STEM-EELS investigations of high-k dielectrics

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Abstract. Metal (e.g. TiN, Ta₂N) inserted hafnia-based high-k gate stacks on Si as well as gallium oxide/gallium gadolinium oxide gate stacks on III-V semiconductors (GaAs, InGaAs, AlInAs) are being investigated for current and future MOSFET applications. The combination of light, medium and heavy elements in multilayers with widths down to a nanometre is a major challenge for high spatial resolution EELS spectrum imaging, especially as the interfaces have significant roughness on this scale. However, it is possible to identify layers significantly thinner than the widths of the profiles that result from beam broadening and/or roughness. To deal with the range of edges present, it is essential to collect edges out to losses of ~3keV. Edges from some elements are difficult to extract but, in suitable case, the ELNES on other edges can be used to give information on such elements. Recording the low loss region of the spectrum as well as the core loss regions at each pixel allows much fuller processing of the data including determination of the absolute numbers of atoms per unit volume. In this way, the distributions of the different phases can be estimated and the overall accuracy is excellent. The results are also in good agreement with earlier work using ELNES.

1. Introduction

In the continued drive to reduce the size of digital semiconductor devices, the traditional SiO₂ gate dielectric has become so thin that leakage due to tunnelling has imposed a limit. This was mitigated to some extent by moving to Si(0,N) but the International Technology Roadmap for Semiconductors (ITRS) has shown the requirement for high-k dielectrics over a number of editions [1]. Much work has gone into their investigation throwing up a wide range of problems. These have led to a steady reduction in the targets in each subsequent ITRS edition. Despite the many problems, INTEL has launched commercial devices which include HfO₂-based dielectrics [2]. There is still intense research activity on HfO₂-based dielectrics but a range of other materials is also being investigated, often involving rare earths [e.g. 3, 4]. The thermal processing required in making a device, including a spike anneal to activate the dopants, causes both physical and chemical changes in the system including re-growth of SiO₂ at the substrate-high-k dielectric interface. While decreasing the effective dielectric constant, if done in a controlled manner, such re-growth can be beneficial as it separates the carriers in the channel from the trapped charge and defects in the high-k dielectric and so stops degradation of the channel mobility [e.g. 5].

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The situation is complicated further by the desire to remove the depletion capacitance associated with the traditional highly-doped poly-Si gate electrode and also to prevent diffusion of the dopant through the thin dielectric to the channel. Both are achieved by replacing the poly-Si electrode with a metallic one. A metal with an effective work function near the conduction band edge is required for n-type devices while one with an effective work function near the valence band edge is required for p-type devices. Again changes of effective work function occur during thermal processing and it seems that one dielectric/metal pair may need to be used for n-MOS devices and a second pair for p-MOS devices [6]. A wide range of metal electrodes has been investigated with TiN, TiAlN, Ta2N, Ta2(C,N) and Ru2O5 being amongst them.

Looking further into the future, semiconductors with higher channel mobility are under investigation including III-V materials for n-type devices and Ge for p-type devices with co-integration of both with Si envisaged to get ultimate CMOS. Passlack et al showed that deposition of Ga2O3 under the correct conditions unpins the Fermi level in GaAs allowing MOSFET action but that it gives an unacceptably high leakage current [7]. Incorporating Gd into the oxide after the first ~1nm, gives gallium gadolinium oxide (GGO) on top of a thin Ga2O3 template layer and this both unpins the Fermi level and gives acceptably low leakage. Using this, world leading MOSFETs have been made on GaAs-based heterostructures incorporating a buried InGaAs channel [8]. However, to scale down such devices to the size needed beyond the 22nm technology node, two issues must be addressed. The first is to increase the mobility in the channel. The second is to achieve sufficiently low access resistance. Only if both are achieved will the necessary drive current be achieved. To increase the mobility, an increase in the In content in the channel seems the most promising approach but raises issues with the dielectric. While the Ga2O3/GGO combination is still a promising candidate, a range of other dielectrics is under investigation [9-12].

The wide range of atomic numbers present and the fine scale on which they are grown produces a challenge for any characterisation technique. Here the information available from nanoanalytical electron microscopy is considered and, in particular, from electron energy loss spectroscopy (EELS) carried out in the scanning transmission electron microscopy (STEM) mode using spectrum imaging (SI) [13]. Not only does this provide information on elemental distributions and composition from the intensities of the ionisation edges but the electron loss near edge structure (ELNES) also gives information on the local chemistry. The challenge is increased significantly by the fact that the interfaces are frequently rough on the nanometre scale rather than being atomically flat.

2. Experimental Methods
The Ga2O3/GGO on the III-V heterostructure discussed below was grown by MBE on a 400 semi-insulating GaAs substrate, using a dual chamber system and the method is described in [14, 15]. The III–V heterostructure was grown in one chamber and consists of a 100 nm GaAs/AlGaAs superlattice layer, a 65 nm AlGaAs buffer layer, a 10 nm InGaAs layer, a 5 nm AlGaAs barrier layer and a 1 nm GaAs spacer layer. The heterostructure was transferred under ultra-high vacuum conditions to a second chamber, where oxide deposition took place. The Ga2O3/GGO gate stack consists of a template layer of ~1nm of Ga2O3 with a carefully controlled interface which unpins the GaAs surface. This is followed by ~10nm of GGO. After removal from the MBE chamber, Pt was deposited on top of the GGO by electron beam evaporation [16].

The metal inserted “hafnium silicate” (HfSiO) dielectric gate stack on silicon sample discussed was grown in the following way. The native oxide was removed from a Si (100) wafer by etching in HF:H2O (1:100) solution. The wafer was then put through an O2/DI water cleaning sequence resulting in the growth of ~1 nm chemical oxide. Metal organic chemical vapour deposition (MOCVD) was used to deposit approximately 4nm of HfSiO (a co-deposition of 70% HfO2 and 30% SiO2). The wafer was then given a standard degas treatment at 330 °C for 40 s and ~10 nm of TiN was deposited by physical vapour deposition (PVD). This was followed by 100 nm of amorphous Si and the resulting stack annealed at 1000 °C for 10 s to simulate dopant activation. During this step, the amorphous Si crystallised into poly-Si and it was likely that the HfSiO phase separated into HfO2 and
SiO$_2$ rich phases. The wafer surface was exposed to the clean room atmosphere as it was moved between process tools. In particular, this occurred after the HfSiO and metal deposition steps. However, there was no vacuum break between the degas stage and the metal deposition.

Cross-sectional TEM specimens were prepared by cutting, grinding, polishing, dimpling and ion milling. A Gatan PIPS ion mill was used for the thinning. The III-V sample was finished by using 200eV energy ions in a Technoorg-Linda GentleMill. The specimens were examined in an FEI Tecnai TF20 operated at 200 keV. Both conventional transmission (CTEM) and scanning transmission (STEM) modes were used. In the latter mode, a probe half angle of ~8 mrad was used with a spot size in the range 0.2 to 0.7 nm depending on the particular experiment. Electron energy loss spectroscopy (EELS) was performed using a Gatan ENFINA spectrometer. The data was collected in the form of spectrum images using the spectrum imaging plug-in of DigitalMicrograph and the scan was provided by DigiScan via DigiScanII hardware. The dispersion, collection angle and the integration time were chosen in the light of the edges that were to be collected.

Where possible, the edge intensities were extracted after removal of the background using a power law model fitted in front of the edge. However, the edge overlaps and edge shapes often require more detailed processing to extract the intensities and the processing parameters required may change as the composition changes. More details are given with the results below.

3. Results and discussion

3.1. Ga$_2$O$_3$/GGO on InGaAs/AlGaAs/GaAs

Analysis of the GGO itself has been reported previously [16, 17]. Study of the dielectric on a range of heterostructures is now underway. Here an InGaAs/AlGaAs/GaAs heterostructure is used as an example [16]. Figure 1a shows a STEM annular dark field (ADF) image of region around the interface. The sensitivity of the ADF signal to the atomic number of the elements clearly delineates the InGaAs, the AlGaAs, the GaAs, the Ga$_2$O$_3$ and the GGO. However, it is also clear that the interfaces are not atomically flat. Figure 1b shows a lower magnification image with the area selected for the spectrum image acquisition. A 0.7 nm probe was scanned over 12x22 pixels with a step size of 1.4 nm. The half angle of collection was 22 mrad and the dispersion was 1 eV/ch. The spectra were recorded over the energy range 480 eV to 1820 eV. The contribution to the spectrum from the stray scattering in the Schottky field emission gun [18] was minimised by using a small non-beam defining C1 aperture and the extraction voltage and the gun lens settings were chosen so that the stray signal stopped ~200 eV before the Ga L$_{2,3}$ edge threshold. GGO undergoes radiation induced crystallisation which is accompanied by some phase separation. To obtain better signal to noise ratio while keeping within an acceptable dose, sub-pixel scanning was used so that the dose was spread over the 1.4 nm x 1.4 nm pixel rather than being concentrated within the area covered by the 0.7 nm probe. To further improve the signal to noise ratio, the spectra from pixels in lines parallel to the interface were summed. This gives a loss of spatial resolution parallel to the interface but maintains it perpendicular to the interface.

![Figure 1. ADF images of the InGaAs/AlGaAs/GaAs/Ga$_2$O$_3$/GGO gate stack: a) high magnification showing details of the layers; b) lower magnification showing the region of the spectrum image.](image-url)
Figure 2. Elemental profiles across the InGaAs/AlGaAs/GaAs/Ga$_2$O$_3$/GGO gate stack: a) profiles normalized to a maximum intensity of unity; b) profiles normalized by cross-sections.

Figure 2a shows a section of the intensity profiles for the As L$_{2,3}$, Ga L$_{2,3}$, Gd M$_{4,5}$, and O K edges. The profiles have been normalised so that the maximum intensity of each is unity and the origin is chosen at the position where the O profile is ~0.5. This is the position where the As profile is also ~0.5 and so is a good indicator of position of the interface between the substrate and the dielectric. Power law backgrounds were removed from under the As, Ga and O edges. The background fitting windows were chosen in regions of the spectrum image where perturbations are not significant. However, their use to process the whole spectrum image results in negative intensities for O in pixels containing little or no O but significant In (from -10 nm to -5 nm in figure 2a) and also for As in regions containing little or no As but significant Ga and Gd (from 3 nm to 11 nm in figure 2a) These negative counts are not shown. Since the Ga L$_{2,3}$ and Gd M$_{4,5}$ are close together, their intensities were separated by scaling the Ga L$_{2,3}$ shape from GaAs to the region of the Ga edge between the Ga and Gd thresholds [17]. More recent work has allowed the extraction of the In M$_{4,5}$ and the Al K edge intensities and this will be reported elsewhere.

It is clear from figure 2a that the Ga extends to the right of the As at the interface while the Gd also rises to the right of the O, showing the presence of the Ga$_2$O$_3$ template layer. Because the Ga is in both the substrate and the GGO layer, it is hard to estimate the width of the template layer from this profile. However, assuming that no Gd penetrates into it, the separation of the O and Gd profiles gives a direct measure of the template layer thickness of 1.2 nm with an uncertainty of less than 0.2 nm. How is this possible given that the probe size is ~0.7 nm and the step size in the spectrum image is 1.4 nm? The answer is that the interface is actually sampled on a much finer scale because of a number of effects. Interface roughness, beam spreading and slight misorientation of the spectrum image with respect to the interface may all play a role in this. Since the displacement of the O and Gd profiles is essentially constant, the relative positions of the two profiles is a measure of the template layer width and the uncertainty involved is controlled by the uncertainty with which the displacement can be measured rather than by the widths.

Using GaAs, Gd$_3$Ga$_5$O$_{12}$ and Ga$_2$O$_3$ as standards, the intensities at each pixel can be made proportional to the number of atoms per unit area. Figure 2b shows the number of atoms per unit area obtained by this method in arbitrary units. Because the low loss intensity was not recorded, it is not possible to correct for the effects of image contrast and so the constant of proportionality will vary with position. Nonetheless, the atomic ratios at each pixel can still be evaluated. Assuming the oxide has the form Gd$_x$Ga$_{4-x}$O$_{0.6}$, it is straightforward to determine the amount of Ga in the oxide (Ga$_{ox}$) and hence the amount in the substrate (Ga$_{sub}$) and these are also shown on figure 2b. Ga$_{ox}$ extends to the left of the Gd at the interface, as expected, because it is obtained from the original O and Gd profiles where the O extends to the left of the Gd. However, the Ga$_{sub}$ profile follows the As profile closely at the interface, which confirms that the interface between the oxide and the substrate is sharp even if rough. There is a clear difference in the Ga:As ratio in the InGaAs and the AlGaAs and the
convergence of the Ga_{sub} and As profiles at the interface hints at the presence of the two monolayers of GaAs at the interface.

3.2. TiN on HfSiO on Si

After the final thermal processing stage, this system is much less radiation sensitive than the III-V system above. HfO$_2$ and HfSiO dielectrics and poly-Si, TiN and Ta$_2$N gates have all been studied successfully and ELNES has been used to separate out the different phases and identify reactions at the metal/poly-Si interface [19-24]. Here data from the Si/SiO$_2$/HfSiO/TiN/poly-Si gate stack are considered. The data were recorded as a line spectrum image using the Glasgow dual EELS system [25]. This allows both the core loss and low loss regions of the spectrum to be recorded at each pixel before the probe is moved to the next pixel. The ADF signal is recorded at each pixel as part of the dataset. The data were acquired using a step size of 0.51 nm, an energy dispersion of 0.5 eV/channel and energy offsets of 56.5 eV and -543.5 eV for the core and low loss parts of the spectrum, respectively. The core loss and low loss data were each recorded using 15 integrations of 0.5 sec. For the low loss part of the spectrum, the fast beam switch was used to limit the time electrons entered the spectrometer to 3 msec per integration.

Recording both the core loss and low loss regions of the spectrum at the same pixel offers a number of advantages [25]. As a first step, the position of the zero-loss peak can be determined to sub-channel accuracy by fitting a peak to it. Combining this with the value of the voltage applied to the drift tube of the spectrometer when recording the core loss region, all of the spectral data can be aligned in energy to this accuracy giving an accurate energy calibration. If the energy ranges overlap, the two sections of the spectrum can be spliced together. Figure 3a shows the elemental profiles extracted from this spliced spectrum using power law backgrounds and the Si L$_{2,3}$, N K, Ti L$_{2,3}$ and O K edges. Each intensity was divided by the appropriate cross-section. Hartree-Slater cross-sections were used for the Si, N and O edges while a SIGMAL cross-section with white line correction was used for the Ti [26]. They were calculated using the routines in DigitalMicrograph. Thus the values are in units of atoms/nm$^2$ multiplied by the number of electron in the spectrum.

There are a number of issues with the values in figure 3a. The negative values for the Si and O in the TiN layer are due to background perturbations caused by the Ti in combination with multiple scattering. The number of atoms/nm$^2$ of Ti obtained using the SIGMAL cross-section is greater than the that for N despite the fact that comparison to a TiN standard shows that they should be the same provided there is no surface oxidation of the specimen [23]. An arbitrary scaling can be applied to force them to be equal, as in the lower Ti profile in figure 3a. However, there is an issue with surface oxidation and this is considered further below. Finally, it was not possible to extract a reliable Hf edge intensity from the Hf edges in the dataset.

It is possible to deal with these problems. Having the spliced spectrum, it can be Fourier logarithmically deconvoluted [26]. This removes the background problem in front of the Si L$_{2,3}$ edges. It also makes it relatively simple to use the Ti L$_{2,3}$ edges from a TiN standard to extract the O K edge when it is sitting on the Ti L$_{2,3}$ edges. However, in this particular case, there was some surface oxidation and so the underlying Ti contribution was subtracted using a Ti edge shape from the TiN layer in the stack. Thus any O present in the TiN layer does not appear in the resulting profiles. The TiN standard also allows scaling of the Ti to the N. The shape of the O K edge can be used to identify the relative amounts of SiO$_2$ and HfO$_2$ using MLLS fitting of edge shapes from standards. Finally, the profiles can be normalised by the low loss intensity to give the absolute values of the numbers of atoms per nm$^2$, as shown in figure 3b.

Now that the composition is known at each point, it is possible to calculate the local inelastic mean free path, $\lambda$, using the parameterised sum rule method of Malis et al [27]. Combining this with the $t/\lambda$ profile from the low loss data gives the absolute thickness ($t$) profile. In fact there is significant thickness variation due to differential thinning between the TiN and the other phases. Dividing the profiles in figure 3b by the absolute thickness profile gives the absolute numbers of atoms per nm$^3$ and
these are shown in figure 3c. The ADF image intensity recorded as part of the data set is also shown with a vertical offset. The position of the peak in the ADF signal clearly matches that of the Hf peak.

**Figure 3.** a) Elemental profiles across the Si/SiO$_2$/HfSiO/TiN/poly-Si gate stack; a) profiles extracted from the un-deconvoluted spectra and normalized by cross-sections; b) profiles extracted from the deconvoluted spectrum normalized by both cross-sections and the low loss intensity to give the absolute numbers of atoms per nm$^3$ along with the Hf profile extracted using ELNES; c) profiles after further normalizing by the local thickness to give the absolute numbers of atoms per nm$^3$ with the ADF signal shown offset; d) the distribution of phases in the dielectric region based on a mixture of stoichiometric compounds plus the calculated fraction of space filled by this distribution.

From figure 3c, the average number of Si atoms/nm$^3$ in the poly-Si region is 47.8 with a standard deviation of 0.9 compared to the bulk value of 49.9 for crystalline Si (c-Si). However, there is also an average of 1.6 O atoms/nm$^3$ in this region with a standard deviation of 0.8 atoms/nm$^3$. This is likely to be the result of oxidation of the TEM sample surface. The presence of this O will reduce the number of Si atoms/nm$^3$ below that for c-Si. If it is assumed that the O is in the form of SiO$_2$, then there are 0.8 units of SiO$_2$ per nm$^3$ and 47.0 atoms/nm$^3$ in the form of c-Si in the poly-Si. From the bulk density of quartz, the volume occupied by 0.8 SiO$_2$ units is 0.031 nm$^3$ while, from the bulk density of c-Si, 47.0 Si atoms occupy 0.941 nm$^3$. Thus both phases together occupy 0.97 nm$^3$ with a standard deviation of 0.02 nm$^3$ compared to the ideal value of unity i.e. on these assumptions, the accuracy is excellent in the poly-Si region.

However, there is clearly a fortuitous cancellation of systematic errors. The accuracies quoted for the mean free paths and the cross-sections are in the range 10% to 20%. The edge intensities are divided by the product of a cross-section and a mean free path, which is essentially a ratio of cross-sections. While this may itself help to cancel the systematic error, the core loss cross-sections and the inelastic mean free paths are evaluated by two very different approaches with a range of possibilities for the introduction of other systematic effects. In addition, the densities and stoichiometries of thin amorphous films can differ significantly from those in the bulk crystalline forms e.g. the density of quartz is higher than that of amorphous SiO$_2$ and it is not clear that the Si on the surface of the TEM
Turning to the TiN layer, in the centre there are 50 TiN units per nm$^3$ while there are 53 in stoichiometric TiN. Here, there are some unresolved issues with oxidation of the TiN and how it should be treated and these are likely to be the reason for the low value. Until these issues are resolved, it is not possible to know whether the level of agreement will match that in the poly-Si region or whether the degree of cancellation of the systematic effects depends on the local composition.

The interfaces between the TiN and poly-Si in Hf-based gate stacks have been investigated previously [20-24]. As in the case of the III-V sample above, there is a separation of the elemental profiles showing an additional layer where the N extends to the right of the Ti in the region from 10-15 nm in figure 3c. In the same region, the O profile also goes through a peak. MLLS fitting using edge shapes from TiN, SiN$_x$ and SiO$_2$ showed the presence of a Si(O,N) reaction layer in such a case [20,22].

In the dielectric region between the c-Si substrate and the TiN, the elemental distribution is quite complex. The O peaks 0.5 nm to the left of the Hf. The N also extends significantly to the left of the Ti in the region of the Hf peak. This was not observed when the dielectric was HfO$_2$ [20-23] and is therefore associated with nitriding of the HfSiO and, in particular, the phase separated SiO$_2$. The c-Si itself shows some degree of oxidation. If it is assumed that the phases present are c-Si, SiO$_2$, HfO$_2$, Si$_3$N$_4$ and TiN, it is possible to obtain the distribution of these phases, as shown in figure 3d. The numbers of molecular units per nm$^3$ are given on the left hand axis. Note that the Si$_3$N$_4$ profile is associated with three times more Si than the SiO$_2$ profile and so its height is “artificially” suppressed in such a plot. As a cross-check, it is possible to calculate the fraction of space filled assuming the bulk densities of these phases, as was done above for the poly-Si. This is plotted as the dashed line and its scale is given on the right hand axis. Its average value is 1.00 with maximum excursions of ±0.07 but it is clear that systematic effects dominate random ones. Given the complexity of the elemental distributions and the uncertainties involved, this result shows excellent self-consistency.

Looking at the distributions of the phases themselves, the c-Si goes negative in the region of the O peak in figure 3c and is most negative at the position of the maximum of the O peak, suggesting a systematic error in the scaling. The HfO$_2$ profile peaks closer to the TiN while the SiO$_2$ profile peaks closer to the c-Si. This is to be expected since there is always some re-growth of SiO$_2$ by oxidation of the substrate during thermal processing. The Si$_3$N$_4$ profile peaks between those of the HfO$_2$ and the SiO$_2$. This is consistent with N diffusing in from the TiN side. The original HfSiO was deposited to give 70mol%HfO$_2$:30mol%SiO$_2$. However, since there is phase separation of the HfSiO in three dimensions and additional growth of SiO$_2$, the original HfSiO composition is not reflected directly in a line profile and it is not clear that it would be reflected even in a two dimensional map. What can be said is that the HfO$_2$ concentration is much higher than the SiO$_2$ concentration adjacent to the TiN. The HfO$_2$ and TiN distributions overlap significantly indicating that they overlap along the beam path. This is due to the roughness of the interface, which has been seen clearly in images e.g. [21]. At the peak of the HfO$_2$ profile, figure 3c shows that the Si:Hf ratio is close to unity. Overall, these observations are consistent with the HfSiO phase separating by SiO$_2$ moving towards the existing SiO$_2$ on the substrate and leaving the HfO$_2$ against the TiN. Such behaviour has been observed in images of HfSiO after annealing [28]. As the substrate is approached, the composition of the dielectric tends towards SiO$_2$.

4. Conclusions
The results presented show that it is possible to analyse nanometre-scale multilayer systems containing elements with a wide range of atomic number using EELS spectrum imaging with great success. Relative shifts of the elemental profiles that are smaller than the scale of the interface roughness or the spectrum image step size can often be detected. This can be particularly useful where radiation sensitivity is a limiting factor. Use of suitable standards can convert edge intensities into relative
elemental concentrations. However, unless the low loss region of the spectrum is collected as well as the core loss region, corrections cannot be made for the effects of image contrast and local thickness. When using sub-nanometre step sizes, it is essential that the core loss and low loss regions of the spectrum are recorded at the same pixel before moving to the next pixel. If this is not done, the spatial registration is unlikely to be accurate enough to cope with the rapid spatial variations of the signals that are frequently encountered. Access to the complete spectrum allows accurate alignment and calibration of the energy scale and deconvolution of the multiple scattering. This often enables the use of a power law background subtraction where it had not previously been possible and allows use of similarly deconvoluted edge shapes from standards as references for separating overlying edges in the data. Once a procedure has been established for extracting the intensities of the edges, they can be converted to the number of atoms per unit area of specimen by dividing by the intensity of the low loss region and by an appropriate absolute cross-section. This procedure also corrects for the image contrast. In some cases, ELNES on an extractable edge can be used to provide information on an element whose edge intensities cannot be measured directly. If the areal concentration of all the elements is obtained, the composition at each pixel can be determined. This in turn allows the inelastic mean free path to be evaluated and used with the low loss spectrum to calculate the absolute thickness at each pixel. Hence the absolute number of atoms per unit volume can be evaluated. The accuracy here is limited not only by potential systematic errors in the processing but, more importantly, by the systematic errors in the cross-sections and inelastic mean free paths. For the former, reasonable values are available for K cross-sections and moderate ones for L cross-sections. Cross-sections for other edges need to be determined from standards containing the edge of interest and a suitable K (or possibly L) edge. Fortuitously, there seems to be reasonable cancellation of the systematic errors involved in calculating the number of atoms per unit volume and there is good agreement for bulk silicon and reasonable consistency in regions with complex elemental distributions. However, this is an area where improved accuracy is required. Whether this can be achieved directly using theoretical approaches or whether a case by case approach using standards is the only way forward is not yet clear.

The ultimate limit is set by the signal to noise ratio that can be obtained in the data without introducing significant electron beam induced change in the samples. To this end, it is essential that signal acquisition is orchestrated in such a way that the maximum information is recorded per incident electron. This means that the coupling of the specimen to the spectrometer is crucial with the collection half angle being big enough to accept a large fraction of the EELS signal. This problem is exacerbated when going to the very small probe size made possible by aberration correction because the larger probe half angle involved requires a larger acceptance angle. In fact, using aberration correction to obtain a higher probe current rather than a smaller probe is likely to be a better approach provided radiation damage is not the factor limiting the spatial resolution.

If it is elemental analysis that is the prime concern, a low dispersion (>1eV/channel) and the use of losses above 1keV is often useful [24]. For the study of ELNES, high dispersion is needed. Current CCD cameras used for EELS can only record a dynamic range of about 100 if the signal to noise ratio at the low intensity end is to be adequate for signal processing. Thus it is necessary to record two or three sections of the spectrum with different energy ranges and integration times in order to acquire data over the energy range 0 to 3keV, which includes useful edges for all elements up to Bi. To do this without introducing excessive deadtime needs a different approach to that used by the Glasgow dual EELS system. Better approaches are possible e.g. that of Tencé et al [29] and a commercial implementation of one of them is urgently need.

Instrumental artifacts also play a significant role in perturbing the data obtained. Stray scattering within a typical Schottky gun leads to perturbations of the background shape [18] as does scattering within the spectrometer itself. Ideally, a spectrum recorded when the probe passes through a hole in the specimen should show only intensity from the zero-loss peak and elsewhere the intensity should be negligible relative to that in an EELS spectrum from the specimen itself. Commercial systems rarely achieve this and there is always significant intensity present over the whole energy range, arising from
scattering within both the gun and the spectrometer. This is an area where manufacturers must continue to strive for improvement.

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