5.2 Relationships between microstructure and mechanical properties

In this section mechanical properties of the LPSSiC materials are discussed, with the view to ultimately relate properties to the microstructure of the material. Mechanical properties that were measured and correlated to microstructure were hardness, fracture toughness, strength and elastic modulus.

The microstructural features that were related to the mechanical properties are as follows:
1) SiC grain size
2) grain boundary phase content
3) grain boundary phase composition

5.2.1 Microstructure to hardness relationships

Figure 5.28 shows the difference in hardness between the hot pressed, gas pressure sintered and heat treated hot pressed materials. Figure 5.28 shows that the hardness of the hot pressed materials are all similar, independent of their grain sizes (the differences in mean silicon carbide grain size can be seen in Figure 5.18). The hardness of the different gas pressure sintered materials was also found to be similar for the different compositions and also independent of SiC grain size. An exception is the GP4Y2Al material, which has a very low hardness due to the fact that it could not be fully densified (discussed in section 5.1).
Figure 5.28 Graph comparing hardness of hot pressed materials with those, which were heat treated and gas pressure sintered.

The hardness of the gas pressure sintered materials was higher than that of the hot pressed materials, as illustrated in Figure 5.29, despite the larger silicon carbide grain size in the gas pressure sintered materials, seen in Figure 5.18. This is contrary to what is predicted by the Hall-Petch relationship\(^{(112)}\), which predicts a decrease in hardness with increasing grain size. Vassen et al.\(^{(112)}\) observed a decrease in hardness with an increase in grain size in SSiC materials, in agreement with the Hall-Petch relationship. SSiC materials do not contain a secondary “grain boundary” phase as LPSSiC materials do. Baud et al.\(^{(165)}\) observed in their pressureless sintered SiC materials with 3 mass-% YAG additive that the Hall–Petch relationship was obeyed and that hardness was independent of grain shape (elongated grains formed during heat treatment). The silicon carbide grain size range in their sintered materials was 1.8 - 3.5 \(\mu\)m, which is similar to the silicon carbide grain size range determined to be in the materials in this research work. This could be because in their
study, the amount of grain boundary phase was too low (3 mass-%) to expect it to have any significant influence on the hardness of their materials. 3 mass% is also too low for the degree of crystallinity and relative phase compositions in the grain boundary phase to be determined accurately.

As seen in Figure 5.28, the hardness of the hot pressed materials increased during the first heat treatment (HT2), and then did no change significantly in subsequent heat treatments (HT3 and HT4). There was a decrease in hardness when the hot pressed material was heat treated at higher temperature (HT5, 1975°C for 5 hours). On the other hand, the hardness of gas pressure sintered materials did not change significantly during any of the post-sintering heat treatments.

Sciti et al.\(^{(107)}\), sintering with 6 – 10 mass-% YAG, also observed such an increase in hardness after heat treatment and attributed it to the decrease in the amount of grain boundary phase due to decomposition reactions during post-sintering heat treatment. They observed mass losses in the
range of 3-5.5 mass-%, which is higher than the mass losses observed in this research work (2.4-4 mass-%). The highest and lowest mass losses (4 mass-% and 2.4 mass-% respectively) were related to a similar increase in hardness after heat treatment of HP materials in this thesis. Therefore, it cannot be concluded that mass loss and consequent increased interconnectivity of silicon carbide is the sole reason for increase in hardness on heat treatment. Sciti et al. detected only YAG in their hot pressed material, but also acknowledged the existence of amorphous silicates in the hot pressed material. They did not however, discuss the effect of the silicates on the hardness of the LPSSiC material.

In this work, it is believed that the main reason for the increase in hardness, was the change of composition of the grain boundary phases during heat treatment. The grain boundary compositions range from silicates ($Y_2Si_2O_7$ and $Y_2SiO_5$) in the as-hot pressed materials, to aluminates ($Y_3Al_5O_{12}$, $YAlO_3$ and $Y_4Al_2O_9$) in the heat treated materials, as well as in the gas pressure sintered materials (given in Table 4.12).

The exact hardness values of all the silicates and aluminates observed in this work could not be found. However, the relative hardnesses of the different grain boundary phases can be arrived at by the following arguments. The hardness of a fused silica glass is 6.2 GPa$^{(207)}$, and the Vickers hardness of yttria-alumina-silica glasses is between 6.36 and 8.28 GPa$^{(177)}$. On the other hand, the hardness of alumina is approximately 20 GPa. Using the Mohs scale of hardness, an example of potassium aluminosilicate is 2-3 and fused silica is around 5.5, compared to YAG which is 8.5 in the Mohs scale of hardness.

These differences in hardness (between the different grain boundary phases in the present work) can also be explained by looking at the crystal structures of the grain boundary phases crystallised at the grain boundary. $Y_2Si_2O_7$ has only two tetrahedra with one common edge of atomic
bonding, with ionic bonding forces holding the structure together. \( \text{Y}_2\text{SiO}_5 \) consists of isolated tetrahedra. The aluminates, on the other hand, consist of a three-dimensional network of alumina octahedra held together by covalent bonding, which is stronger than the ionic bonding in the silicates. Therefore it can be concluded that the structures and bonding in the aluminates, makes them harder phases than the silicates. This could account for the fact that hardness of GPS materials is higher than HP materials, and that the hardness of HP materials increased after heat treatment.

During the 5 hour heat treatment at 1975°C (HT5), the grain growth is much higher than in any of the other heat treatments, resulting in the HT5 materials having more and much larger elongated grains, than any of the other heat treated materials. These coarse grains could be the reason for the slight decrease in hardness during the heat treatment at 1975°C. Another reason for the decrease in hardness, could be the higher mass losses observed in the materials heat treated at higher temperature, which gives rise to more porosity in these materials, as shown in the example microstructure in Figure 5.30.
The hardness of gas pressure sintered materials, did not change significantly during heat treatment for a given grain boundary additive composition, as shown in Figure 5.31. Mass loss of grain boundary material (Figure 5.14) during heat treatment of gas pressure sintered materials is not as high as in heat treated hot pressed materials. This is related to the fact that there is no significant change in the amount (only small mass loss) and composition of the grain boundary phase of the gas pressure sintered materials during heat treatment. Table 4.12 shows the aluminates grain boundary phases in the GPS materials remaining in the GPS heat treated materials and only changing slowly to phases with less Al₂O₃ content. The microstructure of the GPS materials, in terms of SiC grain size (Figure 5.21) and in terms of grain boundary phase composition (Table 4.12), does not undergo as big a change as the HP materials (Figure 5.19 and Table 4.12) do during heat treatment.

Figure 5.30  FESEM image of HT5HP1Y4Al material, showing the increased porosity after heat treatment at 1975°C
Figure 5.31 Hardness of gas pressure sintered materials and materials after post-sintering heat – treatment

The results of the Vickers indentation tests done at high temperatures are shown in Figure 4.13. These results show that hardness decreases monotonically up to 1200°C. At this temperature neither crystallite size nor grain boundary phases change. This decrease in hardness can be attributed to two factors. Firstly, in LPSSiC materials, such as the HP3Y5Al material studied in this project, the volume fraction of grain boundary is high, around 9 vol.-% (values can be seen in Table 5.1), and secondly distance between silicon carbide grains is large enough that grain boundary deformation takes place. The HP3Y5Al material contains YAG and most likely some amorphous silicate in its grain boundary. At higher temperatures, up to 1200°C, at which the hardness was measured, the silicates in these grain boundary phases will soften, i.e. decrease in viscosity. Keppeler et al.\textsuperscript{(183)} observed a decrease in strength above 1000°C, which they attributed to softening of the grain boundary phase. For example SiO\textsubscript{2} glasses have a viscosity of 10\textsuperscript{4}dPa.s at 1300°C\textsuperscript{(208)}. This
softening results in increased grain boundary diffusion, decreasing the bulk hardness of the material at higher temperatures. This decline in hardness of the material is because of grain boundary softening and sliding\(^{(209)}\). This effect of temperature on the hardness of the LPSSiC HP material, is also evidence for the fact that the grain boundary has a significant influence on the hardness of LPSSiC materials.

Another contribution to the decline in hardness at high temperatures could be dislocation movement within silicon carbide grains during indentation. In solid-state sintered materials in which grain boundary phases are not prevalent, thermally activated processes such as dislocation climb and dislocation glide reduce hardness of the materials\(^{(112)}\). Vassen et al.\(^{(112)}\) observed such a decrease in hardness with increase in temperature of their SSiC materials. The dislocation glide and climb as observed by Vassen et al.\(^{(112)}\) can contribute to the decrease in hardness of LPSSiC materials.

### 5.2.2 Microstructure to fracture toughness relationships

The fracture toughness of the different composition hot pressed materials are similar, as is shown in Figure 5.32. This is because the grain sizes are similar in the different hot pressed materials, except for the HP1Y4Al material, which has slightly higher mean SiC grain size than the other HP materials, and contains β-Y\(_2\)Si\(_2\)O\(_7\) (3.3 vol-%) and Al\(_2\)O\(_3\) (6.3 vol-%) not found in the other HP materials. The grain size distributions of all the hot pressed materials are narrow (Figure 5.22), and this is not expected to favour toughening mechanisms, even with a slightly higher grain size. According to Figure 5.32, the fracture toughness of the hot pressed materials heat treated for 5 hours at 1925\(^\circ\)C, only appears to slightly increase for the 1Y\(_2\)O\(_3\):4Al\(_2\)O\(_3\) composition.
There was no significant difference in fracture toughness between the different compositions in the gas pressure sintered materials, as shown in Figure 5.33. A significant increase in $K_{IC}$ after heat treatment at 1925°C (5 hours) was observed in the 1Y$_2$O$_3$:4Al$_2$O$_3$ and the 3Y$_2$O$_3$:5Al$_2$O$_3$ (YAG) compositions. The microstructural analyses of these materials show that grain growth was higher in these compositions (especially 1Y$_2$O$_3$:4Al$_2$O$_3$).

Figure 5.32 Indentation fracture toughness of as-hot pressed materials and hot pressed materials which were heat treated at 1925°C for 5 hours (V2)
The fracture toughness of the hot pressed, gas pressure sintered and heat treated hot pressed materials (up to 5 hour heat treatment at 1925°C) are similar despite their very different grain boundary phases (Table 4.12). The grain boundary phases in the gas pressure sintered and heat treated gas pressure sintered materials are similar, yet there is a definite increase in fracture toughness when the gas pressure sintered materials are heat treated for 5 hours at 1925°C. If the grain boundary material had a significant influence on fracture toughness, an increase in fracture toughness would have been observed after heat treatment of the hot pressed materials at 1925°C even after the shorter heat treatment times, which was not observed. The relationship between fracture toughness and silicon carbide mean grain size, illustrated in Figure 5.34, indicates an increase in fracture toughness with increase in silicon carbide mean grain size within a particular series of materials, namely the hot pressed and gas

**Figure 5.33**  Indentation fracture toughness of gas pressure sintered and materials post-sintering heat treated at 1925°C for 5 hours
pressure sintered materials. The discrepancies in this trend could be due to the broad grain size distributions observed especially for the gas pressure sintered and heat treated materials (seen in Figures 5.22, 5.23, 5.24).

![Graph showing relationship between fracture toughness and SiC mean grain size](image)

**Figure 5.34** Relationship between fracture toughness and silicon carbide mean grain size

It appears that the grain boundary phase composition does not significantly influence the fracture toughness of the LPSSiC material. The increase in silicon carbide grain size in the high alumina content materials (shown in Figure 5.18, 5.21, 5.22 and 5.24) could help to enhance toughening mechanisms such as crack deflection or crack bridging\(^{108}\), which could be the reason for increased fracture toughness in the high-alumina content hot pressed and gas pressure sintered materials which were heat treated at 1925°C for 8 hours. The grain boundary phase therefore indirectly influenced fracture toughness by increasing grain growth in the higher alumina-content grain boundary materials.
Figure 5.35 shows that for the HP3Y5Al material, there is an increase in fracture toughness with increase in heat treatment time, at 1925°C (HT2 – HT4), however the most significant increase in fracture toughness was for the 5 hour heat treatment at 1975°C. This is the same heat treatment time as HT3, but at a higher temperature. For the higher alumina content HP1Y4Al material, the most significant increase in fracture toughness was in the HP material heat treated at 1925°C for 8 hours. The increase in temperature of heat treatment also increased the fracture toughness of this hot pressed material significantly, but had the same effect as the 8 hour heat treatment at 1925°C.

Heat treatment at 1975°C (HT5) appeared to cause a significant increase in fracture toughness for all the compositions HP and GPS materials (Table 4.15 and 4.16). As illustrated in Figures 4.5, 4.9, 5.25 and 5.26, these materials all contained many large elongated grains (platelets), compared to all the other heat treatment results, which would enhance fracture toughness by increasing crack deflection and crack bridging.
The trends in fracture toughness can be related to the observed microstructures in these materials. Figures 5.22, 5.23 and 5.24 illustrate that the as-hot pressed, as-gas pressure sintered, as well as the heat treated materials, HT2, HT3 (1.5 and 5 hour heat treatments at 1925°C) all showed unimodal grain size distributions, which broadened as the materials were gas pressure sintered and heat treated. Studies of the crack paths of a few samples under FESEM showed that increase in fracture toughness was associated with an increase in crack path roughness (selected examples are given in Figure 5.36), which indicates more crack deflection in this material. Although standard deviations of measurements (given in Table 4.18 in the results chapter) were quite high, the general trend showed an increase in fracture toughness from the GP1Y4Al and GP3Y5Al materials (which have similar mean SiC grain
size) to the heat treated HT3GP1Y4Al (which has a higher mean SiC grain size). This was associated with an increase in intergranular fracture, shown in Figure 5.37, with increased crack bridging per unit length, shown in Figure 5.38.

As illustrated in Figure 5.39, the SiC grain size distribution in the GP1Y4Al material broadened after heat treatment for 5 hours at 1925°C (HT3). These results, together with the SiC mean grain size analysis results, indicate that the increased grain size and platelet formation in the HT3GP1Y4Al material over the un-heat treated GP1Y4Al material, activated some increased crack deflection and crack bridging toughening mechanisms, associated with larger grain sizes, as observed by Padture\(^{(10)}\). This resulted in the increase in fracture toughness also illustrated in Figure 5.33.

![Graph showing the relationship between fracture toughness and roughness of the crack path in three materials (GP3Y5Al, GP1Y4Al and HT3GP1Y4Al)](image)

**Figure 5.36** The relationship between fracture toughness and roughness of the crack path in three materials (GP3Y5Al, GP1Y4Al and HT3GP1Y4Al)
**Figure 5.37** Figure showing increased crack deflection resulting in increased fracture toughness

**Figure 5.38** Fracture toughness in relation to the number of crack bridges per unit length
A qualitative investigation of some cracks in the materials heat treated at 1975°C indicated a highly intergranular path, with crack bridging, as illustrated in Figure 5.40.

Figure 5.39 Silicon carbide grain size distributions for the GP1Y4Al and subsequent heat treatments
(a) Crack path in the HT5HP3Y5Al material

(b) Crack path in the HT5HP1Y4Al material

**Figure 5.40** Crack paths observed in (a) the HT5HP3Y5Al material, and (b) the HT5HP1Y4Al material. The arrows indicate examples of crack bridging.
Figures 5.23 and 5.24 show that the silicon carbide grain size distribution in the HP1Y4Al materials broadened more than in the HP3Y5Al materials after 8 hour heat treatment at 1925°C. However, more platelet-like SiC grain formation was observed in the HP3Y5Al composition after 8 hour heat treatment at 1925°C (HT4HP3Y5Al), than in the HT4HP1Y4Al material, as illustrated in Figures 5.41 a and 5.41 b. Microstructures of the materials heat treated at 1975°C show that platelet formation is higher in the 3Y2O3:5Al2O3 composition (Figure 5.41 c)) compared to the microstructure of the HP3Y5Al material heat treated for 8 hours at 1925°C (Figure 5.41 a)), and the microstructure of the HP1Y4Al material heat treated at 1975°C (Figure 5.41 d)). This explains the increase in fracture toughness (11) of the HP3Y5Al composition during heat treatment at 1925°C (8 hours) and a more significant increase during the heat treatment at 1975°C. The microstructure of the HP1Y4Al material heat treated at 1975°C is coarser (seen in Figure 5.41d) than that of the material heat treated for 8 hours at 1925°C (Figure 5.41 c), however there is not an increase in platelet formation and this would explain why there is no difference in fracture toughness between these two materials, as seen in Figure 5.35. This is illustrated for the heat treated HP materials in Figure 5.41, and can also be observed in the microstructures of the heat treated GPS materials in Figures 4.4, 4.5, 4.8 and 4.9.
a) FESEM micrograph of HT4HP3Y5Al material

b) FESEM micrograph of HT4HP1Y4Al material
Figure 5.41  FESEM micrographs of the YAG composition \((3Y_2O_3:5Al_2O_3)\) materials after heat treatment at 1925°C for 8 hours (a and b) and at 1975°C for 5 hours (c and d).
5.2.3 Microstructure to strength relationships

Four-point bending strength tests were carried out on hot pressed and gas pressure sintered materials, followed by a study of some of the fractured surfaces. The results of the strength tests are given in Figure 5.42.

![Figure 5.42 Four-point bending strength results for HP and GPS materials (HP(as-fin) indicates results on as-finished un-tempered HP materials, and (temp) denotes the results of materials which were tempered for 2 hours at 1250°C prior to bending strength tests)](image)

For the HP materials, the strength for the different compositions did not differ significantly. The HP3Y5Al composition does however appear to have a higher bending strength than the other materials. Rietveld analysis in Table 4.12 shows that this is the material with the highest YAG content. This is in agreement with observations by Cheong et al.\(^{108}\), who reported that when the amorphous glassy phases crystallized to form the YAG phase, the strength was increased.

In the gas pressure sintered materials there was a decrease in strength with a decrease in Al\(_2\)O\(_3\) content. This corresponds to the densification
behaviour observed for the gas pressure sintered materials, namely that higher \( \text{Y}_2\text{O}_3 \)-content materials were more difficult to densify. The Griffith equation\(^{(132,182)}\), equation 5.4, gives the relationship between fracture strength and flaw size in the material.

\[
\sigma_f = Y \frac{K_{\text{IC}}}{\sqrt{a}},
\]

where \( \sigma_f \) = fracture strength,
\( Y \) = geometric factor,
\( a \) = flaw size.

The \( K_{\text{IC}} \) for the GPS materials is the same for the different compositions. Considering equation 5.4, this would imply that the decrease in strength must be due to an increase in defect size in the material with increased \( \text{Y}_2\text{O}_3 \)-content. The cause of failure in all the examined GPS materials was porosity, with porous regions up to 100 \( \mu \text{m} \) in size (see Figure 5.43, 5.44 and 5.45). As was previously discussed, this is related to the difficulty in densifying especially the higher \( \text{Y}_2\text{O}_3 \) materials.
**Figure 5.43** SEM micrograph of mirror region of failure in GP1Y1Al material

**Figure 5.44** SEM micrograph of the origin of failure seen in the encircled area in Figure 5.43, in GP1Y1Al material – a large pore (about 35 µm in length)
Figure 5.45 SEM micrograph of GP1Y4Al material. The critical porosity (100 µm size porous region), which was found to be the cause of failure is observed in this micrograph.

5.2.4 Microstructure to elastic modulus property relationships

The elastic moduli of the as-hot pressed, gas pressure sintered and ultra-high pressure sintered materials, as well as selected heat treated materials were measured. The elastic modulus of a given material is indicative of the phase composition and defect density of the material, where the defects may be pores or segregation of phases.

Figure 5.46 shows that the elastic moduli of the hot pressed materials are between 380 and 403 GPa, slightly increasing with decreasing Al₂O₃ content. This is slightly inferior to values reported in the literature (420 GPa). The gas pressure sintered materials of the same compositions have elastic modulus values of between 427 and 435 GPa. Figure 5.47 shows that in the gas pressure sintered materials, the elastic modulus
values of the three higher Al₂O₃ content materials do not differ significantly, whereas the lowest Al₂O₃ content material has an elastic modulus value which is considerably lower than the others due to the high porosity-content in these materials.

![Graph showing elastic modulus results of as-hot pressed and consequent heat-treated materials](image)

**Figure 5.46** Elastic modulus results of as-hot pressed and consequent heat-treated materials

After heat treatment, all the compositions in the hot pressed (Figure 5.46) and gas pressure sintered materials (Figure 5.47) reached approximately the same range of elastic modulus values (between 410 and 440 GPa), increasing significantly in the case of the hot pressed materials, and in general decreasing slightly in the case of the gas pressure sintered materials. The exception in the case of the heat treated gas pressure sintered materials is the highest Y₂O₃ content material, in which the elastic modulus progressively decreased after each of the three heat treatments (going from 311 to 287 GPa after eight hour heat treatment).
Figure 5.47 Elastic modulus results of as-gas pressure sintered and subsequent heat-treated materials

The HP4Y2Al has the highest elastic modulus out of the HP materials, which is opposite to the trend observed in compositions in the GPS materials. This is related to full densification of this composition during hot pressing, which could not be achieved during GPS.

The elastic modulus of a composite is given by the rule of mixtures in equation 5.5\(^{(211)}\):

$$E_c = E_1V_1 + E_2V_2 + E_3V_3$$  \hspace{1cm} (5.5)

If “1” is considered to be SiC phase and $E_1 = 420$ GPa, then $E_1V_1$ can be assumed to be nearly constant, neglecting dissolution of SiC in grain boundary phase. If “2” and “3” are considered to be the grain boundary phase components, then the difference in composition and volume of the...
grain boundary phase between the different sintered and heat treated materials must be considered. \( E_{\text{SiO}_2} = 60 - 65 \) GPa (for fused silica)\(^{(207)}\), compared to the elastic modulus of \( \text{Al}_2\text{O}_3 \) which ranges between 318 and 416 GPa. Since the grain boundary phases in the HP materials change from containing silicates and some aluminates to containing only aluminates during post-sintering heat treatments (shown in Table 4.12), and decrease in \( V_2 \) is only between 2 and 4 volume-%, there is an overall increase in \( E_c \) in silicate-poor gas pressure sintered and heat treated materials, compared to hot pressed materials. Even in the HP3Y5Al material, in which only YAG is detected as the only grain boundary phase by XRD, some amorphous silicates are present. This is known due to the low YAG content calculated in the Rietveld analysis, being about 6.5 mass-%, and also seen in electrical measurements discussed in section 5.3. Even small amounts of amorphous silicate phases are known to affect the elastic modulus\(^{(212)}\).

Sciti \textit{et al.}\(^{(107)}\) also observed an increase in elastic modulus on heat treatment of hot pressed materials, also reaching a maximum value of 438 GPa. They also attributed this to the reduction or modification of the secondary phases (silicates).

\textbf{SUMMARY}

The hardness appeared to be most influenced by the composition of the grain boundary present in the 10 mass-% LPSSiC material. There was an increase in hardness of the HP materials after heat treatment at 1925°C due to loss of softer silicate phases, but a decrease in hardness is observed after heat treatment at higher temperature (1975°C) due to decomposition of aluminates, resulting in porosity in the materials.

For fully dense microstructures, hardness and elastic modulus were found to be related to the grain boundary phase composition. Silicates resulted
in lower hardness and elastic moduli in the HP materials, whereas aluminates resulted in higher hardness and elastic moduli in the GPS and heat treated materials.

A steady decline in hardness was observed in the HP3Y5Al material measured from room temperature to 1000°C and 1200°C. This could be due to thermally activated dislocation glide and climb, but is more likely due to a softening of the grain boundary phases, in particular the amorphous silicates.

The fracture toughness was not as sensitive to the increase in grain size, as it was to an increase in the aspect ratio of silicon carbide grains. Studies of the crack path in selected materials, showed an increase in crack deflection and crack bridging in a heat treated material, which had increased grain size and some platelet formation, compared to the un-heat treated material. This is in agreement with work in the literature\(^{(13)}\), which has shown that platelets in the microstructure enhance fracture toughness by mechanisms such as crack bridging. Although the fracture toughness appears to be dependent on the SiC grain size distribution and platelet formation, and not the grain boundary chemistry, there is a relationship between fracture toughness and grain boundary chemistry after heat treatment, because the microstructures of the heat treated materials appear to be dependent on the grain boundary composition. A more “in-situ” toughened microstructure, in the form of platelet-shaped grains, was observed for the materials with higher alumina content (1Y\(_2\)O\(_3\):4Al\(_2\)O\(_3\) and 3Y\(_2\)O\(_3\):5Al\(_2\)O\(_3\)).

Broader silicon carbide grain size distributions with more platelet-like SiC grain formation was observed when heat treating at higher temperature (1975°C), compared to longer heat treatment at lower temperature (8 hours at 1925°C). This resulted in a large increase in fracture toughness of all the HP and GPS materials.
There appears to be a maximum in strength observed for the hot pressed material with YAG as its only crystalline grain boundary phase (HP3Y5AI). This is in agreement with work in the literature\(^{108}\). A stronger composition dependence of strength is observed in the GPS materials, where the strength decreased with a decrease in alumina content, due to the larger defect size caused by difficulty in densifying the higher yttria-content materials by gas pressure sintering (discussed in section 5.1).