Behaviour of Al-Mg-Si alloys at a wide range of temperatures and strain rates

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Abstract

The thermo-mechanical behaviour in tension of three as-cast and homogenized Al-Mg-Si alloys, distinguished by their Mg and Si content, was investigated at a wide range of strains (the entire deformation process up to fracture), strain rates (0.01-750 s–1) and temperatures (20-350°C). The alloys were shown to have isotropic plastic behaviour. Low and medium strain-rate tests were performed in a universal testing machine, while a split-Hopkinson tension bar (SHTB) system was used for higher strain rates. The samples were heated with an induction-based heating system. In all tests, local recordings of the specimen geometry with digital cameras allowed determining the true stress-strain curve also after the onset of necking. In addition, the failure strains of all samples were measured after the tests. The three alloys had high ductility, even at room temperature, which increased with increasing temperature. It was shown that both the yield strength and the work-hardening decreased with increasing temperature. The materials exhibited negligible strain-rate sensitivity (SRS) for temperatures lower than 200°C, while they revealed strong positive SRS at higher temperatures. The experimental data obtained for the Al-Mg-Si alloys were used to identify the parameters of a physically-based constitutive model proposed in the literature, and reasonable agreement with the experimental stress-strain behaviour was achieved.

*Keywords*: *Quasi-static and dynamic tensile tests; Split-Hopkinson tension bar; Local strain measurements; Thermo-mechanical behaviour; Strain-rate sensitivity*

# Introduction

Aluminium alloys are frequently used by the industry because of their good mechanical properties and their relatively low density. Such alloys are of particular interest in different fields of engineering where lightweight designs are required, as in automotive vehicles and protective structures [1,2,3]. A large amount of the profiles often used in such structures are made of Al-Mg-Si alloys, also called AA6xxx class alloys, due to their high extrudability, good appearance and rather good mechanical properties [4]. Since safety components are designed to resist rapid loading, a number of studies have been devoted to the strain rate sensitivity of various aluminium alloys [5,6,7,8,9]. High strain rate in combination with elevated temperature occurs in situations related to structural impact, but also in all kinds of metal forming operations like extrusion, rolling and forging. Very few systematic studies of such coupled effects for AA6xxx alloys exist in the literature. An investigation requires experimental tests at a wide range of strain rates and temperatures, calling for test rigs and instrumentation systems that are capable of producing reliable data. A particular feature associated with high-temperature tests is that necking occurs at a very early stage of the deformation process. Local measurements of the strains in the neck are necessary to determine the true stress also in the comparatively large phase of the test after the onset of necking.

Thermo-mechanical studies of materials are usually performed using uniaxial compression or uniaxial tension tests and a heating device that allows investigating the material response at different temperatures. High strain-rate tests are often achieved with split-Hopkinson pressure bar (SHPB) or split-Hopkinson tension bar (SHTB) systems coupled with a heating device [10], while low strain-rate tests are frequently performed in universal testing machines in combination with an oven [11]. In dynamic tests involving SHPB or SHTB systems, the nominal stress and strain are normally obtained by measuring the propagation of the waves in the bars with strain gauges and thereafter using the Kolsky equations [12]. It was shown by Vilamosa et al. [13] that tests performed at high temperature on aluminium alloys in an SHTB necessitate local strain measurements due to the early necking of the specimen. Recent improvements in high-speed camera technology make it possible to perform such measurements by edge detection techniques [13] or digital image correlation (DIC) [14]. These methods also provide an accurate determination of the true stress, using the force level found by measuring the stress wave in the transmission bar and the diameter of the necked section of the sample.

Different heating techniques such as ovens [15] or heat guns [16] can be used with SHPB systems, while induction heating seems to be more relevant when tests are accomplished in SHTB systems [10]. Induction systems heat the specimen rapidly, thereby reducing the heat conduction in the bars that could affect the stress wave propagation and subsequently the strain measurements [17]. In this study, the modified SHTB system presented by Vilamosa et al. [13] was used to investigate the thermo-mechanical behaviour of three different Al-Mg-Si alloys at high strain rate. The SHTB was equipped with an induction heating system and a high-speed camera, and thus provided true stress, local strain and strain rate measurements valid for isotropic materials. The same heating system, now with a standard digital camera, was used for the tests performed at low and moderate strain rates in a universal testing machine.

Numerous studies have been performed during the last decades to understand the mechanical behaviour of metallic materials at different temperatures and strain rates. This has resulted in some well-known and widely used constitutive relations such as the phenomenological Johnson-Cook model [18] or the more physically-based Zerilli-Armstrong model [19]. Metals often exhibit a strongly coupled temperature and strain-rate sensitivity (SRS) of the yield strength and the work-hardening. The modelling of complex SRS is often a key factor in many constitutive relations that attempt to describe the material behaviour as function of strain rate and temperature [20]. Indeed, the SRS can change with increasing temperature as shown for aluminium alloys [21] and steel alloys [22]. The SRS might evolve as well with plastic strain [23] and applied strain rate [24]. The SRS of Al-Mg-Si alloys is often rather small at room temperature [6], but increases for temperatures exceeding a threshold temperature [23]. In addition, the SRS is likely to depend on the grain size and the temper of the aluminium alloy [25].

Material models may be broadly classified in two main categories: phenomenological and physically-based models. Phenomenological models have a rather limited number of parameters that makes them easy to use in finite element simulations, while the physically-based models are assumed to give an improved prediction of the material behaviour for a wider range of temperature and strain rate. Numerous phenomenological models have been proposed and applied in the thermo-mechanical studies of metallic alloys. The strain rate and temperature sensitivity of the flow stress are often uncoupled in phenomenological models [26]. This drastically restrains the strain rate and temperature range that is feasible to model for the material at hand. Liang and Khan [27] showed that the coupled effect of strain rate and temperature on the stress-strain behaviour is not correctly predicted by some of the most used phenomenological models. An improved model was proposed by Khan and Liu [21] to predict the stress-strain behaviour of aluminium alloys at a wide range of strain rate and temperature, exhibiting complex strain rate and temperature sensitivity. Physically-based models are often based on dislocation theory. Bergström and Hallén [28] proposed a model of dislocation density evolution with a storage term that is directly related to the plastic strain and a recovery term that is strain-rate and temperature dependent. A similar model was developed by Estrin and Mecking [29]. Another physically based model was proposed by Nes [30], where the evolution of the microstructure, such as misorientation and size of the cells, with plastic deformation is taken into account. Other models aiming to describe the stress-strain behaviour of aluminium alloys at different temperatures and relative low strain rates were more recently proposed by van den Boogaard and Huétink [31] and Kurukuri et al. [32], respectively using Bergström’s framework or the model proposed by Nes. Some physically-based models are based on dislocation theory, but expressed as a constitutive relation for the flow stress in terms of plastic strain, plastic strain rate and temperature [33, 34].

In this study, the thermo-mechanical response of three Al-Mg-Si alloys is characterized under tensile loading. The local deformations in the neck are measured up to large strain levels applying the technique developed by Vilamosa et al. [13]. The tests are performed at nominal strain rates from 0.01 s–1 to 750 s–1 and temperatures from 20°C to 350°C. Two of the investigated aluminium alloys (Al-0.45Mg-0.40Si and Al-0.5Mg-0.45Si) have similar Mg and Si content, and were tested to disclose any significant differences in their stress-strain behaviour at elevated strain rates and temperatures. The third alloy (Al-0.8Mg-0.76Si) has a considerably higher content of Mg and Si, and serves to evaluate the effect of alloy composition on the thermo-mechanical behaviour. In this work, a comprehensive experimental database on the stress-strain behaviour and failure strain of these alloys at the selected strain rates and temperatures is established as a basis for future development of improved thermoelastic-thermoviscoplastic constitutive models and fracture criteria. By use of a local strain measurement technique and the Bridgman correction, the stress-strain curves to failure are determined. All three alloys exhibit significant strain-rate and temperature sensitivity of the yield strength and the work-hardening. The SRS was found to be low for temperatures below 200°C, but increased considerably with higher temperatures and then also with increasing plastic strain. To investigate the accuracy of existing constitutive models, the physically-based model of Voyiadjis and Abed [33] was calibrated to the experimental results and shown to provide a reasonable description of the thermo-mechanical response of these Al-Mg-Si alloys.

# Materials and experimental method

## Materials

The chemical composition of the three Al-Mg-Si alloys investigated in this study is provided in Table 1. Two of the alloys have similar composition with only a slight difference in Mg and Si content, while the last alloy has a much higher content of these alloying elements. The alloy Al-0.8Mg-0.76Si contains also more of the secondary elements Cu, Mn and Cr. The two first materials are within the window of the AA6060 alloy, while the last one is an AA6082 alloy. In this article, the materials are addressed by their Al-Mg-Si content to differentiate the two AA6060 alloys. The three materials were delivered as cast and homogenized billets by Hydro ASA. Micrographs of the grain structure taken in an optical microscope are shown in Fig. 1. The three alloys all have an equi-axed grain structure. The grain size of the Al-0.45Mg-0.4Si and Al-0.5Mg-0.45Si alloys was found to be similar and roughly 95 μm, while the grain size of the Al-0.8Mg-0.76Si alloy is slightly smaller and about 84 μm.

In the homogenization process, the dissolved Mg and Si will form Mg2Si precipitates during cooling on heterogeneities like particles and grain boundaries. The content of Mn and Cr in the Al-0.8Mg-0.76Si alloy will result in formation of nano-sized dispersoids which are not present in the AA6060 alloys. These dispersoids will act as effective nucleation sites for Mg2Si precipitates during cooling. After homogenization, the two AA6060 alloys and the AA6082 alloy were naturally aged for about eighteen and six months, respectively, before the mechanical testing. During natural aging, solute Mg and Si will form clusters and Guinier–Preston (GP) zones that contribute to the strength and the work-hardening of the materials.

## Experimental programme

The experimental programme involves tension tests at different strain rates (between 0.01 s–1 and 750 s–1) and temperatures (between 20°C and 350°C). The tests at low to moderate strain rates were carried out in a universal testing machine, while a SHTB system, described in detail by Chen et al. [7], was applied in the tests at high strain rates. The tension test sample shown in Fig. 2 was used in all tests.

A common feature of the materials at hand is that the onset of necking at elevated temperatures occurs at a very early stage of the test. After necking, determination of the true stress and logarithmic (true) strain calls for local measurements of the deformation of the sample inside the neck. All tests in the experimental programme for evaluation of the thermo-mechanical response were therefore instrumented with a digital camera. Subsequently, information from the pictures was used to find the true stress and the logarithmic strain during the entire test to fracture, see Vilamosa et al. [13]. This technique requires that the material is isotropic during plastic deformation, which was expected for these materials since they were tested in the as-cast and homogenized condition. However, to ensure that the three alloys are isotropic, some additional quasi-static tension tests were carried out at room temperature. These tests are reported in Section 3, while the thermo-mechanical test programme follows in Sections 4 and 5.

# Evaluation of isotropy

## Experimental set-up and instrumentation

The specimens were taken in three different directions at 0°, 45° and 90° with respect to the longitudinal axis of the cylindrical billets. The sample geometry is the same as the one applied in the thermo-mechanical tests, see Fig. 2. A total of 18 tests were carried out at room temperature, i.e. two samples were tested in each direction and for each material. The initial diameter  of the gauge section of each sample was measured before testing.

The tests were performed in a Zwick-Roell testing machine with a maximum load capacity of 30 kN. The samples were screwed into special adaptors attached to the clamps of the test machine. The tests were run in displacement control at a cross-head velocity of 0.01 mm/s. This corresponds to a nominal strain rate of 2⋅10–3 s–1. The tests were instrumented with an AEROEL XLS13XY laser gauge, projecting two perpendicular laser beams of 13×0.1 mm2 towards detectors at the opposite side of the sample [35]. The detectors measure the minimum cross-section diameter of the specimen with a resolution of 1 µm and a frequency of 15 Hz. The laser beam system was mounted on a mobile frame to ensure that the measurements were done at the minimum cross-section also after the onset of necking. The data were transferred by the built-in electronics to the remote computer via fast ethernet.

## Results

The measured diameters in the two perpendicular directions were similar until failure, which implies that the plastic deformation of the three materials is isotropic. The two measured diameters were therefore averaged to obtain the local diameter  of the minimum cross-section. The local logarithmic (true) strain  was then determined by assuming constant volume during plastic straining, i.e.



where  is the initial area of the sample’s cross-section and  is the minimum cross-section area of the sample during testing. The true (Cauchy) stress  was computed using the force  provided by the load cell and the current area , viz.



The plastic strain  was then determined from the following relation



where  is Young’s modulus at room temperature for the aluminium alloys.

The repeatability between the two tests for each direction and alloy was found to be good. The two stress-strain curves obtained in a given set of duplicate tests were more or less coincident, i.e. they differed with less than 5 MPa at all deformation levels. Fig. 3 shows the true stress versus plastic strain curves up to the maximum true stress for one representative test of each kind. The three materials do not show any anisotropic behaviour with respect to specimen orientation regarding yield strength and work-hardening. Thus, the alloys are indeed isotropic, which is as expected in the as-cast and homogenized condition. The Al-0.45Mg-0.4Si and Al-0.5Mg-0.45Si alloys show identical behaviour at room temperature and a rather similar strain to fracture in the three directions investigated.

Since the deformation of the materials was found to be isotropic at room temperature, it is feasible to use the local strain measurement technique presented by Vilamosa et al. [13] in the subsequent study of the thermo-mechanical behaviour of these materials.

# Thermo-mechanical tension tests

## Quasi-static tests

The tests at low to moderate nominal strain rates were carried out in a Zwick-Roell testing machine. Two nominal strain rates of 0.01 s–1 and 1 s–1 were used in the quasi-static tests on the Al-0.45Mg-0.4Si and Al-0.5Mg-0.45Si alloys. The nominal strain rate of 1 s–1 was reduced to 0.33 s–1 in the tests on the Al-0.8Mg-0.76Si alloy in order to avoid the recurrent logging problems experienced in the two first test series. The tests were run in displacement control, and the nominal strain rates were obtained by applying a cross-head velocity of 0.05 mm/s, 1.65 mm/s and 5 mm/s. The material behaviour was investigated at 20°C, 200°C, 250°C, 300°C and 350°C.

A water-cooled induction heating system from MSI Automation was applied in the tests. During heating, the conductive sample is subjected to a magnetic field created by an alternating current in the coil surrounding it. This magnetic field induces an electric current in the sample that is turned into heat due to the thermal resistivity of the metallic material. The heating device provides a power of 5 kW at 180 kHz giving heating rates up to 10°C/s. The induction apparatus is controlled in a feed-back loop provided by the temperature measurement system.

A laser-based pyrometer (IP 140 MB12) from LumaSense Technologies was used to measure the temperature in the specimen. This pyrometer has a temperature range of 160°C to 1200°C, with an uncertainty of ±2°C until 400°C. The pyrometer measurements are based on the emissivity of the sample which is supposed to be constant during a test and equal to 0.95 due to the black paint applied on the surface. The high temperature acquisition frequency of 666 Hz allows for an accurate temperature control of the induction heater, but it is not fast enough to record the temperature increase in the specimen caused by adiabatic heating during a high strain-rate test.

The deformation of the samples was captured by a Prosilica GC2450 digital camera equipped with a 5 megapixel Sony ICX625 CCD sensor. In contrast to the laser gauge described in Section 3.1, the camera-based system can be used together with the heating apparatus and is therefore applicable at all strain rates and temperatures. It also allows investigating the shape of the necked section for each frame. The digital camera was synchronised with the Zwick-Roell testing machine through its gigabit ethernet output, and recorded frames of the minimum cross-section of the sample until fracture with a framing rate of 2 Hz (nominal strain rate of 0.01 s–1) or 15 Hz (nominal strain rate of 0.33 s–1 or 1 s–1). The samples were painted black to increase the contrast between the sample and the background. An edge detection script based on the grey gradient level in the frame and developed in MATLAB [36] was applied to detect the minimum cross-section diameter  of the specimen. Thus, a large grey gradient level facilitates the edge detection and improves the accuracy of the measurement [13]. The pixel size was determined before the test by measuring the initial diameter  of the samples both with the camera and a digital calliper. The logarithmic strain in the necked section was then calculated from Equation , and the true stress was determined according to Equation .

## Dynamic tests

A schematic drawing of the SHTB system is presented in Fig. 4. It consists of an 8140 mm long incident bar (ABC) and a 7100 mm long transmission bar (DE). The two bars have a diameter of 10 mm and are made of the high strength steel Tibnor 52SiCrNi5. Young’s modulus of the bars is  = 210 GPa. This value was determined from the strain wave propagation in the bars, and was also verified from quasi-static tension tests instrumented with an extensometer. The heating system and instrumentation protocol for temperature measurement were the same as in the quasi-static tests, see Section 4.1. The dynamic tests were also carried out at 20°C, 200°C, 250°C, 300°C and 350°C.

During a test, the specimen is first attached between the two bars (C and D). The incident bar is then clamped at B and loaded in tension at A. The force level  in the incident bar is monitored by strain gauge station ➀. The required force  was dependent on the temperature in the sample and the desired strain rate. The dynamic tests reported herein were performed at two levels of nominal strain rate: 350 s–1 and 750 s–1. Simultaneously with the pre-tensioning of part AB of the incident bar, the sample was heated to the selected temperature. Thereafter, a stress wave in tension was released by abruptly breaking the clamp at B. The stress wave propagation was measured by the strain gauge stations at position ➁ and ➂, located 600 mm from the sample in respectively the incident and transmission bars. Using the Kolsky equations, the nominal stress, strain and strain rate can then be determined from the strain signals in gauges ➁ and ➂ [12]. Subsequently, it is straightforward to find the true stress and logarithmic strain by assuming plastic incompressibility and uniform deformation in the gauge part of the sample.

It was, however, pointed out in Section 2.2 that necking occurs at an early stage of deformation at elevated temperatures. Therefore, it was required for this study to determine the local strain inside the neck by use of a camera. All dynamic tests were monitored with a SA1.1 Photron high-speed video camera, equipped with a Sigma lens having a focal length of 105 mm. The frame rate was set to 100 kHz. In the same way as for the quasi-static tests, the samples were painted black prior to testing, and the minimum diameter  of the specimen during the test was determined from the pictures. Equation was thereafter applied to find the logarithmic strain. Vilamosa et al. [13] provide more details on the instrumentation and data processing for the tests in the split-Hopkinson tension bar. It was demonstrated that the logarithmic strains determined from the Kolsky equations and the digital pictures compared well before the onset of necking. Further, it was shown that the local heating of the sample and the subsequent temperature increase of the part of the bars that was closest to the specimen did not give any significant disturbance of the propagation of the stress wave or the measurements during the test.

The force in the sample was found by use of the conventional method for analysis of split-Hopkinson bar experiments. The bars are assumed to be elastic during the test, and it was shown by Chen et al. [7] that the dispersion in the applied system is negligible. Therefore, the strain  measured at position ➂ of the transmission bar is proportional to the force  in the sample, i.e.



where  is Young’s modulus and  is the cross-section area of the transmission bar. Having determined the force and the minimum diameter , the true stress was again calculated from Equation .

## Validation of local strain measurements

A camera-based technique for local measurements of the strain was applied to all thermo-mechanical tests reported in Sections 4.1 and 4.2. The method, which records the minimum diameter of the sample during the test, was originally developed for SHTB tests by Vilamosa et al. [13]. They also evaluated the accuracy of this instrumentation method at high strain rates. Concerning validation in the quasi-static strain rate regime, Fig. 5 compares the results from a test monitored with the camera against the results from a similar test instrumented with the laser gauge system described in Section 3.1. The tests addressed in this figure were performed on Al-0.8Mg-0.76Si samples at room temperature and a nominal strain rate of 0.002 s–1. The agreement between the curves is very good, thus validating the accuracy of the camera-based measurements also at quasi-static strain rates.

# Results of the thermo-mechanical study

## Data processing

The digital images from the tests showed pronounced necking that increased with increasing temperature, and the resulting three-dimensional stress state in the neck must be accounted for. This can be done by using geometrically-based techniques, such as Bridgman’s analysis, that take the shape of the neck and the diameter of the minimum cross-section into account [37]. In the local strain measurement technique proposed by Vilamosa et al. [13], a least squares method is used to estimate the mean radius of curvature  of the neck for each frame. The equivalent stress  is then determined using the Bridgman relation [37]



where  is the true stress, as determined from Equation for all tests, and  is the diameter of the minimum cross-section in the neck. By comparing numerical simulations and experimental results for the entire SHTB system, Vilamosa et al. [13] showed that Equation (5) provides a reasonably accurate prediction of the equivalent stress.

In all data sets from the experimental tests, and in particular the ones from the SHTB, small oscillations in the stress-strain curves were observed. As the subsequent analysis in Section 5.3 requires the values of  at certain levels of plastic strain , these oscillations should be removed in order to avoid artificially high or low values of  when the data point happens to occur close to a local maximum or minimum of the oscillating signal [6]. For this purpose, all equivalent stress versus plastic strain curves obtained from the tests were fitted to a two-term Voce rule extended with a linear hardening term, i.e.



where the plastic strain  is assumed equal to the equivalent plastic strain in uniaxial tension. The parameters ,  and  were fitted to the part of the -curve with plastic strain larger than 0.01 owing to the lack of equilibrium at lower plastic strains in the SHTB tests. The yield strength  was found by back-extrapolation to = 0. Considering a representative test on Al-0.8Mg-0.76Si carried out at a nominal strain-rate of 350 s–1 and a temperature of 350 °C, a fit obtained with Equation (6) is shown together with the test data in Fig. 6. It appears that the smooth curve provides a good representation of the experimental data. A similar quality of the fit was obtained in the other tests. In order to enhance the readability of the curves and the changes of response with increasing strain rate and temperature, all subsequent stress-strain curves presented in this article are fitted to Equation (6).

In all tests, the local plastic strain rate  in the necked section was obtained by numerical differentiation of the plastic strain versus time curve, where the time increment was defined by the camera’s acquisition frequency.

## Stress-strain curves

Tensile tests were performed at different temperatures and nominal strain rates following the experimental program described in the previous sections. The tests completed successfully for each combination of material, temperature and strain rate are compiled in Table 2. The closed loop between the pyrometer and the induction heating system gave satisfactory control for the high rate tests when the light conditions and the emissivity of the black paint remained constant during the test. On the other hand, the quasi-static tests performed at a strain rate of 0.01 s–1 were challenging due to the temperature control. Several tests were therefore performed until a good repeatability was obtained. Some acquisition problems were also observed for the tests performed at a strain rate of 1 s–1, since the limits of both the quasi-static testing machine and the camera were approached. This could affect the accuracy of the measurements of the radius of the neck and the subsequent calculation of the equivalent stress given by Equation . As a result of these problems, the strain rate was reduced to 0.33 s–1 for the Al-0.8Mg-0.76Si alloy. In addition, inverse modelling with the nonlinear finite element code LS-DYNA [38] was performed for the tests with strain rate of 1 s–1 in order to optimise the parameters of the fitting relation given in Equation , using the local plastic strain and the force measurements from the tests as input data.

The response of the Al-0.45Mg-0.4Si and Al-0.5Mg-0.45Si alloys was rather similar over the entire range of temperatures and strain rates, and the equivalent stress versus plastic strain curves for these two alloys are therefore presented together in Fig. 7. The neck was hidden behind the coil of the induction heater for some tests and these curves are therefore only plotted until necking (see e.g. the dotted lines at temperatures 250°C and 300°C in Fig. 7 a)). The alloys exhibit a decrease in both yield strength and work-hardening with increasing temperature for the applied strain rates. At quasi-static strain rates the work-hardening is found to decrease with increasing temperature until it vanishes at a temperature of 350°C. At room temperature, a slightly positive SRS of the flow stress is found for medium to large strains . This positive SRS increases with increasing temperature and extends to the entire plastic domain including the yield stress for temperatures higher than 250°C. At the two highest strain rates, significant work-hardening is observed even at a temperature of 350°C.

The equivalent stress versus plastic strain curves of the Al-0.8Mg-0.76Si alloy at different temperatures and nominal strain rates are plotted in Fig. 8. Also this material shows strong temperature sensitivity. At room temperature the SRS is practically negligible, but it increases significantly for higher temperatures at both low and high levels of plastic strain.

Fig. 9 a) and b) present the strain-rate sensitivity of the flow stress for representative tests at a temperature of 350°C for respectively the Al-0.5Mg-0.45Si and the Al-0.8Mg-0.76Si alloys. The tests performed at the highest nominal strain rate  of 750 s–1 exhibit a work-hardening of respectively 60 and 80 MPa at a plastic strain of 0.75. On the other hand, the tests performed at the lowest nominal strain rate of 0.01 s–1 show negligible work-hardening for the Al-0.5Mg-0.45Si alloy and a work-hardening of only 10 MPa for the Al-0.8Mg-0.76Si alloy at a plastic strain of 0.5. At 350°C, all three alloys exhibit a high positive SRS of the flow stress over the entire plastic strain range.

## Data analysis

To analyse the strain-rate and temperature sensitivity of the materials in more detail, maps of the equivalent stress as a function of temperature and strain rate are plotted at given levels of plastic strain. Two levels of plastic strain, namely  and , were used when calculating the equivalent stress from Equation , and the corresponding stresses are denoted  and , respectively. The stress  is close to the yield stress, while  is the flow stress after a considerable amount of plastic straining. Moreover, it is convenient to work with dimensionless stresses, and, accordingly, the equivalent stresses  and  were normalised with a reference stress , which represents the back-extrapolated yield strength at room temperature and reference strain rate . The reference strain rate  was taken as  for the Al-0.45Mg-0.4Si and Al-0.5Mg-0.45Si alloys and  for the Al-0.8Mg-0.76Si alloy. The maps are plotted using the observed plastic strain rate  occurring locally inside the neck at the considered plastic strain instead of the nominal strain rate.

Fig. 10 a) presents the normalised equivalent stress  as function of plastic strain rate for the Al-0.45Mg-0.4Si and Al-0.5Mg-0.45Si alloys. It was shown in the previous section that the response of these two materials does not differ much, and it is therefore not differentiated between them in the figure. A similar map for the normalised equivalent stress  is provided in Fig. 10 b). It should be noted that the plastic strain rate is larger in the latter figure, indicating that there is a significant local increase of  in the neck. The same information is given for the Al-0.8Mg-0.76Si alloy in Fig. 11. From these plots it is seen that the flow stress of all three alloys has an almost log-linear dependence on the local plastic strain rate. The slopes of the lines depend on the temperature. It is therefore possible to express the combined strain-rate and temperature sensitivity of the flow stress as



where  represents the temperature-dependent slope of the lines in Fig. 10 and Fig. 11, and  gives the stress ratio  at the reference strain rate . The superscript  indicates the level of plastic strain, i.e.  or . The slope  defines the SRS of the materials [39]. Fig. 10 and Fig. 11 show that the overall trend is that  increases with increasing temperature for all three alloys, while the SRS can be neglected at room temperature. It also appears that the positive SRS increases significantly at temperatures above 200°C.

Considering the Al-0.45Mg-0.4Si and Al-0.5Mg-0.45Si alloys, the slope  as found from the experimental tests, see Equation (7), is plotted as function of temperature in Fig. 12 a) at plastic strain levels of  and . The results obtained with the Voyiadjis-Abed model presented in Section 6.2 are also included. The surface plot in Fig. 12 b) presents  as function of  and  over the entire range of plastic strains. The projections at  and , corresponding to Fig. 12 a), are indicated with dashed lines in Fig. 12 b). The parameter  is represented in the same way in Fig. 13 a) and b). Two distinct regimes can be observed from Fig. 12 for the parameter  with respect to its temperature sensitivity. As noted previously, the SRS can, independently of plastic strain level, be neglected for temperatures lower than 200°C, while there is a strong positive SRS for higher temperatures. The SRS is also seen to increase with the level of plastic strain for temperatures above 200°C. The values of  and  are found to decrease with temperature, and seem to converge for the highest temperature, see Fig. 13 a) and b). The reason for this convergence is the low work-hardening at high temperatures and quasi-static strain rates. Similar results are found for the Al-0.8Mg-0.76Si alloy, see Fig. 14 and Fig. 15. Here, the parameter  seems to saturate at a temperature around 300°C at the lowest level of plastic strain, as shown in Fig. 14 a). The coupling effect between strain-rate and temperature is therefore lower for the yield stress than for the work-hardening. Moreover, the SRS remains rather low at yielding. The values  and  approach each other with increasing temperature, see Fig. 15 a), but for this alloy the reduction of the work-hardening is somewhat lower.

At room temperature, the surface  in Fig. 12 b) and Fig. 14 b) has the shape of an aerofoil with increasing plastic strains. This is believed to be an artefact due to the adiabatic conditions that induce a significant temperature increase at large plastic strains for the tests performed in the SHTB. Indeed, the maps plotted in Fig. 10 and Fig. 11 represent the flow stress level as a function of the initial temperature. This artefact is negligible for higher temperatures where the adiabatic heating is comparatively small due to the low stress level.

Fig. 16 shows maps representing the temperature dependence of the normalised equivalent stress at , where the four levels of nominal strain rate  are indicated with different symbols. Fig. 16a) is concerned with the Al-0.45Mg-0.4Si and Al-0.5Mg-0.45Si alloys, while Fig. 16b) addresses the Al-0.8Mg-0.76Si alloy. In these maps adiabatic heating is taken into account at high strain rates. The increase  due to adiabatic heating was estimated by



where  is the density,  is the specific heat capacity and  is the Taylor-Quinney coefficient that represents the proportion of plastic work converted into heat [40]. In accordance with normal practise in the literature, it is assumed that  and  are independent of temperature, and common room temperature values of these two parameters are applied. According to Fig. 16, the specimens tested at the highest strain rates experience a temperature increase of 25 to 50°C, somewhat depending on the initial temperature and the corresponding work-hardening. This increase of temperature affects  and is clearly visible in Fig. 16.

## Failure strains

After each test, the fracture area of the two parts of the ruptured sample was measured with a digital microscope. The average value of the two measured areas is denoted . Assuming incompressible plastic deformation, the failure strain  is obtained as



where the initial area  was determined before each test. Fig. 17 shows the failure strains as function of initial temperature, where different symbols address the four levels of strain rate. The alloys Al-0.45Mg-0.4Si and Al-0.5Mg-0.45Si are treated in sub-figure a), while sub-figure b) pays attention to Al-0.8Mg-0.76Si. For all alloys, a significant increase of failure strain with temperature is observed. Moreover, the strain rate plays a more important role at elevated temperatures, in particular for the two AA6060 alloys in Fig. 17a). The failure strains for the three alloys are in general comparable except in the quasi-static tests at the highest temperature, where Al-0.8Mg-0.76Si exhibits a much lower .

# Constitutive modelling

## Introductory remarks

It transpires from the previous section that there is a complex coupling of the temperature and strain-rate sensitivity of the flow stress for the considered aluminium alloys. The coupling is also dependent on the plastic deformation as it is different at yielding and at substantial plastic strains. As pointed out by Mecking and Kocks [41], the flow stress at constant microstructure depends on temperature and strain rate, as plastic flow by dislocation motion is a thermally activated process. In addition, the evolution of the microstructure depends on temperature and strain rate. While the storage of dislocations is weakly temperature dependent, and possibly also rate dependent [42], the dynamic recovery (or dislocation annihilation) is thermally activated and markedly dependent on temperature and strain rate; i.e., the dynamic recovery increases with temperature and decreases with strain rate [41]. At high temperatures, the microstructure of the 6xxx series alloys considered here will also change due to dissolution of solute clusters and rapid precipitation on heterogeneities, such as dispersoids and dislocations [43]. It emerges that developing physically-based constitutive models including all these features for AA6xxx alloys is a formidable task. An attempt in this direction was made by Gouttebroze et al. [44] for local hot forming. A constitutive model was proposed that describes the flow stress as a function of temperature, strain rate, and internal variables. The internal variables represent the dislocation density and strengthening contributions from solutes and precipitates and develop with time according to some evolution equations. The constitutive model was coupled with a precipitation model developed by Myhr et al. [45].

In this study, given the lack of information on microstructure evolution with temperature and strain rate, it has been sought for a semi-empirical or physically-based constitutive relation that describes the flow stress as a function of plastic strain, strain rate and temperature for the considered AA6xxx alloys. A large number of constitutive relations have been proposed to model the thermo-mechanical response of metals in the literature. The phenomenological Johnson-Cook model [18] is widely applied. It takes the strain-rate and temperature sensitivity of the equivalent stress versus plastic strain curve into account, but it does not capture any significant change of the shape of the curve as the strain-rate and temperature increases. Thus, the attempt to fit the Johnson-Cook model to the experimental data presented above was not successful. Some physically-based models, such as the one proposed by Zerilli and Armstrong [19], were explored as well, but none of these models were capable of describing the coupled strain-rate and temperature sensitivity of the three alloys over the entire temperature range. However, the model proposed by Voyiadjis and Abed [33] was found to provide a reasonable representation of the flow stress as a function of temperature, strain rate and plastic strain, and, accordingly, this model will be described and calibrated to the experimental data in the following.

## Voyiadjis-Abed model

Voyiadjis and Abed [33] based their constitutive model on the well-known Zerilli-Armstrong model [19], but introduced several modifications. The flow stress is assumed to be additively decomposed into athermal and thermal contributions, and for FCC metals it is expressed as



where  and  are dimensionless temperature and strain rate with  and ;  is the athermal yield stress; ,  and  define the yield stress and work-hardening at 0 K; and , ,  and  govern the coupled strain-rate and temperature sensitivity of the yield stress and work-hardening. The reader is referred to Voyiadjis and Abed [33] for a detailed derivation of the model.

## Identification of model constants

As the Al-0.45Mg-0.4Si and Al-0.5Mg-0.45Si alloys exhibited similar thermo-mechanical behaviour, see Section 5.2, the Voyiadjis-Abed model was identified only for the Al-0.5Mg-0.45Si and Al-0.8Mg-0.76Si alloys.

Using the method of least squares and MATLAB [36], the parameters of the Voyiadjis-Abed model, see Table 3, were estimated by fitting the model constants to the obtained equivalent stress versus plastic strain curves over the entire plastic strain range covered by the measurements. The extended plastic strain range obtained with the local measurement set-up facilitated the use of the local plastic strain rate in the parameter identification. During the SHTB tests, there is a substantial temperature increase due to adiabatic heating [46] that softens the material, and this should be taken into account in the parameter identification. The acquisition frequency of the pyrometer was not high enough to measure this temperature increase during testing, and it was therefore estimated with Equation instead.

A comparison between the model and representative experimental results for Al-0.5Mg-0.45Si and Al-0.8Mg-0.76Si is provided in Fig. 18. Given the wide range of plastic strains and nominal strain rates, the model describes the overall stress-strain behaviour for temperatures between 20°C to 350°C with reasonable accuracy, but there are also significant discrepancies. At the lowest strain rate, the model underestimates the flow stress at 200°C and 250°C for Al-0.5Mg-0.45Si and at 200°C for Al-0.8Mg-0.76Si, see Fig. 18 a) and b), respectively. Fig. 18 c) and d) show that the model is capable of capturing the experimental stress-strain behaviour with rather good accuracy for Al-0.5Mg-0.45Si at the intermediate strain rate, while the flow stress at 200°C for Al-0.8Mg-0.76Si is underestimated. At the highest strain rates, i.e. under assumed adiabatic conditions, the model shows good agreement with the experimental data for both alloys for plastic strains below, say, 1.0, see Fig. 18 e) and f). However, significant softening is predicted for the two aluminium alloys at larger strains owing to the adiabatic heating. The general impression is that the stress-strain behaviour at 20°C is in good agreement for the Al-0.5Mg-0.45Si alloy and Al-0.8Mg-0.76Si alloy over the nominal strain rate range covered in this investigation, but there is some discrepancy around yielding.

The normalised equivalent stresses  and , as obtained from representative experiments and the constitutive model, are compared in Fig. 19 for the Al-0.5Mg-0.45Si and Al-0.8Mg-0.76Si alloys. The shaded areas at the highest strain rates in Fig. 19 b) and d) represent the difference in the predicted flow stress under assumed isothermal and adiabatic conditions; i.e., the upper limit of the shaded area is valid for isothermal conditions and the lower limit represents adiabatic conditions. These maps disclose that the combined strain rate and temperature effect is not well captured at temperatures of 20°C and 200°C for the Al-0.5Mg-0.45Si alloy, while better results are obtained for the higher temperatures. The model also gives better agreement with the experimental data for the Al-0.8Mg-0.76Si alloy. The parameters  and  were computed for the Voyiadjis-Abed model in the same way as in Section 5.3 for the experimental data. While rather good agreement with the experimental data is obtained for , see Fig. 13 a) and Fig. 15 a), the temperature sensitivity of the SRS, as represented by , is not particularly well described, see Fig. 12 a) and Fig. 14 a). The temperature sensitivity of the SRS in the Voyiadjis-Abed model increases linearly with temperature and does therefore not capture the sudden increase in SRS above 200°C observed in the tests.

# Conclusions

The thermo-mechanical behaviour of three as-cast and homogenized Al-Mg-Si alloys has been investigated by performing tensile tests over a wide range of temperatures and strain rates. Applying a camera-based technique for determination of the local deformations in the neck, the true stress-strain curve was found up to high levels of plastic strain at all rates and temperatures. The failure strain of all samples was also measured after each test with a digital microscope. The following main conclusions can be drawn from this study:

* As expected, the three Al-Mg-Si alloys exhibit isotropic behaviour, since they were tested in the as-cast and homogenized condition.
* No noticeable difference in behaviour was found between the Al-0.45Mg-0.4Si and Al-0.5Mg-0.45Si alloys within the investigated range of temperature and strain rate.
* The alloys show a complex and highly coupled strain-rate and temperature sensitivity of the yield stress, work-hardening and failure strain.
* The work-hardening is reduced with increasing temperature for the three alloys, while it increases with increasing strain rate at elevated temperature. At a nominal strain rate of 0.01 s–1 and a temperature of 350°C, the work-hardening vanishes for the Al-0.45Mg-0.4Si and Al-0.5Mg-0.45Si alloys, while Al-0.8Mg-0.76Si still exhibits some work-hardening. However, all alloys show considerable work-hardening at the highest strain rate at 350°C.
* The strain-rate sensitivity of the flow stress is low at room temperature, but increases considerably with increasing temperature above 200°C for the three alloys. The strain-rate sensitivity is also found to increase with increasing plastic strain; i.e., the coupling effect between strain rate and temperature is lower for the yield stress than for the work-hardening.
* The Voyiadjis-Abed model is found to reproduce the overall stress-strain behaviour of the alloys for temperatures between 20°C and 350°C with reasonable accuracy considering the relative simplicity of the model.
* The failure strain increases as expected with temperature. It does not vary much with rate at low and medium temperatures, while it increases significantly at the quasi-static strain rate at the highest temperature.

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Table 1: Chemical composition of the aluminium alloys (weight %).

|  |  |  |  |  |  |  |  |  |  |
| --- | --- | --- | --- | --- | --- | --- | --- | --- | --- |
|  | Mg | Si | Fe | Cu | Mn | Zn | Ti | Cr | Al |
| Al-0.45Mg-0.4Si | 0.45 | 0.40 | 0.18 | 0.004 | 0.02 | 0.009 | 0.009 | 0.002 | balance |
| Al-0.5Mg-0.45Si | 0.50 | 0.45 | 0.18 | 0.004 | 0.02 | 0.009 | 0.009 | 0.002 |
| Al-0.8Mg-0.76Si | 0.80 | 0.76 | 0.20 | 0.24 | 0.56 | - | 0.01 | 0.16 |

Table 2: Number of successful tests for each combination of material, temperature and strain rate.

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| --- | --- | --- | --- | --- | --- | --- | --- | --- | --- | --- | --- | --- |
|  | Al-0.45Mg-0.4Si | | | | Al-0.5Mg-0.45Si | | | | Al-0.8Mg-0.76Si | | | |
| [s–1]  T [°C] | 0.01 | 1 | 350 | 750 | 0.01 | 1 | 350 | 750 | 0.01 | 0.33 | 350 | 750 |
| 20 | 1 | 1 | 1 | 1 | 1 | 1 | 1 | 1 | 1 | 1 | 1 | 1 |
| 200 | 0 | 1 | 2 | 1 | 2 | 1 | 1 | 1 | 1 | 2 | 1 | 1 |
| 250 | 1 | 1 | 1 | 1 | 2 | 1 | 2 | 1 | 1 | 2 | 2 | 2 |
| 300 | 1 | 2 | 1 | 1 | 2 | 1 | 2 | 2 | 1 | 2 | 3 | 2 |
| 350 | 1 | 0 | 1 | 0 | 2 | 1 | 2 | 2 | 2 | 2 | 2 | 1 |

Table 3: Parameters of the Voyiadjis-Abed model.

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| --- | --- | --- |
| Parameter | Alloy | |
| Al-0.5Mg-0.45Si | Al-0.8Mg-0.76Si |
| [MPa] | 20.0 | 35.0 |
| [MPa] | 87.34 | 64.07 |
| [MPa] | 307.1 | 211.8 |
|  | 0.355 | 0.275 |
|  | 2.11⋅10–2 | 2.12⋅10–2 |
|  | 1.72⋅10–2 | 1.32⋅10–3 |
|  | 3.02⋅10–5 | 4.07⋅10–5 |
|  | 0.955 | 0.901 |

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|  | (b) |
| Fig. 1: Grain structure of (a) Al-0.5Mg-0.45Si and (b) Al-0.8Mg-0.76Si. | |

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| Fig. 2: Geometry of the tensile specimen (measures in mm). |

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| (a) | (b) |
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| Fig. 3: True stress versus plastic strain curves until the maximum true stress for samples cut at 0°, 45°, 90° with respect to the longitudinal axis of the cylindrical billet: (a) Al-0.45Mg-0.4Si, (b) Al-0.5Mg-0.45Si and (c) Al-0.8Mg-0.76Si. | |

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| Fig. 4: Sketch of split-Hopkinson tension bar (SHTB) system with camera (measures in mm). |

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| Fig. 5: Comparison of true stress versus plastic strain curves obtained with the camera-based and laser-based measuring techniques for Al-0.8Mg-0.76Si at nominal strain-rate of 0.002 s–1 and room temperature. |

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| Fig. 6: Equivalent stress-plastic strain curves computed from a test and Equation for a representative test performed on an Al-0.8Mg-0.76Si aluminium alloy at nominal strain-rate of 350 s–1 and temperature of 350°C. |

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| (a) | (b) |
| G:\PHD\articles\paper2\run1\figures\number\Fig7_c.tif | G:\PHD\articles\paper2\run1\figures\number\Fig7_d.tif |
| (c) | (d) |
| Fig. 7: Equivalent stress as function of plastic strain for Al-0.45Mg-0.4Si (dotted lines) and Al-0.5Mg-0.45Si (solid lines) at all temperatures and nominal strain rate levels of a) 0.01 s–1, b) 1 s–1, c) 350 s–1 and d) 750 s–1. | |

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| (a) | (b) |
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| (c) | (d) |
| Fig. 8: Equivalent stress as function of plastic strain for Al-0.8Mg-0.76Si at all temperatures and nominal strain rate levels of a) 0.01 s–1, b) 0.33 s–1, c) 350 s–1 and d) 750 s–1. | |

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| (a) | (b) |
| Fig. 9: Representative equivalent stress versus plastic strain curves at 350°C and different levels of nominal strain rate for a) Al-0.5Mg-0.45Si and b) Al-0.8Mg-0.76Si. | |

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| (a) | (b) |
| Fig. 10: Normalised equivalent stress  as function of plastic strain rate and temperature for Al-0.45Mg-0.4Si and Al-0.5Mg-0.45Si at plastic strain  of a) 0.02 and b) 0.5. | |

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| (a) | (b) |
| Fig. 11: Normalised equivalent stress  as function of plastic strain rate and temperature for Al-0.8Mg-0.76Si at plastic strain  of a) 0.02 and b) 0.5. | |

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| (a) | (b) |
| Fig. 12: Temperature dependence of  for Al-0.45Mg-0.4Si and Al-0.5Mg-0.45Si at a) plastic strain level of 0.02 and 0.5 from the tests (blue lines) and Voyiadjis-Abed model (red lines), and b) over the entire range of plastic strains. | |

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| (a) | (b) |
| Fig. 13: Temperature dependence of  for Al-0.45Mg-0.4Si and Al-0.5Mg-0.45Si at a) plastic strain level of 0.02 and 0.5 from the tests (blue lines) and Voyiadjis-Abed model (red lines), and b) over the entire range of plastic strains. | |

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| (a) | (b) |
| Fig. 14: Temperature dependence of  for Al-0.8Mg-0.76Si at a) plastic strain level of 0.02 and 0.5 from the tests (blue lines) and Voyiadjis-Abed model (red lines), and b) over the entire range of plastic strains. | |

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| (a) | (b) |
| Fig. 15: Temperature dependence of  for Al-0.8Mg-0.76Si at a) plastic strain level of 0.02 and 0.5 from the tests (blue lines) and Voyiadjis-Abed model (red lines), and b) over the entire range of plastic strains. | |

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| (a) | (b) |
| Fig. 16: Normalised equivalent stress  at plastic strain  as a function of temperature and nominal strain rate, where the temperature is corrected for adiabatic heating, for a) Al-0.45Mg-0.4Si and Al-0.5Mg-0.45Si, and b) Al-0.8Mg-0.76Si. | |

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| (a) | (b) |
| Fig. 17: Failure strain as a function of initial temperature and nominal strain rate for a) Al-0.45Mg-0.4Si and Al-0.5Mg-0.45Si, and b) Al-0.8Mg-0.76Si. | |

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| (a) | (b) |
|  |  |
| (c) | (d) |
|  |  |
| (e) | (f) |
| Fig. 18: Comparison of the Voyiadjis-Abed model (solid black lines without symbols) with experimental data from representative tests at different temperatures: a), c), e) Al-0.5Mg-0.45Si at nominal strain rates 0.01, 1, 350  and b), d), f) Al-0.8Mg-0.76Si at nominal strain rates 0.01, 0.33, 350 . | |
|  |  |
| (a) | (b) |
|  |  |
| (c) | (d) |
| Fig. 19: Comparison of normalised equivalent stresses  and  obtained from representative tests and the Voyiadjis-Abed model (solid lines): a), b) Al-0.5Mg-0.45Si and c), d) Al-0.8Mg-0.76Si. The shaded area represents the difference in stresses between isothermal (upper limit) and adiabatic (lower limit) conditions. | |