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The effect of intermetallics on the fracture mechanism in AlSi1MgMn alloy

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Properties

ABSTRACT

Purpose: Fracture toughness in aluminium alloys is one of the main obstacles to using these materials in widespread ways and, therefore, various aspects of fracture mode would be examined closely, pointing out the microstructure influence. In the present paper, fracture nucleation and propagation of 6082 aluminium alloy was studied.

Design/methodology/approach: Tensile tests, crack resistance test and tensile test in the presence of sharp notch in room temperature tests were executed on the samples in the peak aged condition. The microstructure of tested samples was evaluated in terms of fracture mechanism using an optical microscope - Nikon 300, scanning electron microscope HITACHI S-3400 (SEM) in a conventional back-scattered electron mode and JEOL - JEM 2100 ARP TEM/STEM electron microscope.

Findings: Nucleation of voids is heterogeneous and most likely occurs by the debonding of the particle matrix interfaces. Other damage modes such as fracture of the intermetallic particles has been observed. These damage modes can significantly affect a macroscopic behaviour (tensile strength, fatigue strength, fracture toughness, and so on) of the investigated aluminium alloy under carried out tests.

Practical implications: In order to predict maximum ductility before fracture of the material it is required to characterize the microstructural parameters for the different mechanism of the nucleation of voids and cracking of intermetallic particles leading to final damage. The paper summarize all potential cracking modes that can occur in the aluminium alloy 6xxx type, tensile tested at room temperature using standard tensile specimens and specimens with the presence of sharp notch. This data can be used in practice for modeling many types of engineering processes.

Originality/value: The damage of the 6082 aluminium alloy tested at room temperature can be clearly attributed to the following mechanisms: propagation of cracks by fracturing the intermetallic particles, crack preceded by the nucleation of voids results from debonding along β particle/matrix interfaces and α particle/matrix due to stress concentration in this region.

Keywords: Metallic alloys; Fracture mechanics; Electron microscopy; Microstructure; Intermetallic phases

1. Introduction

The 6xxx series aluminium alloys are commonly used for structural engineering applications, in transportation - aircraft and automotive industries and in civil engineering. The main alloying elements in 6xxx series are Si and Mg. These elements are partly dissolved in the primary α-Al matrix, and to a certain extent present in the form of intermetallic phases. A range of different intermetallic phases may form during solidification, depending on alloy composition and solidification condition. Relative volume fraction, chemical composition and morphology of structural constituents exert significant influence on their useful properties [1-5].

The formability, mechanical properties and fracture behaviour of polycrystalline materials e.g. 6xxx alloys are influenced by the presence of intermetallic phases precipitation at the grain boundaries. Researches have shown that as the level of precipitation at the grain boundaries increases, its fracture toughness and formability decreases [5-15]. The medium strength AlMgSi aluminium alloys are commonly processed by extrusion.

Their extradubility depends to a large extent on chemical composition, casting condition and heat treatment parameters (eg. homogenization treatment) which determine the microstructure of the billet before extrusion. The microstructure of this alloys has an effect not only on the extrusion proces but also on the fracture behaviour of AlMgSi alloys. The 6xxx alloys contain a wide range of various intermatallic particles with size typically ranging between 1 to 10 micrometers. Generally, in microstructure of all 6xxx series alloys, two populations of large particles and dispersoids can be observed [5-8]. The most typical are intermetallic compound such as the spherical of α-$Al_{12}(Fe,Mn)_3Si$ and the plate-like β-Al_5FeSi particles. The α-phase (Fe-rich coarse particles) is arranged in interdendritic channels as small dispersoids. The brittle, monoclinic β-Al_5FeSi phase, which is insoluble during solution heat treatment, is associated to reduced workability and cause a poor surface finish [7-12, 16-23].

Data in the literature [7-10] shows that damage during extrusion begins by decohesion or fracture of the inclusions. Therefore, the resistance to fracture and damage may be strongly influenced by the shape, distribution and volume fraction of the second phase particles. The intermetallic precipitates can fracture and also debond from the matrix, and this modes of damage depends on theier size, morphology, etc. Hence, the particles, coarsened one in particular, are assumed to operate as nucleation sites for cracks. The two varieties of particles, α and β, give rise to damage, with different void nucleation modes. Thus, damage evolution is characterized by following sequence: nucleation, growth and coalescence of different populations of voids. It has been found that the strength, strain hardening capacity and strain rate sensitivity of the matrix surrounding the voids have a significant effect on the damage evolution [7-13, 16-28].

The examination of microstructure of fractured samples in the peak aged condition after static tensile tests, crack resistance tests and tensile test in the presence of sharp notch R_m^k allow to define two damage modes. One of them is characterized by nucleation of voids that results from debonding along some particle matrix interfaces and the other one is typical for fracture mechanism initiated by cracking of brittle intermetallic phases.

2. Material and experimental

The chemistry of the investigated commercial 6082 aluminum alloy is shown in Table 1. The alloy was delivered in as-cast condition. The samples for static tensile tests, crack resistance tests and tensile test in the presence of sharp notch R_m^k were machined from the hot extruded profiles (40 mm x 100 mm) subjected to homogenization, solution heat treatment, quenching and artificially aging process. My previous research have focused on studies how artificial aging affected the microstructure**,** mechanical properties and fracture toughness of aluminium alloys of the 6xxx series [2, 3, 12].

Table 1.
Chemical composition of the AlSi1MgMn alloys, %wt

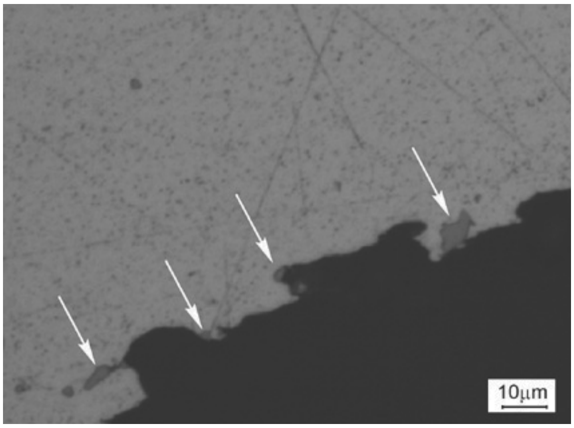
Alloy	Si	Fe	Cu	Mn	Mg	Cr	Zn	Others	Al
6082	1.2	0.33	0.08	0.50	0.78	0.14	0.05	0.15	bal

In order to investigate the fracture mechanisms of 6082 alloy, microstructure and fractografic observation has been carried out on the tensile tested samples in the peak aged condition using optical microscope - Nikon 300 on polished sections etched in Keller's solution (0.5 % HF in 50 ml H_2O). The surface of fracture of the damaged samples were prepared to microscopic examination by scanning electron microscopy HITACHI S-3400 (SEM), operating at 6-10 kV in a conventional back-scattered electron mode. Chemistry of the intermetallics phases was made by EDS technique using the software of Thermo Noran. Discs for thin foils, 3 mm in diameter were punched from the 0.5 mm-thick slice of the sample and then grinded manually to a thickness of about 0.1 mm. The thinned discs were finally electropolished into TEM thin foils employing a standard Struers jet polishing machine, with a solution (by volume) CH_3OH (84 cm^3), HClO4 (3.5 cm^3) and glycerin (12.5 cm^3), operating at -10°C and U=28 V. The thin foils were examined in a JEOL - JEM 2100 ARP TEM/STEM operated at 200 kV electron microscope.

3. Results and discussion

Fracture surface profile of the specimen of AlSi1MgMn alloy in the peak aged condition after static tensile test is present in Figs. 1 and 2.

a)



b)

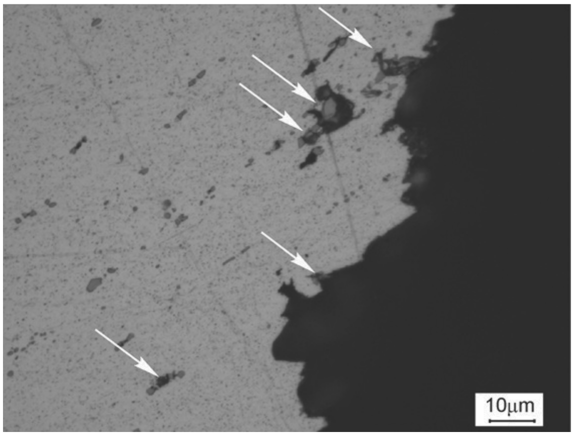


Fig. 1. Fracture surface of the tensile specimen of 6082 alloy (propagation of cracks along interfaces of matrix - precipitates of α and β intermetallic phases is marked with white arrows in the figures (a) and (b)

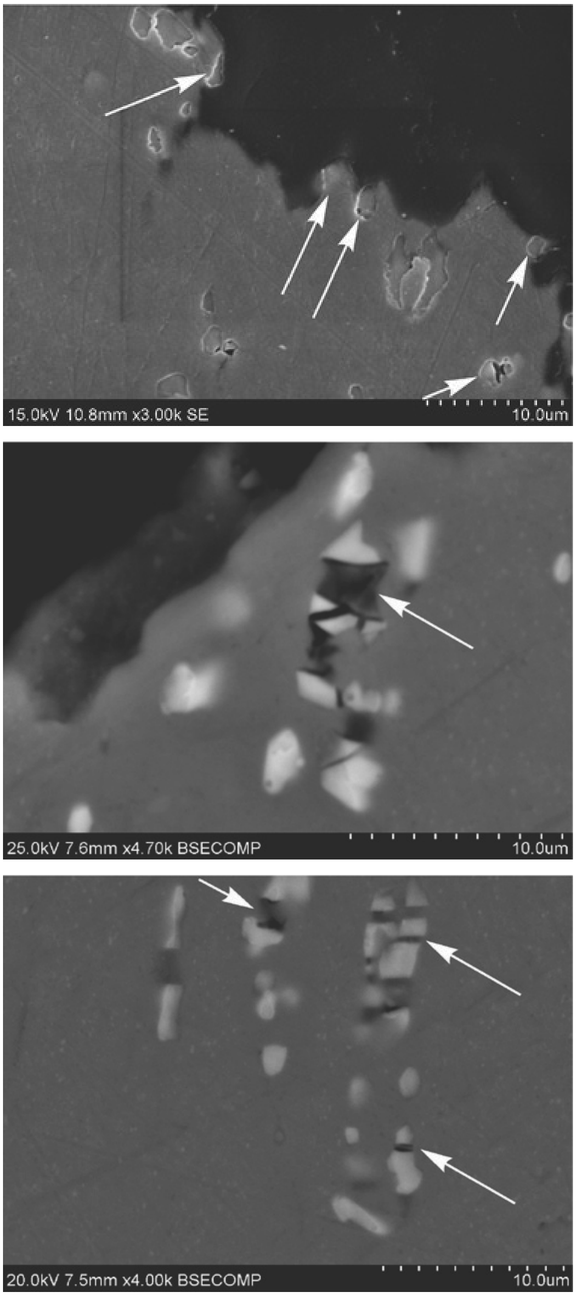
a)
b)
c)


Fig. 2. Fracture surface of the tensile tested specimens of 6082 alloy (a); debonding of the intermetallics precipitates in the matrix (b) and (c)

As can be seen in Fig. 1a formation of cracks is much more favored at the interface between matrix and intermetallics β-Al_5FeSi and precipitates of α-Al(FeMn)Si phases. On the other hand, the microstructure of the tested alloy exhibit the appearance of the hard-brittle intermetallics (β and α) precipitates fractured in the matrix (Figs. 1b, 2b, c). The elongated particles aligned along the main loading direction break into several fragments. The number of broken fragments depends on the particles length.

As the sample is loaded, the stresses on the reinforcement become large. Furthermore, if the reinforcement volume fraction is increased then the stresses within the particle increase more. If the stresses in the particle are large enough, fracture can occur (Fig. 3). Moreover, the likelihood of finding pre-cracked particles rises with increasing volume fraction. Thus, with additional straining the partially broken particles completely fail. As particles become fractured, the stress on the remaining unbroken particles increases, thus accelerating the damage process. Consequently, this can trigger other damage mechanisms such as ductile separation by void growth, thereby leading to premature failure and reduced ductility with respect to the matrix of the material.

Transmission electron microscope examination of microstructure of the aged samples of AlSi1MgMn alloy with the highest tensile strength in the peak aged state [12] (Fig. 4) confirmed results of optical and scanning microscope observation (Figs. 1, 2). One can see in Fig. 4 mechanism of decohesion followed trough nucleation of voids at the interface between matrix and intermetallics α-Al(FeMn)Si phase precipitates.

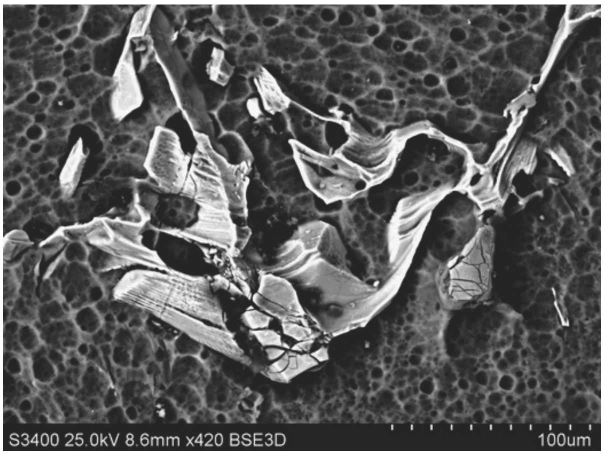


Fig. 3. Partially broken particles observed in the sample of 6082 alloy subjected to the static tensile test

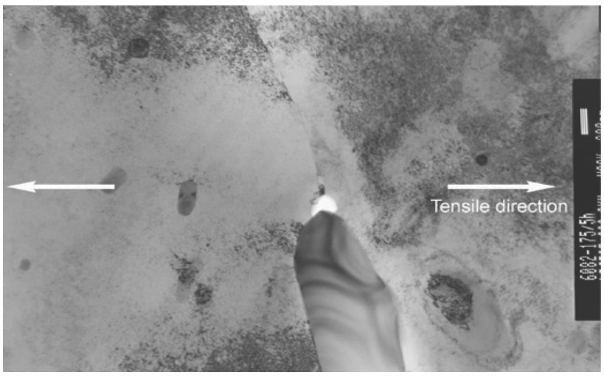


Fig. 4. TEM microstructure of the precipitation hardened 6082 alloy after static tensile test. One of the main source of voids nucleation by decohesion at particle/matrix is presented

a) b)

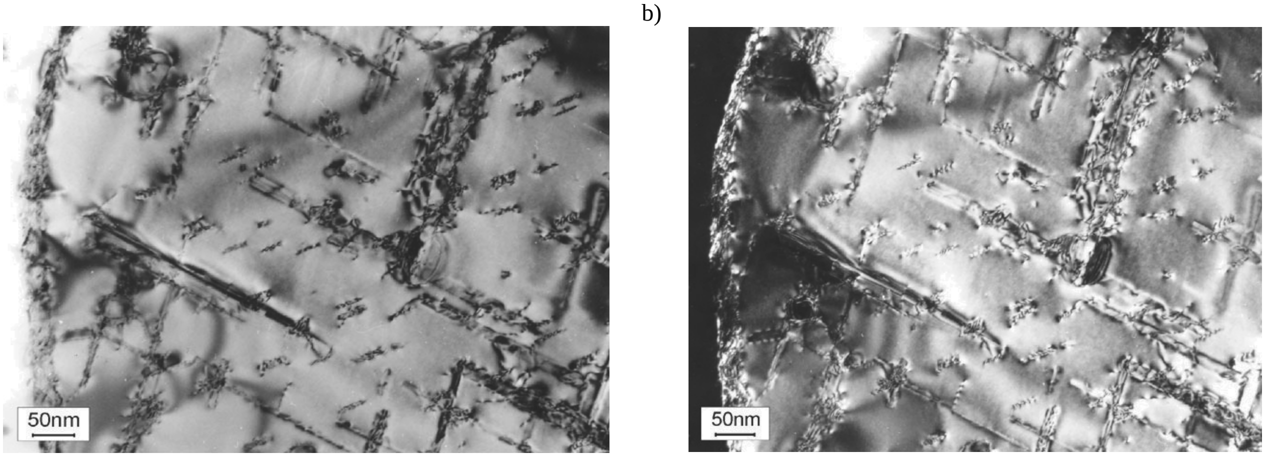


Fig. 5. Bright (a) and dark field (b) micrograph of the post-tested microstructures for the tensile test samples after strengthening process

TEM examination of the post-tested microstructures for the tensile test samples in the highest strengthening state (hardness peak, maximum R_m value) (Fig. 5) revealed that fine, plate-like precipitates of Mg_2Si played significant role in the hardening process. It has been reported that volume fraction of the strengthening phase precipitates in a 6xxx type aluminium alloys resulted in a substantial increase in R_m, $R_{p0.2}$ values. At the same time, the crack toughness of these alloys has been decreasing [1-5, 12, 27]. However, TEM observations showed also that while the fine plate in shape particles of Mg_2Si phase which inhibit dislocation movement and cause an increase of strength, larger particles work as void nucleation sites and so trigger damage (Fig. 4).

SEM observation of fracture processes in the sample with the highest tensile stress in the presence of sharp notch (Fig. 6) confirmed that fracture initiates within void clusters as a result of a sequence of void nucleation, void growth, and void coalescence.

Figure 7 presents the classical ductile profile with existence of dimples with different sizes which can be related to the presence of the two populations of voids. Inside the dimples the presence of different particles are visible. EDS measurements have confirmed that these particles are Mg_2Si, β-Al_5FeSi and α(Al_8Fe_2Si). Large dimples around hard intermetallic α(Al_8Fe_2Si) and β(Al_5FeSi) precipitates and also smaller ones around dispersive hardening β-Mg_2Si and α-Al(FeMn)Si precipitates were formed.

Figure 8 showed the fracture surface of the samples after static tensile tests. In this sample plastic striations were observed on fatigue fracture surface. Fatigue striations are often present in dispersion - hardening aluminium alloys. They are the traces of the crack propagation reflecting the changes in the rate of the main crack propagation. Each fatigue line may be composed of thousand of fatigue striations.

The change of fracture mode as a function of the particle orientation illustrates Figure 9. The results of SEM observation connected with EDS analysis showed the particles of β-Al_5FeSi [5, 6] (Fig. 9a, b) and Mg_2Si [5, 6] (Fig. 10a, b) phases aligned along the main tensile loading direction breaking into several fragments.

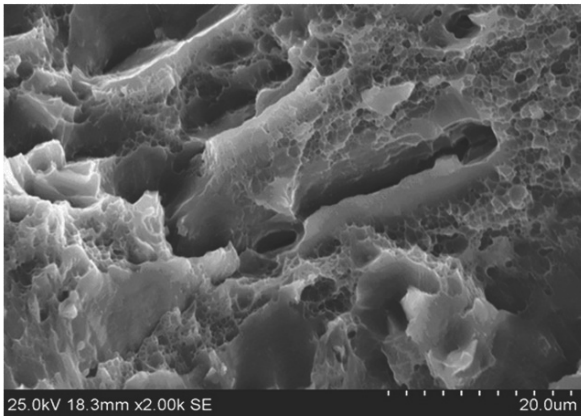


Fig. 6. Fracture surface of the 6082 alloy after static tensile test in the presence of sharp notch R_m^k. Visible shear oval dimples formed after coalescence of the linear void sequence

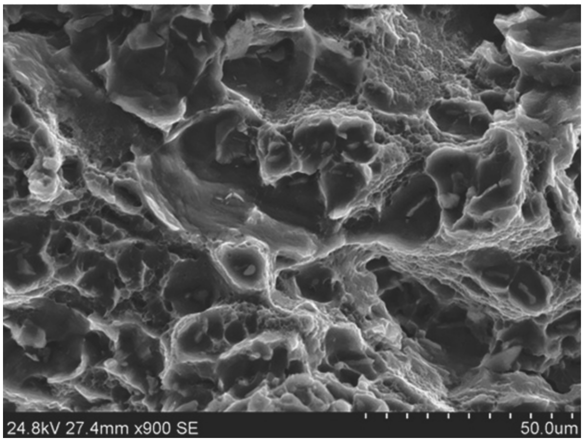


Fig. 7. Fracture surface of the 6082 alloy after static tensile test in the presence of sharp notch R_m^k. Visible large dimples around large intermetallics α(Al_8Fe_2Si) and β(Al_5FeSi) precipitates and smaller around dispersive hardening β-Mg_2Si precipitates

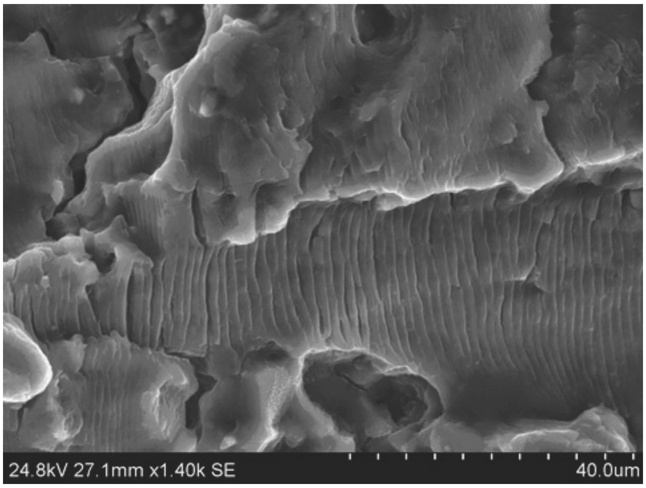
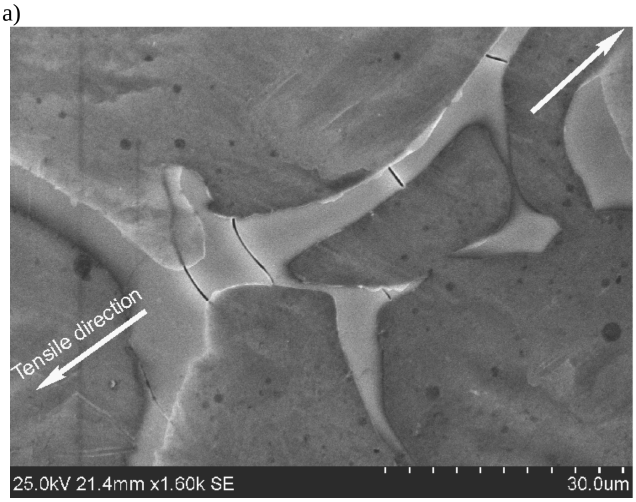


Fig. 8. Morphology of the fracture surface in the deformed 6082 alloy. Typical plastic striations on the fatigue fracture surface are shown

a)



b)

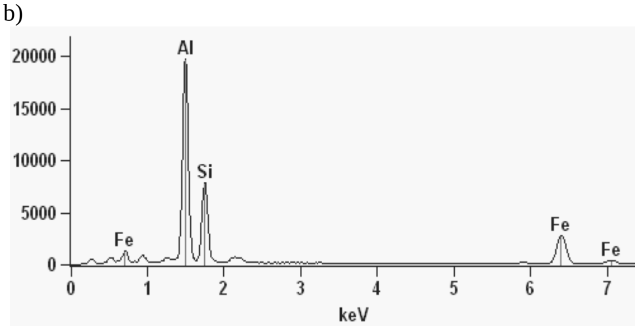
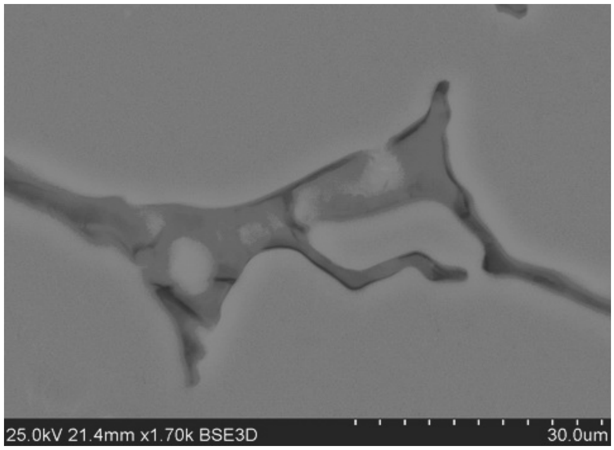


Fig. 9. SEM micrographs of specimens after static tensile test. The main cracks are present in the brittle intermetallic (a) β-Al_5FeSi and (b) EDS spectra of β-Al_5FeSi phases

The cracks in the particles are normal to the macroscopic tension axis. The number of fragments increases with increasing particle length. Figures 9 and 10 show also how the fracture mode depends on the particle orientation. It was observed that the majority of the particles oriented in the range of 0^o to 45^o to the loading direction lead to particle fractures. If the particles were oriented in the range of 45° to 90° fracture occurred on particle / α-aluminium interface decohesion (Figs. 1, 2, 4).

One can see in Figures 11 and 12 the fracture in two-phase region of the investigated alloy subjected to the static tensile test in the presence of sharp notch R^k_m. In this sample, nucleation of voids is heterogeneous and most likely occurs around the cracked α-Al(FeMn)Si particles within deformed matrix band. The edge decohesion is present on the interface between α-aluminium and α-Al(FeMn)Si particles. The numerous cleavage cracks were also visible in the α-Al(FeMn)Si particles. In the microregion of the solid solution. The oval and open shear dimples were revealed within microzones of the solid solution α-Al (Fig. 11a). In the surface of the tested samples the microzones with the fracture of a mixed morphology were discovered. The cleavage facets of α-Al(FeMn)Si precipitates and shear dimples of α-Al matrix were arranged in parallel bands.

a)



b)

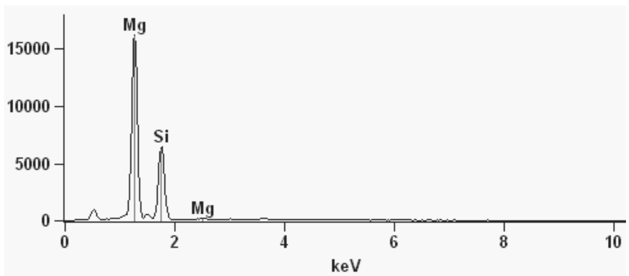


Fig. 10. SEM micrographs of specimens after static tensile test. The main cracks are present in the brittle intermetallic (a) Mg_2Si phases (b) EDS spectra of Mg_2Si phases

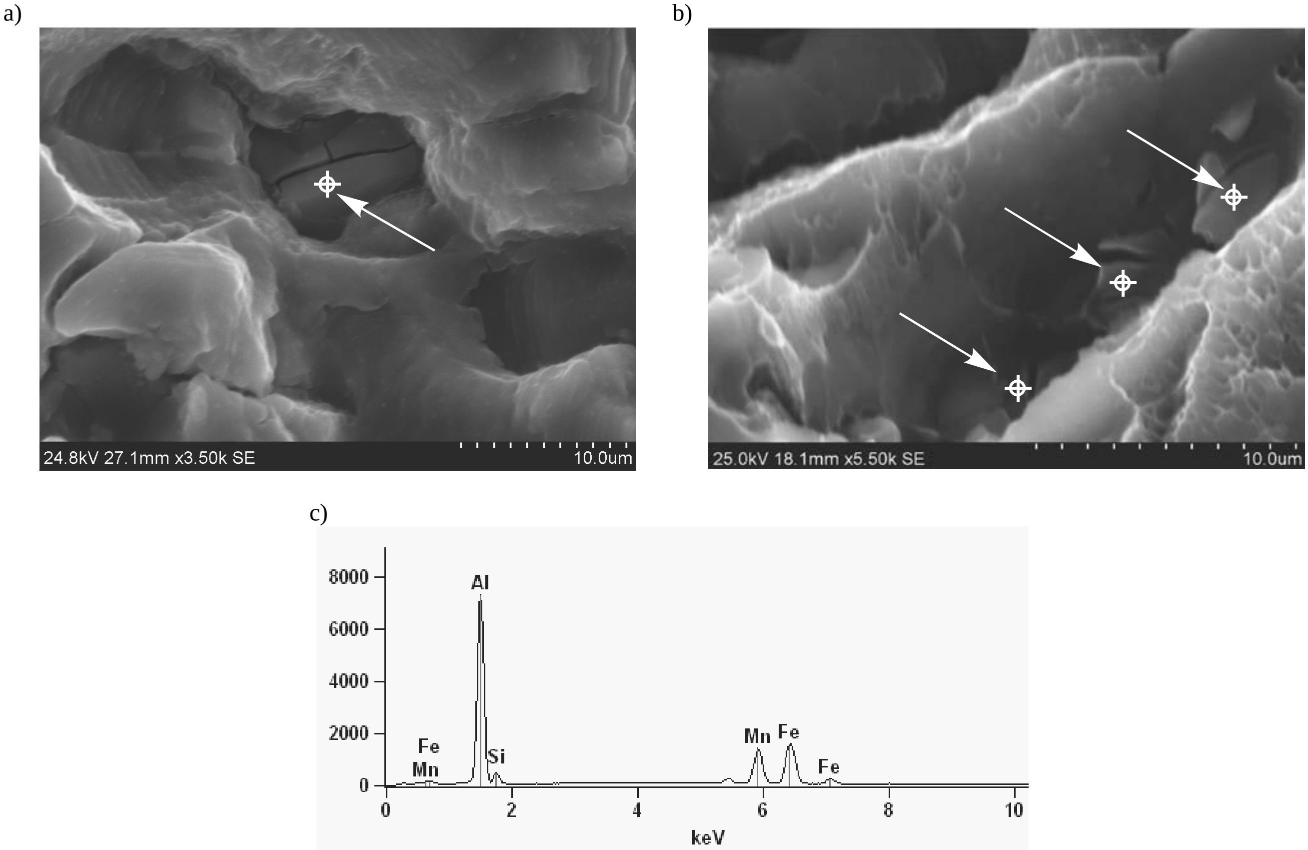


Fig. 11. Fracture in two-phase region of the 6082 alloy after static tensile test in the presence of sharp notch R_m^k (a, b) EDS spectra of cracked α-Al(FeMn)Si particles (points of the EDS analysis are marked with white spots and arrows in the figure- (c)

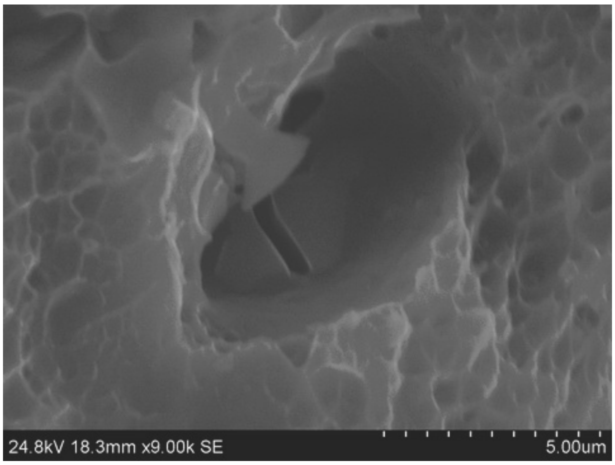


Fig. 12. Fracture in two-phase region of the 6082 alloy after static tensile test - cracked α-Al(FeMn)Si particles.

4. Summary

The fracture behaviour of 6082 alloy at room temperature is mainly influenced by the presence of intermetallics particles along the grain boundaries. Optical and electron (SEM, TEM) microscopic observations of microstructure and fracture of the samples in the peak aged condition after static tensile tests, crack resistance tests and tensile test showed that the presence of the most typical intermetallic phases such as: Mg_2Si, β-Al_5FeSi and α-Al(FeMn)Si had a significant effect on fracture behaviour of the examined alloy. The results confirmed that the nucleation of voids at room temperature in 6082 aluminium alloy occurs either by decohesion or fracture of second phase particles (particle fracture and particle/matrix decohesion) (Fig. 13).

Figure 13 shows how the fracture mode depends on the particle orientation. The majority of the particles oriented in the range of 0^o to 45^o with respect to the loading direction lead to particle fracture whereas the majority of the particles oriented in the range of 45^o to 90^o lead to particle/matrix interface decohesion. The elongated particles aligned along the main loading direction break into several fragments. The cracks in the particles are normal to the macroscopic tension axis. The number of fragments increases with increasing particle length.

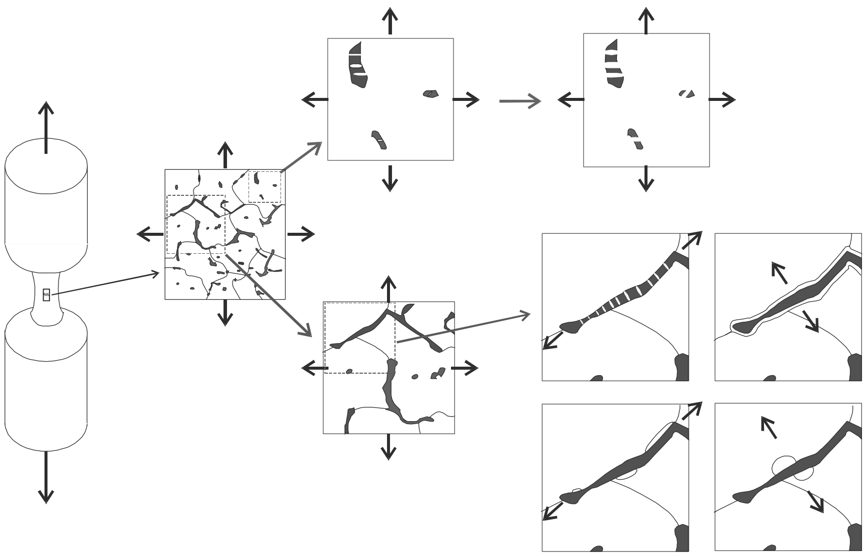


Fig. 13. Sequence of damage mechanism in the 6082 alloy

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