the phases PtGe, Pt₂Ge₃ and PtGe₂ respectively. The germanium in these three regions is seen at the surface position. It is seen from the solid line in the figure that the region A consisted of unreacted Ge and the phase PtGe₂. The Pt peak of the solid line lies at the surface position. This shows that there was some Pt at the surface. The ‘shoulder’ marked in the figure however indicates that there was less Pt at the surface than deeper down. Region A therefore consisted of
PtGe₂ at the bottom while at the top there was unreacted Ge together with the phase PtGe₂.

Fig. 14. Superposition of selected RBS spectra from each of the five regions. Pt peak heights of the various phases and the surface positions of Ge and Pt are indicated.

(c) Pd-Ge System

(i) Thin Film Couples

Of major concern while determining the best thickness of our Pd-Ge sample structure, was the likelihood of overlap between Pd and Ge RBS signals. This is by virtue of their having atomic numbers that lie relatively close to each other in the periodic table. At the same time, samples needed to comprise of elemental layers that were thick enough to induce an appreciable X-ray yield. The structure used was SiO₂/Ge (500 nm)/Pd (70 nm).

In this system, reaction was induced at relatively low temperatures. Figs. 15 and 16 show RBS spectral data for samples annealed at temperatures of 100 °C and 150 °C for 2 hour and 80 minutes respectively. Data from XRD analysis were also obtained, these are
displayed along-side the corresponding RBS data. Layer thicknesses obtained by RUMP simulations (solid lines) are shown.

Fig. 15. X-ray diffraction and corresponding RBS spectrum, of a sample of composition SiO₂/Ge (500 nm)/Pd (70 nm) after annealing at 100 °C for 2 hours. The insert shows the result of the RUMP simulation with thickness in units of 10¹⁵ at/cm². The phase Pd₂Ge is observed to form first.

Fig. 16. X-ray diffraction and corresponding RBS spectrum of the sample of composition SiO₂/Ge (500 nm)/Pd (70 nm) after annealing at 150 °C for 80 minutes. The insert shows the result of the RUMP simulation with thickness in units of 10¹⁵ at/cm². The phase PdGe is observed to form after Pd₂Ge.

Rather straight forward behavior is observed in this system with the two congruent phases Pd₂Ge and PdGe being the only ones observed. Pd₂Ge was the first to form at around 100 °C.
(ii) Lateral Diffusion Couples

Samples for lateral diffusion study were prepared by deposition of thick Ge islands on thin Pd films. This configuration was chosen on the basis of the results observed in the Ir-Ge and Pt-Ge systems. Several lateral diffusion samples were annealed at various temperatures for different lengths of time. The Pd-Ge system exhibited relatively low temperature reaction. It was therefore necessary to carry out the investigation for this system at much lower temperatures than those used for the other systems.

Fig. 17 shows an SEM micrograph of one representative sample with a 100 nm thick Ge island on a 20 nm thick Pd film, annealed at 325 °C for 2 hours, showing three distinct regions labeled A to C.

![Fig. 17. SEM micrograph of a Ge island (100 nm) on a Pd film (20 nm) annealed at 325 °C for 2 hours showing the different reaction regions.](image)

Areas which were chosen to include all the reaction regions observed were scanned on the nuclear microprobe for analysis by μRBS. A spectrum picked from each of the three regions of the Pd-Ge lateral diffusion sample is shown in Fig. 18.
Fig. 18. Superposition of selected RBS spectra from each of the regions of the Pd-Ge lateral diffusion sample. Pd peak heights of the various phases and surface positions of Ge and Pd are indicated.

The spectrum from region C shows a peak of unreacted Pd and no Ge. The peak height of the spectrum taken from the region B shows the phase Pd₃Ge. This phase is seen at the surface position. From the solid line in the figure, it can be seen that the region A consisted of unreacted Ge and PdGe. The Pd peak of the solid line lies at the surface position, showing that there was some Pd at the surface. The ‘shoulder’ marked in the figure indicates that there was less Pd at the surface than deeper down. Region A therefore consisted of PdGe at the bottom while at the top there was unreacted Ge intermingled with PdGe.

(d) Rh-Ge System

(i) Thin Film Couples

Only four equilibrium phases exist in the Rh-Ge system, viz., Rh₂Ge, Rh₃Ge₃, RhGe and Rh₇Ge₂. The chance of getting excessive overlap of RBS peaks was greater in this system than in any other yet reported on in this study; therefore great care was taken to abate this. The sample structure used in this study was, SiO₂/Ge (500 nm)/Rh (60 nm).
Figs. 19, 20 and 21 show XRD results alongside RBS spectra with RUMP simulations for samples with the structure, SiO₂/Ge (500 nm)/Rh (60 nm) annealed between 320 °C and 400 °C.

Fig. 19. X-ray diffraction and corresponding RBS spectrum of a sample of composition SiO₂/Ge (500 nm)/Rh (60 nm) after annealing at 320 °C for 2 hours. The insert shows the result of the RUMP simulation with thickness in units of 10¹⁵ at/cm².

Our RBS data strongly suggests the formation of the non-congruent phase Rh₂Ge as the first phase but there is no firm evidence of this from the X-ray data. RBS data showed

https://doi.org/10.1595/205651318X696639
that the Rh$_2$Ge started to form around 280 °C and later gave way to RhGe. Fig. 21 shows that Rh$_{17}$Ge$_{22}$ was formed after the RhGe phase.

Fig. 21. X-ray diffraction and corresponding RBS spectra of a sample of composition, SiO$_2$/Ge (500 nm)/Rh (60 nm) after annealing at 400 °C for 20 minutes. The insert shows the result of the RUMP simulation with thickness in units of $10^{15}$ at/cm$^2$. The phases Rh$_{17}$Ge$_{22}$ is observed. Only RhGe and Rh$_{17}$Ge$_{22}$ peaks were observed in the X-ray results, probably because the Rh$_2$Ge layer was too thin to induce a good X-ray yield.

(ii) Lateral Diffusion Couples

Samples for lateral diffusion study were prepared by deposition of thick Ge islands on thin Rh films. This configuration, as for the Pd-Ge lateral diffusion samples, was chosen on the basis of the results observed in the Ir-Ge and Pt-Ge systems. Several samples were annealed at various temperatures while time was monitored in the usual way. Shown in Fig. 22 is an SEM micrograph of one representative sample, with a 100 nm Ge island on a 20 nm Rh film, annealed at 600 °C for 15 minutes. The figure shows four reaction regions labeled A to D.
Fig. 22. SEM micrograph of a Ge island (100 nm) on an Rh film (20 nm) annealed at 600 °C for 15 minutes.

Areas which were chosen to include all the reaction regions observed were scanned on the nuclear microprobe for analysis by µRBS.

Fig. 23. Superposition of selected RBS spectra from each of the four regions of the Rh-Ge lateral diffusion sample. Rh peak heights of the various phases and surface positions of Ge and Rh are indicated.
RBS spectra picked from each of the four regions, A, B, C and D are shown in Fig. 23. The spectrum from region D shows a peak of unreacted Rh and no Ge. Peak heights of the spectra taken from regions C and B show the phases RhGe and Rh$_{17}$Ge$_{22}$ respectively. The germanides in these two regions are seen at the surface position. It can be seen from the solid line in the figure that the region A consisted of unreacted Ge and the phase Rh$_{17}$Ge$_{22}$. The Rh peak of the solid line lies below the surface position. This shows that there was no Rh at the surface. Region A consisted of unreacted Ge overlaying the phase Rh$_{17}$Ge$_{22}$. For very long annealing times at relatively high temperatures (around 600 °C and above) the phase RhGe was observed to slowly stretch across the original interface into the island region.

4. Discussion

In this work we used conventional thin film as well as lateral diffusion couples to study the germanide systems of the four Platinum Group Metals: Ir, Pt, Pd and Rh.

(a) Ir-Ge System

(i) Thin Film Couples

Using conventional thin film couples, IrGe and Ir$_4$Ge$_5$ where observed to be the first phases to form in the Ir-Ge system and co-existed at annealing temperatures of around 350 °C. Ir$_3$Ge$_7$ formed after these two phases while the IrGe$_4$ phase only appeared above 800 °C. By interposing a thin layer of Ti (1.2 nm) to act as an inert marker between coupling layers in the Ir-Ge system, the direction of atomic mobility was successfully monitored during the initial stages of the reaction. In the marker samples, IrGe and Ir$_4$Ge$_5$ were again found to coexist from the first stages of reaction. The movement of the marker indicated that Ge was the sole moving species during IrGe$_5$ formation. It was not certain whether Ge was also the sole moving species during IrGe formation.
(ii) Lateral Diffusion Couples

The phases observed to form in lateral diffusion couples of the Ir-Ge system were the same as those observed in the thin film study on this system with the exception of IrGe, i.e. IrGe₄, Ir₃Ge₇ and Ir₄Ge₅. The phase Ir₃Ge₇ was seen to stretch across the original island interface at all temperatures. As in the results of the thin film couples, the phase IrGe₄ was only observed to nucleate at temperatures above 800 °C.

The graph in Fig. 24 is a plot of the growth width with time of annealing for a sample annealed at 800 °C. In this figure the growth width of the Ir₃Ge₇ phase region is labeled as $X_\beta$ while the width of the Ir₄Ge₅ region is labeled as $X_\gamma$.

![Graph](image)

Fig. 24. Plot of growth width with time of anneal for the phases Ir₃Ge₇ and Ir₄Ge₅, for a sample annealed at 800 °C. $X_\beta$ refers to the growth width of the Ir₃Ge₇ region while $X_\gamma$ refers to that for Ir₄Ge₅.

It should be pointed out that the origin in Fig. 24 does not correspond to the original island interface but refers to the interface between the IrGe₄ and Ir₃Ge₇ regions. The growth curve for the region labeled as Ir₃Ge₇ was a result of Ir₃Ge₇ grown from Ir₄Ge₅ and that from the decomposition of IrGe₄. The growth characteristics observed are parabolic with time.


https://doi.org/10.1595/205651318X696639
The growth of the phases $\text{Ir}_3\text{Ge}_7$ and $\text{Ir}_4\text{Ge}_5$ was monitored at the temperatures of 700 °C, 750 °C and 800 °C. Different annealing times were chosen to obtain a reasonable range of growth widths at each of the three temperatures. The squares of the growth widths were plotted against the times of annealing at each temperature and the diffusional growth constants, $K_\beta$, were obtained from the slopes. The logarithms of the diffusional growth constants were then plotted against the reciprocals of the product, $k_B T$, of the Boltzmann constant and the absolute temperatures, the resulting Arrhenius plot for the phase $\text{Ir}_4\text{Ge}_5$ is shown in Fig. 25.

![Arrhenius plot](image)

Fig. 25. Arrhenius plot, $\ln K_\beta$ versus $1/k_B T$, showing temperature dependence of Ge diffusion rate through $\text{Ir}_4\text{Ge}_5$, yielding an average activation energy of $1.6 \pm 0.1$ eV.

The average activation energy, $E_a$, was obtained from the slope of the straight line fit of the Arrhenius plot. The value determined for the diffusion process in the $\text{Ir}_4\text{Ge}_5$ phase was $E_a = 1.6 \pm 0.1$ eV.

Unfortunately the results from the samples annealed at the temperatures 700 °C, 750 °C and 800 °C could not be used to obtain a value of the activation energy for the $\text{Ir}_3\text{Ge}_7$ phase. This was because the mechanisms at play during $\text{Ir}_3\text{Ge}_7$ growth were not the same.
at 800 °C as at temperatures below. At 800 °C the Ir$_3$Ge$_7$ observed inside the island region was from the decomposition of IrGe$_4$ while that outside was formed by the interaction between Ir$_4$Ge$_5$ and the outward diffusing Ge atoms. Below 800 °C the Ir$_3$Ge$_7$ outside was also formed by the interaction of Ge and Ir$_4$Ge$_5$ but that inside was observed as a result of exposure as Ge was consumed from the source region. On the other hand, Ir$_4$Ge$_5$ was formed by the interaction of Ge with the unreacted Ir film, both below and above 800 °C and thus at all three temperatures.

(b) Pt-Ge System

(i) Thin Film Couples

In our work on the Pt-Ge system, Pt$_2$Ge was the first phase formed. The second phase observed was Pt$_3$Ge$_2$. The next phase detected was PtGe and the last was PtGe$_2$. The non-congruent phase Pt$_2$Ge$_3$ was skipped between the last two phases.

The marker technique was also applied to the Pt-Ge system. Pt was found to be the dominant diffusing species during Pt$_2$Ge formation. Some Ge diffusion was also observed to take place, we measured the atomic diffusion ratio of Pt to Ge as being about 4 to 1.

(ii) Lateral Diffusion Couples

Three phases were observed to form in the lateral diffusion couples of the Pt-Ge system; PtGe$_2$ and Pt$_2$Ge$_3$ inside the original island region and PtGe outside. The Pt$_2$Ge and Pt$_3$Ge$_2$ phases which were observed in the thin film study of the system were absent in the lateral diffusion couples.

The graph in Fig. 26 is a plot of the various growth widths against the time of annealing for a sample annealed at 550 °C. The origin in this figure does not correspond to the original island interface but refers to the interface between the regions labeled as A and B in Fig. 12. The growth widths of the phase regions, PtGe$_2$, Pt$_2$Ge$_3$ and PtGe are labeled as, $X_\alpha$, $X_\beta$ and $X_\gamma$ respectively.
Fig. 26. Plot of annealing time against reaction widths for the phases PtGe₂, Pt₂Ge₃ and PtGe at a constant annealing temperature of 550 °C. The growth widths of the phase regions, PtGe₂, Pt₂Ge₃ and PtGe are labeled as, \(X_\alpha\), \(X_\beta\) and \(X_\gamma\) respectively.

It must be pointed out that whereas the lateral growth of the Pt₂Ge₃ and PtGe in the regions labeled as C and D in Fig. 12 respectively are due to reaction mechanisms, the growth of region B is due to exposure of PtGe₂ by the consumption of overlaying Ge. The growth characteristics observed for all regions are parabolic with time.

The lateral widths, \(X_\alpha\), \(X_\beta\) and \(X_\gamma\) were monitored at the temperatures, 450 °C, 500 °C and 550 °C. Different annealing times were chosen to obtain a reasonable range of growth widths at each of these three temperatures. The squares of the growth widths were plotted against the times of annealing at each temperature and the diffusional growth constants, \(K_\beta\) where obtained from the slopes. The logarithms of the diffusional growth constants were then plotted against the reciprocals of the product, \(kBT\) of the Boltzmann constant and the absolute temperatures, the resulting Arrhenius plots for the phases PtGe₂, Pt₂Ge₃ and PtGe are shown in Fig. 27.
Fig. 27. Arrhenius plot, $\ln K_\beta$ versus $1/k_BT$, showing temperature dependence of the Ge diffusion rate through Pt$_2$Ge$_3$, PtGe and PtGe$_2$.

The activation energies, $E_a$ were obtained from the slopes of the straight line fits. The average activation energies determined from the lateral growth rates of Pt$_2$Ge$_3$ and PtGe were, $0.9 \pm 0.1$ eV and $1.3 \pm 0.4$ eV respectively. The activation energy corresponding to the apparent lateral growth rate of the PtGe$_2$ region was $1.5 \pm 0.2$ eV.

(c) Pd-Ge System

(i) Thin Film Couples

The only phases observed to form in the thin film study of the Pd-Ge system were the two congruent phases PdGe and Pd$_2$Ge.

(ii) Lateral Diffusion Couples
The two phases, PdGe and Pd$_2$Ge, which were observed to form in the thin film study of the Pd-Ge system were also the only two observed in the lateral diffusion couples.

The growth region outside the original island interface, labeled as region B in Fig. 17, consisted of the phase Pd$_2$Ge. The original island region, labeled as region A in Fig. 17, consisted of PdGe at the bottom while at the top there was unreacted Ge intermingled with PdGe. There was no region of completely exposed PdGe without intermingled unreacted Ge, it was therefore not possible to obtained data for the growth or exposure rates of the phase PdGe.

The growth of the Pd$_2$Ge region was monitored at the temperatures 275 °C, 300 °C and 325 °C. The periods of annealing were chosen so as to obtain a reasonable range of growth widths at each of the three temperatures; results are presented in Fig.28. Parabolic growth characteristics were observed.

![Graph](image)

**Fig. 28.** A plot of reaction length against the time of annealing for the phase Pd$_2$Ge at temperatures 275 °C, 300 °C and 325 °C.

Arrhenius plots were obtained from the data presented in Fig. 28, in the same way as explained for the Ir-Ge and Pt-Ge lateral diffusion couples. Fig. 29 is an Arrhenius plot showing the temperature dependence of the Ge diffusion rate through Pd$_2$Ge.


https://doi.org/10.1595/205651318X696639
Fig. 29. Arrhenius plot, $\ln K_\beta$ versus $1/k_B T$, showing temperature dependence of Ge diffusion rate through Pd$_2$Ge, yielding an average activation energy of $1.0 \pm 0.1$ eV.

The average activation energy determined from the plot in Fig. 29 was $E_a = 1.0 \pm 0.1$ eV.

(d) Rh-Ge System

(i) Thin Film Couples

Our RBS data strongly suggested the formation of Rh$_2$Ge as the first phase in the Rh-Ge system but there was no firm evidence of this from the X-ray data because the thickness of this phase was too thin to give a good X-ray yield. This was a consequence of the fact that the samples had to be kept thin enough to avoid excessive RBS peak overlap. The Rh$_2$Ge phase appeared to give way to RhGe while Rh$_{17}$Ge$_{22}$ was formed as the last phase.

(ii) Lateral Diffusion Couples

The phase Rh$_{17}$Ge$_{22}$ was observed inside the original island region in lateral diffusion couples of the Rh-Ge system while RhGe grew outside in most cases. Under certain
conditions, the latter phase was observed to slowly stretch across the original interface into the island region. This suggested a slow decomposition of \( \text{Rh}_{17}\text{Ge}_{22} \) into \( \text{RhGe} \).

There were four distinct regions observed in the lateral diffusion couples of the Rh-Ge system; these are represented schematically in Fig. 30.

![Diagram showing different phase regions observed in lateral diffusion couples of the Rh-Ge system.](image)

Fig. 30. Diagram showing different phase regions observed in lateral diffusion couples of the Rh-Ge system.

The relative position of the original island interface and different phases are shown in Fig. 30. Knowledge of the position at which the original interface lay was vital in the analysis of the reactions taking place between different phase regions. This was particularly so in this system with wide reaction regions and a slight shift of the reaction interface between the \( \text{Rh}_{17}\text{Ge}_{22} \) and \( \text{RhGe} \) regions with respect to the original interface for different times of annealing.

Our RBS data from the thin film study of this system strongly suggested the formation of the non-congruent phase \( \text{Rh}_2\text{Ge} \) but this phase was not observed in the lateral diffusion couples. The growths of \( \text{Rh}_{17}\text{Ge}_{22} \) and \( \text{RhGe} \), in the lateral diffusion couples, were monitored at the temperatures of 450 °C, 500 °C and 600 °C. The temperature range and annealing times were chosen in such a way that the decomposition of \( \text{Rh}_{17}\text{Ge}_{22} \) into \( \text{RhGe} \) was not significant. Results for carefully chosen annealing times at 500 °C are presented in Fig. 31. The growth characteristics were observed to be parabolic.
Arrhenius plots obtained from the data shown in Fig. 31 are presented in Fig. 32, showing the temperature dependence of the Ge diffusion rate through RhGe and Rh$_{17}$Ge$_{22}$.

An average activation energy of 1.2 ± 0.3 eV was obtained for both Rh$_{17}$Ge$_{22}$ and RhGe.
5. Summary and Conclusion

The results of our thin film study are summarized in Table 1. Temperatures at which the first reactions were observed to begin are indicated.

Table 1: Summary of the phase formation sequence results for the four systems studied. The temperatures at which the first reactions were observed to start are indicated.

<table>
<thead>
<tr>
<th>SYSTEM</th>
<th>PHASE FORMATION SEQUENCE</th>
<th>TEMPERATURE AT WHICH FIRST REACTION BEGINS</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ir-Ge</td>
<td>1st IrGe and Ir₄Ge₅ (Co-existing) 2nd Ir₃Ge₇ 3rd IrGe₄</td>
<td>350 °C</td>
</tr>
<tr>
<td>Pt-Ge</td>
<td>1st Pt₂Ge 2nd Pt₃Ge₂ 3rd PtGe 4th PtGe₂</td>
<td>190 °C</td>
</tr>
<tr>
<td>Pd-Ge</td>
<td>1st Pd₂Ge 2nd PdGe</td>
<td>100 °C</td>
</tr>
<tr>
<td>Rh-Ge</td>
<td>1st Rh₂Ge 2nd RhGe 3rd Rh₁₇Ge₂₂</td>
<td>280 °C</td>
</tr>
</tbody>
</table>

The lateral diffusion samples used in this study were prepared with the configuration of Ge islands evaporated onto metal films. In all systems, the germanide phases were seen to spread out from the source region in their decreasing order of germanium content. The growth characteristics observed in all phases of the four systems studied were parabolic with time. This is indicative of a diffusion controlled process which, as modeled by Kidson [37], results in parabolic growth even in multi-phase systems.

Table 2 gives a summary of the various phases observed in the lateral diffusion couples of each system and the corresponding activation energies obtained.
Table 2: Summary of the germanide phases seen to spread out from the source region in their decreasing order of germanium content during lateral diffusion. Corresponding activation energies obtained are included.

<table>
<thead>
<tr>
<th>SYSTEM</th>
<th>PHASES OBSERVED DURING LATERAL DIFFUSION AND ACTIVATION ENERGIES OBTAINED</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ir-Ge</td>
<td>IrGe₄</td>
</tr>
<tr>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
</tr>
<tr>
<td>Pt-Ge</td>
<td>PtGe₂</td>
</tr>
<tr>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
</tr>
<tr>
<td>Pd-Ge</td>
<td>PdGe</td>
</tr>
<tr>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
</tr>
<tr>
<td>Rh-Ge</td>
<td>Rh₁₇Ge₂₂</td>
</tr>
<tr>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
</tr>
</tbody>
</table>

The magnitudes of the activation energies, \( E_a \) calculated for all phases, as shown in Table 2, suggest that the lateral diffusion reactions in all four systems were not driven by surface diffusion but rather by diffusion through the interior of the lateral diffusion couples; typical values for surface diffusion being around 0.6 eV [38].

In the current design and processing of transistors the contact material should exhibit low sheet and contact resistances, form at a low temperature and be stable over a wide temperature range. A previous systematic study of the thermally induced reaction of a large number of transition metals with germanium substrates revealed that NiGe and PdGe are the most promising candidates when taking the above requirements into account [7, 8]. From Table 1 we see that palladium germanides form at lower temperatures than any of the other platinum-group metal germanides. Platinum germanides were also found to be promising candidates for microelectronic applications [7]. There has been very little work carried out on the electrical properties of Ir/Ge junctions [32]. Our survey of previous work in this research field showed no evidence of any systematic study of the electrical properties of Rh/Ge junctions. Our work has looked at several aspects of the interfacial phase growth and inter-diffusion kinetics at platinum-
group metal/germanium junctions. However, more work needs to be carried out regarding the electrical properties of these junctions. Our current results could then be used in conjunction with the results based on the study of electrical properties in order to draw more insight and to make comprehensive suggestions on the ideally suited choice of platinum-metal/germanium combinations for particular applications in semiconductor technology applications such as gates for metal-semiconductor field-effect transistors, solar cells and detectors.

Acknowledgements

The authors wish to thank the University of Cape Town, the South African National Research Foundation and the University Science, Humanities and Engineering Partnerships in Africa (USHEPIA) for financial assistance. They also wish to thank the Materials Research Group at the iThemba Labs at Faure for the use of their facilities, Miranda Waldron in the Electron Microscope unit at the University of Cape Town, Ms. Terry Davies of the X-ray unit in the Geological Science Department, University of Cape Town.
References


Authors’ information and photos for the article: Inter-diffusion of iridium, platinum, palladium and rhodium with germanium.

Prof. Craig Comrie is an Emeritus Associate Professor at the University of Cape Town. His research areas are Solid State Physics and Nanotechnology.

Email: craig.comrie@uct.ac.za
Tel: 021 650 3322

Dr. Adrian Habanyama is a Senior Lecturer at the Copperbelt University. His research areas are Solid State Physics and Nanotechnology.

Email: adrian.habanyama@cbu.ac.zm
Tel: +260 963313923