Ductile to Brittle Transition in Magnesium Alloy with a Crack

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Abstract. We evaluated the strength of a cast non-combustible magnesium alloy that is intended for use in automobile engines. This alloy contains thin sheet-like oxides because of its production process. The size of these thin sheet-like oxides varies, and this means that the temperature dependency of the strength characteristics of the alloy cannot be determined by a tensile test alone. Initially, specimens of the alloy having an artificial pre-crack of 5 [mm] to simulate the sheet-like oxides were subject to tensile testing. The testing was carried out at room temperature (23 [°C]), 150 [°C], 200 [°C] and 250 [°C]. 250 [°C] is the temperature of the environment in which the alloy will be used in automobile engines. All the fractures propagated from the artificial pre-crack and the alloy exhibited its maximum strength at 150 [°C]. To investigate the temperature dependency of the fracture mechanism, a scanning electron microscope (SEM) was used to observe the region near the tip of the pre-crack on the fracture surface of the specimens tested at room temperature and at 250 [°C]. On the fracture surface of the specimen tested at 250 [°C], some dimples were observed over some of the surface near the tip of the artificial pre-crack; however, the fracture surface of the specimen tested at room temperature showed brittle fracture surface morphology over the entire area. For the magnesium alloy, the critical resolved shear stress for non-basal slip, which is a characteristic of a hexagonal close-packed metal, is dependent on temperature. Therefore, ductile to brittle transition was considered to be occurred at a temperature between room temperature and 250 [°C]. Mechanical restraint is also considered to be affected the ductile to brittle transition.

Introduction

Magnesium alloys have a high specific strength and are excellent for absorbing mechanical vibrations. For these reasons, the use of magnesium alloys in transport equipment will save weight and reduce noise. For this study, the objective is to use cast magnesium alloys in automobile engines. A non-combustible magnesium alloy is a magnesium alloy that can be produced in the atmosphere instead of in an inactive gas environment. The production of magnesium alloy in the atmosphere is achieved by significantly raising the ignition point by adding a few percent of Ca. The environment temperature of the actual automobile is about 523 [K] (250 [°C]). This is a high-temperature environment for a magnesium alloy whose melting point is about 923 [K] (650 [°C]), i.e. $523/923 > 0.5$. A new magnesium alloy with a high Vickers hardness number, indicating that the alloy has excellent static or fatigue strength at elevated temperatures, was developed [1]. This new magnesium alloy is a non-combustible magnesium alloy AZX1211 + 1%Si to which 1% Si by mass is added. In this study, from the perspective of cost, we propose the use of cast material in automobile engines, instead of plastic forming material such as extruded material or wrought material, whose static or fatigue strength is actually better than cast material. Fig. 1 shows the microstructure of the cast non-combustible magnesium alloy AZX1211 + 1%Si and Table 1 shows its mechanical properties. Dendrite morphology α-Mg as well as hard crystalline objects of CaMgSi

and eutectic Mg_2Si and $Al_{12}Mg_{17}$ is present in Fig. 1. The Vickers hardness number of CaMgSi at elevated temperatures is high (about 500), and this is why the Vickers hardness number of this new alloy is high at elevated temperatures compared with a common magnesium alloy such as AZ91. The problem with the new magnesium alloy is that oxides of magnesium are formed in liquid metal during casting because of its production method. These oxides look like thin sheets and in tensile or fatigue tests, fractures propagate from these oxides. Fig. 2 shows the largest size of the oxide that appeared on the fracture surface in the tensile test. For an actual automobile engine, casting methods are being investigated by using casting simulations in which large oxides are not interfused at regions of stress concentration or other high-risk regions. However, in this study, considering the interfusion of oxides could not be avoided, and the strength evaluation of a material with defects was carried out. To evaluate the temperature dependency of the static strength characteristic of the new magnesium alloy, tensile tests of specimens with an artificial pre-crack to simulate the oxides were carried out.

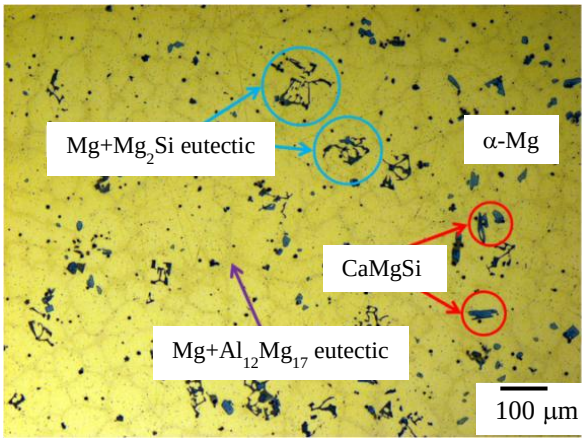


Fig. 1 Microstructure of cast non-combustible magnesium alloy AZX1211 + 1%Si.

Table 1. Chemical composition of non-combustible magnesium alloy AZX1211 + 1%Si (mass%).

Al	Zn	Mn	Ca	Si	Fe	Cu	Ni	Mg
11.41	1.00	0.16	1.28	0.63	0.03	0.041	0.011	Bal.

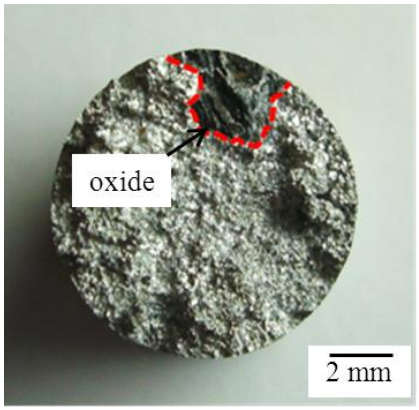


Fig. 2 Sheet-like oxide appeared on the fracture surface in tensile test.

Temperature dependency of strength characteristic in AZX1211 + 1%Si

Experiment method. To understand the static strength characteristics of the cast magnesium alloy AZX1211 + 1%Si, tensile tests of specimens with a pre-crack to simulate an oxide interfused in the material were carried out. Fig. 3 shows the shapes and dimensions of the tensile test specimen. After machining, the test section surface of the specimen was buffed until it becomes a mirror plane. Next, two small artificial holes with diameters of 1 [mm] and depths of 1.2 [mm] were drilled in the central part of the specimen. Fig. 4 shows the artificial holes and the pre-crack on the fracture surface. The size of the artificial holes was determined to introduce an artificial pre-crack from which a fracture would definitely propagate in the tensile test instead of propagating from a random oxide interfused in the alloy. Finally, a pre-crack with a surface length of 5 [mm] (±5 [%]) was introduced using a fatigue test machine, and this cracked specimen was used in the tensile test. The length of the pre-crack on the surface was observed by a replica method. The tensile test was carried out at temperatures of 150 [°C], 200 [°C], and 250 [°C], which is the environment temperature of the actual machine by using a constant temperature reservoir. The tensile test was also carried out at room temperature (23 [°C]) in the atmosphere. The tensile speed was 0.5 [mm/min]. Furthermore, as a comparison, a test was also conducted on extruded AZX1211 + 1%Si, whose chemical composition is the same as the cast AZX1211 + 1%Si, which is shown in Table 1. The test temperature for the extruded material was room temperature (23 [°C]), 150 [°C], and 250 [°C].

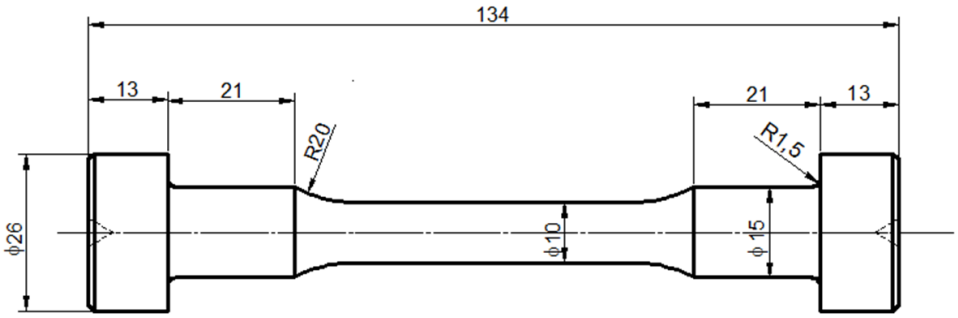


Fig. 3 Shape and dimensions of tensile test specimen.

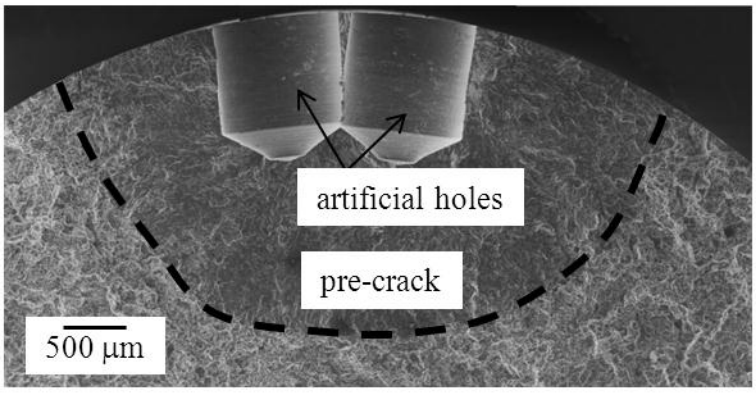


Fig. 4 Artificial holes and pre-crack on the fracture surface.

Test results. All of the fracture propagated from the artificial pre-cracks. The shapes of the pre-cracks were observed after the tensile test. All of the surface lengths of the pre-cracks were 5 [mm]; however, the depths varied. For this reason, the temperature dependency of the strength characteristic could not be determined by comparing the tensile strengths because the defect sizes of

each specimen were different. To exclude the influence of the defect size, K_{IB}* was introduced as a strength parameter. The factor K_{IB}* is a stress intensity factor when the fracture loading is applied to the pre-crack. The comparison of the tensile strengths, which excluded the difference in the defect sizes, can be performed using K_{IB}*. Fig. 5 shows the relationship between K_{IB}* and the test temperature. Here, K_{IB}* is calculated using the Eq. 1 below.

$$K_{IB}* = F\sigma_B\sqrt{\pi s} \tag{1}$$

where F is a correction coefficient determined by Shiratori [2], σ_B is the tensile strength, and s is half the surface length of a pre-crack. For the cast material, the value of K_{IB}* at 250 [°C] compared with the value at room temperature is about 86 [%]; therefore, the cast material has not only a high Vickers hardness number but also excellent static strength at elevated temperatures. In the seven test results shown in Fig. 5, necking was observed only in the test result at 250 [°C] for the extruded material. All of the other test results showed fracturing taking place before the phenomenon of plastic instability, at which nominal stress starts to decrease.

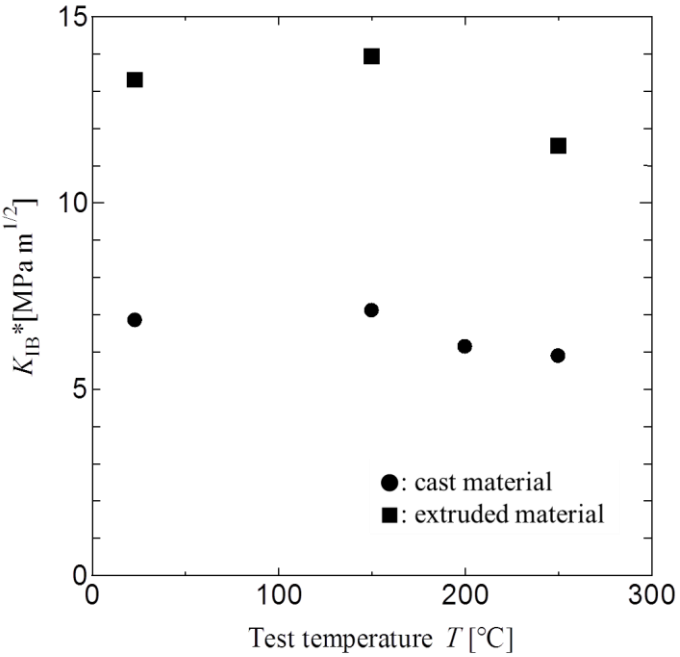
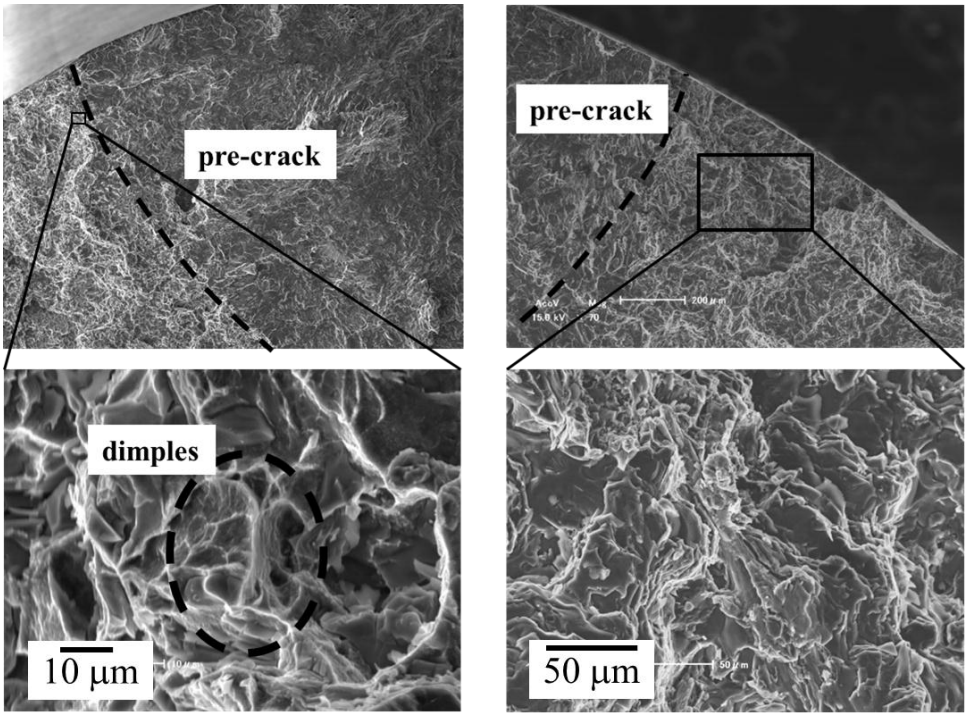


Fig. 5 Relationship between strength parameter K_{IB}* and test temperature.

Fracture mechanism at elevated temperatures

Even if the material whose fracture mechanism is a ductile fracture, which is the result of growth and coalescence of the microvoids in a smooth specimen, the presence of a crack in the material causes the fracture strain to become lower and the ratio of the elastic strain to all of the fracture strain becomes higher. For this reason, even if a material has a fracture mechanism that is ductile when it is a smooth specimen, the fracture mechanism becomes brittle when a crack is present. However, the ratio of a plastic strain to the fracture strain becomes higher when the material becomes easier to deform plastically at an elevated temperature. In this case, it is assumed that plastic strain easily occurs at the tip of the crack where the strain is concentrated. For this reason, even if the fracture strain becomes lower owing to the presence of a crack, it is assumed that a ductile fracture due to the growth and coalescence of the microvoids occurs at a region close to the

tip of the crack. The fracture strain was not measured in the tensile test of the AZX1211 + 1%Si, in which a pre-crack was introduced; however, all the fractures occurred before the phenomenon of plastic instability occurred, except for the extruded material tested at 250 [°C]. Therefore, the local fracture mechanism cannot be determined by the stress-strain curve or the observation of the surface appearance of the specimen. To investigate the variation in the local fracture mechanisms at elevated temperatures, the region near the tip of the pre-crack on the fracture surface was observed using a scanning electron microscope (SEM). Facture surfaces at room temperature (23 [°C]) and at 250 [°C] were observed. Fig. 6 shows the SEM images near the tip of the pre-crack on the fracture surfaces at room temperature (23 [°C]) and at 250 [°C]. The fracture surface tested at room temperature (23 [°C]) shows brittle morphology; however, some dimples were observed on the fracture surface tested at 250 [°C]. In addition, a similar observation was also carried out on the extruded material. Fig. 7 shows the SEM images near the tip of the pre-crack on the fracture surface at 250 [°C]. Many more dimples were observed on the extruded material than on the cast material. Although the observed dimples on the fracture surface of the cast material tested at 250 [°C] were localized near the tip of the pre-crack, they were taken as evidence of local ductile fracture due to growth and coalescence of the microvoids. In the extruded material, more dimples were observed and they existed over the entire area of the fracture surface. From the above results, it can be assumed that ductile to brittle transition occurred at a certain temperature between room temperature (23 [°C]) and 250 [°C]. It appeared that the transition to ductile fracture had clearly taken place in the extruded material, in which dimples were observed over the entire area on the fracture surface and in which fracture occurred after necking at 250 [°C].



(a) Tested at 250 [°C] (b) Tested at 23 [°C]

Fig. 6 Image taken using an SEM of the fracture surface near the pre-crack of the cast material.

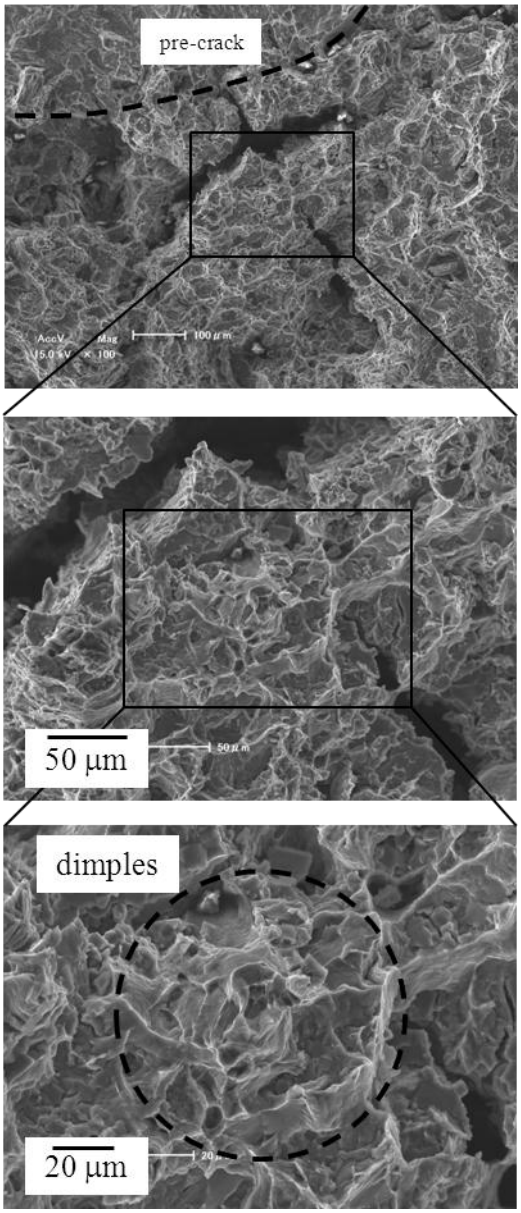


Fig. 7 Image taken using an SEM of the fracture surface near the pre-crack of the extruded material tested at 250 [°C].

Discussion

In general, tensile strength decreases as temperature increases because the yield stress or strain hardening exponent decreases. However, in our tensile tests, the tensile strength (K_{IB}*) had a the maximum value at 150 [°C], which is an elevated temperature compared to room temperature. Ductile to blittle transition is a likely factor behind the temperature dependency of the tensile strength and the yield stress not corresponding to the general relationship stated in the first sentence of this paragraph. At room temperature, plastic deformation does not occur as easily as it does at elevated temperatures; then the presence of a crack causes the ratio of the elastic strain to all of the fracture strain to rise, and the fracture surface morphology becomes brittle, as shown in Fig. 6 (b). Furthermore, the dimples that were observed on the fracture surface of the specimen tested at 250 [°C] are considered to be an indication that the transition to ductile fracture due to growth and coalescence of the microvoids took place. By elevating the temperature, plastic deformation occurs

more easily. A magnesium alloy such as AZX1211 + 1%Si is a hexagonal close-packed metal, and it is known that the critical resolved shear stress of non-basal slip depends on temperature [3]. Ductile to brittle transition is considered to be caused by an existing a crack and the temperature dependency is due to non-basal slipping. It is assumed that the transition temperature is higher than room temperature because the mechanical restraint is greater than a smooth material owing to the presence of a crack.

Conclusion

To evaluate the temperature dependency of the static strength characteristic of cast non-combustible magnesium alloy AZX1211 + 1%Si, tensile tests on specimens with an artificial pre-crack to simulate an oxide were carried out. In addition, to investigate the temperature dependency of the fracture mechanism, a scanning electron microscope was used to examine the region near the tip of the pre-crack on the fracture surface of specimens tested at room temperature (23 [°C]) and at 250 [°C]. The results are as follows:

(1) The tensile strength (K_{IB}*) had a maximum value at 150 [°C]

(2) Some dimples were observed on the fracture surface of the cast material tested at 250 [°C].

The authors intend to investigate in more detail the ductile to brittle transition in magnesium alloy with a crack via using the impact tests or by using a transmission electron microscope to perform a microanalysis.

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