# Nanoindentation-induced Deformation Mechanisms in Germanium

David John Oliver

A thesis submitted for the degree of **Doctor of Philosophy** of The Australian National University

November 2008

ii

### CERTIFICATE

This thesis, to the best of my knowledge and belief, does not contain any results previously published by another person or submitted for a degree or diploma at any university except where due reference is made in the text.

David J. Oliver

### Acknowledgements

I firstly wish to thank my supervisors, Dr Jodie Bradby and Prof Jim Williams, for their expert guidance, patience, and assistance in all aspects of research. Their enthusiasm for doing and communicating quality research has been inspirational. I acknowledge and thank Dr Bradby for performing most of the TEM contained in this thesis.

I sincerely thank Prof Mike Swain of the University of Sydney for sharing the wealth of his expertise on indentation with me, which greatly helped to inform the theoretical interpretation of the results presented here.

I would like to thank Dr Simon Ruffell for carrying out the ion implantation in this thesis, and for helpful discussions of results. I also thank Martin Conway for assistance with ion implantation.

I thank Dr Rui Rao for her assistance in using the Hysitron nanoindenter and Quesant AFM.

I acknowledge Prof Paul Munroe at the University of New South Wales for access to the FIB system and helpful scientific input. Thanks also go to Dr Damien McGrouther and Dr Charlie Kong at the University of NSW for their cheerful and highly capable assistance in the use of the FIB.

My thanks to Dr Brian Lawn and Dr Robert Cook for hosting me at the National Institute of Standards and Technology and sharing with me their formidable combined experience on the topic of indentation fracture. I thank Dr Mark Reitsma, Dr Sanjit Bhowmick, and Dr James Lee for support, company, and experimental assistance during my stay at NIST. I acknowledge the Australian Research Network for Advanced Materials for providing travel support for the visit.

I acknowledge Prof Peter Simpson of the University of Western Ontario for the positron measurements in this study.

I acknowledge Prof Arne Nylansted Larsen for generously providing the thin film Ge on Si samples examined in this thesis.

I am deeply grateful to the people who have supplied support and encouragement over the course of this project. These include my parents, Anna Correll and Nick Oliver, my brother Michael and sister Beth, and my good friends at Graduate House and University House. I am also grateful to my fellow students in Electronic Materials Engineering, and I must particularly thank Bianca Haberl for stimulating discussions and good company on weekend FIB trips, and for performing some of the TEM.

Finally, my deepest gratitude goes to my girlfriend and best friend Hannah Joyce for her solidarity, companionship, and support throughout my PhD, and for proof-reading this thesis.

#### Abstract

Germanium (Ge), a Group IV elemental semiconductor, is an important electronic material used in many technological applications. Although it is frequently considered to be a classic brittle material, deforming elastically under mechanical stress up to the point of fracture, in practise this is not the case. Instead, under indentation with a sharp tip, plastic deformation plays a dominant role and other deformation mechanisms may be activated. In the literature there is some controversy as to what is the dominant indentation response of Ge at room temperature, shear-induced plasticity or high-pressure phase transformation. This thesis addresses that controversy by investigating the indentation response of germanium over a range of loading regimes and sample preparation conditions. A diverse range of responses is observed, shedding light on the behaviour of Ge at nano- and microscale contact events.

A wide range of techniques has been employed in this work to investigate the sharp contact response of Ge. Instrumented nanoindentation with a sharp diamond tip has been used to introduce mechanical damage at small scales. Features of the indentation forcedisplacement (P-h) curve can be linked to changes induced in the material. A number of techniques have been applied to characterise the damage produced, including crosssectional transmission electron microscopy (XTEM), micro-Raman spectroscopy, atomic force microscopy (AFM), scanning electron microscopy (SEM), and focussed ion beam (FIB) analysis. In addition, high-energy ion implantation has been used to introduce structural defects and disorder or to completely amorphise the material.

Loading conditions are found to profoundly effect the deformation response of Ge. Rapid loading rates promote the formation of high-pressure phases during indentation, due to the rate-limited nature of shear plasticity mechanisms. These high-pressure phases transform to amorphous Ge (a-Ge) or metastable crystalline phases on load release. At high maximum load values, cracking becomes an important response. Lateral cracking in the vicinity of the indent is found to cause spallation and debris expulsion, resulting in a dramatic 'giant pop-in' event observed in the P-h curve.

Implantation-induced disorder is found to have a pronounced effect on the mechanical properties of Ge. Implantation-induced defects in crystalline Ge lower the hardness and elastic modulus, suppressing cracking and causing enhanced plasticity and quasi-ductile extrusion. In ion-implanted a-Ge, high-pressure phase transformation is the dominant indentation response. Intriguingly, this phase transformation results in the formation of crystalline Ge on unloading.

Finally, it is found that the deformation response can be altered by confining Ge in the form of a thin film. Thin films of crystalline Ge on Si deform by high pressure phase transformation, resulting in the formation of a-Ge on unloading. The threshold film thickness at which this occurs is associated with the geometry of the stress fields under the indenter.

These results show that a diverse range of indentation responses are possible in Ge and that the dominant response can be controlled via loading conditions and sample preparation. End phases of a-Ge and Ge-III are obtained under appropriate conditions with novel electronic, optical, and chemical properties. Furthermore, many of the findings here should be generalisable to other technologically important covalent semiconductors, opening new avenues of research.

### Contents

1	Intr	oduction 1	
	1.1	Literature review	
		1.1.1 High-pressure behaviour of Ge and Si	
		1.1.2 Nanoindentation studies of Ge and Si	
	1.2	Thesis structure 5	
2	Exp	erimental Techniques 7	
	2.1	Indentation	
		2.1.1 Historical background of hardness testing	
		2.1.2 Nanoindenter operation	
		2.1.3 Nanoindentation Theory	
		2.1.4 Indentation fracture	
	2.2	Ion implantation	
		2.2.1 High-energy ion implantation	
		2.2.2 Theory of ion implantation	
	2.3	Raman microspectroscopy	
		2.3.1 Raman penetration depth	
	2.4	Scanning electron microscopy	
	2.5	5 Transmission electron microscopy	
	2.6	Focussed ion beam system	
	2.7	Atomic Force Microscopy	
	2.8	Positron Annihilation Spectroscopy	
3	Rate	e-dependent indentation behaviour of Ge 29	
	3.1	Introduction	
	3.2	Experimental details	
	3.3	Results	
	3.4	Discussion	
	3.5	Concluding remarks	

4	Nan	oindentation of implanted crystalline Ge	41
	4.1	Introduction	41
	4.2	Experimental details	42
	4.3	Results	43
	4.4	Discussion	52
	4.5	Concluding remarks	55
5	Nan	oindentation of amorphous Ge	57
	5.1	Introduction	57
	5.2	Experimental details	59
	5.3	Results	60
	5.4	Discussion	63
	5.5	Concluding remarks	66
6	Nan	oindentation of germanium thin films on silicon	67
	6.1	Introduction	67
	6.2	Experimental and Modelling Procedures	67
	6.3	Results	69
	6.4	Discussion	80
	6.5	Concluding remarks	82
7	Giaı	nt pop-ins in Ge and Si	85
	7.1	Introduction	85
	7.2	Experimental details	86
	7.3	Results and interpretation	86
		7.3.1 Giant pop-ins in Ge	86
		7.3.2 Observational details and mechanisms of giant pop-in	90
		7.3.3 Phase transformation in Ge after giant pop-in	99
	7.4		104
8	Syn	opsis and concluding remarks	107
	8.1	General observations	107
	8.2	Deformation mechanisms in germanium	109
	8.3	Future directions	

# **List of Figures**

1.1	Schematic of the high-pressure phase transformations of crystalline Ge	
	under diamond-anvil cell (DAC) loading.	3
1.2	Metallic transition pressure vs indentation hardness for selected elemental	
	and compound covalent semiconductors.	6
2.1	Load vs time functions illustrating (a) continuous loading cycle and (b)	
	partial unload loading cycle.	10
2.2	Schematic of the UMIS-2000 indentation instrument. (Taken from UMIS	
	manual.)	11
2.3	Schematic of the Hysitron transducer. (Taken from Triboindenter manual.)	11
2.4	Schematics of (a) indentation geometry and (b) force-displacement curve,	
	showing key parameters in the Oliver and Pharr analysis	13
2.5	Schematic of PL test force-displacement data, showing parameters in the	
	Field and Swain analysis.	15
2.6	Schematics of the main modes of cracking during indentation: (A) Hertzian	
	cone cracking, (B) radial cracking, (C) median cracking, (D) full half-	
	penny cracking, and (E) lateral cracking.	17
2.7	Schematic showing the key features of the tandem accelerator	19
2.8	An energetic ion entering a sample will trigger a displacement cascade.	20
2.9	(a) Raman spectrum of diamond cubic structure Ge (Ge-I). (b) Raman	
	spectrum of amorphous Ge. (c) Raman spectrum of st-12 structure Ge	
	(Ge-III)	22
2.10	Schematics of focussed ion beam (FIB)-milled transmission electron microsc	ору
	(TEM) cross-sections: (a) H-bar cross-section, (b) cross-section for pluck-	
	out	26
2.11	Schematic of FIB-milled cross-section for viewing with the scanning elec-	
	tron microscopy (SEM) column.	27
2.12	Definition of S-parameter for a Doppler-broadened positron annihilation	
	spectroscopy (PAS) spectrum, $S = A/C$ .	28

3.1	Load vs time functions for the Hysitron with (a) a slow loading rate $(dP/dh = 0.5 \text{ mN} \cdot \text{s}^{-1} \text{ on loading, } dP/dh = 10 \text{ mN} \cdot \text{s}^{-1} \text{ on unloading}),$ and (b) a fast loading rate $(dP/dh = 900 \text{ mN} \cdot \text{s}^{-1} \text{ on loading, } dP/dh = 10 \text{ mN} \cdot \text{s}^{-1} \text{ on unloading}).$	30
3.2	Load vs time plots for UMIS high loading rate indents. (a) Test with a rapid unloading rate $(dP/dh \approx 140 \text{ mN} \cdot \text{s}^{-1} \text{ on loading, } dP/dh \approx 15 \text{ mN} \cdot \text{s}^{-1}$ on unloading), and (b) a slow unloading rate $(dP/dh \approx 165 \text{ mN} \cdot \text{s}^{-1} \text{ on loading, } dP/dh \approx 0.7 \text{ mN} \cdot \text{s}^{-1} \text{ on unloading})$ .	31
3.3	<i>P</i> - <i>h</i> curves for Hysitron Berkovich indents to 9 mN at different loading rates: (a) loading rate of 1 mN·s <sup>-1</sup> , (b) loading rate of 150 mN·s <sup>-1</sup> , with elbow, (c) loading rate of 150 mN·s <sup>-1</sup> , with pop-out. For all curves the unloading rate is 10 m·s <sup>-1</sup> . (d) Dependence of measured hardness on loading rate.	32
3.4	P- $h$ curves from a high loading rate indent with the Hysitron with a hold	52
	period of 5 s at maximum load.	33
3.5	Raman spectra from indents made with the UMIS. (a) Spherical tip, $dP/dh \approx 15 \text{ mN} \cdot \text{s}^{-1}$ on loading, $dP/dh \approx 1 \text{ mN} \cdot \text{s}^{-1}$ on unloading. (b) Spherical tip, $dP/dh$ 100 mN $\cdot \text{s}^{-1}$ on loading, $dP/dh \approx 1 \text{ mN} \cdot \text{s}^{-1}$ on unloading. For all plots, solid lines are Raman spectra taken from indents (made under	2.4
3.6	the same conditions in each plot), dotted line is the spectrum of pristine Ge. Raman spectra from indents made with the UMIS. (a) Spherical tip, $dP/dh$ 100 mN·s <sup>-1</sup> on loading, $dP/dh \approx 15$ mN·s <sup>-1</sup> on unloading. (b) Berkovich	34
	tip, $dP/dh = 100 \text{ mN} \cdot \text{s}^{-1}$ on loading, $dP/dh \approx 15 \text{ mN} \cdot \text{s}^{-1}$ on unload- ing. For all plots, solid lines are Raman spectra taken from indents (made under the same conditions in each plot), dotted line is the spectrum of pristine Ge	35
3.7	XTEM from a high loading rate UMIS Berkovich indent ( $dP/dh \approx 60$ mN·s <sup>-1</sup> on loading, $dP/dh \approx 15$ mN·s <sup>-1</sup> on unloading). (a) BF micrograph of the whole indent, showing phase transformation and shear damage. (b) SADP from the phase-transformed zone. (c) SADP from pristine	
	material for comparison.	37
3.8	BF micrograph of a different rapid loading rate UMIS Berkovich indent $(dP/dh \approx 140 \text{ mN} \cdot \text{s}^{-1} \text{ on loading, } dP/dh \approx 15 \text{ mN} \cdot \text{s}^{-1} \text{ on unloading}).$	
	Inset: SADP from the phase-transformed zone	38
3.9	BF image of a high loading rate UMIS spherical indent ( $dP/dh \approx 165$ mN·s <sup>-1</sup> on loading, $dP/dh \approx 1$ mN·s <sup>-1</sup> on unloading). Inset: SADP	
	from the phase-transformed zone.	38

4.1	TRIM simulation results for 800 keV Ge ions implanted in Ge to $3 \times 10^{13}$	
	ions $\cdot$ cm <sup>-2</sup> . (a) Distribution of implanted ions. (b) Distribution of vacan-	
	cies generated during implantation.	42
4.2	XTEM of implanted layers. (a) BF image of $1 \times 10^{13}$ ions·cm <sup>-2</sup> dose	
	implanted layer. (b) BF image of $3 \times 10^{13}$ ions $\cdot$ cm <sup>-2</sup> dose implanted layer.	43
4.3	Results of positron analysis on as-implanted and annealed samples. (a) $S$	
	parameter vs. positron energy for all doses, as-implanted and annealed at	
	200 °C. (b) S parameter vs. positron energy for $3 \times 10^{13}$ ions cm <sup>-2</sup> dose,	
	annealed at 3 different temperatures	45
4.4	AFM micrographs of indents to 100 mN. (a) Indent in unimplanted Ge	
	(10×10 $\mu$ m image). (b) Indent in sample implanted to 3×10 <sup>13</sup> ions·cm <sup>-2</sup>	
	and unannealed (7×7 $\mu$ m image)	46
4.5	Raman spectra from pristine Ge and from the $3 \times 10^{13}$ ions·cm <sup>-2</sup> dose	
	as-implanted sample	47
4.6	XTEM of indent to 100 mN in $3 \times 10^{13}$ ions·cm <sup>-2</sup> as-implanted sample.	
	(a) BF image of indent. (b) SADP from indent damage region within	
	implanted layer. (c) SADP from implanted layer outside indented region.	
	(d) SADP from underlying pristine Ge	48
4.7	(a) BF TEM image of indent to 100 mN in $3 \times 10^{13}$ ions·cm <sup>-2</sup> implanted	
	200 $^{\circ}$ C annealed sample. (b) Enlargement of damage region, showing	
	microcrack. (c) BF TEM image of a different 100 mN indent in the same	
	sample	49
4.8	Nanoindentation hardness vs implanted ion dose for as-implanted and an-	
	nealed Ge samples.	50
4.9	Elastic modulus $E$ vs implanted ion dose for as-implanted and annealed	
	Ge samples.	51
4.10	(a) $P$ - $h$ curve to 50 mN in unimplanted Ge. (b) $P$ - $h$ curve to 50 mN in	
	$3 \times 10^{13}$ ions·cm <sup>-2</sup> as-implanted Ge.	51
4.11	Raman spectra from sample regions that were indented to 100 mN and	
	subsequently implanted.	52
5.1	Schematic of high-pressure phase transformations of a-Ge observed in	
0.11	diamond-anvil cell (DAC) experiments.	58
5.2	P-h curves for indents in a-Ge samples, (a) unrelaxed and (b) relaxed	60
5.3	P-h curve to 100 mN in unrelaxed a-Ge sample.	60
5.4	P-h curve to 50 mN, with a 30 s hold at maximum load, in unrelaxed	
	a-Ge sample.	61
	-	

5.5	Hardness vs implanted dose for unrelaxed and relaxed a-Ge samples	62
5.6	Raman spectra taken from indents in a-Ge, (a) unrelaxed and (b) relaxed. Undeformed a-Ge spectra shown for comparison.	62
5.7	TO linewidth measured from Raman spectra for unindented specimens plotted vs implantation dose, for both unrelaxed (as-implanted) and re- laxed (250 °C annealed) specimens.	63
5.8	(a) BF XTEM micrograph of an indent to 60 mN in unrelaxed a-Ge. (b) SADP from the amorphous layer, away from the indent. (c) SADP from recrystallised region under indent.	64
5.9	(a) BF XTEM micrograph of an indent to 60 mN in relaxed a-Ge. (b) SADP from the amorphous layer away from the indent. (c) SADP from recrystallised region under indent.	64
6.1	Load-50% partial unload $P$ - $h$ curve for bulk Ge, showing departure from elastic response at yield load $P_c$ .	69
6.2	Raman spectra for indents in thin film samples: (a) 50 nm film, (b) 100 nm film, and (c) 200 nm film	70
6.3	XTEM micrograph of 100 mN indent in 50 nm thin film Ge on Si sample, (a) whole indent and (b) close-up of film. Inset: selected area diffraction pattern (SADP) from transformed region (Si and Ge)	72
6.4	XTEM micrograph of 100 mN indent in 100 nm thin film Ge on Si sample, (a) whole indent and (b) close-up of film, with SADP inset	73
6.5	XTEM micrograph of 100 mN indent in 200 nm thin film Ge on Si sample, (a) whole indent and (b) close-up of film, with SADP inset	74
6.6	XTEM of 100 mN indent in 400 nm Ge film on Si. (a) Bright-field (BF) image, showing twinning and punched-out dislocations. (b) Dark-field (DF) image taken using the boxed reflection in the inset diffraction pat-	
	tern, showing twins	75
6.7	SEM micrographs of 100 mN indents in thin film Ge on Si samples. (a) 50 nm film. (b) 100 nm film. (c) 200 nm film	76
6.8	<ul><li>P-h curves to 50 mN and 100 mN in thin Ge films on Si. (a) 50 nm film,</li><li>(b) 100 nm film, (c) 200 nm film, and (d) 400 nm film.</li></ul>	77
6.9	Derivative of 100 mN $P$ - $h$ curve in Fig. 6.8(b), plotted versus depth	77

6.10	Isobaric plots of (a), (c), (e) hydrostatic stress and (b), (d), (f) von Mises stress (shear stress) beneath a spherical indenter, for different combinations of tip radius, film thickness, and applied load. (g) Shear yield point and hydrostatic yield point plotted as a function of Ge film thickness, for the 4.3 $\mu$ m tip	78
6.11	Deformation mechanism diagram, showing dominant initial deformation mechanism in Ge films on Si as a function of indenter tip radius and Ge film radius. Circles show critical transition points obtained from Elas- tica simulations. The boundary between the two regions is given by the relationship $R_c \approx 70h_f$	79
7.1	(a,b) Force-displacement curves for 350 mN indents in Ge, created using identical test parameters. Curves in (a) feature a giant pop-in event. Curve in (b) features only small pop-ins.	87
7.2	P-h curves for Ge to 500 mN	88
7.3	Histograms of (a) the load at which the giant pop-in occurs, and (b) the magnitude of the giant pop-in for 400 indents made with a spherical tip of radius 4.3 $\mu$ m loaded to 500 mN.	88
7.4	Partial unload results for Ge. (a) Force-displacement data. (b) Contact pressure as a function of load, calculated by Field and Swain method	89
7.5	SEM images of 350 mN load indents: (a) one of the indents in Fig. 7.1(a) (giant pop-in) and (b) indent in Fig. 7.1(b) (no giant pop-in).	90
7.6	(a) Force-displacement curve for 350 mN Ge indent featuring giant pop- in. (b) FIB ion-beam image (prior to Pt deposition) of the indent. (c) FIB electron-beam cross-sectional image of indent (a). (d) Force-displacement curve for a 350 mN indent with no giant pop-in. (e) FIB ion-beam image	0.1
	of the indent. (f) FIB electron-beam cross-sectional image of indent (d).	91
7.7	(a) <i>P</i> - <i>h</i> curves for Si to 900 mN. (b) Top-down and (c) cross-sectional FIB images of an indent in Si to 550 mN that has undergone a giant pop-in.	93
7.8	Optical micrographs of 350 mN indents: (a) with a large maximum pop-in (1.96 $\mu$ m), (b) with a small maximum pop-in (0.17 $\mu$ m).	94
7.9	The amount of debris around 350 mN indents plotted against the size of the largest pop-in for the indent. The dotted line is a guide for the eye	94
7.10	Pop-in size as a function of pop-in load for (a) Si and (b) Ge. Points are experimental data. The solid line is the indenter contact radius, calculated from $a_* = (P_*/\pi H)^{1/2}$ . Inset: Schematic showing $P_*$ and $h_*$	96

7.11	7.11 Schematic representation of the deformation process. (a) a shallow latera	
	crack opens, which (b) triggers material removal (shaded area), causing	
	the indenter depth to increase by $h_x$ . Initial unloading occurs by elastic	
	recovery in the bulk. (c) When bulk recovery is complete, the tip loses	
	contact with the base of the indent. Beyond this point, the force on the tip	
	is due to lateral plates of material, until the tip is fully unloaded [diagram	
	(d)]. (e) Force-displacement curve schematic, with points in the loading	
	cycle corresponding to the diagrams marked	99
7.12	(a) Raman spectra from indents in Fig. 7.1(a) (giant pop-in), with spec-	
	trum from undamaged Ge for comparison, (b) Raman spectrum from in-	
	dent in Fig. 7.1(b) (no giant pop-in)	100
7.13	Cross-sectional bright-field TEM of a 400 mN indent without a giant pop-	
	in. Inset: diffraction pattern from deformed region	102
7.14	a) A multiple-loading (5 x 200 mN) P-h curve for Ge, featuring a giant	
	pop-in. b) Raman spectra from the indent and from undeformed Ge. c)	
	XTEM image of the indent, with inset SADP from region directly below	
	indent showing the presence of a phase-transformed amorphous zone	103
8.1	Schematic of elastic stresses (shear stress and hydrostatic stress) as a func-	
	tion of indenter load during an indentation test and point of intersection	
	with yield stresses, illustrating the case for c-Ge at moderate loading rates.	110
8.2	Schematic of elastic stresses and yield stresses for various cases: (a) c-Ge	
	indented at rapid loading rates, (b) ion-implanted c-Ge with defects, (c)	
	amorphous Ge, (d) thin film Ge.	111

# Table of acronyms

a-Ge	amorphous Ge
a-Si	amorphous Si
AFM	atomic force microscopy
BSE	back-scattered electron
BF	Bright field
c-Ge	crystalline Ge
c-Si	crystalline Si
DAC	diamond anvil cell
DF	Dark field
DP	diffraction pattern
DLTS	deep-level transient spectroscopy
FESEM	field-emission SEM
FIB	focussed ion beam
hda-Ge	high-density amorphous Ge
LVDT	linear variable differential transformer
MEMS	microelectromechanical systems
NEMS	nanoelectromechanical systems
PAS	positron annihilation spectroscopy
SADP	selected area diffraction pattern

- **SEM** scanning electron microscopy
- **TEM** transmission electron microscopy
- **UMIS** ultra-micro indentation system
- **XTEM** cross-sectional transmission electron microscopy

### **Publications**

- D. J. Oliver, J. E. Bradby, J. S. Williams, M. V. Swain and P. Munroe. Thicknessdependent phase transformation in nanoindented germanium thin films. *Nanotechnology* 19(47):475709, 2008.
- D.J. Oliver, B.R. Lawn, R.F. Cook, M.G. Reitsma, J.E. Bradby, J.S. Williams and P. Munroe. Giant pop-ins in nanoindented silicon and germanium caused by lateral cracking. *Journal of Materials Research* 23(2):297-301, 2008.
- D. J. Oliver, J.E. Bradby, J.S. Williams, M.V. Swain, D. McGrouther, and P. Munroe. Indentation-induced damage mechanisms in germanium. *Mater. Res. Soc. Symp. Proc.* 983E:0983-LL08-02, 2007.
- 4. D.J. Oliver, J.E. Bradby, J.S. Williams, M.V. Swain and P. Munroe. Giant pop-ins and amorphization in germanium during indentation. *Journal of Applied Physics* 101(4):043524, 2007.