Microstructure and mechanical properties of high strength Al–Mg–Si–Cu profiles for safety parts

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Abstract.
Aluminium extrudate used for safety parts in cars need to exhibit high yield strength and ductility, a combination that is not easily achieved. In this work, the mechanical properties and microstructure of profiles with a yield strength greater than 280 MPa achieved by two different artificial ageing treatments were studied. Profiles from one of the heat treatments performed well in quasi-static compression testing while those from the other heat treatment clearly failed. The batch of profiles that failed showed higher uniform elongation in tensile testing but a lower reduction in area. However, the difference in bending angles in the three-point-bending test were not as pronounced. Microscopic investigation of polished sections and fracture surfaces revealed that failure is dominated by the fracture of intermetallic phases resulting in voids. The growth and coalescence of these voids is facilitated by another population of smaller voids within the matrix, presumably nucleating at secondary phases.

1. Introduction
Extruded profiles of Al–Mg–Si–Cu alloys are used for safety parts in the front section of cars to absorb kinetic energy in the case of an accident. This is ideally achieved by uniform concertina–style deformation. In a constant quest for lightweight design the automaker’s specifications for strength are ever–increasing; the requirements for ductility, however, remain on a constant high level [1]. Since high strength and ductility are often mutually exclusive the production of profiles with appropriate properties is not trivial.

1.1. Microstructure of Al–Mg–Si–Cu extruded profiles
For a given geometry the mechanical properties of extruded Al–Mg–Si–Cu profiles are determined by their microstructure. Characterization of microstructural features and understanding their contribution to the mechanical properties is crucial for the design of new materials with improved properties.
The grain structure of extruded high strength Al–Mg–Si–Cu profiles for safety parts is developed during extrusion and consists of a fine, fibrous core of grains elongated along the extrusion direction and a peripheral coarse grain layer formed due to dynamic recrystallization. Throughout this matrix intermetallic phases of the $\alpha$–AlFeSi or $\alpha$–Al(FeMn)Si type with sizes of typically a few $\mu$m are dispersed.

The peripheral coarse grain layer can lead to orange peel, bending failure and various optical defects [2]. Thus, attempts are generally made to reduce its thickness by proper alloy design and choice of manufacturing parameters, e.g. sufficient cooling. An important approach to maintain the fibrous grain structure formed during extrusion is the facilitation of the formation of secondary dispersoid phases that inhibit recrystallization and grain growth. These phases with a size under 500nm form during the homogenization treatment and only have a minor impact on strength; thus, they are usually referred to as non-hardening phases. The chemical composition and crystal structure of the dispersoids are strongly dependent on alloy composition [3].

During artificial ageing Mg and Si form nano-scale hardening precipitates, significantly increasing the hardness of the profiles by dispersion hardening. The main hardening precipitate in alloys of the 6xxx series is $\beta''$ [4]. When the alloy contains considerable amounts of Cu, $Q'$ and precursors of $Q'$ also play an important role as hardening precipitates [5, 6].

The micromechanics of ductile fracture of 6xxx Al alloys have been extensively studied by Lassance et al. [7]. They showed that the fracture of brittle intermetallics as well as particle–matrix decohesion was observed even at low strains in in–situ tensile tests in a scanning electron microscope. It was found that particles with the major axis oriented at an angle of less than 45° with respect to the loading direction undergo fracture while particles oriented at an angle of more than 45° lead to particle–matrix decohesion. The examination of fracture surfaces showed a ductile profile with dimples containing particles.

Mrówek–Nowotnik et al. published various papers on the mechanical properties of EN AW–6082, e.g. [8, 9]. It was found that the mechanical properties and fracture toughness of this alloy strongly depend on the artificial ageing temperature and time, or more specifically, the size and number density of $\beta''$ and $\beta'$ precipitates formed by different ageing regimes. Scanning electron micrographs of fracture surfaces of tensile samples revealed that cracks initiated by nucleation, growth and coalescence of voids. Two void populations were found through observation of the fracture surfaces. The nucleation sites for large dimples were identified as primary intermetallics, while for smaller dimples dispersion hardening phases and dispersoids were suggested as nucleation sites.

1.2. Assessment of crash performance

It is generally accepted that quasi–static compression testing of whole profiles shows good correlation with high–speed crash behaviour. Thus, quasi–static compression testing is the main criterion used by automakers to assess crash performance of extruded profiles. Although rating systems for the results of quasi–static compression testing have been proposed [10], the interpretation of quasi–static compression testing remains subjective and difficult to quantify. Furthermore, the reproducibility of quasi–static compression testing is often poor for high strength profiles. These problems greatly impair comparability of results, and therefore hinder purposeful R&D efforts.

It is desirable to use small scale mechanical tests like hardness, tensile, and three–point–bending tests, and correlate their results to those of quasi–static compression testing. For example, three–point–bending of sheet specimens cut out of extruded profiles has been shown to correlate well with quasi–static compression testing for Al sheet [11]. Furthermore, stress–strain curves from small scale mechanical tests are used for the development of FEM models. For the improvement of these models, correct interpretation of the tests is important.
1.3. Aim of this work
This work is part of a project aiming to understand the relationships between processing conditions, the resultant microstructure and crash properties of extruded profiles. We studied the microstructure of extruded high–strength Al–Mg–Si–Cu profiles artificially aged using different regimes and performed hardness, tensile, and three–point–bending tests. We seek to correlate parameters from these tests with the outcomes of quasi–static compression testing. The results of mechanical testing are interpreted with regard to the observed microstructure and fracture surfaces.

2. Materials and Methods
Extruded double chamber hollow profiles of a modification of the EN AW–6082 alloy were received as fabricated from Hammerer Aluminium Industries Extrusion GmbH (HAI) with a wall thickness of 2.7mm and outer dimensions of approximately 140mm × 80mm. These were cut to a length of 300mm. Table 1 shows the composition limits of the non–standardized alloy.

|        | Mg | Si | Mn | Cr | Zn | Cu | Fe | Al |
|--------|----|----|----|----|----|----|----|----|
| min    | 0.6| 0.7| 0.4| -  | -  | -  | -  | balance |
| max    | 1.2| 1.3| 1.0| 0.25| 0.2| 0.5| 0.5| balance |

The samples were divided into two batches and proprietary artificial ageing treatments were conducted. According to their ageing conditions, the designations of the two batches are “low temperature” (LT) and “high temperature” (HT). HT was aged for a shorter time than LT. The temperatures used for ageing were in the range usually used for alloys of the 6xxx series.

Quasi–static compression testing was conducted on a 1200kN hydraulic press with a rate of 100mm/min to an end length of 100mm where possible (i.e. if no disintegration of the profile occurred).

Vickers micro–hardness testing was conducted on polished sections of the profiles using a test load of 100g.

Tensile and three–point–bending tests were conducted on a Zwick Z100 materials testing machine. Specimens were cut parallel to the extrusion direction from the outer shell of the profiles. They had a width of 5mm, a reduced section length of 14mm and a gauge length of 10mm. An initial loading rate of 8MPa s\(^{-1}\) was used up to a strain of 1% and then the strain rate was increased to 0.008s\(^{-1}\). Using a micrometer caliper the reduction of area was measured on the failed samples.

The design of the three–point–bending test was modeled similar to VDA 238–100 [12]. Specimens had a width of 15mm and a length of 60mm, with the long side being parallel to the extrusion direction. The diameter of the supports for three–point–bending was 30mm, the distance between the supports was 5.4mm (i.e. twice the specimen thickness) and the radius of the punch was 0.4mm. Testing was conducted at a rate of 20mm/min and stopped when a drop in force of at least 15N occurred. The bending angle \(\alpha\) (see Figure 1) was measured on the unloaded specimens.

For optical microscopy and scanning electron microscopy (SEM), sections were ground and mechanically polished using standard metallographic techniques. Grains were revealed by etching with Barker’s reagent. SEM investigations were performed on a Zeiss Ultra Plus 55 field emission gun SEM.
3. Results and Discussion

3.1. Microstructure

As expected, no difference between LT and HT profiles could be observed by optical microscopy since the microstructural features visible, namely primary intermetallic phases and grain size, are not affected by the artificial ageing treatment. Figure 2 shows a polished longitudinal section of a profile. The intermetallics are mostly smaller than 10µm and exhibit rounded edges typical for α-Al(FeMn)Si–particles, indicating proper alloy design and homogenization treatment to avoid β–particles or convert them to α–particles. Most particles have an aspect ratio of approximately 1, however, some particles have aspect ratios as low as 0.2. Virtually all particles with a low aspect ratio are orientated more or less horizontally, i.e. parallel to the extrusion direction.

Figure 3 shows a longitudinal section etched with Barker’s reagent to reveal individual grains. The core of the profile exhibits fine, elongated, non–recrystallized grains whereas the outer layer shows coarse, recrystallized grains that are less elongated and more equiaxed.

3.2. Mechanical properties

The outcomes of quasi–static compression testing are illustrated in Figure 4. While the HT profile shows good folding behaviour, the test of the LT profile had to be stopped due to beginning of disintegration of the profile.
No difference in Vickers micro-hardness HV0.1 could be found. The mean value of five measurements was 107.8±1.5 for LT and 107.8±2.2 for HT (plus/minus values represent standard deviation).

Average stress-strain curves from the tensile tests are shown in Figure 5 and the results given in Table 2. Since the yield strength (\(R_{y0.2}\)) determines the yield strength class of a safety part it is considered the most important value by manufacturers. Both batches of profiles exhibit yield strengths over 280MPa whereat the values of HT exceed those of LT. However, LT profiles show a higher ultimate tensile strength (\(R_m\)) and, interestingly, a higher elongation at break (\(A_{10}\)). However, the reduction of area (Z) is higher for HT indicating higher localized fracture strain.

Table 2. Mean tensile and three-point-bending test results. \(A_{10}\) is the technical elongation for a gauge length of 10mm and Z is reduction of area measured on broken specimens. \(\alpha\) is the bending angle measured on unloaded specimens as defined in Figure 1. Plus/minus values represent standard deviation from three measurements (except for Z – only two measurements).

| Tensile test          | Three-point-bending test |
|-----------------------|--------------------------|
| \(R_{y0.2}\) [MPa]    | \(R_m\) [Mpa]            |
| LT                    | 286.8±0.7                | 347.0±0.3 |
| HT                    | 297.5±0.4                | 330.5±0.4 |
| \(A_{10}\) [%]        | 21.0±0.4                 | 19.7±0.0  |
| Z [%]                 | 41.9±0.6                 | 51.0±2.2  |
| \(F_{max}\) [N]       | \(\alpha\) [°]          |
| LT                    | 1958±29.6                | 71.5±3.3  |
| HT                    | 1817±15.4                | 78.2±2.8  |

Although the quasi-static compression testing results of the two types of profiles were very different, the difference in bending angle (\(\alpha\)) in the three-point-bending test was not as pronounced. However, HT shows a lower maximum bending force (\(F_{max}\)) than LT (see Table 2).
3.3. Microstructural investigation of failure

Figure 6 shows breakage and particle–matrix decohesion of intermetallics due to the applied stress, resulting in the nucleation of voids. These mechanisms were observed in tensile as well as in three–point–bending specimens. In both cases breakage of intermetallics was found more frequently than particle–matrix decohesion. Since all of the particles with a low aspect ratio are oriented near parallel to the load direction, this is in accordance with the findings of Lassance et al. [7]. When, however, decohesion was observed it was not on the interface of particles oriented
perpendicular to the load direction – since such particles did not exist – but rather at the sharp edges of particles oriented in line with the extrusion direction.

The number of voids was highest near the fracture surface in tensile test and near the curved outer boundary on the tensile side of the three-point-bending specimens. In Figure 7 the coalescence of voids facilitating the propagation of a crack in the three-point-bending test can be seen.

Investigation of fracture surfaces from the tensile specimens (Figure 8) revealed large dimples. Primary intermetallic phases were found in the large dimples. Therefore, it is expected that the voids nucleated at these phases. In addition, small dimples in the matrix between the large dimples were found. These dimples suggest the existence of a secondary population of voids as described in the literature for other alloys [7, 9]. It is expected that these voids nucleate at non-hardening dispersoids due to their incoherence with the matrix or near dispersion hardening precipitates due to their effect on the matrix.

The only difference between LT and HT is the post-extrusion artificial ageing regime. However, primary intermetallic phases are hardly affected by artificial ageing. Their chemical composition, volume density, form and size are defined during casting and homogenization of the billet and their final distribution and form is created by subsequent extrusion. Therefore, no significant difference in primary intermetallics between LT and HT should exist and, in fact, no difference in intermetallics or in their breakage or decohesion behaviour could be found between LT and HT. Thus, these mechanisms cannot explain the differences in ductility.

The only factor remaining that is different between the two batches of profiles is the artificial ageing treatment. A possible explanation is that HT, which was aged at a higher temperature for a shorter time, contains fewer, larger dispersion hardening phases than LT, allowing for more dislocation mobility yielding higher ductility. Further work is needed to clarify the mechanisms involved.

4. Conclusions
The results show that care must be taken when interpreting small scale mechanical tests of extruded aluminium profiles for safety parts. Understanding the microstructural failure mechanisms is important for the proper design of high strength profiles for safety parts in cars.

The main findings of this work are:

- Small variations to artificial ageing can dramatically alter quasi-static compression testing performance, even though yield strength, ultimate tensile strength, and uniform elongation are similar and microhardness is the same.
- For the studied profiles, strong necking in tensile testing indicated good quasi-static compression testing behaviour.
- Higher uniform elongation did not result in better quasi-static compression testing properties.
- Differences in bending angle were rather small, however the maximum bending force was lower for profiles with good quasi-static compression testing properties.

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