Temperature-dependent changes in structural and magnetic properties of heavy ion irradiated nanoscale Co/Pt multilayers

T Som\textsuperscript{1,4}, S Ghosh\textsuperscript{1,5}, M Mäder\textsuperscript{2,6}, R Grötzschel\textsuperscript{2}, S Roy\textsuperscript{1}, D Paramanik\textsuperscript{1} and A Gupta\textsuperscript{3}

\textsuperscript{1} Institute of Physics, Sachivalaya Marg, Bhubaneswar 751005, India
\textsuperscript{2} Research Center Rossendorf, P O Box 510119, D-01314 Dresden, Germany
\textsuperscript{3} University Grants Commission-Department of Atomic Energy Consortium for Scientific Research, Khandwa Road, Indore 452017, India
E-mail: tsom@iopb.res.in

\textit{New Journal of Physics} 9 (2007) 164
Received 20 April 2006
Published 15 June 2007
Online at \url{http://www.njp.org/}
doi:10.1088/1367-2630/9/6/164

\textbf{Abstract.} We report the changes in 100 keV Kr ion-irradiated nanoscale Co/Pt multilayers at different temperatures and fluences. We observe irradiation-induced changes in their structural and magnetic properties. Ion beam-induced mixing across the interfaces leads to the formation of a CoPt ordered/disordered phase. A large increase in the coercivity is also observed when the irradiation is performed at room temperature. Such an enhancement in the coercivity is attributed to defect clustering and CoPt ordered/disordered phase formation. The mechanism of phase formation is explained in light of the heat of formation rule.
Ion beam processing of nanoscale thin film materials offers both appealing advantages and new challenges in modern science. The nanoscale Co/Pt multilayer system shows giant magnetoresistance properties. Therefore, this system is of immense technological importance in the field of high-density magnetic recording media [1]. The state-of-the-art research on magnetic media has already approached nanoscale and subnanoscale engineering in the lateral and vertical directions. However, when the grain size is reduced to the nanoscale, superparamagnetism places a limit on the magnetic properties. This problem is avoided by using magnetically patterned and laminated antiferromagnetically coupled systems as magnetic recording media [1]. Ion irradiation is a viable route to enhance the magnetic memory by patterning magnetic media [2]. Low energy ion irradiation provides a good handle to tune the magnetic properties of multilayer systems [1]–[5]. For instance, 30 keV He+ ion-irradiation of a Pt/Co/Pt ultra thin trilayer leads to a reduction in the coercivity from 250 to 10 Oe, which is associated with a reduction in the Curie temperature to below room temperature (RT) [4, 5]. On the other hand, irradiation of thicker Co/Pt bilayer samples with 200 keV Kr and 50 keV Ar ions leads to an enhancement of the coercivity [6]. Some of the earlier reports on ion-irradiation of Co/Pt multilayers also have common features like reduction of the coercivity and increase in the Kerr intensity [7]–[10]. Although these results are explained on the basis of collisional mixing, the bilayer results could not be explained in the same way. Thus, a lot more needs to be learnt about the mechanism of the ion beam-induced interface modification process in the case of ultra thin Co/Pt multilayers, its behavioural dependence on the experimental parameters, and its effect on the structural and the magnetic properties.

In this paper, we report the results of temperature-dependent heavy ion-irradiation of Co/Pt multilayers. We focus on the irradiation-induced phase formation across the interfaces and changes in the coercivity. Complementary studies of Rutherford backscattering spectrometry (RBS), x-ray reflectometry (XRR), x-ray diffraction (XRD), atomic force microscopy (AFM), and magneto optical Kerr effect (MOKE) are performed to correlate the irradiation-induced changes in the structural and magnetic properties of the Co/Pt multilayers.

2. Experimental details

Ultra high vacuum e-beam evaporation was employed for the sequential growth of Co/Pt multilayers on a native oxide covered Si(100) at room temperature RT under a vacuum of
1 × 10⁻⁸ mbar; the deposition rate was 0.01 nm s⁻¹. To avoid any possible oxidation of the Co layer, we chose Pt as the top layer. We deposited eight bilayers of Pt and Co with a period of 4.1 nm (2.3 nm Pt and 1.8 nm Co). The total thickness of the multilayer stack is thus 32.8 nm. The samples were uniformly irradiated by 100 keV Kr⁺ ions to fluences in the range of 0.5–2 × 10¹⁶ ions cm⁻² and at three different temperatures, viz. RT, 473 and 573 K.

Monte Carlo SRIM-2006 simulation code [11] shows that 100 keV Kr ions cover the entire multilayer stack and should cause significant atomic displacements in the system due to the large deposited energy. The ion distribution and the deposited energy in the multilayer stack are shown in figure 1.

After ion bombardment, selective samples were analysed by x-ray reflectivity (XRR) to measure the change in the layer configuration due to Kr ions. For this purpose, the Mo-Kα1 line (λ = 0.0709 nm) was used as incident radiation and a NaI detector was used to detect the reflected x-rays. RBS measurements were performed on all samples using 1.7 MeV He⁺ ions to study the possible interface modification. XRD studies (using a Cu-Kα radiation) were performed for the phase analysis. AFM measurements were employed using a Nanoscope-III microscope to study irradiation-induced changes occurring in the surface morphology and its statistical properties. Longitudinal MOKE measurements were performed to study the irradiation-induced changes in the magnetic properties.

### 3. Results and discussion

In order to determine the accurate thickness of the multilayer stack, XRR measurements were performed. Figure 2 shows the experimental and the simulated XRR spectra. The reflectivity curves for the pristine sample shows prominent Bragg peaks, which appear due to the well ordered periodic nature of the Pt/Co multilayer. Simulation for the pristine sample is performed on a multilayer model-dependent approach following Parratt’s formalism [12]. This yields a surface roughness of 0.7 ± 0.01 nm whereas Pt and Co layer thicknesses are found to be within...
Figure 2. XRR spectra of pristine and RT-irradiated Co/Pt multilayers. The vertical scale has been shifted for the irradiated spectra to maintain clarity.

Figure 3. RBS spectra for pristine and irradiated samples: (a) RT and (b) 573 K.

3–4% of the nominal ones. On the other hand, the experimental spectra for the irradiated samples show that the periodicity of the multilayer structure is completely destroyed. This is caused by the complete mixing across Pt and Co interfaces. Another noteworthy feature is the reduction in the total film thickness, which results from the sputtering.

RBS spectra were recorded for all the Co/Pt multilayer samples before and after irradiation at different temperatures. Figure 3(a) shows the RBS spectra collected from the pristine sample and those irradiated at RT. The signature of the multilayer peak configuration is clear for the pristine sample. However, not all the peaks corresponding to the individual Pt and Co layers are clear enough due to poor depth resolution. A qualitative analysis of the RBS spectra corresponding to the RT irradiated samples indicates that the interface mixing starts even at the lowest fluence. With increasing ion fluence, we observe a reduction in the total thickness, which could be due to
sputtering. SRIM-2006 simulation shows that corresponding to 100 keV Kr ions, the sputtering yield of Pt is high (≈10 at. ion⁻¹), which is close to our RBS measurements. The predicted interface mixing and reduction in the total film thickness due to sputtering are consistent with our XRR measurements as described earlier. The RBS spectrum corresponding to the highest fluence is quite complicated in nature and is difficult to simulate in a practically interpretable manner. The most notable features in the spectrum are the reduction in the Pt signal height and its broadening to a significant extent. From these observations, the most likely possibility is ion-induced surface roughening [13] and formation of big and asymmetric nanostructured Co–Pt alloy islands at the sample surface. This can be explained as follows. For each individual island, the thickness is expected to decrease from the centre outward. Summing up such RBS signals for Pt from individual points over all the islands would result in a distributed signal as seen in figure 3(a). Likewise, the nature of the Co signal can be explained (as the other constituent element of the Co–Pt alloy islands).

In this context, looking into the nature of the RBS spectra (figure 3(b)) of the multilayer samples irradiated at the highest temperature, it is clear that sputtering and the formation of big, asymmetric alloy islands at the surface (as predicted for RT irradiation at the highest fluence) are either absent and/or become gradually less although the signature of complete mixing is evident right from the lowest fluence. In fact, ‘RUMP’ simulation [14] shows the formation of a layer having a composition of CoPt although the broadened rear edges of the Pt and Co signals could be indicative of the presence of small islands and/or a rough surface.

To address the issue of change in the surface morphology and formation of nanosized islands due to irradiation, the samples are imaged ex situ using AFM. Figure 4(a) shows the AFM image of the pristine sample which is smooth and featureless with a small rms roughness value of 0.63 nm. This matches well with that obtained from the XRR measurements. We observe distinct changes in morphology and rms roughness (3.88 nm) due to RT irradiation at the lowest fluence (image not shown). The morphology of the multilayer sample irradiated at the highest fluence is shown in figure 4(b). It is observed that big isolated islands of arbitrary shapes are formed in this case and there is a good amount of variation in the lateral size of these islands (with an average value of 465 nm). These islands cover ~65% area which corroborates well with the RBS data where the Pt signal reduces by a factor of ~1.45 (figure 3(a)). Change in the morphology is accompanied by a substantial increase in the rms roughness up to 9.42 nm. On the other hand, the sample irradiated to the same fluence but at a temperature of 573 K shows a different morphology (figure 4(c)). In this case, much smaller grains (average island size of 246 nm) are observed to grow with a much higher coverage (~80%) and a much smaller rms roughness of 4.39 nm.

We have also evaluated the lateral correlation length for different irradiated samples. This length corresponds to the repeat distance between dominant features on the surface. The correlation length, corresponding to the highest fluence, for the samples irradiated at RT and 573 K is given by 563.2 and 312.7 nm, respectively [15].

Figure 5(a) shows the XRD pattern of the pristine film and the one irradiated to a fluence of 2 × 10¹⁶ ions cm⁻² at RT. The XRD pattern of the pristine sample indicates a strong Pt(111) texture. It is observed that the satellite peak disappears for RT-irradiation using the lowest fluence. This is due to the diffusion of atoms across the interfaces. In addition, evolution of some new peaks is observed due to irradiation at RT for all fluences. These peaks match well with different reflections corresponding to the CoPt phase [16]. A similar trend is observed for other elevated temperatures and fluences. These observations are consistent with our aforementioned RBS results where we observe the complete disruption of the interfaces, which results in a
Figure 4. AFM images of Co/Pt multilayers: (a) pristine (10 µm × 10 µm) and irradiated to the fluence of $2 \times 10^{16}$ ions cm$^{-2}$ corresponding to (b) RT (5 µm × 5 µm) and (c) 573 K (5 µm × 5 µm).

completely mixed layer (figure 3(b)). Figure 5(b) shows the representative XRD pattern obtained from a sample irradiated at 573 K to a fluence of $2 \times 10^{16}$ ions cm$^{-2}$ where peaks corresponding to face centred tetragonal (fct) CoPt (101), (110), (111), (112) and (211) are observed. It can be mentioned that since the CoPt (110) reflection is a superstructure one and is forbidden for a disordered fcc CoPt phase, its presence would indicate that at least some fraction of the CoPt phase has the chemically ordered fct L1$_0$ microstructure [17]. It is known that in a two-phase ordered/disordered composite, the superlattice diffraction intensities arise only from the ordered regions, while both the ordered and the disordered regions of the crystal contribute to the other allowed intensities [18]. In addition, the (101) fundamental peak of the CoPt alloy arising out of the irradiated multilayers gets shifted to the higher angle side followed by a reduced peak width. This indicates the partial transformation to the fct ordered phase [19, 20] followed by a grain growth.

The ordering factor is directly deduced from the intensity ratio of the superlattice peak (110) to the fundamental peak (101) [21] and is shown in figure 6. It is clear that at higher
Figure 5. XRD patterns of Co/Pt multilayer samples: (a) pristine along with irradiated at RT and (b) irradiated at 573 K to a fluence of $2 \times 10^{16}$ ions cm$^{-2}$.

Figure 6. Variation in the chemical ordering factor as a function of ion fluence for Co/Pt multilayers irradiated at different temperatures.

fluenes, the amount of ordering reduces with increasing irradiation temperature although it is known that for higher ion fluences more energy is deposited in the system to initialize ordering [20]. Similar diminution in the ordering factor was observed in FePd films irradiated at temperatures between 420 and 570 K [22].

The observed changes in the magnetic properties due to irradiation at RT are depicted in figure 7. It is observed that the conventional square loop shape of the hysteresis gets highly distorted due to irradiation at RT to a fluence of $2 \times 10^{16}$ ions cm$^{-2}$. The loop increases in area with a higher coercive field as the ion fluence increases. On the other hand, the changes are much less for the samples irradiated at 573 K using the same fluence (Kerr plots not shown).

Our MOKE data show substantial enhancement in coercivity values from 100 (pristine) to 221 ($5 \times 10^{13}$ ions cm$^{-2}$) to 291 ($1 \times 10^{16}$ ions cm$^{-2}$) to 483 Oe ($2 \times 10^{16}$ ions cm$^{-2}$) at RT.
Figure 7. Hysteresis loops for the Co/Pt multilayer samples irradiated with Kr ions at RT to different fluences. The MOKE plot for the pristine sample shows the conventional square loop but is not shown here to maintain the clarity.

An enhancement in the coercivity for increasing fluence is also observed for irradiations performed at other temperatures and shows a very similar trend to the ordering factor. However, enhancement in the coercivity is the least for samples irradiated at the highest temperature (245 Oe for a fluence of $2 \times 10^{16}$ ions cm$^{-2}$). In addition, the Kerr hysteresis loops maintain their square shapes (figure not shown).

The large increase in coercivity may be caused by the formation of big nanostructured islands at RT or by the defect clustering, which corresponds to an increase in the magnetic domain wall pinning site density [1, 9]. In fact, for magnetic materials, grain boundaries, phase boundaries, line defects, and defect clusters are known to form pinning sites that impede the movement of magnetic domain walls leading to a high coercivity [18, 23]. In addition, the coercivity is always high when there is a coexistence of ordered/disordered CoPt phases [18, 20]. This is supported by the calculated values of the ordering factor for RT irradiation. On the other hand, irradiation at higher temperatures results in enhanced defect mobility as radiation enhanced diffusion (RED) leads to less point defect clustering. As a matter of fact, for Co/Pt systems, the critical temperature for RED to occur is $\leq 473$ K [6] which supports the above argument. This is further strengthened by our RBS measurements (figure 3(b)) and the lower degree of structural changes, viz. ordering factor and surface roughness.

In order to find out the possible mechanism leading to the phase formation across the interfaces and the defect evolution under Kr-ion irradiation, we make use of the Monte Carlo SRM-2006 simulation. Figure 8 shows the vacancy distribution profile in a Co/Pt multilayer used in the present study. It is clear from this figure that a large number of Pt and Co vacancies are created. In addition, there exists a steep gradient in Pt and Co vacancy concentrations across the interfaces. Pt and Co vacancies are even seen to be driven into the neighbouring layers. Thus, the interfaces can be considered to be good defect clustering sites. It is observed that an average of $50$ ion$^{-1}$ nm$^{-1}$ vacancies are created in Pt layers and an average of $75$ ion$^{-1}$ nm$^{-1}$ vacancies are
created in Co layers. This would mean that corresponding to a fluence of $5 \times 10^{15}$ ions cm$^{-2}$, the vacancy concentration in Pt and Co will be $1.1 \times 10^{23}$ and $2.1 \times 10^{23}$ ions cm$^{-3}$, respectively. It is obvious that all of them will not survive because of the recombination/trapping process. In addition, many of the vacancies will form clusters under continuous ion bombardment.

In general, to know the growth and dissolution of vacancy clusters, one needs to establish a rate equation which assumes that only monovacancies are mobile and undermines the role of interstitials. This necessitates knowledge of a few parameters, viz. vacancy generation rate, diffusion coefficient, and emission and absorption rates of the clusters. While vacancy generation rate is available from SRIM simulation, the emission and absorption rates are related to cluster formation or binding energies and, in the diffusion limited regime, the vacancy migration barrier. Unfortunately, not all such values are available for Pt and Co and hence it is difficult to quantify the number of vacancy clusters. However, to have some idea about the formation of vacancy clusters and their stability, it can be mentioned that Haley et al [24] have shown in their simulation studies on Si that greater vacancy concentrations lead to more clusters, with fewer vacancies. At lower temperatures, clusters grow by dissociation and aggregation of smaller clusters whereas at higher temperatures, fewer clusters form and more vacancies are likely to be free, not bound in a cluster. In addition, the lifetime of clusters at lower temperatures is significantly greater than clusters of the same size at higher temperatures. A similar trend is also observed for metals [25]. The above discussion explains why the irradiation-induced changes in the structural and magnetic properties are more prominent for RT irradiation.

Let us now try to understand the mixing across the Pt and the Co interfaces due to irradiation up to a fluence of $2 \times 10^{16}$ ions cm$^{-2}$. In a head-on collision, from energy momentum conservation, one can estimate that $\sim 84\%$ of the Kr-ion energy is transferred to a Pt atom while...
\[ \sim 97\% \] of the same is transferred to a Co atom in the first collision [26]. These atoms can be readily displaced from their original positions. These displaced Pt and Co atoms cause further displacements by repeated collisions with other atoms. We have taken eight layer pairs of Pt (2.3 nm)/Co (1.8 nm) on Si and used Monte Carlo SRIM-2006 simulation to calculate the number of displaced Pt and Co atoms in each layer. The results shown in figure 9 are obtained from 10 000 incident Kr ions. We observe that more Co atoms \((84 \text{ ion}^{-1} \text{ nm}^{-1})\) are displaced than those of Pt atoms \((55 \text{ ion}^{-1} \text{ nm}^{-1})\). This happens as a result of higher energy transfer from Kr ions to the Co atoms. Figure 9 shows that Pt atoms not only get displaced from their regular sites within the Pt layers but also a fraction is pushed into the Co layers. While the Pt distribution across the interfaces may be described by error functions, this is not enough to describe the Pt distribution within the Co layers (distribution in figure 9(a)). This can be reasonably approximated to be a uniform concentration of Pt in Co layers \(\sim 4 \text{ Pt ion}^{-1} \text{ nm}^{-1}\). Therefore, for a fluence of \(5 \times 10^{15} \text{ ions cm}^{-2}\), this amounts to \(\sim 4 \times 10^{16} \text{ Pt at. cm}^{-2}\) (within the 1.8 nm Co layer). Similarly, many Co atoms are pushed into the neighboring Pt layers although the number is low (compared to those of Pt atoms which are pushed into the Co layers). These recoil atoms lead to mixing across the interfaces. This picture matches well with our experimental results described above.

It can be mentioned that in different temperature regimes the ion beam mixing is governed by different processes [27]. For example, at lower temperatures, it is driven by the ballistic collision and the thermal spike processes, whereas at high temperatures the thermally activated long-range diffusion process (in other words RED) plays a dominant role. The last process is influenced by chemical driving force and temperature. In the present case, for RT irradiation, we have only considered collisional (ballistic) mixing where the initial ion energy is distributed via a set of collision cascades and subcascades randomly relocating atoms in the solid through ballistic jumps. In this framework, after the passage of the ions, the displaced (freely moving) atoms are

Figure 9. SRIM-2006 simulated distributions of displaced (a) Pt and (b) Co atoms due to 100 keV Kr-ion-irradiation of a Co/Pt multilayer deposited on Si. Pt/Co layer boundaries are marked with vertical lines.
quenched to RT in a very short period of time ($\sim 10^{-11}-10^{-12}$ s). This ultrafast quenching process freezes the atoms from an excited state to the metastable state with an expanded lattice spacing, since the timescale is insufficient for recrystallization to the equilibrium state [28]. On the other hand, for high temperature irradiations ($\leq 473$K), RED will be the dominant process where the defects generated by the incoming ions become easily mobile. According to the heat of formation rule [29], the sequence of phase formation should follow the most negative heat of formation as it has the largest chemical driving force. The heats of formation for the CoPt ordered and disordered phases are $-7$ and $-6$ kJ g$^{-1}$ at.$^{-1}$, respectively, while that of the CoPt$_3$ phase is $-2.3$ kJ g$^{-1}$ at.$^{-1}$ [30]. Therefore, the CoPt ordered/disordered phase formation is energetically more favourable and is consistent with our present XRD results. However, the origin of the dewetting process, which leads to nanostructured island formation of different sizes, is not clear.

4. Summary

In summary, we have shown that 100 keV Kr-ion-irradiation of nanoscale Co/Pt multilayers leads to the formation of CoPt ordered/disordered phases at different temperatures. RT irradiation leads to a significant enhancement in the values of coercivity with increasing ion fluence. This is attributed to the defect-induced pinning of the magnetic domain walls. Increase in the coercivity is less at higher temperatures possibly due to thermally activated higher defect mobility caused by the RED.

Acknowledgments

We acknowledge S R Potdar and Dr V R Reddy (UGC-DAE, CSR, Indore) for their help during film growth and MOKE measurements. Dr M Posselt at Forschungszentrum Rossendorf, Dresden is acknowledged for his valuable comments on vacancy clustering. Thanks are also due to J K Tripathi for his help during XRD measurements.

References

[1] Fassbender J, Ravelsona D and Samson Y 2004 J. Phys. D: Appl. Phys. 37 R179
[2] Lohau J, Moser A, Rettner C T, Best M E and Terris B D 2001 Appl. Phys. Lett. 78 990
[3] Devolder T, Chappert C, Chen Y, Cambrill E, Bernas H, Jamet J P and Ferré J 1999 Appl. Phys. Lett. 74 3383
[4] Chappert C et al 1998 Science 280 1919
[5] Ferré J and Chappert C, Bernas H, Jamet J P, Meyer P, Kaitanov O, Lemerle S, Mathet V, Rousseaux F and Launois H 1999 J. Magn. Magn. Mater. 198–199 191
[6] Ghosh S, Mäder M, Grötzschel R, Gupta A and Som T 2006 Appl. Phys. Lett. 89 104104
[7] Ravelsona D, Chappert C, Mathet V and Bernas H 2000 Appl. Phys. Lett. 76 236
[8] Devolder T 2000 Phys. Rev. B 62 5794
[9] Rettner C T, Anders S, Baglin J E E, Thompson T and Terris B D 2002 Appl. Phys. Lett. 80 279
[10] Mougin A, Mewes T, Jung M, Engel D, Ehresmann A, Schmoranzer H, Fassbender J and Hillebrands B 2001 Phys. Rev. B 63 060409
[11] Online at http://www.srim.org/
[12] Parratt L G 1954 Phys. Rev. 95 359
[13] Som T, Satpati B, Satyam P V, Kabiraj D, Gupta A and Mishra N C 2004 J. Appl. Phys. 96 7141
[14] Doolittle R 1985 Nucl. Instrum. Methods Phys. Res. B 9 344

New Journal of Physics 9 (2007) 164 (http://www.njp.org/)
[15] Karr B W, Petrov I, Cahill D G and Greene J E 1997 Appl. Phys. Lett. 70 1703
[16] 2000 JCPDS—International Center for Powder Diffraction Data, Card No. 29-0499 (Newton Square, PA: ICDD)
[17] Withrow S P, White C W, Budai J D, Boatner L A, Sorge K D, Thompson J R and Kalyanaraman R 2003 J. Magn. Magn. Mater. 260 319
[18] Ristau R A, Barmak K, Lewis L H, Koffey K R and Howard J K 1990 J. Appl. Phys. 86 4527
[19] Spada F E, Parker F T, Platt C L and Hasard J K 2003 J. Appl. Phys. 94 5123
[20] Lai C-H, Yang C-H and Chiang C C 2003 Appl. Phys. Lett. 83 4550
[21] Warren B E 1969 X-ray Diffraction (Reading, MA: Addison-Wesley) p 208
[22] Ravelsona D, Devolder T, Bernas H, Chappert C, Mathet V, Halley D, Samson Y, Giles B and Marty A 2001 IEEE Trans. Magn. 37 1643
[23] Livingston J D 1981 J. Appl. Phys. 52 2544
[24] Haley B P, Beardmore K M and Jensen N G 2006 Phys. Rev. B 74 045217
[25] Bacon D J, Colder A F and Gao F 1997 J. Nucl. Mater. 251 1
[26] Feldman L C and Mayer J-W 1986 Fundamentals of Surface and Thin Film Analysis (Amsterdam: North-Holland) p 16
[27] Dhar S, Som T, Mohapatra Y N and Kulkarni V N 1995 Appl. Phys. Lett. 65 1700
[28] Som T, Satpati B, Prokert F, Cantelli V and Kabiraj D 2006 Nanotechnology 17 5248
  Chang G S, Lee Y P, Rhee J Y, Lee J, Jeong K and Whang C N 2001 Phys. Rev. Lett. 87 067208
[29] Was G S 1990 Prog. Surf. Sci. 32 211
[30] de Boer F R, Boom R, Mattens W C M, Miedema A R and Niessen A K 1988 Cohesion in Metals ed F R de Boer and D Pettifor (Amsterdam: North-Holland) p 627