Controlling oxygen coordination and valence of network forming cations

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Understanding the structure-property relationship of glass material is still challenging due to a lack of periodicity in disordered materials. Here, we report the properties and atomic structure of vanadium phosphate glasses characterized by reverse Monte Carlo modelling based on neutron/synchrotron X-ray diffraction and EXAFS data, supplemented by Raman and NMR spectroscopy. In vanadium-rich glass, the water durability, thermal stability and hardness improve as the amount of P2O5 increases, and the network former of the glass changes from VOx polyhedra to the interplay between VOx polyhedra and PO4 tetrahedra. We find for the first time that the coordination number of oxygen atoms around a V4+ is four, which is an unusually small coordination number, and plays an important role for water durability, thermal stability and hardness. Furthermore, we show that the similarity between glass and crystal beyond the nearest neighbour distance is important for glass properties. These results demonstrate that controlling the oxygen coordination and valence of the network-forming cation is necessary for designing the properties of glass.

Oxide glass components are basically classified into network formers, network modifiers and intermediates by Zachariasen1 and Sun2. Typical network formers satisfy Zachariasen’s rules1, but V2O5 is classified as a network former or intermediate3. This is because it is hard for this oxide to form glass on its own because the oxygen coordination number is five in a crystalline phase, which is larger than the coordination number of typical network formers of three and four. V2O5 based glasses have been widely studied for decades because V2O5 glass is a typical semiconducting glass originating from hopping conduction. Indeed, not only fundamental research on electrical properties4–8 but also applied research on cathode materials for lithium, sodium and magnesium ion batteries have been reported9–12. This glass also has a low glass transition temperature and relatively low thermal expansion4. These are quite attractive properties for low-melting glass used for sealing. In this field, lead borate glass with a high percentage of PbO has been applied to sealing below 400 °C in electronic devices, such as IC ceramic packages, crystal oscillators and micro-electro-mechanical systems. Since lead components are a hazardous substance for human health and the environment, it is necessary to avoid the use of PbO, and hence, V2O5 based glass is a promising material for overcoming this problem13–19. However, there are two critical problems that face the practical application of V2O5 based glass: poor water durability and low thermal stability.

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Several components, i.e., P2O5, TeO2, GeO2, BaO and PbO, can vitrify V2O5 in binary systems. Above all, the most common vitrification component is P2O5 because the V2O5-P2O5 system has the widest range of vitrification in binary systems. Moreover, the properties of this binary system change drastically with the amount of V2O5 content decreases25. In addition, the average V-O coordination number (N(V,O)) for VPx glasses changes from 4.5 (x = 10) to 5.1 (x = 50) as the amount of V2O5 content decreases25.

Table 1. Composition and properties of VPx glasses.

| Sample No. | V2O5/mol% ± 0.5 | P2O5/mol% ± 0.2 | [V4+]/[Vtotal] ± 0.02 | Density/g cm−3 ± 0.0004 | Atomic number density/atom Å−3 | Melting temperature/°C | Thermal stability/°C ± 2 | ΔT |
|------------|----------------|----------------|----------------------|-------------------------|-----------------------------|----------------------|-------------------------|----|
| VP10       | 90.2           | 9.8            | 0.06                 | 2.9712                  | 0.0702                      | 900                  | 224                    | 253 |
| VP19       | 81.0           | 19.0           | 0.14                 | 2.9315                  | 0.0705                      | 950                  | 244                    | 379 |
| VP28       | 71.5           | 28.5           | 0.24                 | 2.8926                  | 0.0709                      | 1000                 | 280                    | 420 |
| VP37       | 63.5           | 37.5           | 0.38                 | 2.8691                  | 0.0716                      | 1000                 | 348                    | 545 |
| VP44       | 55.8           | 44.2           | 0.57                 | 2.8613                  | 0.0723                      | 1150                 | 413                    | None |

Results

Glass compositions, the ratio of V4+/Vtotal, densities and melting temperatures of VPx glass samples are summarized in Table 1. The V4+/Vtotal ratio of the prepared samples increased as the amount of P2O5 increased. This trend is supported by the shift in the energy of the absorption edge of the X-ray absorption near-edge spectroscopy (XANES) spectra at the V-K edge (Fig. S1) and is consistent with the results of a previous study26. However, the values of the previous study differed from those of this study22 because the results were affected greatly by the melting conditions.

The glass transition temperature Tg and crystallization temperature Tc determined from differential thermal analysis (DTA) curves of these samples are shown in Fig. 1a,b and are summarized in Table 1. It is demonstrated from Fig. 1b that these characteristic temperatures and thermal stability ΔT = Tc − Tg increased as the P2O5 amount increased, that P2O5 is an element for improving thermal stability in the V2O5 glass system. In particular, no crystallization peak was observed in the VP44 glass, and the glass softened and flowed to the melting temperature without crystallization. The normalized weight loss in water and Vickers microhardness of these samples, shown in Fig. 1c, suggest that the water durability improved drastically, whilst the glass hardened as the P2O5 amount increased. The atomic number densities and apparent molar volume of O ions for these samples are shown in Fig. 1d. It was confirmed that the atomic number density increased, whereas the apparent molar volume of O ions decreased as the amount of P2O5 increased. Both indicate that the packing density of atoms increased with the addition of P2O5. Indeed, the atomic number density and apparent molar volume of O ions for crystalline V2O5 were 0.0779 Å−3 and 10.83 cm3 mol−1, respectively, suggesting that the packing densities of these glass samples were smaller than that of crystalline V2O5.

Diffraction and EXAFS data. Data on X-ray structure factors, S(Q), neutron total structure factors, S(Q) and EXAFS k3χ(k) measured at the V-K edge for a series of VPx glasses are summarized in Fig. 2a–c, respectively. It is confirmed that all samples are homogeneous, because we cannot observe any small angle scattering...
at $Q < 1 \text{ Å}^{-1}$ in Fig. 2a,b. Three-peak structure was observed for the neutron $S^N(Q)$ for VP100 [glassy (g)-P2O5]26 glass, for which we can assign the peak at $Q = 1.25 \text{ Å}^{-1}$ to a split first sharp diffraction peak (FSDP)27, and the peak at $Q = 2.95 \text{ Å}^{-1}$ to the principal peak (PP)27. A split FSDP was not observed for glassy
a series of VPx glass; Blue, VP44 glass; Cyan, VP100 glass26,29. V-O increased, whereas V-V correlations and O-O correlations in the corner sharing VOwhilst such a doublet feature was not observed in the VP0 glass. Moreover, it was found that the bond length ofthe connectivity of VOxtransformed (FT) EXAFS spectra for VPx glasses. Black, VP0 glass; Red, VP10 glass; Green,VP28 glass; Blue, VP44 glass; Cyan, VP100 glass26,29.

Figure 3. X-ray/neutron diffraction and EXAFS data in real space for VPx glasses together with c-V2O5data. (a) X-ray total correlation functions, \(T^x(r)\). (b) Neutron total correlation functions, \(T^n(r)\). (c) Fourier-transformed (FT) EXAFS spectra for VPx glasses. Black, c-V2O5; Purple, VP0 glass30; Red, VP10 glass; Green,VP28 glass; Blue, VP44 glass; Cyan, VP100 glass26,29.

(q)-SiO2 nor glassy (g)-GeO228, but it was observed for the VP100 glass due to the formation of a Q1 network, in which a phosphorous atom has three bridging oxygen atoms and one bridging oxygen atom. In contrast, weobserved only a two-peak structure (FSDP at \(Q = 1.4 \text{ Å}^{-1}\) and PP at \(Q = 2.1 \text{ Å}^{-1}\)) in the X-ray \(S^x(Q)\) for the VP100 glass29 because a PP was not observed in the X-ray data due to small weighting factors of oxygen-related correlations for X-rays. Intriguingly, this complicated peak structure disappeared in the other VPx glasses, and it showeda sharp PP at \(Q = 2.7 \text{ Å}^{-1}\) in the neutron \(S^n(Q)\) and a relatively sharp FSDP at \(Q = 1.8 \text{ Å}^{-1}\) in the X-ray \(S^x(Q)\). The EXAFS data measured at the V-K edge differed substantially, suggesting that the local environment of V atomschanged with the glass composition.

The X-ray total correlation functions, \(T^x(r)\) and neutron total correlation functions, \(T^n(r)\), for a series of VPx glasses are shown in Fig. 3a,b, respectively, together with the data for crystalline (c)-V2O5. We can see an excellent contrast between the X-ray and neutron diffraction data because X-rays are sensitive to vanadium, whilst oxygen can be easily detected with neutrons, and it is very difficult to detect vanadium with neutrons. In addition, it ispossible to detect both P-O and V-O correlations with X-rays, as can be seen in the Fig. 3a, where a tiny negativeV-O correlation peak can be observed at \(r = 1.7 \text{ Å}\) in the neutron \(T(r)\). Indeed, only oxygen atoms can be detectedwith neutrons, and this allows us to observe both corner-sharing and edge-sharing O-O correlations in c-V2O5.

The most striking feature in the X-ray \(T(r)\) is the big difference in V-O correlation peaks between the VP0 glass[glass] (g)-V2O5 and c-V2O5. In the case of c-V2O5, both V-O and V = O bonds were clearly distinguished,whilst such a doublet feature was not observed in the VP0 glass. Moreover, it was found that the bond length ofV-O increased, whereas V-V correlations and O-O correlations in the corner sharing VO4 polyhedra, and theP-O distance decreased as the P2O5 fraction increased. These results are consistent with the results reported byHoppe et al. for the VP27 and VP50 glasses25, indicating that the structural units of VO4 and PO4 polyhedra and theconnectivity of VO4-VO4 and VO4-PO4 in each glass change with the glass composition. Figure 3c shows Fourier-transformed (FT) EXAFS spectra for the VPx glasses, in which significant composition-dependent modification of the local environment around the V atoms was observed in real space, too.

Structural model of the glasses. We performed RMC modelling on the basis of experimental data touncover the relationship between glass structures and properties. As can be seen in Fig. S2, RMC-modelled X-ray\(S^n(Q)\), neutron \(S^x(Q)\) and V-K edge EXAFS \(k^x(\chi)(k)\) data for the VP0 (g-V2O5)30, VP10, VP28, VP44, and VP100(g-P2O5)31 glasses agreed well with experimental data. Figure 4 compares the partial structure factors, \(S_{ij}(Q)\), of aseries of VPx glasses calculated from the RMC models. First of all, we address partial structures in the VP0(g-V2O5) and VP100 (g-P2O5) glasses. It is well known that the FSDP of network formers, e.g., g-SiO2 and g-GeO2,shows up a positive peak in each \(S_{ij}(Q)\)28. However, the FSDP of the \(S_{PP}(Q)\) and \(S_{OO}(Q)\) was a doublet28, and both


Figure 4. RMC-generated partial structure factors, $S_g(Q)$, for VP$x$ glasses. Purple, VP0 glass; Red, VP10 glass; Green, VP28 glass; Blue, VP44 glass; Cyan, VP100 glass.

the X-ray and neutron $S(Q)$ did not exhibit a well-defined FSDP for VP100 glass. The reason for this is due to the different length scale that arose from both the P-O bonds (1.58 Å) and P=O bonds (1.43 Å) in the glass (see Fig. 3b), as discussed by Hoppe. In comparison, a sharp positive FSDP observed at $Q \approx 1.8$ Å$^{-1}$ for three $S_g(Q)$ in the VP0 glass, which is related to a V-O correlation peak that is relatively symmetrical in comparison with the VP0 glass. Other interesting behaviour is the very sharp negative and positive peaks observed at $Q = 1.38$ Å$^{-1}$ and $Q = 2.60$ Å$^{-1}$ in the $S_{g,v}(Q)$ of the VP44 glass, suggesting that the network was formed by the interplay between PO$_4$ tetrahedra and VO$_x$ polyhedra.

Partial pair-distribution functions, $g_{ij}(r)$, of a series of VP$x$ glasses calculated from the RMC models are shown in Fig. 5. The first P-O correlation peaks observed at $r = 1.6$ Å for the VP10, VP28 and VP44 glasses were more symmetrical in comparison with the VP100 glass, suggesting that electrons were more delocalized in the VP10, VP28 and VP44 glasses. A significant composition-dependent change was observed for $g_{O-O}(r)$, $g_{V-O}(r)$ and $g_{V-V}(r)$, and it was confirmed that this reflected the changes in the experimental $T(r)$. The first P-V correlation peak observed at 3.2 Å became sharp as the P$_2$O$_5$ fraction increased, which was in line with the behaviour in Q space, suggesting the interplay between PO$_4$ tetrahedra and VO$_x$ polyhedra in forming the network.

The average coordination numbers calculated up to 2.5 Å and the ratio of edge-sharing VO$_x$ polyhedra calculated from the RMC models are summarized in Table 2. The oxygen–cation coordination number $N_{O,M}$ ($M = V, P$), the V-O coordination number $N_{c-O}$ and the ratio of edge-sharing VO$_x$ polyhedra increased as the amount of P$_2$O$_5$ increased. It is worth mentioning that 15% of VO$_x$ polyhedra were edge-shared in the VP44 glass, which is outside of Zachariasen’s rule. These behaviours are consistent with the fact that the packing density of atoms in glass increased as the P$_2$O$_5$ fraction increased. The atomic configurations together with the cavity volume of a series of VP$x$ glasses are shown in Fig. 6. It is noted that the cavity volume increased as the amount of P$_2$O$_5$ increased, whilst packing density increased.

The $Q^2$ distribution of V and P species calculated from the RMC models are shown in Fig. 7a,b. Although the $Q^2$ distribution of V species shifted to a smaller number of $n$ as the amount of P$_2$O$_5$ increased, the V species had a high proportion at $n \geq 2$ (Fig. 7a). In comparison, the P species were almost isolated PO$_4$ ($Q^2$) tetrahedra not only in the VP10 glass but also in the VP44 glass, as shown in Fig. 7b. These behaviours are supported by the Raman spectra as shown in Fig. S3, in which a peak assigned to $Q^2$ was observed for the VP10 and VP44 glasses.

The ring statistics of VP$x$ glasses are shown in Fig. 7c. It was demonstrated that the normalized number of V-O rings decreased as the amount of P$_2$O$_5$ increased, whereas P-O rings did not form in the range of glasses VP10 to VP44. In contrast, the normalized number of M-O rings increased as the amount of P$_2$O$_5$ increased. These results strongly indicate that the network former of the VP$x$ glasses changed from VO$_x$-VO$_4$ networks to the interplay between VO$_x$ and PO$_4$ for the VP28 and VP44 glasses. These changes in the network structure and edge-shared VO$_x$ polyhedra were clearly visible in the three-dimensional atomic configurations obtained from the RMC models, as shown in Fig. 7d,e.
Figure 5. Short-range structural analysis on RMC-generated models for VPx glasses. Partial pair-distribution functions, $g_{ij}(r)$. Purple, VP0 glass; Red, VP10 glass; Green, VP28 glass; Blue, VP44 glass; Cyan, VP100 glass.

Table 2. Coordination numbers of VPx glasses obtained by RMC models.
The fraction of VO\(_x\) polyhedra derived from the RMC models is shown in Fig. 8a. The figure clearly shows that the majority of short-range structural units in the VP\(_x\) glasses consisted of VO\(_4\) and VO\(_5\) units. It was also found that the fraction of VO\(_4\) units decreased, whilst the fraction of VO\(_5\) units increased systematically as the amount of P\(_2\)O\(_5\) increased. Taking the vanadium valence into account, VO\(_x\) polyhedra were divided into six units, V\(^4+\)O\(_4\)\(-6\) and V\(^5+\)O\(_4\)\(-6\). To obtain information on the local structure around the V\(^5+\) in the VP\(_x\) glasses, NMR spectroscopic measurements were performed. \(^{51}\)V NMR spectroscopy can selectively detect the diamagnetic V\(^5+\) state because the V\(^4+\) state cannot be analysed directly due to its paramagnetism. The \(^{51}\)V MAS-NMR spectra of the VP\(_x\) glasses are shown in Fig. 8b. The spectra were obtained by RMC modelling and selected by the P\(_2\)O\(_5\) content.

**Figure 8.** Distribution of oxygen coordination number and valence of vanadium. (a) Total fraction of VO\(_4\), VO\(_5\) and VO\(_6\) units obtained by RMC modelling. (b) Fraction of V\(^{5+}\)O\(_{4,6}\) and V\(^{4+}\)O\(_{4,6}\) units calculated by RMC modelling and \(^{51}\)V NMR spectroscopy.

The fraction of VO\(_x\) polyhedra derived from the RMC models is shown in Fig. 8a. The figure clearly shows that the majority of short-range structural units in the VP\(_x\) glasses consisted of VO\(_4\) and VO\(_5\) units. It was also found that the fraction of VO\(_4\) units decreased, whilst the fraction of VO\(_5\) units increased systematically as the amount of P\(_2\)O\(_5\) increased. Taking the vanadium valence into account, VO\(_x\) polyhedra were divided into six units, V\(^{4+}\)O\(_{4,6}\) and V\(^{5+}\)O\(_{4,6}\). To obtain information on the local structure around the V\(^{5+}\) in the VP\(_x\) glasses, NMR spectroscopic measurements were performed. \(^{51}\)V NMR spectroscopy can selectively detect the diamagnetic V\(^{5+}\) state because the V\(^{4+}\) state cannot be analysed directly due to its paramagnetism. The \(^{51}\)V MAS-NMR spectra of the VP\(_x\) glasses are shown in Fig. 8b.
glasses are shown in Fig. S4. The peaks observed at around $-750$, $-540$, $-490$ and $-300$ ppm can be attributed to symmetric-V$^{2+}$O$_5$ ($s$-VO$_5$), distorted-V$^{3+}$O$_5$ ($d$-VO$_5$), V$^{5+}$O$_4$ (VO$_4$) and V$^{5+}$O$_3$ units (VO$_3$), respectively$^{1-3}$. The fractions of these units, as calculated from the area of each peak, are summarized in Table S1. It is suggested for the V$^{5+}$-state in the VPx glasses that the fraction of V$^{5+}$O$_3$ units decreased as the amount of P$_2$O$_5$ increased, whereas the fraction of V$^{5+}$O$_4$ units increased. The fractions of V$^{5+}$O$_4$ and V$^{5+}$O$_3$ units calculated from the RMC models, $^{31}$V NMR spectroscopy and the fraction of V$^{4+}$/V$_{total}$ are shown in Fig. 8b. The result clearly indicates that the unusually small coordination number of V$^{4+}$O$_4$ units existed in the VPx glasses and the fraction increased as the amount of P$_2$O$_5$ increased, although it has been cited for the V$^{4+}$ state in glass that only V$^{4+}$O$_4$ or V$^{4+}$O$_3$ units are taken as the same crystals due to the larger ionic radius of V$^{4+}$ 35–38. Such unusually small coordination numbers in oxide glasses are also reported for MgO-SiO$_2$ glass, CaO-Al$_2$O$_3$, and ZnO-P$_2$O$_5$ glasses and, hence, it is suggested that this may be a characteristic feature in some oxide glasses with atypical network formers.

Discussion

We set out to model the glass structure to understand the water durability, thermal stability and hardness, which are important in practical applications. The water durability of VPx glasses improves as the amount of P$_2$O$_5$ increases, even though VP0 (g-V$_2$O$_5$) and VP100 (g-P$_2$O$_5$) glasses readily dissolve into water$^{31,42,43}$. Gin et al. reported on the water durability of glass divided into several stages$^{44}$. This improvement of VPx glasses in terms of water durability was determined to be due to the suppression of the hydrolysis process, which is affected by the structure of VO$_4$ and PO$_4$ polyhedra and the valence of vanadium. The cavity volume in the glass did not seem to significantly affect the water durability because, when the cavity volume increased, the water durability improved. Regarding the structure of VO$_4$ polyhedra and valence of vanadium, Nabavi et al. suggested that the V$^{5+}$O$_4$ units in amorphous V$_2$O$_5$ are highly reactive towards water, whereas the V$^{5+}$O$_4$ units in crystalline V$_2$O$_5$ are not. Moreover, they also suggested that the water durability of V$^{5+}$O$_4$ units decreased, which implies that water durability is affected by V-O coordination$^{35}$. Our results also show that there is a clear correlation between normalized weight loss in water and the fraction of V$^{5+}$O$_4$ units in all cations (see Fig. S5). In other words, we suggest that V$^{4+}$O$_4$ units are stable against water, whereas V$^{5+}$O$_4$ units are highly reactive towards water. We conclude that this might be affected by the polarizability of VO$_4$ polyhedra, as in the case of PO$_4$ polyhedra. Therefore, the increase in edge-sharing VO$_4$ polyhedra might improve the water durability. Regarding the structure of PO$_4$ polyhedra, several papers reported that phosphate glass that has PO$_4$ tetrahedra with Q0 or Q1 units shows good water durability, suggesting that the low polarizability in PO$_4$ tetrahedral units makes it hard to attract polar water molecules$^{45-47}$. Therefore, the reason that the water durability improved in the range of glasses VP0 to VP44 is because the structures of the PO$_4$ tetrahedral units were only Q0 and Q1, whereas the Q3 and Q5 units in all cations (see Fig. S5).

The thermal stability, an indirect indicator of glass forming ability, of VPx glasses improves as the amount of P$_2$O$_5$, which indicates that the driving force of crystallization is reduced. Since a driving force arises from the difference in chemical potential between the glass state and crystalline state, our result indicates that the difference becomes smaller as the amount of P$_2$O$_5$, increases. We conclude that there are two reasons for this. One reason is that the atomic structure of the glasses is very far away from the conventional oxide glasses classified by Zachariasen$^1$ and Sun$^2$, and the similarity with a crystalline structure increases as the amount of P$_2$O$_5$ increases. The similarity between glass and crystal is often discussed for the fast phase-change materials used for DVD/Blu-ray media, where the topology in terms of the ring-size distribution between glass and crystal is similar, whilst coordination in the glassy phase is significantly lower than that in the crystal phase. Therefore, it is concluded that this similarity is the crucial reason for the rapid phase change between two phases$^{48}$. Indeed, the increased packing density and large fraction of edge-sharing in the glassy state are a remarkable signature of P$_2$O$_5$-rich glass. Accordingly, these behaviours demonstrate that the similarity between glass and crystal beyond the nearest neighbour distance (for example, polyhedral connection and atomic packing density) is rather important for the glass properties because the first V-O coordination number and the valence of vanadium of VPx glasses are different from those of crystal. In other words, this indicates that the similarity beyond the nearest neighbour distance is important for decreasing the difference in chemical potential between the glassy state and crystalline state. The other reason is the high ratio of V$^{4+}$/V$_{total}$ in the VPx glasses. Sakurai et al. also reported that the thermal stability of VPx glasses improves as the amount of P$_2$O$_5$ increases, but that of their VPx glasses is significantly lower than that of our VPx glasses$^{25}$. Although the packing densities of their VPx glasses were not mentioned in ref. 25, the ratio of V$^{4+}$/V$_{total}$ in their VPx glasses was lower than that of our VPx glasses. Accordingly, we suggest that the high ratio of V$^{4+}$/V$_{total}$ is an important factor for improving thermal stability, too. However, it is still not certain whether V$^{4+}$ states themselves reduce the chemical potential between glass and crystal or that they increase the atomic packing density and edge-sharing VO$_4$ polyhedra.

The hardness of the VPx glasses also improves as the amount of P$_2$O$_5$ increases. We conclude that this should be explained by an increased packing density associated with a higher V-O-M coordination number and increase in edge-sharing VO$_4$ polyhedra. Indeed, Rosales-Sosa et al. reported that a high atomic packing density and dissociation energy per unit volume of components increases the values of hardness and the elastic modulus$^{49}$. We calculated the atomic packing density in Fig. S6. It was found that the increase in P$_2$O$_5$ content caused the glass structure to have a higher packing density with increased V-O-M coordination numbers and increase
in edge-shared VO\(_\text{4}\) polyhedra. The dissociation energy per unit volume of components should increase with the number of bonds per unit volume of components. Furthermore, the reduction in vanadium ions might affect the dissociation energy, whilst the cavity volume in glass should be independent of hardness because it does not affect the packing density and dissociation energy.

In addition, we have previously reported that the thermal expansion coefficients of VP\(_x\) glasses did not change significantly despite an increased glass transition temperature as the amount of P\(_2\)O\(_5\) increased\(^{21}\). Generally, the thermal expansion coefficient of glass materials increased as the glass transition temperature decreased. Such an anomalous coefficient was also identified with the network configuration in ZnO-P\(_2\)O\(_5\) glass systems\(^{41}\). We calculated the fractions of M-O ring distributions as shown in Fig. S7. The distribution remained the same in the range of glasses VP10 to VP44. Hence, our results imply that this anomalous thermal expansion coefficient was related to the fractions of the distributions only because of a structural feature that does not change in the range of glasses VP10 to VP44. These results for the structure-property relationship demonstrate that the properties of glass are strongly affected by the structure and valence of network forming cations.

### Conclusion

In this article, we discussed the relationship between the properties and atomic structure of V\(_2\)O\(_5\)-P\(_2\)O\(_5\) glass as characterized by RMC modelling on the basis of neutron and synchrotron X-ray data. The present findings indicate that the structure and valence of network formers is important for designing their properties. Adding P\(_2\)O\(_5\) causes the packing density of atoms to increase and the amount of vanadium ions to decrease in V\(_2\)O\(_5\)-rich glass, resulting in a glass structure that is associated with an increase in edge-sharing VO\(_\text{4}\) polyhedra. We find that these are important for improving the water durability, thermal stability and hardness. In particular, the valence and structural change of vanadium ions affect the change in the water durability and thermal stability. We are confident that the unusually small coordination number of V\(^{4+}\) is especially important for water durability and that the similarity between glass and crystal beyond the nearest neighbour distance is important for thermal stability. The results presented in this study are a significant advance in understanding the fundamental properties of glass materials. Furthermore, this work paves the way towards glass sealing materials being completely lead-free and cathode materials for secondary batteries being improved.

### Materials and Methods

#### Sample preparation.

Glass samples with a nominal molar composition of (100 \(−\) \(x\))V\(_2\)O\(_5\)-xP\(_2\)O\(_5\) were synthesized by melt quenching of V\(_2\)O\(_5\) and P\(_2\)O\(_5\) powders (Kojundo Chemical Laboratory Co., Ltd.). 100 g of the mixture was melted in a platinum crucible and kept at melting temperature for one hour. The molten glass was cast onto a stainless-steel plate at 100°C. The prepared glass samples were annealed at 10°C higher than the glass transition temperature and slowly cooled at 1°C min\(^{-1}\) for 40 min to relieve residual internal stress. V\(_2\)O\(_5\) glass was synthesized using a twin roller method. The oxide powder was melted at 800°C for 5 min in a platinum crucible and then rapidly quenched by rotating rollers.

#### Characterization.

The cation compositions of the resulting material were determined by wavelength dispersive X-ray fluorescence spectrometers (Rigaku, ZSX Primus II). The resulting material was fully amorphous, and this was confirmed by X-ray diffraction (XRD) using a diffractometer system equipped with a monochromatic Cu Kα radiation source (Rigaku, RINT-2000). The fraction of the reduced amount of V ion ([V\(^{4+}\)/[V\(_\text{total}\)]) was estimated by measuring the quantities of pentavalent vanadium ions [V\(^{5+}\)] and total vanadium ions [V\(_\text{total}\)] by oxidation-reduction titration, assuming that the V ions in the glass consisted of V\(^{5+}\) and V\(^{4+}\). The densities of the glass samples were measured using a dry pycnometer (Micromeritics, AccuPyc II 1340). The atomic number densities of these samples were calculated from the density and V\(^{4+}/V\(_\text{total}\)] ratio. The apparent molar volumes of O ions were also calculated from the density and V\(^{4+}/V\(_\text{total}\)] ratio by using the formula described by Drake et al.\(^{51}\). The glass transition temperature (\(T_g\)) and crystallization temperature (\(T_c\)) were measured by using differential thermal analysis (DTA) (Advance Riko, DT-1500) at a heat rate of 5°C min\(^{-1}\). The weight loss in water at 70°C for 30 min for the glass sample plates was measured to determine the water resistance. Vickers microhardness measurements were made on the sample surfaces at room temperature by using an auto hardness test system (Matsuzawa, AMT-X7FS) with an accuracy of ±0.04 GPa. A load was applied for 0.98 N, 15 s.

#### Structural analysis of glass.

High-energy X-ray diffraction experiments (HEXRD) were carried out at room temperature by using the BL04B2 beamline of SPring-8\(^{35}\). The incident X-ray energy was 61.6 keV as obtained from a Si(220) crystal monochromator. Diffraction patterns of the samples, placed in a silica glass tube, were measured in a transmission geometry. The collected data were corrected by using a standard program\(^{32}\). X-ray absorption fine structure (XAFS) experiments were carried out by using the BL14B2 beamline of SPring-8\(^{36}\). The XAFS samples were ground with boron nitride and made into pellets. V-K edge spectra were measured by using a Si(111) double-crystal monochromator in transmission mode. Ionization chambers were used to measure the intensity of the incident and transmitted X-rays, and the quick scan technique (QXAFS) was used for this measurement. These spectra were normalized and analysed by using Athena\(^{54}\). Time-of-flight neutron diffraction (ND) experiments were conducted by using the total scattering spectrometer NOVA at the BL21 beamline of the Materials and Life Science Experimental Facility (MLF), Japan Proton Accelerator Research Complex (J-PARC)\(^{55}\). The samples were placed in a cylindrical vanadium cell (6 mm in diameter). The observed scattering intensities for the samples were corrected for instrumental background, absorption of samples and cell\(^{56}\), and multiple\(^{57}\) and incoherent scatterings and then normalized by the incident beam profile, which obtained from the scattering intensity for a vanadium rod. \(^{51}\)V spectra were measured at 11.7 T (JEOL, ECA-500 FT-NMR). A NaVO\(_3\) aqueous solution (0.16 mol dm\(^{-3}\)) was taken as a reference (\(b = −574.28 \text{ppm}\)).
The pulse delay was 1 s. Raman spectra were recorded by using a LabRam spectrometer (Jobin-Yvon). A laser with an emission wavelength of 514.5 nm and a power of 0.2 mW was used.

**Structural modelling.** The RMC modelling was performed on an ensemble of 3500, 4139, 3898 and 3818 particles for VP0, VP10, VP28 and VP44 glass, respectively. The starting configurations were generated by hard-sphere Monte Carlo simulations with constraints applied to avoid physically unrealistic structures. The constraints on the P-O connectivity were that all phosphorus atoms were coordinated for four oxygen atoms for atomic distances of up to 1.7 Å. X-ray S(Q), neutron S(Q) and k^3(Q) EXAFS data measured at the V-K edge were fitted simultaneously by using RMC + code58. EXAFS back scattering tables were obtained from FEFF calculations59. The ring statistics were calculated by primitive rings analysis60–62 using R.I.N.G.S. code63. Cavity analysis was carried out by employing pyMoDyn code64. The cut-off distance r_{c} for cavity calculation was 2.3 Å.

**Data availability**

The datasets generated during and/or analysed during the current study are available from the corresponding author upon reasonable request.

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Author contributions

T.A., S.K. and T.N. designed and directed this study. T.A. and T.M. prepared and characterized the samples. S.K., Y.O. and K.O. carried out the synchrotron X-ray diffraction experiment. T.A., M.K., T.O., D.T. and T.U. carried out XAFS measurements and analysed data. T.A. and K.S. conducted the neutron diffraction experiment. T.A., S.K. Y.O. and S.T. performed RMC modelling and analysed the results. O.S., T.M., Y.H. and H.T. advised in this study. T.A., S.K. and Y.O. wrote the paper.

Competing interests

The authors declare no competing interests.

Additional information

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